Mechanical Properties of Metals at Low Temperatures

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Mechanical Properties of Metals at Low Temperatures

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Foreword

A knowledge of the behavior of metals at low temperatures is of importance in understanding their fundamental rheological properties. Many mechanical devices and equipment are required to operate at low temperatures. Embrittlement of materials and accompanying brittle failures at low temperatures is of concern to designers, manufacturers, and users. Moreover, the importance of the subject has grown as the applications of metals at low temperatures have increased with the steady growth of the refrigeration industry and the expanding demand for the liquefaction, transportation, and storage of many gases.

The papers in this volume present some of the latest results of studies conducted both in industry and government. They were initially presented at the Symposium on the Influence of Low Temperatures on the Mechanical Properties of Metals, the second of a series of twelve symposia held by the National Bureau of Standards during its Semicentennial in 1951. The cooperation of the Office of Naval Research in making possible the symposia series is gratefully acknowledged. The Symposium on the Influence of Low Temperatures on the Mechanical properties of Metals was planned and conducted by the Bureau's Metallurgy Division—in particular, T. G. Digges and G. A. Ellinger, co-chairmen of the symposium and G. W. Geil, who arranged the program and prepared this volume.

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1. Recent European Work on the Mechanical Properties of Metals at Low Temperatures

By N. P. Allen[1]

A brief account is given of work in Europe since 1940 on the mechanical properties of metals at low temperatures. Elasticity and plasticity are dealt with, and the development of brittleness at low temperatures is discussed in relation to brittle fracture in mild steel, the behavior of heat-treated alloy steels at temperatures prevailing in the stratosphere, and the reliability of austenitic steels at liquid-air temperature. The properties at low temperatures of pure alloys of iron made at the National Physical Laboratory are described, and it is shown that some alloying additions exert influences on the rupture strength that are independent of their influence on the resistance to plastic deformation.

Introduction

The growth of interest in the mechanical properties of metals at low temperatures is rather recent, and has been stimulated principally by the advent of stratosphere flying, and by the rapid development of chemical processes at very low temperatures. Laboratories devoted to systematic experimental work at the lowest temperatures have generally been concerned principally with the thermal, electric, and magnetic properties of metals, and the problem of superconductivity has attracted much attention, without a complete explanation having yet been found. During the past 10 years, most European laboratories have experienced a greater or less disruption of their work. It is not surprising, therefore, that little investigation of the theory of the mechanical properties of metals at low temperatures has yet been carried out. Most of the experimental work carried out in Europe has been done to find out how the common metals and alloys of engineering are likely to behave when the temperature is lowered. Researches of the same kind have been made in America, and the conclusions are generally similar. In this review of European work on the mechanical properties of metals at low temperatures simple exploratory investigations are avoided in favor of investigations concerned with general ideas. The European reactions to some important practical problems are also described. It is impossible to be comprehensive. The choice of material is a personal one, essentially of those investigations that have most interested the author.

Elastic Properties of Metals

There is at present a revival of interest in the elastic properties of metals, partly because they provide the most direct approach to the study of the cohesive forces within the crystals, partly because the

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development of supersonic techniques has made it possible to measure elastic constants with accuracy, a precision within 0.1 percent being now common. Some 15 years ago it was found possible to calculate the elastic properties of monovalent metals by the methods of wave mechanics. The constants obtained applied to 0° K, and theory suggests that the constants should fall continuously as the temperature rises. Although work is in progress, calculations have not yet been completed for other metals on account of the mathematical difficulties of the problem, and the effect of temperature has not been derived. There has, however, been an interesting attempt by Born and his collaborators at Edinburgh, notably Furth [1, 2, 3, 4] [2] to arrive at a complete expression of the elastic behavior of cubic crystals on the basis of the assumptions that the energy of the bond between adjacent atoms is given by

$$\theta = -ar^{-m} + br^{-n}, \tag{1}$$

where r is the distance between atoms, and a, b, m, n are constants, that the forces act directly between the centers of the atoms, and that the forces between nearest neighbors alone are important. These assumptions are of course not quite correct as they lead to relations between the elastic constants that are not observed, and the metallic bonding forces are known not all to be central. However, the departure from truth is thought not to be large, and the theory serves as a starting point for investigation. Figure 1.1 shows how the elastic constants would be expected to vary with temperature in face-centered cubic lattices having the values $m=4$, $n=8$, which are two not unlikely figures. The general behavior is the same for other probable values of m and n, and Furth [4] has described methods of arriving at the values which should be chosen in any particular case. In these figures, θ is a fixed temperature, and p_0 is a fixed pressure, each characteristic of the lattice—their values are dependent on the values of the constants in

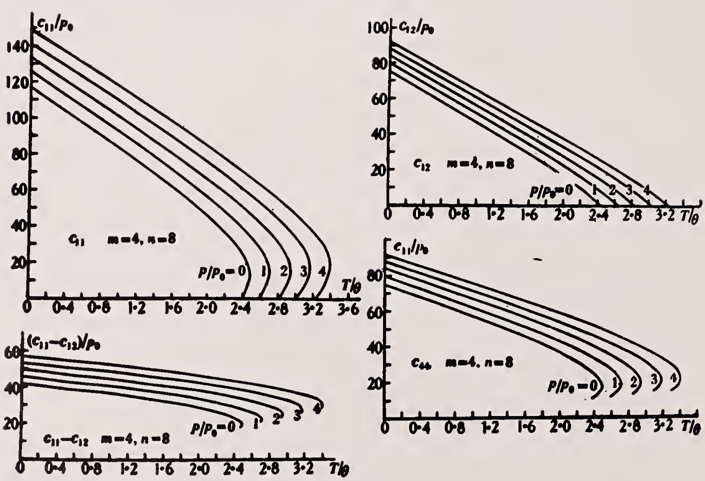

FIGURE 1.1. *Elastic constants for $m=4$, $n=8$.*

[2] Figures in brackets indicate the literature references on p. 35.

eq. (1). It is seen that all the elastic constants decrease at first almost linearly with increasing temperature, and that the rate of fall increases gradually as the temperature rises. The curve for the modulus of rigidity (C_{44}) changes direction at a certain temperature, which indicates that the lattice at this point would become unstable, but this occurs at a temperature well above the melting point of the material.

Recently, Köster [5] published a very comprehensive series of measurements of the temperature variation of Young's modulus for the pure metals. Some of the results are given in figures 1.2 and 1.3, and, for the sake of comparison with the theoretical curves given in figure 1.1, a few of them are replotted in figure 1.4, adjusting the scale of elasticity so that all the curves start from the same point at 0° K and using for θ the melting point of the metal. Some of the metals follow very closely the expected behavior, but with others, the modulus of elasticity falls rapidly from a given temperature. When this occurs the change is accompanied by an increase of damping capacity, in-

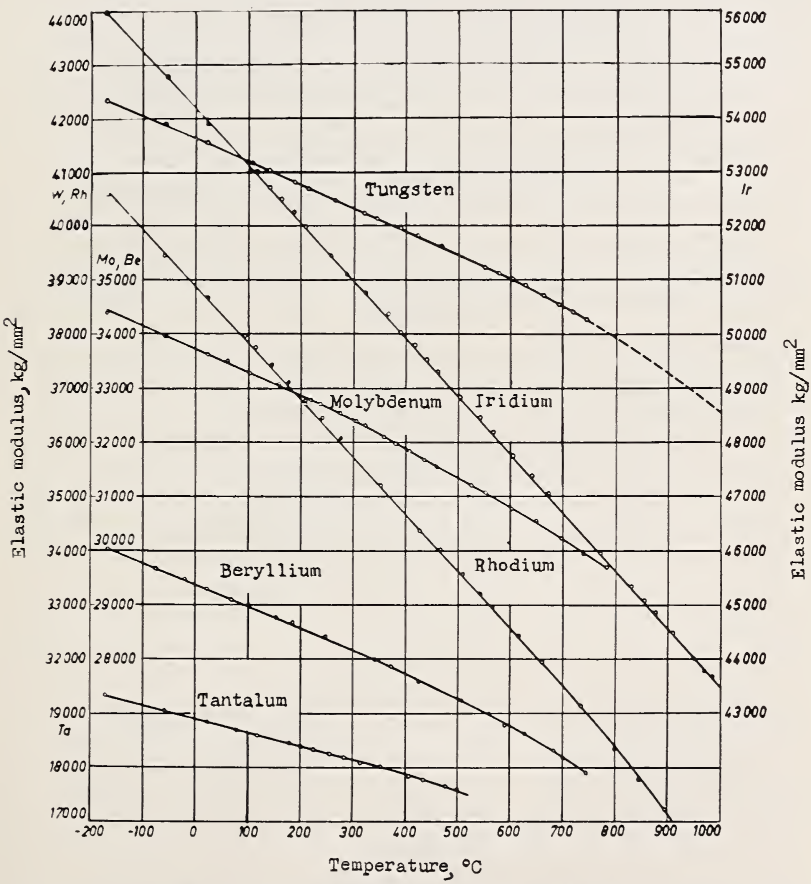

FIGURE 1.2. *Variation with temperature of elastic modulus for various metals.* (Köster)

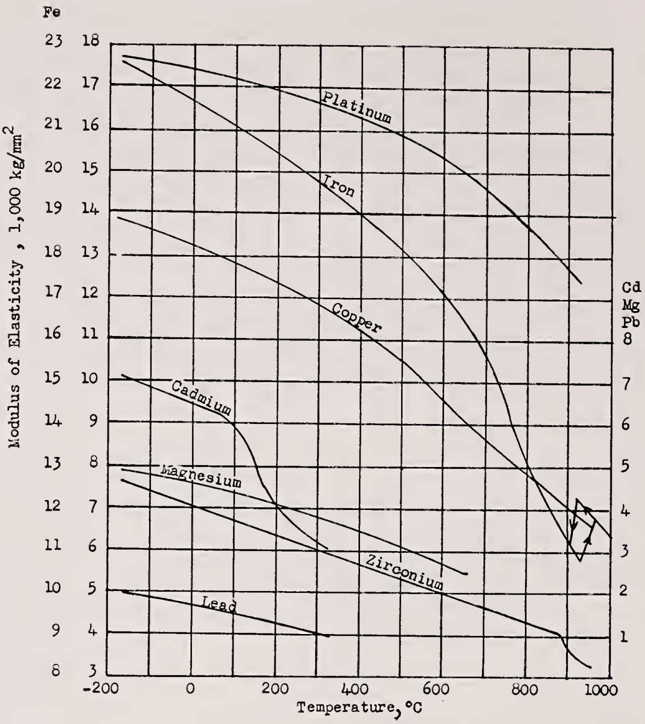

FIGURE 1.3. *Modulus of elasticity plotted against temperature far various metals.*
(Köster)

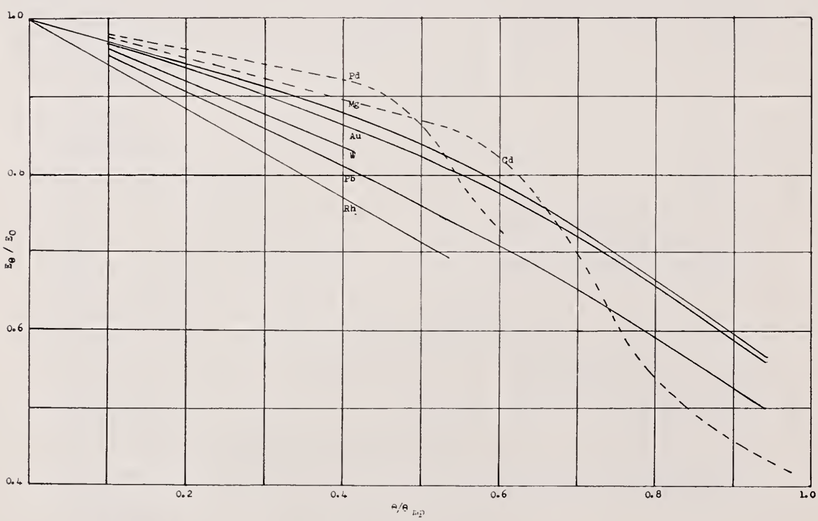

FIGURE 1.4. *Fall of Young's modulus with rise of temperature.*
$E\theta$ is Young's modulus at θ°K; Eo is Young's modulus at 0° K; θ_{mp} is melting point of metal.
(Köster)

dicating that some energy-absorbing process is affecting the measurements. These processes seem to be important only above room temperature. Figure 1.5 shows that in practically all metals the temperature coefficient of elasticity has settled down to a constant value at 0° C and below. The same is not necessarily true in alloys, for there are well-established examples of irregular variation of the elastic constants of alloys at low temperatures [6]. Figure 1.3 shows cases in which the value of the elastic constant is affected by allotropic or magnetic changes in the metal. They are included to show that the appearance of ferromagnetism affects the elastic constants, and that in those alloys in which ferromagnetic changes occur at low temperature, or which contain carbides with low Curie points, corresponding changes of elastic constants are to be expected. It may be noted that Köster's data extend to −183° C and include only two temperatures of measurement below 0° C. There may well be more detail to be revealed by more complete experiments. Furthermore, as the changes of elastic constants are an expression of the changes of the energy of the crystal, it may be expected that the elastic constants will change very little when the specific heat approaches zero at very low temperatures, but no observations at such low temperatures appear to be available.

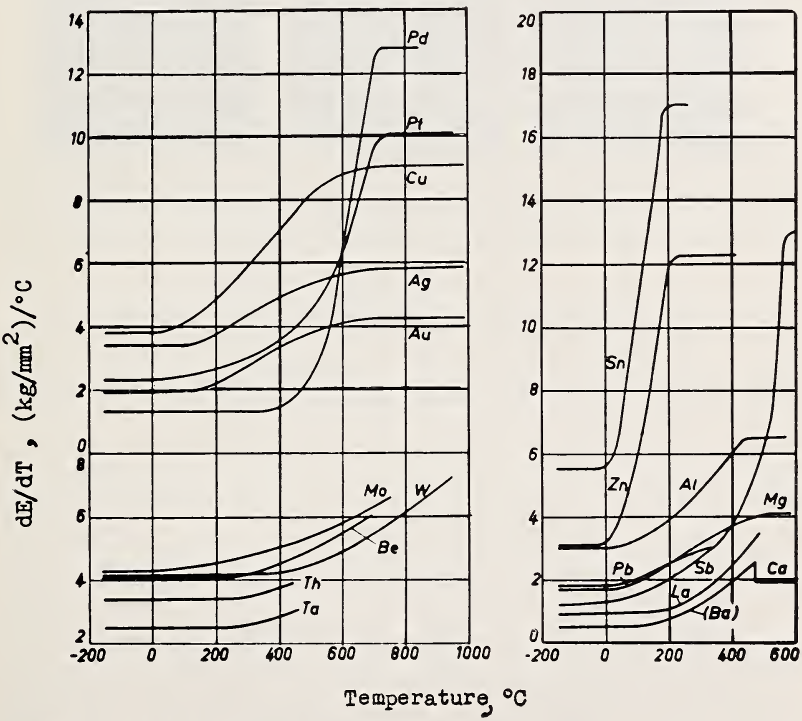

FIGURE 1.5. *Temperature coefficient of elasticity plotted against temperature for various metals.*

(Köster)

Without exception, lowering the temperature increases the resistance of metals to plastic deformation; but the increase is not large, and there is nothing to suggest that plastic deformation is impossible at the absolute zero of temperature. At 20° K the critical shear stress of a cadmium single crystal is only four times its critical shear stress just below the melting point. The crystallographic mechanism of slip is considerably modified by lowering the temperature, and there are two types of change (1) a change in the mode of slip, and (2) a change in the degree of disruption of the crystal structure accompanying slip, associated with a change in the work-hardening capacity of the material. Both changes are illustrated in a study by Andrade and Chow [7] of the influence of temperature on the deformation of crystals of body-centered cubic metals. It was found for these metals that whereas the slip was always in the octahedral direction, the plane of slip altered systematically with temperature, being the (112) plane at the lowest temperatures, (110) at intermediate temperatures, and

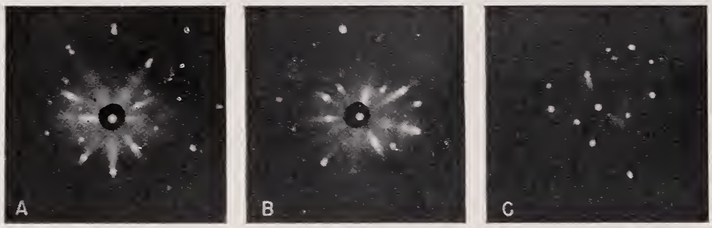

FIGURE 1.6. *Laue pictures showing asterism is greater at low temperatures for equal elongation in single crystals of sodium*

A: (−185° C) e, 2 percent, $\Delta\theta$, 2.2°; B: (−82° C) e, 2 percent, $\Delta\theta$, 1.8°· C: (+20° C) e, 1.9 percent, $\Delta\theta$, 1.6°.
(e, Elongation; $\Delta\theta$, rotation of crystal fragments.)
(Andrade and Chow)

(123) at high temperatures. This sequence is illustrated in table 1.1, which is set out in such a way as to demonstrate that the choice of slip plane is decided by the ratio of the testing temperature to the melting point of the metal, both measured on the absolute scale. No explanation has yet been given of this relationship, and it does not apply to iron. Andrade and Chow observed that the glide planes were more closely spaced at the lower testing temperatures, and also that the amount of asterism produced by a given amount of plastic deformation was greater at low temperatures (fig. 1.6). They attached great importance to the latter observation: the asterism measures the amount of rotation of the crystal fragments in the neighborhood of the glide planes, and they were able to show that the amount of strain hardening in the case of sodium was proportional to the amount of crystal rotation (fig. 1.7). Their emphasis upon the importance of fragmentation and disorientation in relation to work-hardening capacity at low temperatures has been justified by later work on the changes taking place during deformation at high temperatures, for it has been shown that deformations at high temperatures, particularly if slow, are characterised by slight fragmentation, and by little disorientation of the structure [8].

TABLE 1.1. *Glide planes in body-centered cubic metals*

(Andrade and Chow)

θ=temperature of test, °K/melting point of metal, °K.

Test temperature	θ	Metal	Glide plane
° C			
20	0.08	Tungsten	(112)
20	.10	Molybdenum	(112)
300	.20	do	(112)
−185	.24	Sodium	(112)
20	.26	β brass	(110)
1,000	.40	Molybdenum	(110)
−82	.50	Sodium	(110)
20	.80	do	(123)
20	.87	Potassium	(123)

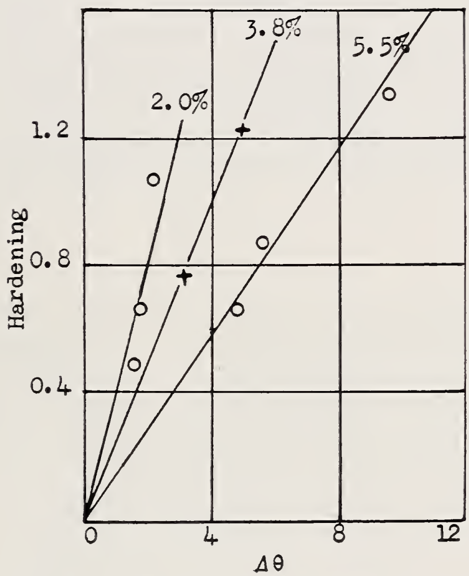

FIGURE. 1.7. *Hardening against Δθ (rotation of crystal fragments) at constant extension.*

More recently Brown [9] has studied with the electron microscope, using an oxide-replica technique, the effect of temperature on the form of the slip bands in aluminum. The microscopically visible slip zones were found to consist of clusters of parallel slip lamellas, as had previously been found by Heidenreich and Shockley [10], and the final distribution of lamellas could be characterized by (1) the mean distance between slip zones, (2) the number of slip lamellas in each zone, (3) the mean distance between the lamellas in the zone. The first stage of deformation was much the same at all temperatures and consisted of the formation of widely spaced zones of simple slip fig. 1.8, a). Further deformation at low temperatures caused the production of more zones of simple slip at closer spacings (fig. 1.8, b) until a minimum spacing was reached, after which additional slip lamellas might appear in each zone. At high temperatures, on the other hand, new zones did not appear so readily—the minimum spacing was wider and further slip took place by the production of many new slip lamellas in each zone, producing the appearance that is represented diagrammatically in figure 1.8, c. Table 1.2 illustrates the

TABLE 1.2. *Spacing and nature of slip zones in aluminium*

(A. F. Brown)

Temperature of straining	Spacing of slip zones	Number of lines in each zone
° *C.*	μ	
−253	0.1	1
−180	0.5 to 1	1 to 2
+20	2	3 to 4
250	4	5 to 6
500	10	~12

behavior of aluminum deformed by 15 percent at various temperatures, and figures 1.9 and 1.10 contrast the appearance of slip zones formed at —180° and +500° C, respectively.

In figures 1.9 and 1.10 the traces of slip zones across cubical etch pits are shown. These enabled the directions and magnitudes of the displacements to be measured. Curiously enough the magnitude of the displacement in each slip lamella was practically independent of temperature—the distance between adjacent lamellas in one slip zone was also independent of temperature. It is not easy to find explanations of this behavior, and perhaps not wise to try to do so until it is known whether the behavior is peculiar to aluminum, or common to a number of metals, but the behavior does suggest that, in agreement with the observations of Andrade and Chow, the disturbance in the neighborhood of a single slip at low temperatures is so violent that no other slip can readily occur in that region, whereas at high temperatures, the disturbance is less, and conditions in the vicinity of an existing slip lamella are rather in favor of further slip in that region than the reverse.

According to more recent results the hardening associated with a single slip zone is much the same whatever the temperature, and the

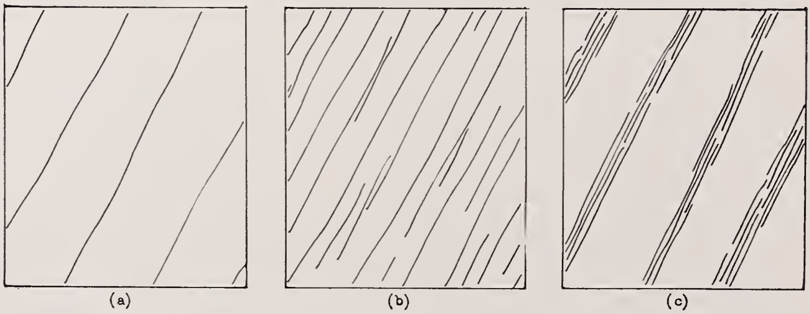

FIGURE 1.8. *Effect of temperature on distribution of slip bands in aluminum (schematic)*

a, Early stage of deformation (high and low temperature); b, later stage of deformation (low temperature); c, later stage of deformation (high temperature).

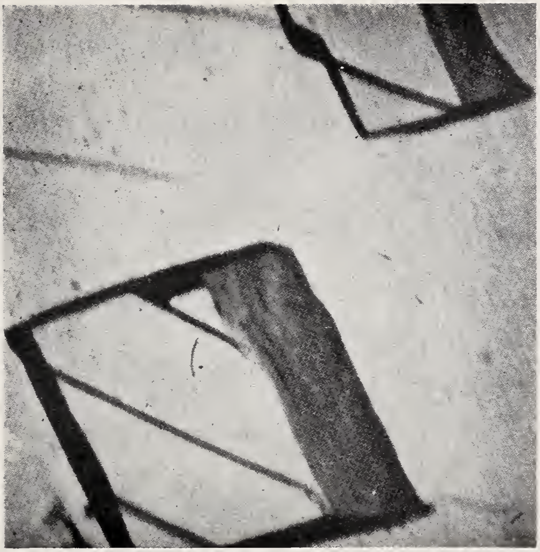

FIGURE 1.9. *Slip zones in aluminum at −180° C, ×10,000.*
(A. F. Brown)

FIGURE 1.10. *Slip zones in aluminum at 500° C, ×25,000.*
(A. F. Brown)

greater strain hardening that occurs at low temperatures is due to the more numerous slip zones that are formed when the temperature is low.

There have been few measurements of the critical shear stresses of single crystals, and of the effect of low temperatures upon them, but Druyvesteyn [11] has measured the rise of proof stress of many polycrystalline pure metals on cooling to −183° C. He found the radio, proof stress at −183° C/proof stress at 20° C to be more reproducible, and more significant, than the actual values of proof stress and has discussed his results in terms of this ratio. There proved to be a strong relation between the type of lattice and the value of the ratio. It was small, below 1.75, for all face-centered cubic metals, and for those hexagonal metals whose structures approximate to the hexagonal close packing of spheres, but it was large for body-centered cubic metals, tetragonal metals, and hexagonal metals whose structures

Table 1.3. *Yield points of pure metals at —183° C*

(Druyvesteyn [11])

Bending tests: T_a is the annealing temperature; N, the number of grains per square millimeter; P_{20} and P_{-183}, the yield point in grams at +20° C and −183° C; and E is the modulus of elasticity

Element	T_a	N	P_{20}	P_{-183}	$\xi=\frac{P_{-183}}{P_{20}}$	$\frac{E_{-183}}{E_{20}}$
	°C					
Nickel	350	--------	2,320	2,740	1.18	1.10
Do	450	3,000	575	700	1.21	--------
Do	550	2,000	400	495	1.23	--------
Do	650	1,500	330	400	1.21	--------
Copper	500	500	140	169	1.21	1.07
Do	700	300	102	122	1.20	--------
Do	900	200	94	117	1.24	--------
Silver	800	500	409	502	1.23	1.10
Aluminum	250	--------	460	537	1.17	1.10
Do	350	100	105	134	1.27	--------
Do	400	20	56	68	1.22	--------
Do	600	2	47	60	1.28	--------
Lead	100	83	--------	--------	1.73	1.18
Do	150	4	26.6	39	1.46	--------
Tantalum	900	10,000	1,630	4,010	2.5	--------
Molybdenum	1,000	600	2,520	>4,500	(3.0)	--------
Iron	620	4,400	1,080	>3,520	(3.9)	1.05
Do	850	30	245	2,020	8.3	--------
Zirconium	600	--------	603	977	1.62	--------
Titanium	700	--------	1,180	2,450	2.1	--------
Do	700	--------	2,120	3,120	1.5	--------
Cobalt	[a]1,100 to 400	300	1,390	1,370	0.99	--------
Magnesium	500	50	122	162	1.33	1.08
Do	600	8	86	121	1.41	--------
Zinc	150	250	146	394	2.70	1.29
Cadmium	200	20	50	136	2.72	--------
Thallium	150	100	23	48	2.1	--------
Tin	150	60	68	313	4.6	1.27
Do	200	1	53	166	3.1	--------

[a] 1 hr at 1,100° C; 5 hr at 400° C.

depart markedly from spherical close packing. When the proportionate changes of proof stress were plotted against a reduced temperature defined by the ratio of the absolute temperature of test to the absolute melting temperature of the metal, the curves for the face-centered cubic metals fell close together, and the curves for tin, cadmium, and zinc fell well above them. These points are illustrated in table 1.3 and figure 1.11. The temperature coefficient of the proof stress was always higher than the temperature coefficient of elasticity, indicating that all the lattices would withstand more elastic deforma-

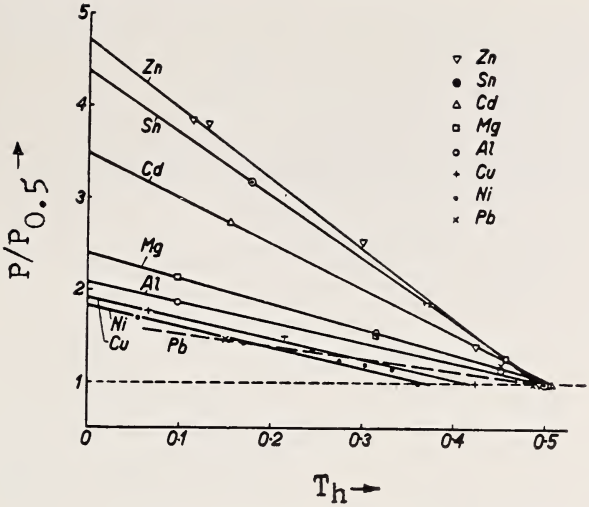

FIGURE 1.11. *Proportionate changes of yield point ($P/P_{0.5}$) plotted against reduced temperature (T_h) for pure metals.*

p is yield point of metal; $P_{0.5}$ is yield point when T_h is 0.5; T_h is temperature of test/melting point of metal (absolute temperature).

tion before yielding when cold. Druyvesteyn also measured the change in elongation before fracture of the samples, and found that those metals having a small increase of proof stress on cooling increased in elongation in liquid air, but that those metals that experienced a large increase in proof stress were embrittled. These results suggest rather strongly that the rise of proof stress on cooling is the predominant factor in embrittlement at low temperatures and that the effect of temperature on fracture stress is not great and is much the same for all metals.

The relationship between the proportionate rise in proof stress and the tendency to embrittlement has also been commented upon by McAdam and his colleagues. While it is probably valid as a broad generalization concerning the behavior of entirely distinct metals, it is of no value as a guide to the behavior of alloys of one metal. In the course of work at the National Physical Laboratory it has repeatedly been observed that there is no relationship of this kind between the yield point and the ductility of alloys of iron at the temperature of liquid air. One example will suffice and is given in table 1.4. When manganese is progressively added to a pure iron alloy containing 0.05 percent of carbon, the proof stress rises at 20° C and a little at −196° C, but the ratio of proof stress at −196° C to proof stress at +20° C does not alter appreciably up to 1 percent of manganese. The reduction of area at −196° C, however, rises markedly with increase of the manganese content. Between 1 and 2 percent of manganese the proof stress at 20° C rises sharply, and that at −196° C does not change. The ratio of the two proof stresses therefore alters, but there is little corresponding change in the reduction of area at −196° C. Observations such as this lead to the suggestion that the

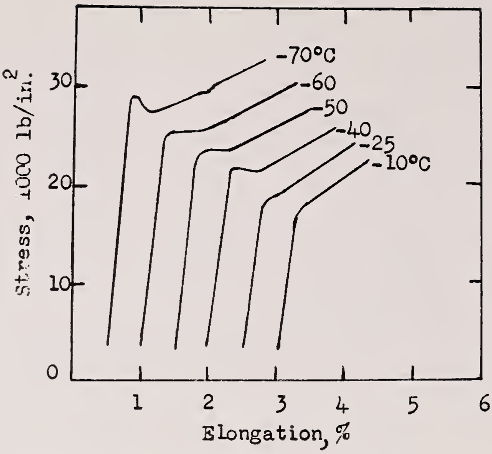

FIGURE 1.12. *Effect of temperature of test on yield point of Swedish iron containing 0.016 percent of carbon, annealed at 450° C and quenched.*

existence of important variations in the fracture stresses of alloys must be admitted.

In the course of this work it has also been observed that temperature has a noticeable effect on the form of the load-elongation curve at the yield point. The sharp yield point of mild steel is associated with the presence of carbon, and disappears if the carbon is less than some rather poorly defined amount, approximately 0.003 percent, or, alternatively, is bound up in some stable form, such as titanium or vanadium carbide. It has been observed that in irons containing small amounts of carbon the yield point commonly does not appear in tests at 20° C, but is very distinct in tests at lower temperatures. Figure 1.12 gives an example of the progressive development of the yield point as the temperature is lowered from −10° to −70° C. In this case the material contains 0.016 percent of carbon, and the yield point, which would ordinarily be present, has been suppressed at room temperature by annealing the iron at 450° C and quenching. The results show that the absence of a yield point at room temperature is no proof of the absence of carbon from the iron. The yield point sometimes becomes less distinct at still lower temperatures, and twinning may become more marked. The British Non-Ferrous Metals Research Association is studying the yield behavior of aluminum

TABLE 1.4. *Properties of normalized iron-manganese alloys containing 0.05 percent of carbon*

[National Physical Laboratory]

Manganese	Proof stress at 20° C	Proof stress at −196° C	Reduction of area at −196° C
Percent	*1,000 lb/in.²*	*1,000 lb/in.²*	*Percent*
0	24.8	98.8	0.3
0.5	23.0	102.0	16
1.0	27.3	104.0	45
2.0	34.7	103.0	55

alloys at various temperatures and in some respects the results resemble those described above for steel. It will be particularly interesting to see, as the work proceeds, whether after making due allowance for difference in melting points the two materials behave similarly.

Theoretical treatments of the variation of the yield stress with temperature have hitherto generally made use of the Becker-Orowan equation

equation (2)

$$R=k \exp \{(\sigma_0-n\sigma)^2 V/2kGT\},$$

where

R = rate of flow
G = shear modulus
σ_0 = yield stress at 0° K
σ = applied stress
V = volume of order of 200 A^3
n = stress concentration factor,

according to which, assuming that yield occurs at a stress resulting in a constant observable rate of flow, the yield stress decreases with increasing temperature from 0° K in a parabolic manner. The equation assumes that the difference between the stress ($n\sigma$) operating at any point where slip occurs and the constant critical stress σ_0 necessary for slip to take place is made up by random thermal fluctuations. Recently, there has been dissatisfaction with this equation, since, on account of the application of the theory of dislocations, it has been increasingly realized that both the critical stress (σ_0) necessary to initiate slip and the stress concentration factor (n) vary from point to point in the metallic crystal, so that it is necessary to consider, instead of one "activation energy of flow," a complex spectrum of activation energies. The form of the relation to be expected between rate of flow and temperature is naturally dependent upon the distribution of activation energies within this spectrum. It is also realized that the general nature of the relation between stress and rate of flow over a wide range of temperature, including both very low and very high temperatures, is such as to be inconsistent with the stresses σ_0 required to initiate slip being independent of temperature [12]. The shear modulus (G) is also subject to important variations. One of the important problems at the present time therefore is to find out how the critical stresses and the associated energies should vary with temperature, but this problem has not yet been attacked.

Brittleness and Fracture

Most of the recent researches on the mechanical properties of metals at low temperatures have been concerned with the problems of brittleness and fracture. The elastic and plastic properties of all metals improve on cooling, but some become brittle, and it is necessary to ensure that such materials as are put into use are employed under conditions in which brittleness cannot arise. There are three main problems: first, can the ordinary mild steel used as the basis of general

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structural engineering be relied upon when the atmospheric temperature falls, and when the steel is employed in large, highly stressed, and rigid structures that may contain great quantities of elastically stored energy; second, how will the heat-treated alloy steel that is used when stresses are high and dynamic loading is severe behave in airplanes flying in the stratosphere or in vehicles subjected to extreme arctic conditions; third, what materials are to be used for chemical engineering at a below liquid air temperatures; and since the austenitic stainless steels seem to have the required general properties, are they to be relied upon under all conditions that may arise in fabrication and in use? It is convenient to deal with the subject under these three heads.

Brittle Fracture of Mild Steel

Most steel-producing countries have taken note of the occasional catastrophic failures of large welded mild-steel structures, and have had to consider the degree of security against this type of failure provided by their ordinary products. The question has generally led to controversy about the test to be applied, and several committees have been formed to examine the question. It seems to be agreed that it is security against cleavage fracture without prior deformation that is primarily required, since this occurs without appreciable absorption of energy, and can take place with disastrous speed under the influence of the stored elastic energy of the structure. Cleavage fracture may occur after considerable plastic deformation has taken place, but this is a far less dangerous phenomenon. As cleavage fractures are occasionally taken as evidence of a dangerous condition, it is important to emphasize that cleavage fractures have frequently been observed in specimens that have extended as much as 40 percent of their length before fracture. Every steel may be supposed to have a flow limit (which is the maximum shear stress that can be withstood without plastic deformation taking place) and a rupture stress, at which brittle fracture occurs. Under conditions of simple tension the rupture stress is generally higher than the flow limit and plastic deformation precedes rupture. If by any means, such as triaxial stressing, rapid loading or change of temperature the flow limit can be raised above the rupture stress, brittle fracture occurs. This is the original conception of Ludwig, and is used for simplicity—the fact that the value of the rupture stress is affected by many variables, including the triaxiality of the stress system, may for the present be neglected. Steel behaves as if the flow stress rises more rapidly with fall of temperature than the rupture stress (fig. 1.13) and a temperature is ultimately reached at which brittle fracture occurs. This condition is reached at a higher temperature when the test is one in which a triaxial stress is produced, for example, by a notch. Controversy arises between those who believe that it is sufficient to adopt a test in which a certain degree of triaxiality is obtained, and measure simply the temperature at which the steel exhibits a brittle behavior, and those who, more practically, and much more inconveniently, insist that the steel should be tested for resistance to the conditions at which it will be used, and therefore think in terms of large test pieces, extremely sharp notches, and conditions of test promoting the rapid spread of the crack.

In the Netherlands, the Schnadt test [13], in which the steel is broken at room temperature with notches of three degrees of severity—the sharpest notch being obtained by pressing a knife edge into the steel—has gained some popularity. There have also been suggestions that a fatigue crack or stress-corrosion crack should be used as the initial notch. In regard to these questions much depends on whether there is or is not a limit to the completeness of triaxiality of stress that can be induced at the tip of an infinitely sharp notch. If, as is suggested by the theory of elasticity applied to a perfectly elastic material, there is no limit, every material should be capable of being broken in a brittle manner if a sufficiently sharp notch can be produced. All materials will then fail in a brittle manner once a crack has been started, and it will be of primary importance to study the conditions under which a crack forms. If, on the other hand, there is a limit, those materials for which the ratio of flow stress to fracture stress exceeds a value depending on the limiting triaxiality ratio (i. e., for which flow stress/fracture stress$>(S_1-S_3)/S_1$) will propagate cracks in a brittle manner while the propagation of cracks in other materials will be accompanied by plastic deformation. Testing will then need to be designed to discover whether the ratio of flow stress to fracture stress is above or below the critical figure. Orowan [14] has put forward reasons for thinking that in plastic materials the flow stress at the root of a notch can never be raised to more than about three times its value under simple tension, and Mott [15], after considering the stresses existing in a lattice of atoms at the end of a crack of atomic dimensions, has concluded that in every case the critical shear stress should be exceeded long before the rupture stress is reached, and that therefore brittle rupture should only occur under conditions such that the spread of the crack is so rapid that plastic deformation cannot take place.

In this connection, some experiments recently made by Mr. Robertson of the Naval Construction Research Establishment are of great

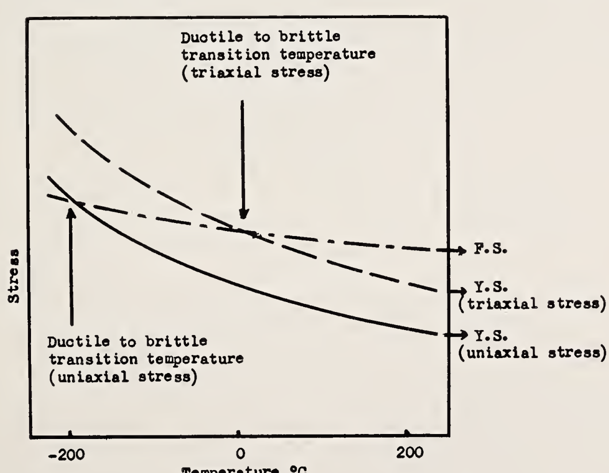

FIGURE 1.13. *Variation with temperature of yield stress and fracture stress of steel (schematic).*

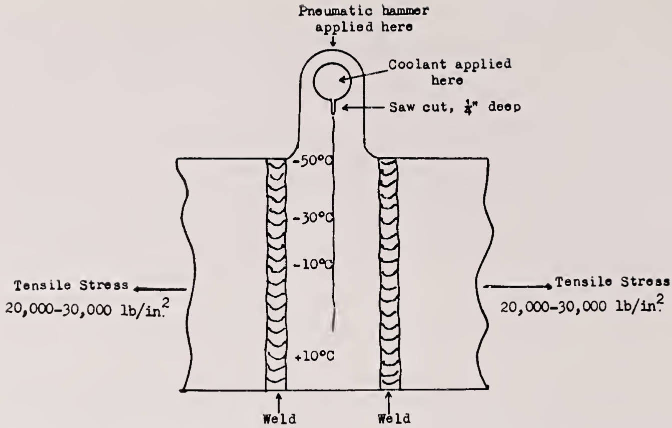

FIGURE 1.14 *Tests for resistance of steel plate to the spread of a crack.* (T. S. Robertson)

interest. A test piece of the form shown in figure 1.14 is prepared. At one end there is a round hole and a saw cut. Two pieces of mild steel are welded to the sides of the bar, by means of which the test piece can be placed in a testing machine, and submitted to a lateral stress of, say, 20,000 to 30,000 lb/in ^{2}. A cooling agent is supplied to the round hole, so as to maintain a temperature gradient along the test piece, from about —70° C at the cold end to +15° C at the other, and the cold end is repeatedly struck with a pneumatic hammer. A crack starts from the bottom of the saw cut, and shoots across the specimen with little or no plastic deformation of the steel, until a point corresponding to a definite temperature is reached. From this point plastic deformation begins, and the crack opens out slowly at a rate determined by the motion of the crosshead of the testing machine. This experiment shows most clearly, first, that there is a limit to the triaxiality of stress at the root of a spreading crack, and second, that the ratio of fracture stress to flow stress in the steel is such that the flow stress under the triaxial conditions set up at the root of the crack is greater than the corresponding fracture stress below the critical temperature and less above that temperature.

It is the fracture stress and flow stress of the undeformed steel that are important in this experiment. Once flow begins, both flow stress and fracture stress alter, and it often happens that after a certain amount of deformation the conditions for the spread of a cleavage fracture are set up again. This has happened in the fracture illustrated in figure 1.15, in which A marks the limit of the first stage of failure, and B the limit of a second stage of cleavage fracture initiated after some plastic deformation had taken place. In relation to the application of steels at low temperatures, the temperature at which the spread of the crack ceases to be entirely deformationless is clearly of the greatest importance.

In Europe, where the basic Bessemer process is much used, attempts to improve the low-temperature properties of mild steel have been directed to the reduction of the nitrogen content of the steel by modifications of the blowing procedure, such as enrichment of the air with

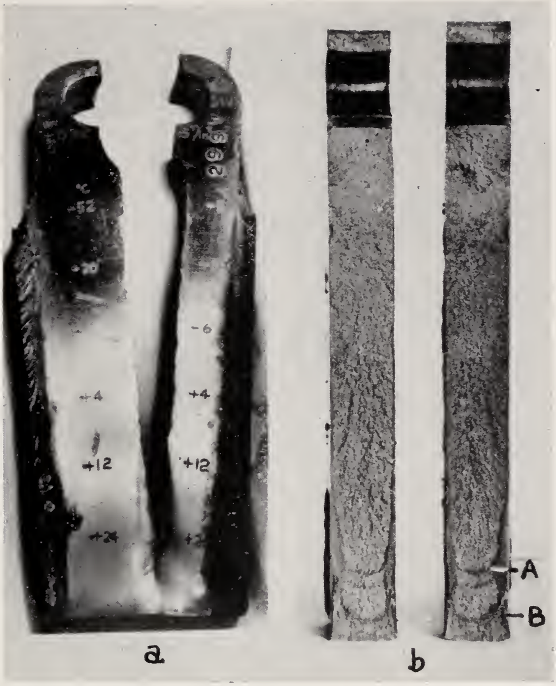

FIGURE 1.15. *Test for resistance of steel plate to the spread of a crack.*

Transverse stress 33,600 lb/in.^{2}; fracture transition +10° C.: (a) General view of test piece after fracture, (b) fracture showing (A) limit of initial crack; (B) limit of subsequent cleavage crack after plastic deformation.

(T. S. Robertson)

oxygen or the use of a mixture of steam and oxygen. Technically, considerable success has been obtained both in eliminating the nitrogen and in lowering the temperature at which embrittlement occurs, as is demonstrated by table 1.5 taken from the work of M. Coheur and

TABLE 1.5. *Improved Bessemer steels*

(Coheur)

Process	Nitrogen	Temperature at which notch-impact value fell below 5 kg-m/cm^2 in strain-aged condition
	Percent	° *C*
Ordinary Thomas steel	0.0115	16
Oxygen enriched air plus limestone	.0060	−5
Oxygen plus steam	.0020	+7

others [16]. The use of oxygen-enriched air and limestone seems to be more effective in lowering the impact transition temperature than the use of oxygen and steam, and it is possible that hydrogen introduced by the steam exerts a harmful effect. It is yet too early to say whether economic success has been achieved.

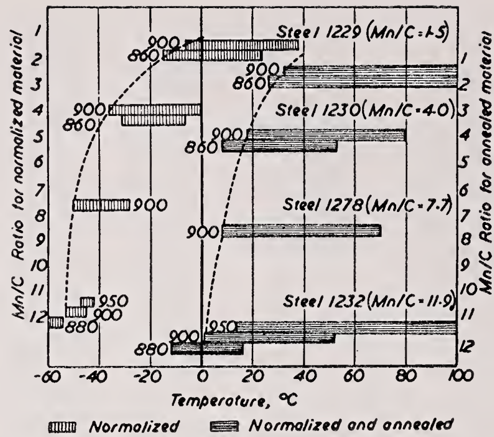

FIGURE 1.16. *Effect of* Mn : C *ratio in mild steel ship plate on transition temperature range in impact.*
(Barr and Honeyman)

In Great Britain, where the basic open-hearth process is more common, nitrogen is seldom troublesome, and more attention has been given to the carbon and manganese contents of the steel. It has been usual to take as a criterion of quality the temperature of transition from tough to brittle in an Izod test or Charpy test made on a specimen with an Izod V-notch. Figure 1.16 gives the results of tests by Barr and Honeyman showing that for equal tensile strength, steels with a high manganese and low carbon content have lower impact transition temperatures than steels with high carbon content and low manganese content [17]. As a result of these and other tests [18] it has been decided to keep the ratio of manganese to carbon in plate steels not less than 3:1. No theoretical significance is attached to this ratio. The specification for tensile strength sets limits upon the percentages of carbon and manganese that can be employed, and a limitation upon the manganese-carbon ratio merely ensures that a steel of the right general type is made.

Since the above step was taken, developments have taken two directions. For plates over about 1 inch in thickness or where exceptional toughness is required, steels with still higher manganese content may be employed and the material normalized before fabrication is commenced. Where normalizing is adopted, it is often advantageous to "grain control" the steel with aluminium. The effect of these measures on the impact properties of ½- and 1½-inch-thick plates is illustrated by the results in table 1.6. A survey has also been set in hand under the auspices of the British Iron and Steel Research Association of the properties of the normal production of mild-steel plates made to the ordinary specification. Opportunity has been taken in a preliminary survey to compare various tests that have been advocated, such as notch impact tests, slow notch bends, and notch tensile tests. The results are not yet available, but the compilation of the facts appears to have caused the divergencies of views to be moderated to a marked extent and in the author's opinion there is reason to hope that a simple Izod test at an appropriate temperature may eventually be agreed upon

TABLE 1.6. *Properties of grain-refined high-manganese plate steel*

(Barr)

Quality	C	Mn	Si	Thickness	V-notch Charpy test Temperature for 50% fibrous (typical results)
	Percent	*Percent*	*Percent*	*in.*	*° C*
Ship quality (as rolled)	0.18	0.6	0.05	½ 1½	+10 +50
Ship quality (as normalized)	.18	.6	.05	½ 1½	0 +30
High manganese-low carbon (as rolled)	.13	1.2	.05	½ 1½	−10 +10
High manganese-low carbon Grain-controlled and normalized	.13	1.2	.15	½ 1½	−40 −30

as satisfactory. It has also been found that firms making plate for the same market may be producing materials of distinctly different standards of quality in respect of impact transition temperature, and this has led firms to reexamine their procedures in order to find out how these differences arise. I am indebted to Mr. Barr for the data given in figure 1.17, which were obtained for the plate mill of a Scottish Works. They show the variation in impact properties for mild-steel plates of different thicknesses and the manner in which the impact value is related to the temperature at which the rolling is completed. It is known that inferior properties of thick plates are partly accounted for by the finishing temperature and partly by the rate of cooling from that temperature. They are no doubt associated with the distribution of minor constituents, of which more is written below. British steelmakers, of course, are not alone in thinking along these lines, and several continental firms are looking into these matters with equal care.

In order to study the factors associated with brittleness with more precision than is possible in a works, an investigation using materials of exceptional purity has been undertaken during the past few years

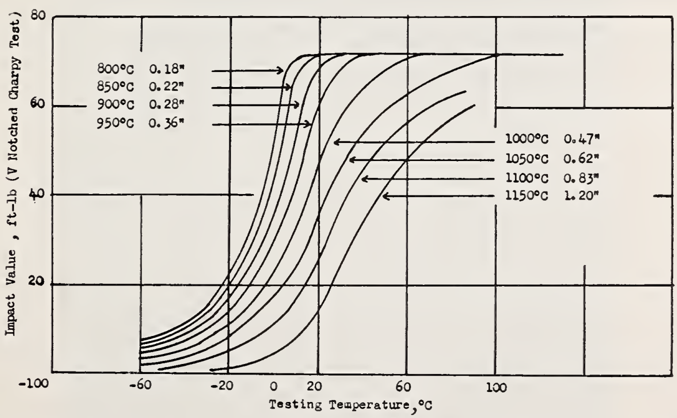

FIGURE 1.17. *Effect of plate thickness and finishing temperature of rolling on impact transition temperature of plate steel.*

(Barr)

at the National Physical Laboratory on behalf of the Alloy Steels Research Committee of the British Iron and Steel Research Association. Figure 1.18 illustrates the vacuum furnace in which the materials are prepared in 25-lb. melts, and an example of the control of composition that is achieved is given in table 1.7. The brittleness of the material is assessed by means of tensile tests at temperatures down to −196° C, and impact tests with a Charpy machine and a

TABLE 1.7. *Chemical analyses of high-purity irons and iron alloys*

Element	Iron	Iron	5%-Mn alloy	5%-Ni alloy	5%-Si alloy	5%-Mo alloy
	Percent	*Percent*	*Percent*	*Percent*	*Percent*	*Percent*
Carbon	0.0025	0.0031	0.0043	0.0022	0.0024	0.0022
Silicon	.003	.001	.005	.005	4.98	.002
Manganese	<.005	<.005	4.9	<.005	0.003	<.005
Sulfur	.0046	.0034	0.0056	.0042	.006	.0051
Phosphorus	.001	<.001	<.001	.001	.001	.0012
Nickel	.007	.005	.006	4.96	.005	.005
Chromium	.001	<.001	.002	0.001	.001	.001
Copper	.005	.005	.005	.005	.005	.005
Aluminum	.004	.004	.001	<.001	.004	.005
Oxygen	.001	.00073	.0007	.0014	.0011	.0012
Nitrogen	.001	.001	.0017	.001	.0006	.001
Hydrogen	<.000005	<.000005	<.000005	<.000005	<.000005	<.000005
Molybdenum	[a]	[a]	[a]	[a]	[a]	5.01

[a] Not detected spectroscopically.

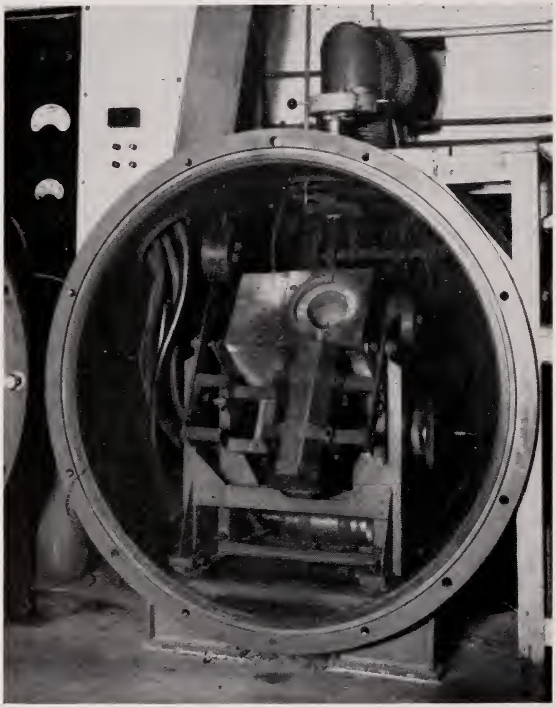

FIGURE 1.18. *Twenty-five-pound vacuum melting furnace.*

specimen with an Izod V-notch. The purest iron that has been made (on this scale) is extremely ductile at room temperature, but it begins to embrittle between −73° and −196° C, and at −196° C is almost completely brittle, breaking with a cleavage fracture and not more than 5-percent elongation under a stress of about 100,000 lb/in.^{2} Its impact transition temperature is very sharp, and occurs at about −15° C. Figure 1.19 shows the effect of additions of carbon on the impact transition temperature. The first 0.01 percent has little effect; greater additions raise the transition temperature sharply, until at 0.03 percent of carbon it is +88° C. At −196° C the alloys are quite brittle, and the fracture stress is, if anything, slightly raised by the presence of carbon. As the solubility of carbon in iron varies from 0.005 percent at 400° C to 0.016 percent at 700° C, it may be guessed that the carbon in solution in the iron is harmless, and it is the carbon out of solution and, perhaps, that which separates from the ferrite during cooling that exerts the embrittling effect. Additions of manganese alone lower the impact transition temperature and render the iron appreciably ductile in tension at −196° C. Although on account of the influence of plastic deformation on cleavage strength it is impossible at this stage to state the cleavage stress, it is certainly higher than 112,000 lb/in.^{2} in a 2-percent-manganese alloy, and may be considerably higher. More important, however, is the effect of manganese in the presence of carbon. Figure 1.20 shows that the high impact transition temperature of iron containing 0.05 percent of carbon is rapidly lowered as manganese is added, and that a 2-percent manganese alloy containing 0.05 percent of carbon has a slightly lower impact transition temperature than one containing 0.003 percent of carbon. It is also appreciably more ductile in a tensile test at −196° C. Evidently the manganese has a profound effect on the condition of the carbon in the iron, and this is at least as important as the effect of the dissolved manganese atoms.

The effect of oxygen additions is still more remarkable (fig. 1.21). As soon as the content rises above 0.003 percent the impact transition

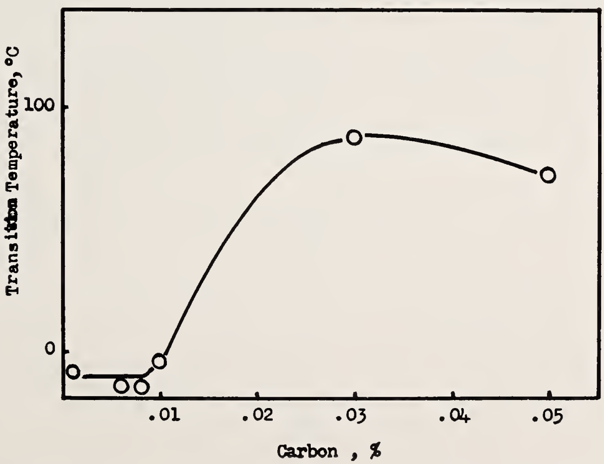

FIGURE 1.19. *Effect of carbon on the impact transition temperature of iron (normalized).*

temperature begins to rise sharply. At the same time, the fractures in the brittle condition begin to show intercrystalline areas, and they eventually become entirely intercrystalline. The tensile properties of a plain bar are unaffected at room temperature, but at −196° C the fracture stress falls sharply, and the fracture again becomes partly or fully intercrystalline. It is evident that the notch brittleness at room temperature and the weakness at −196° C are in this case both due to a weakening of the cohesion at the grain boundaries. Iron containing oxygen thus provides a clear example of a material containing a serious weakness, which is not detected by a tensile test at room temperature, and is detected by a notched bar test or a test at low temperature.

The fracture stresses are moderately reproducible. No attempt has yet been made to find out whether a measurable range of values dependent upon the size of specimen can be observed, as was done by Davidenkov, Shevandin, and Wittmann [19], but as the specimens are exceptionally pure, it would be expected that the range would be small and it would be of considerable interest to know whether any "size effect" remains in material of this kind.

Brittle fractures that are partly intercrystalline and partly due to transcrystalline cleavage have been found to be rather common, and in some circumstances they have been found in irons containing very little oxygen. Whether they occur when the iron has been "deoxidized", for example, with manganese or silicon has not yet been determined, but the question is clearly very important, as is also the question whether the favorable effects of aluminum additions are due to the removal of the last traces of oxygen. Some very curious observations have been made with respect to the influence of nickel. It

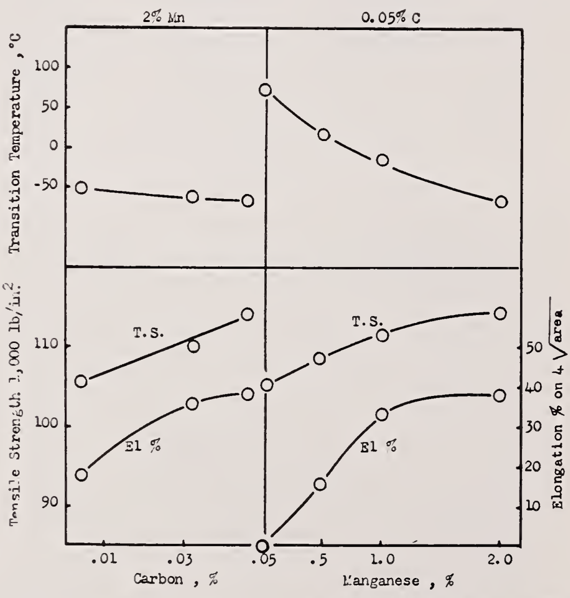

FIGURE 1.20. *Effect of manganese and carbon on transition temperature and tensile properties at —196° C of high-purity iron (normalized)*

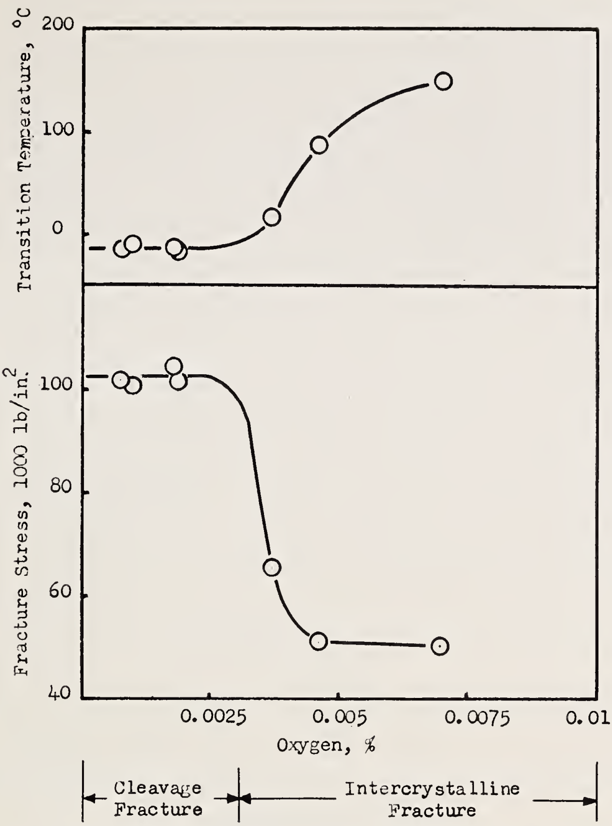

FIGURE 1.21. *Effect of oxygen on the impact transition temperature and tensile fracture stress at —196° C of high-purity iron (normalized).*

is well known that the presence of nickel favors ductility at low temperatures. Tests of iron-nickel alloys at −196° C show this to be so (table 1.8) but in notch-impact tests of the same alloys at room temperature and above, the impact transition temperature rises progressively with increasing nickel content (table 1.8) and the alloys with 2

TABLE 1.8. *Influence of nickel on the impact transition temperature and tensile properties at —196° C of high purity iron (normalized)*

Material	V-notched Charpy impact tests		Tensile tests at −196° C				
	Transition temperature	Impact value at 0° C	Lower yield	Tensile strength	Elongation in $4\sqrt{A}$	Reduction of area	Fracture stress
	° C	*ft-lb*	*1,000 lb/in.²*	*1,000 lb/in.²*	*Percent*	*Percent*	*1,000 lb/in.²*
High-purity iron	−14	188	84.2	95.5	4.0	5.0	100.8
2%-Ni alloy	140	8	62.9	74.4	0.5	1.0	74.8
3%-Ni alloy	170	10	80.0	83.6	3.5	4.0	87.2
4%-Ni alloy	210	36	87.6	96.1	27.0	23.0	122.5
5%-Ni alloy	210	10	88.5	96.7	29.0	21.0	121.8

to 5 percent of nickel break at 0° C with a comparatively small energy absorption. The materials in this case are very pure, although they do contain some 0.001 percent of oxygen and 0.004 percent of sulfur, but it is hard to believe that the embrittlement is the true effect of nickel. It is more likely that it is due to some impurity in combination with nickel; but in this case it is apparent that the embrittling effect is exerted in notch tests at room temperature and above, and not obvious in the tensile tests of the high nickel alloys at −196° C.

In this connection reference may be made to a series of papers that have lately been published by Bastien and Azou [20]. They found that when steel wire is charged with hydrogen, electrolytically or by corrosion with acid, a reduction of elongation and of true stress at rupture occur, but that the embrittlement disappears rather abruptly on cooling below −110° C. The effect, which is illustrated in table 1.9, is reversible. Furthermore, at room temperature the speed of testing exerts an influence of such a kind that the embrittlement is observed only within certain speed limits. Above and below these speed limits, little or no embrittlement is found. It seems that it may be necessary to recognize that the rupture strength of metals is influenced by extremely small proportions of such elements as carbon, oxygen, hydrogen, and nitrogen, and that these influences for some reason not yet understood may come and go as the temperature rises and falls.

Table 1.9. *Effect of temperature on the embrittlement of 0.15 percent C Steel by hydrogen*

(Bastien and Azou)

State of metal	Rupture stress at +18° C	Elongation at +18° C	Rupture stress at −70° C	Elongation at −70° C	Rupture stress at −160° C	Elongation at −160° C
	1,000 lb/in.²	*Percent*	*1,000 lb/in.²*	*Percent*	*1,000 lb/in.²*	*Percent*
Uncharged	150	74	152.3	69	183.5	48
Charged with H_2 by corrosion	112.2	56	107.4	37.5	182	47
Electrolytically charged with H_2	112.2	56	103.8	39	184.5	48

The influence of an element such as molybdenum is in striking contrast. As much as 2 percent of molybdenum has little effect on the transition temperature of iron and on its general behavior and rupture stress at −196° C (table 1.10). The known effect of molybdenum on the susceptibility of steels to temper-brittleness must be due to the influence that the metal exerts upon the distribution of some third element.

Table 1.10. *Effect of molybdenum on impact transition temperature and tensile properties at —196° C of normalized pure iron*

Molybdenum	Impact transition temperature	0.2% proof stress	Ultimate strength	Fracture stress	Elongation
Percent	*° C*	*1,000 lb/in.²*	*1,000 lb/in.²*	*1,000 lb/in.²*	*Percent*
0	−15	78.8	95.6	101.0	4
0.5	−20	85.8	93.0	96.5	1.5
2.0	−36	82.0	93.4	97.5	4

It is worth while to point out a general consequence of conclusions of this kind. It may be profitable, in view of the shortage of the major alloying elements, and the need to use them in applications for which their special properties are indispensable, to spend more effort on detecting and removing the impurities responsible for embrittlement and less to making special additions to compensate for their presence. As an example of the possibilities, alloys of iron with 25 and 30 percent of chromium, which for so long have been retarded in their development on account of their brittleness, have recently been reported by Hochmann [21] to be perfectly ductile, and to have impact values of 20 to 30 kg-m/cm^2 down to $-30°$ C if their carbon, nitrogen and oxygen contents are reduced to 0.002 to 0.005 percent by melting in a good vacuum.

There has been some difficulty in establishing the true effect of grain size on embrittlement. Petch [22] has determined the brittle fracture stress at liquid-air temperature of a series of moderately pure irons in which the grain size was varied by cooling at different rates from different temperatures above the critical range. He found that the fracture stress increased as the grain size decreased, being inversely proportional to the square root of the grain diameter; but when the grain size was varied in the same irons by straining and annealing them, the brittle strength at $-196°$ C was found to be quite unaffected. The latter conclusion is supported by some results obtained with purer iron at the National Physical Laboratory. The impact transition temperature was not altered when the grain size was varied by straining and annealing, but all the strained and annealed specimens had transition temperatures higher than that of the same iron in the normalized condition. The explanation of these phenomena is not yet clear.

Properties of Heat-Treated Alloy Steels at Low Temperatures

The results of investigations of the properties at low temperatures of typical ferritic alloy steels have been published in Great Britain, Germany, and Russia, and these confirm the early work of Hadfield and de Haas, and the American work, in showing that all ferritic steels become embrittled at low temperatures. They show also that the heat-treated alloy steels are less easily embrittled than non-heat-treated plain carbon steels, and that the high nickel steels maintain their toughness best. The above mentioned investigations have mostly been made on small wrought bars, and much interest is now being taken in the properties of bars of large section, of castings and of forgings, and in the transverse properties of large forgings. The work of Pomp and Krisch [23], carried out at the Kaiser Wilhelm Institute für Eisenforschung (now the Max Planck Institute) when the properties of the German economy steels were being investigated, is particularly helpful. It deals with a chromium-molydenum steel of a not very deep hardening type (0.3% C, 1% Cr, 0.2% Mo) and a deeper hardening nickel-chromium-molybdenum steel (0.3% C, 1.8 to 2.5% Cr, 1.5 to 2.5% Ni, 0.2 to 0.4% Mo), and shows (a) that when the same steel is hardened and tempered to a given level of tensile strength by oil hardening and by water hardening, the water-hardened steel retains its toughness to lower temperature, (b) that when large bars are heat treated to a given strength level, the insides of the bars lose their toughness on cooling more readily than the outsides of the bars, and large bars suffer more than small bars; this is a particularly important conclusion, for the differences between the outside and center of large forgings has been found to be considerable, (c) that the transverse properties are more seriously affected than the longitudinal properties, (d) that different samples of the same type of steel suffer to different extents, in a way not easily explained by the composition. Figures 1.22 and 1.23 give examples. These are clearly hardenability effects, and suggest that steel must be quenched to a fully martensitic condition to secure the greatest possible toughness at low temperatures. At one time it was confidently suggested that a knowledge of the hardness of a quenched and tempered steel was sufficient to enable all its other mechanical properties to be deduced. The proposition failed when the results of notched-bar impact tests were taken into consideration, and it fails more completely in its application to the resistance of materials to complex stresses at low temperatures. The reason is that hardness tests are influenced mainly by flow stress and work-hardening capacity, and do not take the considerable variations of fracture stress into account. The fracture stress becomes more important as the temperature falls. In Europe the notch bar impact test has been widely used as a guard against the presence of undue brittleness. This may be regarded broadly as the presence of an unduly low fracture stress, and the results of recent investigations suggest that the experience that has been accumulated in obtaining a high impact value at room temperature is in the main applicable to methods of retaining great toughness at low temperature.

It has been feared that brittleness at low temperatures in notch impact tests would be accompanied by great sensitivity to notch effects in fatigue, but a comprehensive investigation by Pomp and Hempel [24] has shown that, at least down to $-78°$ C, this is not so. The ratio of the fatigue limit to the tensile strength is at $-78°$ C much the same as it is at $+20°$ C, and the ratio of the fatigue limit of a notched

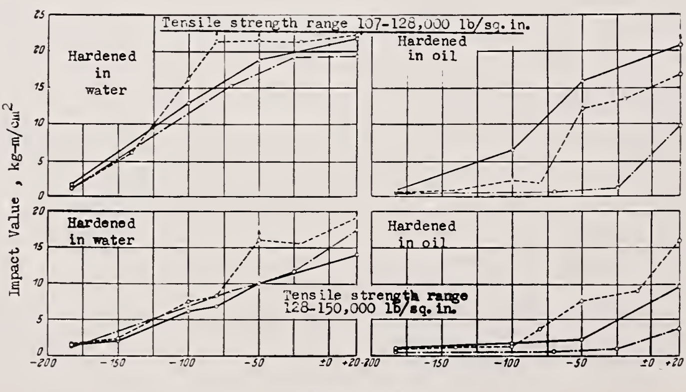

FIGURE 1.22. *Impact values at various temperatures of three steels of 1 percent chrominum, 0.2 percent molybdenum, 0.25 percent carbon type.*
(Pomp and Krisch)

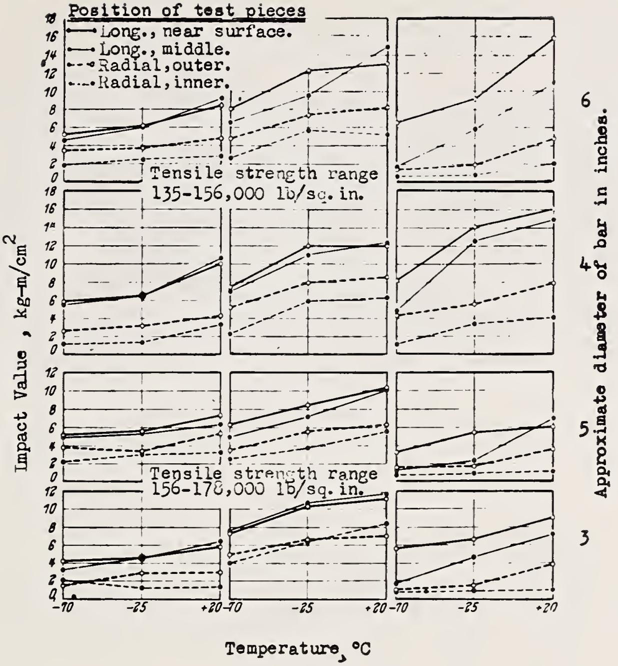

FIGURE 1.23. *Impact values at various temperatures of three steels of 2 percent nickel, 2 percent chromium, 0.35 percent molybdenum, and 0.3 percent carbon type.*

(Pomp and Krisch)

bar to the fatigue limit of an unnotched bar is only slightly less at −78° C than it is at +20° C.

From the foregoing it would be concluded that temper brittleness would exert a great influence on the behavior of heat-treated steels at low temperature. In France it has been strongly emphasized that the best method of determining sensitivity to temper brittleness is to study the effect of embrittling treatments upon the temperature at which, on cooling, the transition from tough to brittle behavior occurs in a notched-bar test. Although this conclusion is not new, the work of Jolivet and Vidal [25] merits attention because of its completeness. Figure 1.24 shows how, in a quenched and tempered 1.38-percent-chromium steel, progressive embrittlement at 525° C causes the impact transition temperature to rise from −10° C to +160° C, and figure 1.25, in which the impact transition temperature is plotted against the time of embrittlement at various temperatures, shows how close is the analogy between this phenomenon and a precipitation hardening phenomenon. Practically all steels show some susceptibility to temper brittleness when studied in this manner, but the susceptibility is much increased when certain impurities, such as phosphorus and arsenic, are present. Very recently Entwhisle [26] has studied at Cambridge the susceptibility induced by additions of arsenic in a nickel-chromium-molybdenum steel, and has made use of tensile tests

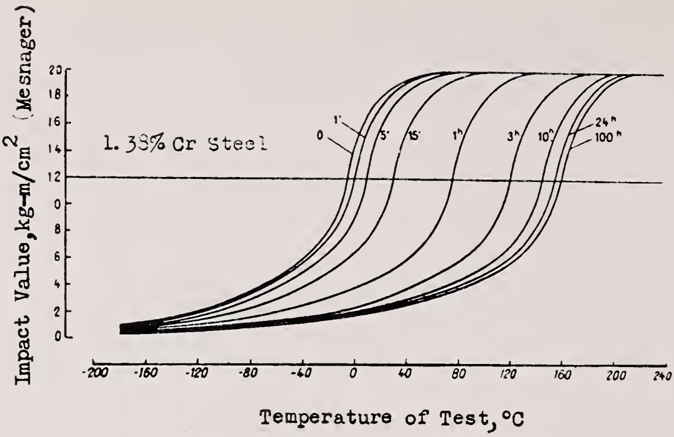

FIGURE 1.24. *Impact temperature curves after different times of embrittlement at 525° C.*
(Vidal)

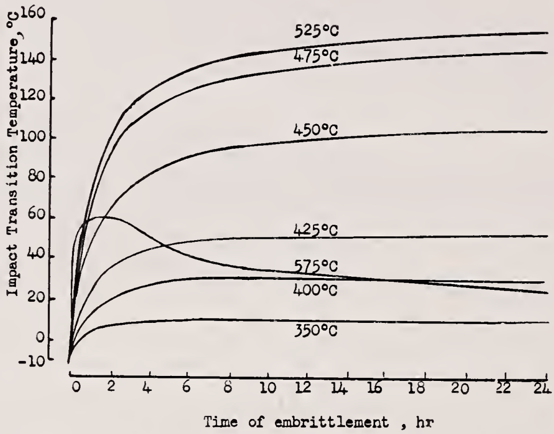

FIGURE 1.25. *Effect of embrittlement time at various temperatures on impact transition temperature.*
(Vidal)

in liquid nitrogen. The steel in the correctly heat-treated state is quite ductile at −196° C, having a reduction of area of about 35 percent, but after embrittling treatments at 525° C the reduction of area falls almost to zero, and the true stress at fracture falls to some 60 percent of its initial value. The yield point of the steel does not alter, but the fracture of the steel becomes partly intercrystalline. It is apparent that during the treatment the embrittling impurities have congregated at the grain boundary and there produced the reduction of fracture stress which is responsible for the loss of ductility at −196° C and the lowered impact value at room temperature. The grain boundary in question is the boundary of the austenite grain that existed before the steel was quenched. It is perhaps surprising that this should be so as the steel is in the ferritic condition when the embrittlement takes place, but there are geometrical relations between the parent austenite grains and the ferrite grains arising from them, and the boundary of the austenite grain is likely to have an exceptionally irregular structure.

Austenitic Steels at Liquid-Air Temperatures

Preliminary explorations have all shown that austenitic nickel-chromium steels possess an excellent combination of strength and toughness at the temperature of liquid air, and these materials are much used in chemical plants intended for operation at low temperatures. The steels are not all equally good. It is necessary to choose a composition that does not become martensitic on cooling, for otherwise a permanent embrittlement may occur, particularly if the carbon content is high, as the following example shows (quoted from Colbeck [27] for 0.41 percent of carbon, 0.95 percent of manganese, 19.8 percent of nickel steel).

Temperature	Shock test (Fremont) *kg-m*
Normal	37.1
—182° C	2.6
On return to normal	4.7

The steel must also be free from the tendency to become martensitic on cold-working, for which reason it is desirable to test the steels for their stability. Mr. Keating supplied the data in figure 1.26, which relate to a steel which, although of a type that is capable of excellent toughness at −180° C in the fully soft condition, was less tough as

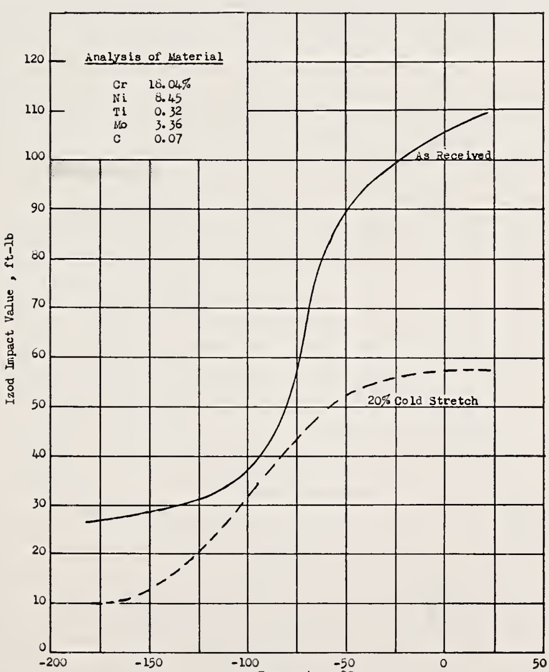

FIGURE 1.26. *Izod impact tests at low temperatures on austenitic steel bar.* (Keating)

received than it should have been and was rather brittle at −180° C after being cold-stretched 20 percent.

Precipitates of carbide or of sigma phase cause a deterioration of the properties at low temperature and are important when, as is often the case, the steels have to be welded. Annealing after welding to restore the embrittled heat-affected zone may be desirable. The conditions of this annealing are particularly important for the steels containing titanium, or titanium and molybdenum, which are often used in the chemical industry on account of their freedom from weld decay and good welding properties. They owe their ease of welding partly to the presence of a proportion of alpha in the structure, which gives them a plasticity at high temperatures and freedom from hot cracks that a fully austenitic steel does not have, but also makes them exceptionally liable to become brittle on cooling and to form sigma phase if wrongly annealed. This is specially true of the weld metal when these steels are welded for, being in the cast state, the weld metal contains more of the alpha phase than wrought and annealed metal of the same composition, and is consequently more liable to embrittlement. Figure 1.27 gives an example of the properties of a typical weld metal as deposited (curve A). Annealing at too low a temperature results in complete embrittlement (curve B), but a correct annealing, designed to produce a structure with the minimum quantity of the alpha phase brings about a much more satisfactory condition (curve C). Much can be done by careful adjustment of the composition of the electrodes to improve the toughness of the weld at −180° C. It is generally thought that 5 to 10 percent by volume of the alpha phase is sufficient to permit welds to be made without cracks, and is tolerable

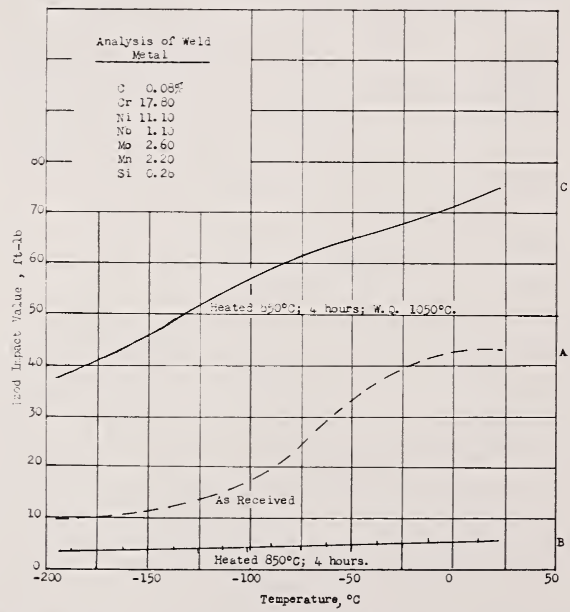

FIGURE 1.27. *Izod impact tests at low temperatures on austenitic steel weld metal in various conditions.*

(Keating)

from the viewpoint of toughness at low temperature. On the other hand, the adoption of other welding methods, avoiding the necessity for using a weld metal containing the alpha phase, might in some cases be a better solution of the problem.

The toughness of austenitic steels at low temperatures is often attributed to the face-centered cubic lattice of the austenite crystal. Hadfield and de Haas [28], however, noted many years ago that certain austenitic manganese steels are severely embrittled at sufficiently low temperatures, although they retain their austenitic structure unchanged, and recover their toughness on reheating to room temperature. The low-carbon austenitic chromium-manganese steels retain their toughness quite as well as the chromium-nickel steel [29]. Colbeck has studied this problem, and confirmed that the formation of martensite is not concerned. He found that by lowering the carbon content, and introducing nickel and chromium to some extent at the expense of manganese, the brittleness can be very materially reduced (fig. 1.28). In case it should be concluded that the brittleness is

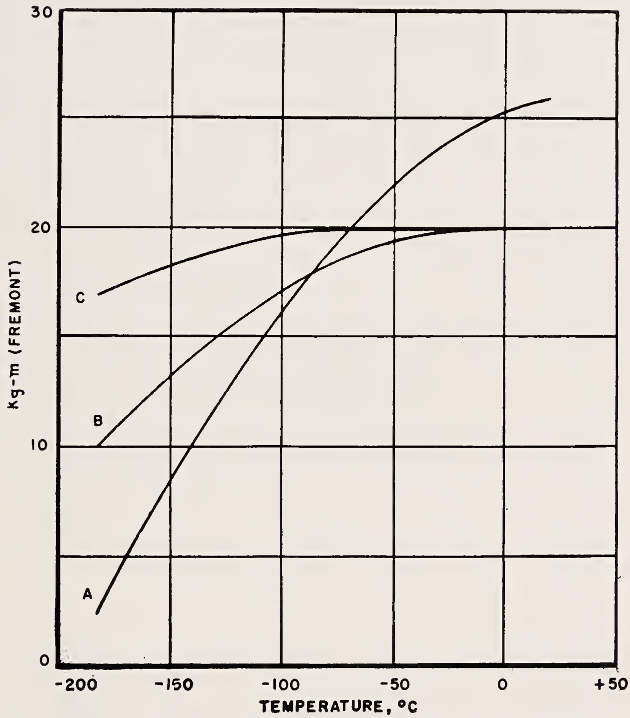

FIGURE 1.28. *Impact values at various temperatures of three austenitic manganese steels.*
(Colbeck)

	C	Si	S	P	Mn	Cr	Ni
	%	%	%	%	%	%	%
A	1.25	0.18	0.008	0.126	12.65	--------	--------
B	0.80	.91	.02	.07	14.0	0.07	3.54
C	.55	.20	.008	.068	8.1	4.0	8.0

essentially due to carbon, it should be pointed out that an austenite containing 35 percent of manganese and only 0.22 percent of carbon is quite brittle at −180° C.

Some light may perhaps be thrown on the occurrence of brittleness in face-centered cubic metals at low temperatures by some recent observations on copper-zinc and copper-arsenic alloys at the National Physical Laboratory. Simple alpha solid solutions of zinc in copper or arsenic in copper in the annealed condition have remarkably similar tensile properties at room temperature if their compositions are so adjusted that the ratios of atoms to valency electrons are the same. Figure 1.29, which compares the stress-strain curves of a copper-zinc alloy with 8.75 atomic percent of zinc and a copper-arsenic alloy with 2.18 atomic percent of arsenic, both of which have an electron atom ratio of 1.087, is an example. This is true throughout the range of the copper-arsenic solid solution, and even when the limit of solid solubility is slightly exceeded, so that a trace of a grain boundary constituent can be detected in the copper-arsenic alloy, the tensile properties of a copper-arsenic alloy and of the equivalent copper-zinc alloy are at room temperature practically identical. When the alloys are cooled, a difference of properties appears. Figure 1.30 shows the tensile properties at +20° C, −73° C, and −196° C of a very slightly supersaturated copper-arsenic alloy and of the equivalent copper-zinc

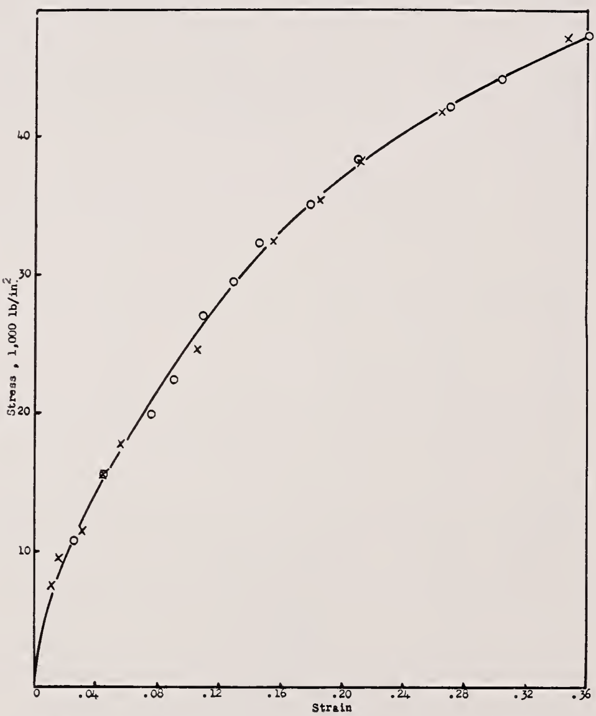

FIGURE 1.29. *Stress-strain curves for copper-zinc and copper-arsenic alloys of equal electron: atom ratio.*

○, Copper-zinc (8.75%) alloy; ×, copper-arsenic (2.18%) alloy; electron concentration is 1.087.

alloy. The proof stresses and ultimate tensile stresses remain alike on cooling, but the reduction of area of the copper-arsenic alloy falls progressively in relation to that of the copper-zinc alloy. At the temperature of liquid air the copper-arsenic alloy is relatively brittle, and breaks without undergoing any local contraction. The changes are reversible—the copper-arsenic alloy on being reheated to room temperature recovers its original ductility. The phenomenon seems to be analogous to that found in iron-oxygen alloys, as the weakness produced by oxygen in iron is also detectable in tensile tests only when the metal is cooled. It is probable that the explanation is the same, namely, that the grain-boundary film in the arsenic alloy produces a lowering of the true breaking stress of the metal. At room temperature the breaking stress is nevertheless so high in relation to the flow stress that the flow stress and strain hardening capacity of the alloy dominate its behavior. At low temperature this is no longer the case, and the breaking stress is sufficiently close to the yield stress to bring about rupture at an early stage in the test. Face-centered cubic metals are thus not entirely free from liability to be embrittled on cooling, and not fundamentally different from body-centered cubic metals. The difference is of degree, the face-centered cubic metals having on the whole a higher ratio of fracture stress to proof stress than the body-centered cubic metals. It seems not unlikely that the brittleness of austenitic manganese steels in liquid air has a similar explanation.

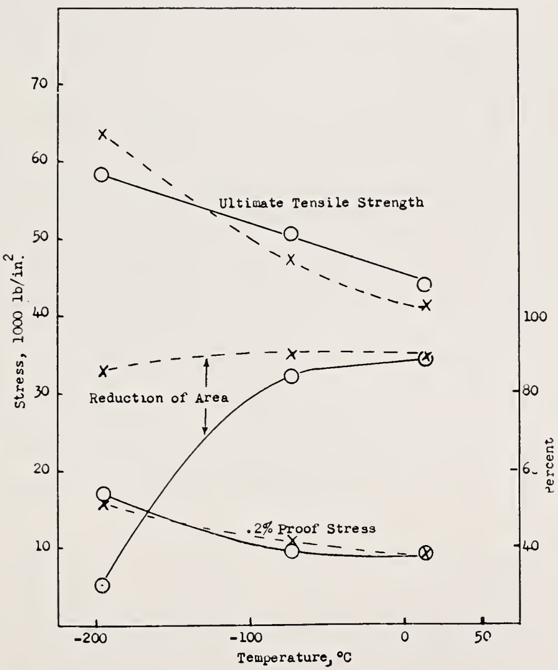

FIGURE 1.30. *Tensile properties at room temperature, —73° C. and —196° C. of copper-zinc and copper-arsenic alloys of similar electron: atom ratio (1.165 and 1.173).*

×, Copper-zinc alloy; ○, copper-arsenic alloy.

This paper was written as part of the General Research Program of the National Physical Laboratory, and is published with the permission of the Director. The author thanks Sir Charles Goodeve, Director of the British Iron and Steel Research Association; Dr. Bailey, Director of the British Non-Ferrous Metal Research Association; Mr. Colbeck, Mr. Keating, Mr. Barr, Mr. Entwhisle, Dr. Petch, and Dr. Brown for information kindly given, and acknowledges the assistance of many colleagues, especially Mr. Rees, Mr. Hopkins, Mr. McLean, and Mr. West.

Discussion

Dr. John R. Low, Jr., Head, Metals Research Section, General Electric Co., Schenectady, N. Y.: Dr. Allen, you showed two plots of transition temperatures versus carbon content, and another versus oxygen content earlier in a group of slides, where the carbon variation was roughly from 0 to 0.03 percent and that of oxygen 0 to 0.0025 percent. The question is whether or not there was any grain-size change occurring from group to group of specimens as you changed the composition, either of carbon or oxygen, which might conceivably contribute to the change in the transition temperature in the impact test.

One other question: You showed that in the presence of 0.016 percent of carbon in iron quenched from 450° C, that no yield point was exhibited. I believe, however, upon aging at room temperature, or even for shorter times at a more elevated temperature, the yield point is again observed after such a treatment. Is this correct? I wondered whether you meant to imply that this permanently eliminated the yield point behavior.

Dr. N. P. Allen: We have had extraordinary difficulty in proving that grain size has any effect at all. If we take cold-worked samples and anneal them below the critical range to get a range of grain sizes, the impact transition temperature and general properties are entirely unaltered, but if, on the other hand, we take the same materials and alter their grain size by taking them to different temperatures above the critical range and cooling them at various rates, considerable changes are effected. The feeling we generate is that the effect of grain size is an apparent one on the distribution of minor impurities in the grain boundaries. We have a rather interesting example. Silicon normally embrittles iron if sufficient is present. We have had alloys containing 1 or 2 percent of silicon that in the normalized condition had an impact transition temperature of about 100° C, but if the same iron is cold-worked and heat-treated in such a way as to produce a small grain size, the impact transition temperature is very much lowered.

Now, it is a curious thing that a silicon alloy has a tensile strength at liquid-air temperatures of about 30 tons per square inch, even though its oxygen content is less than 0.001 percent. The silicon content in this case may make the material exceptionally sensitive to very small quantities of oxygen. If the amount of oxygen concerned is very small and the grains are very large, the effect may be very much more serious than if the same amount of oxygen is distributed around small grains.

Although we feel that the effect of grain size is bound up with the distribution of impurities at grain boundaries, we do not in any way consider that we have proved the point.

It is commonly assumed in the literature that if no yield point is observed no carbon is present. I made the remark on yield point to show that it is possible in some cases to have carbon present and not observe a yield point.

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2. Manufacture of Steels for Low-Temperature Service

By J. B. Austin [1]

The temperature of transition from tough to brittle behavior as derived from notched-bar impact tests is generally taken as an index of the low temperature performance of a steel. This transition temperature depends in large measure on the microstructure, and the best structure appears to be austenite. In ferritic steels, tempered martensite has the lowest transition temperature, coarse pearlite the highest and mixed structures containing pearlite and tempered martensite have an intermediate transition temperature. Other significant factors influencing the low temperature behavior are grain size, addition of alloying elements and mode of deoxidation. The best means of achieving the desired end product depends to a considerable extent on the precise conditions of service and on economic factors such as relative cost and availability.

In the present state of our knowledge, any discussion of the manufacture of steels for service at low temperature must, of necessity, be in the nature of a progress report. For, although we have identified a number of variables in manufacturing processes that have a significant influence on the behavior of steels at low temperature, and although we know in a qualitative way what their influence may be, we have very few good data on the quantitative effect of these factors. Nor are we always certain as to how to achieve the desired ends most economically in practice. The best that one can do, therefore, is to point out certain facts that have been fairly well established and to indicate trends in this field as they appear on the basis of the evidence now available.

One of the greatest handicaps is our lack of an unequivocal criterion for the suitability of a given material for service at low temperature. It is obvious that a steel intended for this purpose must meet many of the same requirements as a steel intended for use at room temperature; it must, for example, have appropriate strength and ease of fabrication. But there is the additional consideration that, at low temperature, steel shows an increased tendency toward brittle behavior in the presence of notches or surface imperfections. This makes good notch toughness one of the most desirable characteristics for low-temperature service, and the notched-bar impact test has been widely used to evaluate the suitability of steels for such service. The precise evaluation of notch toughness is neither easy nor simple, as has been pointed out by Stout and McGeady [1][2] and by Vanderbeck and Gensamer [2]. Considerable confusion still exists because of the use of different criteria to select the transition temperature. Some selections are based upon the transition from tough to completely brittle behavior, whereas others are based upon the transition from shear to predominantly cleavage fracture. In most specimens, the latter transition occurs at a higher temperature than the former.

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[1] Director of Research, Research Laboratory, United States Steel Co., Kearney, N. J.
[2] Figures in brackets indicate the literature references on p. 47.

Moreover, the type of notch and size of specimen employed, as well as the rate of straining, influence the test results.

But when the same methods of test are used and the results interpreted, using the same criteria, many of the discrepancies disappear and trends begin to become evident. Furthermore, the increasing use of statistical methods in the design of test programs and in the interpretation of test results promises to provide data that can be accepted with greater confidence.

Another difficulty is that, with the significant exception of ship service, there is as yet no well-established correlation between the results of notched-bar impact tests and behavior in use. In the case of ship service, Williams and Ellinger [3] have made a notable contribution by showing that the temperatures at which brittle failures occurred in plates in actual ship service correlate quite well with the temperatures obtained from the V-notched Charpy curves for these plates at low energy levels, that is, at about 10 ft.-lb. This correlation has not only been of direct value in the study of steels for use in ships but offers hope that further investigation will lead to correlation in other types of service. In general, however, the most that can be said is that if two steels have different transition temperatures, the one with the lowest temperature is the one less likely to fail in a brittle manner in service.

In this paper will be discussed the manufacture of steels that have good notch toughness, as indicated by the transition from tough to brittle behavior in the Charpy keyhole-notched-bar impact test. More specifically, the influence of chemical composition, mode of deoxidation, heat treatment, and microstructure, including grain size, will be dealt with. No attempt will be made to give a complete review of the literature in this field. The discussion will be confined to certain data that reveal, or confirm, a general trend.

The influence of microstructure will be considered first, as the properties we seek to attain appear to depend in large measure on this structure. It can be stated at once that the best miscrostructure is austenite. Many austentic alloys, such as those of the low-carbon chromium-nickel stainless type, show no transition, even at liquid-air temperatures (about $-315°$ F), and so are useful down to this temperature. Such alloys are expensive, however, and at the present time are subject to governmental restrictions on their use.

In ferritic steels, full martensitic hardening, followed by tempering to the desired hardness level, seems to produce the lowest transition temperature of any heat treatment capable of producing the given hardness. On the other hand, a completely pearlitic structure obtained by continuous cooling at a rate required to attain the desired hardness seems to have the highest transition temperature. In such a structure, the interlamellar distance is a factor as the transition temperature appears to be lower as the mean distance between lamellas decreases. At a given hardness level, mixtures of tempered martensite and pearlite have in general an intermediate transition temperature.

Another significant structural factor is grain size, which has been studied by a number of investigators, including Rosenberg and Gagon [4], National Bureau of Standards. Görrissen [5], as well as Hodge, Manning, and Reichhold [6], have reported quantitative data on this effect. The latter authors state that the keyhole-notched Charpy transition temperature of steel containing 0.02 percent of carbon varies as a linear function of grain size, there being an increase of 30° F for an increase in ferrite grain size of one ASTM number.

R. W. Vanderbeck has kindly supplied some further data for steel plate, which are presented in figures 2.1 and 2.2. These figures present data on the variation of transition temperature, as measured by keyhole-notched Charpy impact tests, with ASTM grain size for semikilled and silicon-killed steel plate containing 0.23 percent of carbon and 0.45 percent of manganese. It will be observed that the transition temperature decreases with decreasing grain size and that the variation is substantially linear. It will also be noted that an increase of one ASTM grain-size number is associated with an increase of about 20° F in the transition temperature of the silicon-killed steel and of about 30° F for that of the semiskilled steel. This is of the same order as the change reported by Hodge, Manning, and Reichhold. For the benefit of those interested in further details, it should be stated that these are the grain-size measurements cited by Vanderbeck in a recent paper [7].

Keeping these facts in mind, we can now consider ways of achieving a given structure. Virtually, the only way to obtain austenite stable at room temperature or below is by adding large amounts of alloying elements, which, as has been noted, makes the steels comparatively expensive and subject to restrictions as to use.

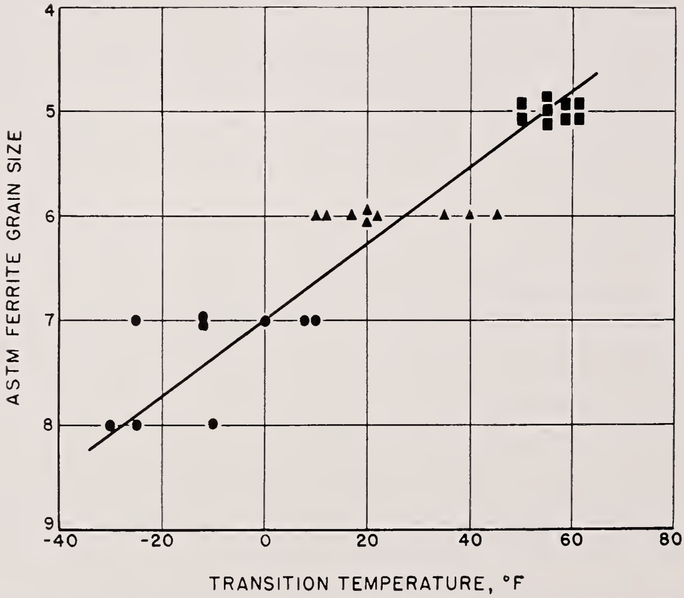

FIGURE 2.1. *Relation between Charpy keyhole-notch transition temperature and ferrite grain size in semikilled steel plate containing 0.23 percent of carbon and 0.45 percent of manganese.*

●, ½ in. thick; ▲, 1 in. thick; ■, 1½ in. thick.

(Vanderbeck)

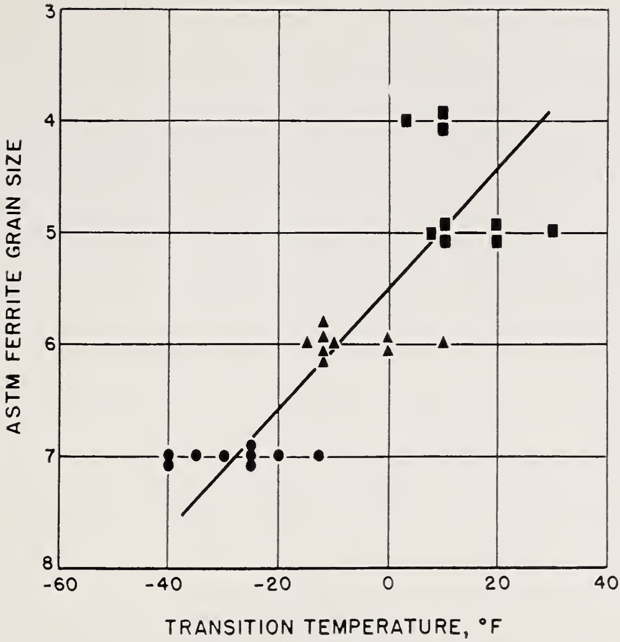

FIGURE 2.2. *Relation between Charpy keyhole-notch transition temperature and ferrite grain size in silicon-killed steel plate containing 0.23 percent of carbon and 0.45 percent of manganese.*

●, ½ in. thick; ▲, 1 in. thick; ■, 1½ in. thick.

(Vanderbeck)

The heat treatment required to obtain tempered martensite is the usual quench-and-temper procedure. It should be noted, however, that in this connection hardenability is a significant factor. Obviously, such a structure is more easily obtainable in alloy steels than in carbon steels and will vary with the particular amounts of alloying element present.

Microstructural uniformity is, however, an important consideration. For example, the structures obtained by normalizing or by normalizing and tempering are likely to be mixtures of ferrite and fine pearlite, together with some martensite or tempered martensite. Steels having such a mixed structure, although not in general desirable for use at very low temperature, may be quite good enough for the moderately low range. A significant exception to this generalization is the recently developed alloy steel containing 9 percent of nickel, which has a mixed microstructure yet is suitable for service at very low temperatures.

Carbon and low-alloy steels in the as-rolled or annealed condition usually contain coarse pearlite, which has generally been associated with a relatively high transition temperature. But the microstructure, and in consequence, the transition temperature in such steels, varies with the finishing temperature, cooling rate, and alloy content, especially the amount of carbon and manganese. A wide range of behavior is therefore available in such steels.

Clearly the thickness of the section considered must also be taken into consideration. Thus, a thin section can often be quenched-and-tempered to give a uniform structure of tempered martensite more easily than a thick section, or a large piece, which cools slowly, will tend to have a coarser pearlite than a small piece of the same steel, which cools rapidly.

A typical illustration of some of these effects is presented in figure 2.3, taken from a forthcoming publication by Seens, Miller, and Jensen [8], which shows data for an SAE 2317 steel in different thicknesses and heat treated in different ways. These data show the marked difference in behavior between specimens 2 in. thick and those ½ in. thick. They also reveal the striking difference between the as-rolled and the quenched-and-tempered microstructures. Further evidence bearing on this point has been supplied by Vanderbeck [7]. He observed that in a number of hot-rolled plates made by different deoxidation practices, the transition temperature decreased markedly as plate thickness decreased, due mainly to the difference in grain size. This trend is evident in the data presented in figures 2.1 and 2.2.

In controling grain size, the finishing temperature is a very significant factor. Thus Banta, Frazier, and Lorig [9] report that in ¾-in. hot-rolled plates, made from 200-lb semikilled experimental heats, raising the finishing temperature from 1,650° to 1,850° F increased the transition temperature in the Navy tear test by 30 to 45 deg F, depending on the carbon and manganese contents. The effect on the Charpy transition temperature, taken as the 20 ft-lb level, appeared to be much less evident. Control of finishing temperature appears therefore to be one means of obtaining a low transition temperature, although there are practical limits to the extent to which such control is feasible with existing equipment.

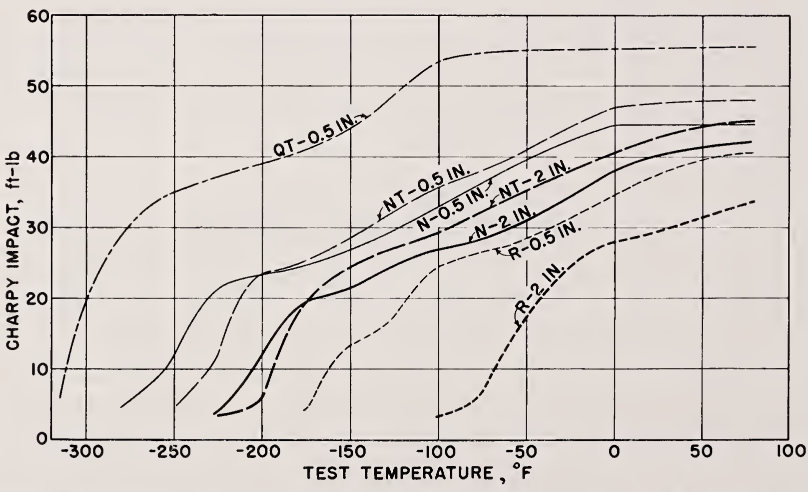

Figure 2.3. *Charpy impact strength (keyhole-notch) of 0.5- and 2-in. plates of 2317 steel (longitudinal specimens) as a function of temperature.*

QT, Quenched and tempered; NT, normalized and tempered; N, normalized; R, as rolled.

(Seens, Miller, and Jensen)

A closely related question is that of deoxidation practice. In a very general way it can be stated that the better the deoxidation the lower the transition temperature, although other factors must sometimes be taken into account. For example, Herty and McBride [10] in their well-known investigation of low-carbon steels showed that the relative position of impact-strength–temperature curves for silicon-killed, rimmed, and semikilled ingots varied somewhat with the heat treatment.

Vanderbeck's data [7] for plate steel in the as-rolled condition are also illuminating in this respect. He reports that killing with silicon plus aluminum produces the lowest transition temperatures, although the manganese content must be high in order to obtain full benefit from this practice. Killing with aluminum alone produced no better results than semikilling.

Seens, Miller, and Jensen [8] have reported that tests made on specimens cut from tubing made of 1022 steel show that the transition temperature is lower when the metal is deoxidized with silicon plus aluminum than when it is deoxidized with aluminum alone. Barr [11] has also commented on the beneficial effect of deoxidation with aluminum but has pointed out certain commercial problems involved, especially in the production of ship plate by such a deoxidation practice.

We come now to the question of chemical composition. It has already been noted that composition has an effect, on hardenability, but it appears to have an influence beyond this. An extensive investigation of the effect of alloying elements on notch toughness of pearlitic steels has been reported by Rinebolt and Harris [12]. Their data include measurements on the influence of aluminum, boron, chromium, copper, manganese, molybdenum, nickel, phosphorus, sulfur, silicon, titanium, or vanadium when added to a base steel containing 0.30 percent of carbon, 1.00 percent of manganese, and 0.30 percent of silicon. All specimens had essentially the same coarse pearlitic structure and the same grain size (ASTM 7–8). It should be noted that their tests were made with Charpy **V**-notched specimens, and that they report the transition temperature determined in four different ways. In discussing their results use will be made of the transition temperature corresponding to the average energy, that is, midway between maximum and minimum energy.

Rinebold and Harris [12] found that a relatively small increase in the content of carbon, molybdenum, or phosphorus is notably effective in raising the transition temperature. The extent to which this behavior is associated with differences in hardness is difficult to determine. Silicon and copper have somewhat less effect on the transition temperature, whereas chromium appears to cause little, if any, change. The effect, if any, of adding small amounts of boron was within the experimental error, so that for practical purposes it can be concluded that boron has no influence under these particular conditions. In these low-hardenability steels in the air-cooled condition, the addition of nickel in amounts up to 3 percent caused a measureable lowering of the transition temperature and manganese produced a much greater effect in the same direction. The notable influence of manganese has been reported by others, as for instance, Barr. The addition of titanium or vanadium up to 0.1 or 0.2 percent increased the transition temperature, but subsequent additions caused a decrease. The effect of aluminum was obscured by its influence on grain size but appears to be small. Likewise, the effect of sulfur was uncertain because of the presence of laminations in the high-sulfur alloys, but it seems reasonably certain that this effect cannot be great. These data also suggest, but do not prove, that the effect of carbon, manganese, and nickel is additive.

Further data on the effect of carbon, manganese, silicon, phosphorus, sulfur, and vanadium on the transition temperature of ¾-in. plates of semikilled steel have been reported by Banta, Frazier, and Lorig, who used both the Navy tear test and the Charpy (keyhole notched) impact test. Although the range of composition covered was not as great as that used by Rinebolt and Harris, the results confirm the marked influence of carbon and phosphorus in raising the transition temperature, the effect being more consistent in the tear test than in the impact test.

They likewise observed that vanadium up to 0.20 percent increased the transition temperature in the Charpy impact test with no indication of the maximum reported by Rinebolt and Harris. In their semikilled steels containing 0.21 to 0.25 percent of carbon and 0.48 to 0.55 percent of manganese, they found that increasing silicon content up to 0.31 percent had no significant effect on the Charpy impact transition temperature. On the other hand, in steels containing about 0.20 percent of carbon and 0.80 to 0.86 percent of manganese, the Charpy impact transition temperature decreased with increasing silicon content, reached a minimum at about 0.16 percent of silicon and then increased. This minimum was not observed by Rinebolt and Harris in their killed steels, in which increasing the silicon content increased the transition temperature. The effect of manganese in decreasing the Charpy impact transition temperature was marked, whereas that of sulfur appeared to be negligible.

In discussing the paper by Banta, Frazier, and Lorig, Halley [13] made two points that bear significantly on the influence of composition on transition temperature. He pointed out that data for alloying elements which should be in solid solution, as reported by investigators using different types of test, are remarkably consistent when considered on the basis of the change in transition temperature produced by each increment of 0.10 percent of alloying element. For manganese there is a decrease of 4 to 10 deg F, for nickel a decrease of 1 to 2 deg F, for phosphorus an increase of 100 to 130 deg F, for vanadium an increase of 80 deg F, and for silicon an increase of 12.5 deg F. In view of the tendency of vanadium to form insoluble carbides, and of manganese to replace iron in iron carbide, a generalization of this kind must be viewed with some caution until the true content of alloying element in solid solution is known. It is, nevertheless, an interesting idea and merits further consideration.

Halley also called attention to the fact that for these elements, which should be in solid solution, those which expand the gamma field lower the transition temperature, whereas those that form a gamma loop raise it. Moreover, the more effective an element is in restricting the range of stability of gamma iron, the greater is its effect in raising the transition temperature. On this basis Halley predicts that tin and antimony should have an effect about equal to that of vanadium, whereas molybdenum should be less effective, and tungsten still less so. The data of Rinebolt and Harris on the effect of molybdenum would appear to confirm this prediction.

It has frequently been stated in the literature that a high ratio of manganese to carbon is beneficial in decreasing the transition temperature of ferritic steels. Barr [11] has reported that the transition temperature of a normalized steel was lowered as much as 108 deg F (60 deg C) by increasing the manganese-carbon ratio from 1.4 to 11.9. The significance of this ratio was also studied by Rinebolt and Harris, whose results, shown in figure 2.4, suggest that the matter is not as simple as it appears at first sight. Their data appear to be represented best by three straight lines, indicating that the absolute level of carbon content is significant, as well as the manganese-carbon ratio.

The beneficial effect of additions of nickel has recently been receiving increased attention. In addition to the data cited above for coarsely pearlitic alloys, Hodge, Manning, and Reichhold have reported that in steel containing 0.02 percent of carbon and constant in grain size, the transition temperature is lowered 60 deg F by the addition of 3.6 percent of nickel.

The recently developed alloy containing 9 percent of nickel appears to be noteworthy. Thus Seens, Miller, and Jensen report that when

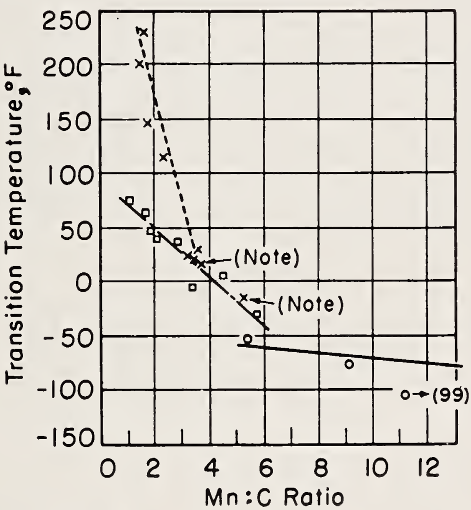

Figure 2.4. *Effect of manganese-carbon ratio on transition temperature of pearlitic steels.*

	Pearlite	Carbon	Yield strength
	Percent	*Percent*	*1,000 psi*
×	40 to 82	0.31 to 0.67	49 to 61
□	18 to 30	.17 to .30	41 to 49
○	0 to 13	.01 to .11	30 to 39

Note.—The carbon content of these points is 0.30 percent, not 0.31 percent.

(Rinebolt and Harris)

quenched and tempered its notch toughness is retained substantially, even at temperatures as low as that of liquid air. Moreover, in the normalized and tempered condition much of its notch toughness is retained at this temperature.

It should be noted that the addition of alloying elements may not only change the transition temperature but may also change the whole shape of the energy-temperature curve. Rinebolt and Harris report that for coarse pearlitic steels, the addition of carbon not only raises the transition temperature but lowers the maximum energy and widens the transition range. An analogous, although less marked, effect was produced by the addition of silicon or phosphorus. Manganese, on the other hand, shifts the whole curve, with no significant change in maximum energy or in the width of the transition range. The mechanism by which the several alloying elements exert their influence on notch toughness is in most cases still uncertain. Some appear to influence the properties of the ferrite, others may cause precipitation.

The foregoing discussion is an attempt to isolate the influence of individual factors but has largely neglected the interactions between these variables. In practice, such interactions are often significant. Moreover, one must take into account other factors such as feasibility of design or relative cost and availability. For example, it must be remembered that a large mass of steel is more prone to brittle fracture than a small mass or a thin section, not only because of the difficulty of obtaining a tought (tempered martensitic) structure, but also because of the increased restraint upon local deformation imposed by the surrounding mass of rigid metal. Again, the notch toughness of a relatively soft material containing coarse pearlite can often be improved by a heat treatment to refine the grain size. The possible effect of heat treatment in stress-relieving or in welding must also be given consideration.

Summary

The suitability of a steel for service at low temperature, judged on the basis of notch toughness as indicated by the transition temperature observed in the Charpy keyhole-notched bar impact test, depends in large measure on its microstructure. The best structure appears to be austenite, which has the lowest transition temperature. In ferritic steels, tempered martensite has the lowest transition temperature, and coarse pearlite the highest. Mixed structures containing ferrite and fine pearlite, bainite, and tempered martensite are intermediate. Another structural factor is grain size, the transition temperature being lower the finer the grains.

In producing steels for low-temperature service there are a number of manufacturing variables that can aid in obtaining a desired structure. These include chemical composition, mode of deoxidation, and mode of heat treatment. The best means of achieving the desired end depends to a considerable extent on the precise conditions of service and on economic factors, such as relative cost and availability.

I express my gratitude to my colleagues, S. C. Snyder and R W. Vanderbeck, for their advice and helpful discussion.

Discussion

Mr. Russell Franks, Manager of Development, Electro Metallurgical Division, Union Carbide & Carbon Corporation, Pittsburgh, Pa.: A few years ago the Union Carbide & Carbon Research Laboratories made a series of impact tests at low temperatures on low-carbon austenitic manganese steels, and the following results are cited to illustrate the influence of alloying elements on the low-temperature toughness of these high-manganese steels. The tests were made on samples taken from forged and annealed 1-in. round bars.

Mn	Cr	Ni	C	Izod impact at −183° C
%	%	%	%	ft-lb
16	--------	--------	0.06	5
16	5	--------	.06	28
16	12	--------	.05	48
16	3	2	.05	40

These results show that the austenitic manganese steel in itself is not particularly tough at −183° C, but as the chromium is increased to 5 and 12 percent the low-temperature impact toughness is increased to a considerable extent. The addition of 3 percent of chromium and 2 percent of nickel greatly improves the low-temperature toughness of the 16-percent manganese steel. It is clear from these results that large percentages of manganese in combination with chromium are instrumental in bringing about high toughness at low temperatures.

Mr. Sam Tour, General Manager, Sam Tour & Co., Inc., New York, N. Y.: The effect of various alloying additions on the impact values of soft steels, such as described by J. B. Austin, should not be extrapolated and applied to steels at higher hardness levels. Free ferrite does not react to alloying additions in the same manner as does tempered martensite. Years of experience with heat-treated spring steels and pneumatic-hammer and chisel steels indicates that some of the manganese-silicon chisel steels at hardness levels of 45 Rockwell C, and harder, are excellent for use at both room and low temperatures. Work completed recently by Sam Tour & Co., Inc. for the Quartermaster Corps on steels for hand tools for use in the Arctic at hardness levels of 45 to 50 and 55 to 60 R_C indicates that molybdenum is a beneficial alloying addition, as well as manganese and silicon.

Many automotive, aircraft, and ordnance parts must be hard to perform their functions properly. If they are to be subject to shock loading at subzero temperatures, they should not be brittle at such subzero temperatures as are expected to be encountered. Carbon and chromium are common alloying elements to obtain required hardnesses but do this at the expense of toughness at low temperatures. Nickel is accepted as an alloying addition to promote toughness at low temperatures but is expensive, is scarce, and does not give hardness directly. The manganese-silicon-molybdenum steels offer great promise for toughness at subzero temperatures in combination with high levels of hardness.

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Dr. Carl A. Zapffe, Research Metallurgist, Baltimore, Md.: How do you distinguish between the effect of plate thickness and grain size in your figure, since both are known to be factors affecting fracture?

Dr. J. W. Spretnak, Associate Professor of Metallurgy, Ohio State University, Columbus, Ohio: Are your results based upon tensile or impact tests?

Mr. E. F. Bailey, Metallurgist, Naval Research Laboratory, Washington, D. C.: Research performed at NRL showed that a 3.5-percent Mn, 2.33-percent Ni, 0.10-percent C steel [14] had austenite formation and decomposition kinetics very similar to the 9-percent Ni steels. This work also demonstrated that retained austenite transformed during the impact test by plastic deformation of the metal, causing a deleterious effect on the transition temperature. This is mentioned to point out that the broad statement "austenite is the best microstructure for impact properties" is not entirely correct, and it would be better to say that fully austenitic steels exhibit the best impact performance.

Mr. J. A. Rinebolt, Metallurgist, Naval Research Laboratory, Washington, D. C.: Further work at the Naval Research Laboratory indicates that 0 to 0.26 percent of Si had but little effect on the transition temperature of normalized steels, but with higher Si content the transition was raised. Other work at the Naval Research Laboratory [15] showed that cobalt is the third alloying element that tends to lower the transition temperature, and it is also one of the elements that widens the gamma loop.

Mr. George F. Comstock, Assistant Director of Research, TAM Div., National Lead Co., Niagara Falls, N. Y.: It should be noted in connection with Dr. Austin's quotation of the ASM paper by Rinebolt and Harris on the "Effect of Alloying Elements on Notch Toughness of Pearlitic Steels" in regard to the effect of titanium on the transition temperature that this statement refers only to steel in the normalized (1,650° F) condition. The effect of titanium on the notch toughness of as-rolled steel is entirely different, as was explained in [16]. In as-rolled steel, titanium around 0.01 or 0.02 percent seems to improve the notch toughness, or lower the transition temperature slightly, but above 0.05 percent it has a serious effect on as-rolled plates in the opposite direction, and plate steels with over 0.05 percent of titanium should be normalized if notch toughness in service is desired.

Dr. J. B. Austin: It is gratifying that the discussion has brought out additional data on the effect of chemical composition on notch toughness at low temperature. More information of this kind is needed, if sound general conclusions are to be drawn.

In view of the remarks by Mr. Franks and by Mr. Bailey, it would appear that my statement classifying austenite as the best microstructure is rather too broad and needs some qualification. The alternative statement proposed by Mr. Bailey seems to be more precise.

In reply to Dr. Spretnak's question, all the data cited are based on impact tests, unless there is a specific statement to the contrary.

With regard to Dr. Zapffe's question, in my opinion, plate thickness and grain size are not usually independent variables. In most cases the significant factor is grain size, which, as is pointed out in the paper, has a considerable effect. The influence of plate thickness enters largely through its effect on the grain size of the steel. The data presented in figures 2.1 and 2.2 show primarily the direct variation in notch toughness with grain size, and, secondarily, illustrate the fact that, in normal processing, grain size varies with plate thickness.

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3. Development and Application of Chromium-Copper-Nickel Steel for Low-Temperature Service

By Walter Crafts[1] and C. M. Offenhauer[2]

An economical low-alloy chromium-copper-nickel steel designed for service at low temperatures has been used successfully for many years in chemical-plant equipment. The steel has required no unusual practices in manufacture or fabrication and has consistently been suitable for use at —80° C, and several heats have been applied at —100° C. Experimental studies under laboratory conditions and observations of commercial heats have suggested that a most significant and immediately applicable improvement of the steel would result from increase in the aluminum content above the amount required for grain refinement to a level of 0.10 to 0.20 percent.

Introduction

In order to meet the needs of the chemical industry for a reliable and economical structural steel in equipment operating at low temperatures, the Union Carbide & Carbon Research Laboratories, Inc. began to investigate the properties of steel at subzero temperatures about 20 years ago. Because of the obviously nonductile character of low-temperature failures and the low notched-bar impact strength of failed steel, this study, in common with collateral work by others, was directed primarily toward notch sensitivity, with the Izod or Charpy test as the main criterion. Although the broader aspects of the low-temperature toughness of steel have made it desirable to study different phases of the general problem continuously ever since, one of the early results was the development of a chromium-copper-nickel steel with good toughness down to at least —80° C. This steel has been applied in several installations operating at temperatures as low as −80° C, but the demand for application at −100° C has led to certain problems. Although most of the heats were suitable for use at −100° C, difficulties with heats not suitable at −100° C have been associated with incomplete grain refinement and indicate a need for more effective treatment with aluminum.

In the early work [1, 2, 3][3] it was found that low-temperature toughness could be produced effectively in economical low-alloy steels by adding balanced amounts of austenite-forming, carbide-forming, and deoxidizing types of alloying elements. It was considered that the basic requirements of such a steel were that it should be low in cost and be conventional with respect to common alloying and melting practices. The rolling, forming, heat-treating, and welding characteristics should also entail no special practices. Finally, it must have adequate toughness, as defined at that time by the ASME Boiler Code for Unfired Pressure Vessels as 15 ft-lb minimum impact strength in the Charpy test with the keyhole notch. These considerations led to the following fine-grained steel composition (in per-

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[1] Chief Metallurgist, Union Carbide & Carbon Research Laboratories, Inc., Niagara Fall, N. Y.
[2] Research Metallurgist, Electro Metallurgical Co., Niagara Falls, N. Y.
[3] Figures in brackets indicate the literature references on p. 66.

cent) : C, 0.12 max; Mn, 0.65 to 0.85; Si, 0.15 to 0.25; Cu, 0.45 to 0.65: Ni, 0.50 to 0.75; Cr, 0.65 to 0.85.

This chromium-copper-nickel steel after normalizing consistently developed high impact strength at −80° C and was also tough after welding and stress relieving. Commercial production began in 1934, and steel of this type has been used up to the present time without failure. Further, as most of the heats developed good properties at −100° C, and in view of a need for steel at that temperature, it has become the practice to divert the better heats to the lower temperature service. As might be expected, this procedure has led to difficulties in borderline cases, especially in parts involving heavier sections. Sometimes these heats can be effectively improved by recourse to special normalizing and stress-relieving treatments. Examination of such steels has confirmed experimental results that the grain size should not only be fine, but that a significant residual aluminum content is desirable to produce toughness at relatively low temperatures.

In view of the increasing demand for low-alloy steels suitable for still lower temperatures, research on the problem has been extended well beyond the limits of the initial investigation, so that data have been obtained over a wide range of compositions from steels of the chromium-copper-nickel type. This work has demonstrated the effects of combinations of alloying elements and some of the principles that govern the extent to which steels, low in strategically critical alloying elements, may be utilized for low-temperature service.

There have been two recent summaries of the effects of individual elements on the toughness of low-alloy steels. Seigle and Brick [4] discussed the elements individually but pointed out that the ideal experiment of adding controlled amounts of single alloying elements to pure ferrite had not yet been performed. This type of experiment was recently reported by Rinebolt and Harris [5], who were able to plot the effects of increasing amounts of the various elements on the transition temperature. The results of the present investigation confirm, for the most part, the trends shown by Rinebolt and Harris.

Review of Experimental Investigation

The original composition, as well as improved modifications of the chromium-copper-nickel steel, was based on empirical testing of small high-frequency furnace heats, followed by larger scale tests in semicommercial or commercial melting furnaces. It has been found that the high-frequency furnace steels tend to be somewhat finer grained and have a characteristic quality that gives them toughness at somewhat lower temperatures than corresponding arc or open-health steels. In some cases, the differences may be very large, but ordinarily it is equivalent to some 10 or 20 deg C. It is also difficult to simulate the mass of a large plate on an experimental scale, and in order to approach a significant result, the laboratory work is usually carried out on 2-in. square bars forged from 3-in. square ingots. The melting practice consists in melting Armco iron, deoxidation with manganese and silicon, addition of the alloys, and final deoxidation. Impact tests were made over a range of temperatures in order to estimate the temperature at which the fracture changes from tough to brittle and the energy drops abruptly. As both tough and brittle fractures occur over a fairly wide range of temperature, it is difficult to estimate the "transition temperature" by a rational criterion, and an arbitrary limit of 15 ft-lb has been used in accordance with commercial practice. However, even this measure of toughness must be interpreted with judgment, as it is statistically probable that, even in those steels that give values in excess of 15 ft-lb at a given temperature, a few surprisingly low values will be encountered if enough tests are made. For this reason, most of the experimental data are reported as the average of a number of tests rather than as the minimum value of a series.

Carbon

Carbon tends to raise the transition temperature. As indicated in table 3.1, normalized experimental steels of the chromium-copper-nickel type were benefited by very low carbon contents.

TABLE 3.1. *Effect of carbon content on low-temperature impact resistance*

[The specimens were normalized at 925° C.]

Chemical composition							Charpy impact at—		
C	Mn	Si	Cu	Ni	Cr	Al	Room temp.	−78° C	−120° C
%	%	%	%	%	%	%	*ft-lb*	*ft-lb*	*ft-lb*
0.008	0.76	0.32	[a] 0.50	[a] 0.75	0.83	[a] 0.10	79 to 80	65 to 84	50 to 56
.014	.55	.10	.57	.86	.87	[a] .10	80 to 85	62 to 82	80 to 90
.064	.83	.35	[a] .50	[a] .75	.89	[a] .10	65 to 67	43 to 47	19 to 27
.090	.84	.34	[a] .50	[a] .75	.87	[a] .10	62 to 62	44 to 44	19 to 31

[a] Approximate analysis.

One reason for the beneficial effect of low carbon, especially in the high-alloy steels, is that tensile strength becomes critical at a room-temperature level of about 75,000 psi. High-strength steels tend to have higher transition temperatures. This imposes a limit on the strengthening types of alloying elements and makes it desirable to have lower carbon contents in steels that tend toward the strength ceiling. In addition to the difficulties of manufacturing very low-carbon steel, a certain amount of carbon is essential to avoid normalizing temperatures that are too high to ensure maintenance of fine grain size.

Manganese

Manganese lowers transition temperature through its influence on ferrite-pearlite "hardenability" and is beneficial in steels of otherwise relatively low alloy content up to somewhat above 1 percent of manganese, as in the aluminum-killed, plain carbon steel used for service to −40° C. Manganese above about 1 percent encourages the retention of austenite. This has an adverse effect on impact strength in the tough range and also raises the transition temperature.

In the transformation of austenite during normalizing, there is first a rejection of ferrite and, ideally for low-temperature toughness, a transformation of the residual austenite to pearlite. When the composition or cooling rate is such as to inhibit pearlite formation, the resulting microstructural constituent has an adverse effect on toughness and is the major factor limiting the amount of alloying elements, especially the carbide formers, that can be employed to lower transition temperature. This austenite tends to transform to martensitic or pseudomartensitic structures, but may be retained in an apparently untransformed condition. Typical examples of the two types of structures are shown in figures 3.1 and 3.2. The appearance of this structure is accompanied by lowered impact strength at room temperature and a higher transition temperature. It is also associated with a moderate increase of strength and a very low ratio of yield strength to tensile strength. The carbide-forming elements (manganese, chromium, and molybdenum) are most powerful in causing the retention of austenite, and also affect the tolerance for noncarbide-forming elements, such as nickel and copper. Grain refinement and high aluminum contents reduce the tendency toward retention of austenite.

Chromium

As in the case of manganese, chromium is beneficial in lowering transition temperature, but above a certain level causes a retention of austenite. This tendency is illustrated in table 3.2 obtained from experimental chromium-copper-nickel steels. The effects of both manganese and chromium are illustrated for comparison in another series of steels. It is notable that, in both the high- and low-man-

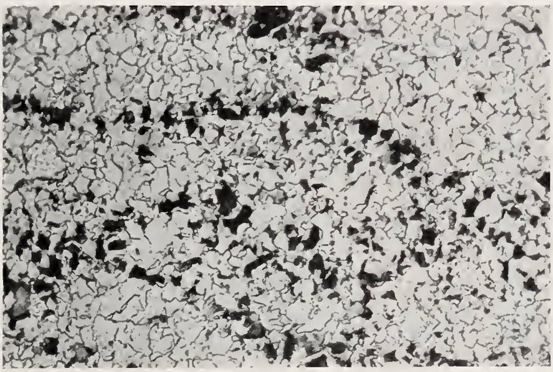

FIGURE 3.1. *Microstructure of steel showing fine-grained ferrite and pearlite, etched with 2% nital, × 250.*

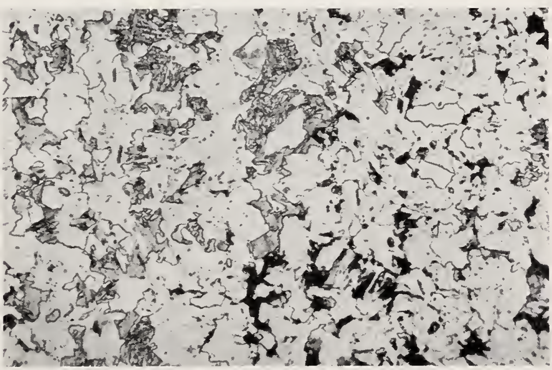

FIGURE 3.2. *Microstructure of steel showing pearlite, pseudomartensite, and ferrite, etched with 2% nital, × 250.*

ganese series, the transition temperature was lowered by less than a critical amount of chromium and that the critical level of chromium was higher in the lower-manganese series (table 3.3). The tendency of chromium toward austenite retention is about half as large as that of manganese, so that about twice as much chromium may be used effectively.

Table 3.2. *Effect of chromium on low-temperature impact resistance*

[The specimens were normalized at 925° C]

Chemical composition							Average values for Charpy impact at—			
Cr	C	Mn	Si	Cu	Ni	Al	−78° C	−100° C	−120° C	−140° C
%	%	%	%	%	%	%	*ft-lb*	*ft-lb*	*ft-lb*	*ft-lb*
0.87	0.09	0.84	0.34	[a] 0.50	[a] 0.75	[a] 0.15	44	----	25	-----
1.06	.08	.73	.19	.40	.75	[a] .15	67	53	42	42
1.31	.11	.71	.25	.41	.71	[a] .15	66	44	39	7

[a] Approximate analysis.

Table 3.3. *Effect of chromium and manganese on low-temperature impact resistance*

[The specimens were normalized at 925° C]

Chemical composition							Average values for Charpy impact at—			
Cr	C	Mn	Si	Cu	Ni	Al	−78° C	−100° C	−120° C	−145° C
%	%	%	%	%	%	%	*ft-lb*	*ft-lb*	*ft-lb*	*ft-lb*
1.07	0.12	0.32	0.23	0.56	0.72	[a] 0.15	63	36	32	32
1.28	.10	.31	.22	.53	.75	[a] .15	47	39	35	32
1.60	.12	.32	.22	.49	.70	[a] .15	38	36	30	11

[a] Approximate analysis.

Molybdenum

Molybdenum is much like manganese in low-temperature steels, but is roughly four times as powerful in retention of austenite. Although small amounts may be beneficial in chromium-copper-nickel steel of the usual composition, molybdenum contents higher than those often found as residuals may raise the transition temperature. Although molybdenum has been used effectively in low-temperature normalized steels, especially in high-nickel—low-manganese types, its utility in the chromium-copper-nickel steel described here is limited to small amounts, as indicated in table 3.4.

Table 3.4. *Effect of molybdenum on low-temperature impact resistance*

[The specimens were normalized at 925° C.]

Chemical composition									Average values for Charpy impact at—	
Mo	C	Mn	Si	Cu	Ni	Cr	Al	N	−160° C	−183° C
%	%	%	%	%	%	%	%	%	*ft-lb*	*ft-lb*
-----	0.045	0.23	0.11	0.51	1.45	0.92	0.18	Low	46	6
0.23	.037	.31	.17	.56	1.45	1.00	.17	Low	5	3
-----	.052	.27	.13	.51	1.48	.98	.16	0.018	55	9
.05	.076	.36	.26	.56	1.49	.76	.23	.017	12	4
.09	.040	.28	.17	.53	1.57	.97	.12	.027	4	3
.14	.071	.30	.22	.55	1.50	.99	.22	.019	6	3

Nickel

Nickel has a strongly beneficial effect on toughness and transition temperature, and, as it is not a carbide-former, produces only minor effects on microstucture. As a result, it has been widely used for low-temperature steels of both simple and complex types. In simple nickel steels that require low transition temperatures after normalizing only, the nickel content is commonly from 2 to 3.5 percent and may extend to 5 percent, if the carbon and manganese are low enough to avoid undue hardening. In complex steels containing strong carbide-forming elements, there is a ceiling to the amount that can be added without inducing excessive austenite retention. Typical values for experimental chronium-copper-nickel steels are given below.

TABLE 3.5. *Effect of nickel in complex steels on low-temperature impact resistance*

[The specimens were normalized at 925° C.]

Chemical composition							Average values for Charpy impact at—		
Ni	C	Mn	Si	Cu	Cr	Al	−120° C.	−140° C.	−160° C.
%	%	%	%	%	%	%	*ft-lb*	*ft-lb*	*ft-lb*
------	0.058	0.45	0.28	0.50	0.74	0.20	5	------	------
0.83	.076	.46	.28	[a] .50	.77	.19	49	36	------
1.51	.088	.48	.33	[a] .50	.76	.20	34	40	54
2.63	.075	.44	.31	[a] .50	.77	.21	16	------	16
3.57	.074	.48	.32	[a] .50	.77	.20	21	17	20

[a] Approximate analysis.

Copper

Copper behaves much like nickel in normalized low-temperature steel, as shown in table 3.6 for experimental chromium-copper-nickel steels. As the nickel content is increased, the tolerance for copper is reduced. Nickel is usually added in conjunction with copper in order to obtain satisfactory surface quality. Although larger amounts are usually beneficial to transition temperature, precipitation hardening effects of copper limit its general utility to about 0.5 percent.

TABLE 3.6. *Effect of copper in complex steels on low-temperature impact resistance*

[The specimens were normalized at 925° C]

Chemical composition							Average values for Charpy impact at—	
Cu	C	Mn	Si	Ni	Cr	Al	−160° C	−183° C
%	%	%	%	%	%	%	*ft-lb*	*ft-lb*
0.52	0.045	0.22	0.05	0.99	0.85	0.13	9	3
.99	.050	.24	.10	.96	.94	.16	47	5
.51	.045	.23	.11	1.45	.92	.18	46	6
1.06	.045	.27	.12	1.45	.97	.15	48	6
.53	.048	.29	.15	1.96	1.00	.26	33	8
1.06	.047	.27	.14	1.93	1.01	.18	13	6

Silicon

Although the primary effect of silicon is unfavorable to low transition temperature, it has secondary effects that sometimes make it a useful component of low-temperature steel in amounts up to 1 percent.

Its principal influence is to protect other deoxidizers and to promote grain refinement. This property is particularly useful in conjunction with strong deoxidizers. Because it raises the critical temperature and tends to restrict austenite formation, it is most useful in the medium (0.10%) carbon, relatively low-alloy types of steels. In chromium-copper-nickel steels, silicon is not added beyond the normal range of steelmaking practice.

Aluminum

Grain refinement is usually essential to development of low transition temperatures, and aluminum is used for this purpose in all but a few special cases. For this effect, a bare minimum of 0.02 or 0.03 percent is usually adequate in fine-grained steel. However, aluminum as an alloying element has a very powerful effect on transition temperature that is seldom utilized and is distinct from its grain-refining effect. Aluminum is beneficial in amounts that exceed the optimum for maximum grain refinement.

The effect of aluminum on low-temperature impact strength in experimental heats of carbon and chromium-copper-nickel steels is given in figure 3.3. Increase of aluminum content from the amount sufficient to ensure grain refinement to the range of 0.10 to 0.20 percent lowered the transition temperature by 40° to 60° C. Since the improvement resulting from relatively high aluminum contents is accompanied by a tendency toward a slight coarsening of the grain size, it would appear that the lower transition temperature results from a true alloying effect of aluminum rather than from a secondary influence on grain size.

Because the most effective aluminum content for the properties in question appeared to be about 0.10 to 0.20 percent, a large part of the

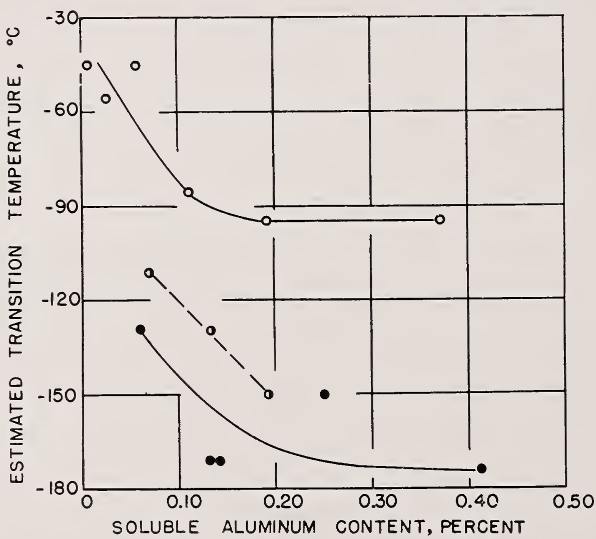

FIGURE 3.3. *Effect of soluble aluminum on the estimated transition temperature of some experimental steels.*

○, Plain carbon steel; ◑, Cr-Cu-Ni steel; ●, modified Cr-Cu-Ni steel.

experimental work has been carried out on steels containing aluminum in that range. It has been observed that high aluminum has a tendency to counteract the retention of austenite in steels containing large amounts of the alloying elements that tend to suppress pearlite formation. It has also been found in a limited amout of experimental work that, if chromium is added to high-aluminum steel as high-nitrogen ferrochrome, the resulting high nitrogen content favors the formation of fine grain size and tends to lower the transition temperature still further.

Service experience with aluminum contents above 0.10 percent in chromium-copper-nickel steel is quite limited. Only one commercial heat of a modified composition is known to have been made with aluminum in that range. This heat contained 0.13 percent of aluminum and, in addition to good low-temperature toughness, was reported to have had a good surface and to have presented no abnormal difficulties in manufacturing or fabricating. As a high aluminum content appears to entail no special manufacturing problems, it would appear to be feasible to utilize 0.10 to 0.20 percent of aluminum in order to effect a substantial lowering of the transition temperature in chromium-copper-nickel steel.

It is apparent that, from the experimental point of view, the chromium-copper-nickel composition is reasonably well-balanced, and that the total alloy content is nearly as high as can be tolerated in a steel that can have no special limitations and no particularly difficult problems in melting, rolling, or fabrication. Further development of experimental steels of this general type for still lower temperatures would require compositions that are somewhat unconventional as compared to commonly used steel compositions, and it is concluded that the chromium-copper-nickel steel is a practically useful steel for general application in the range −80° to −100° C. One modification of practice strongly suggested by the experimental work is the increase in the residual aluminum content to an amount, 0.10 to 0.20 percent, that will produce a significant alloying effect on the lowering of transition temperature.

Commercial Application of Chromium-Copper-Nickel Steel

Chromium-copper-nickel steel has been employed in a number of pressure vessels and other chemical-processing equipment for service at low temperatures. The experience gained in the production and use of this equipment has demonstrated that this steel can be fabricated in the same manner as other steels of this general type. All the equipment was constructed in accordance with the governing ASME Boiler Code requirements and has performed satisfactorily for as long as some 15 years in low-temperature service.

Although the steel was originally recommended for use at temperatures down to −80° C, it was observed that certain heats exhibited satisfactory toughness at −100° C, and these heats were set aside for use at the lower temperature. In a recent group of steels tested for possible use at the lower temperature, about 65 percent exhibited satisfactory properties at −100° C. The chemical compositions and mechanical properties of these steels are shown in table 3.7.

TABLE 3.7. *Chemical composition and mechanical properties of chromium-copper-nickel steel*

Heat Number	Chemical composition [1]								Yield strength	Tensile strength	Elongation in 2 in.	Reduction of area	Charpy (keyhole) impact at—		Remarks
	C	Mn	P	S	Si	Cu	Ni	Cr					−80° C.	−100° C.	
Group I—Steels satisfactory for service at −80° C, but not at −100° C.															
	%	%	%	%	%	%	%	%	*psi*	*psi*	%	%	*ft-lb*	*ft-lb*	
1	0.11	0.66	0.015	0.023	0.25	0.56	0.75	0.69	40,500	68,500	32	----------	----------	2 3 3	Normalized at 900° to 925° C.
2	.11	.75	.020	.040	.23	.53	.65	.74	41,000	70,000	31	----------	----------	2 6 4	Normalized at 900° to 925° C.
3	.09	.49	.009	.020	22	.64	.64	.74	45,100	67,400	41.0	63.3	52 49 45	5 12 6	Charpy specification from 3- by 5-in block. Double normalized at 900° and 885° C. Stress relieved at 650° C.
4	.08	.86	.010	.022	.25	.54	.62	.79	48,000	65,000	36.0	76.1	84 84 87	4 5 15	Heat treatment not reported.
avg	0.10	0.69	0.014	0.026	0.24	0.57	0.66	0.74	43,650	67,700	36.5	69.7	----------	----------	

Group II—Steels satisfactory for service at −100° C.

5	0.11	0.47	0.019	0.029	0.22	0.57	0.63	0.56	37,400 40,600 39,600	66,100 66,300 65,300	30 28.5 29			13 14 11	Specification required minimum 10 ft-lb.
6	.09	.44	.014	.021	.18	.44	.70	.58	40,300	62,300	29			39 28 39	Normalized at 870° to 900° C; stress-relieved 620° to 650° C.
7	.10	.62	.013	.028	.23	.48	.61	.68	46,300	68,000	29.5			29 28 32	Normalized at 870° to 900° C; stress-relieved 620° to 650° C.
8	.10	.65	.015	.030	.20	.52	.68	.67	39,900	63,700	27			31 29 26	Normalized at 870° to 900° C; no stress-relief reported.
9	.10	.65	.013	.030	.21	.52	.81	.72	40,500	67,000	31.0			29 15 25	Normalized at 870° to 900° C; stress-relieved 620° to 650° C.
10	.11	.75	.030	.024	.25	[2] .50	.69	.83	38,800	74,200	34.0			36 29 34	Normalized and stress-relieved; temperatures not reported.
11	.10	.74	.012	.023	.28	.59	.63	.76	61,500	75,000	36.0			64 67 67	Heat treatment not reported.
12	.11	.78	.024	.033	.28	.55	.63	.74	52,600	73,500	26.0			39 28 29	Normalized and stress-relieved; temperatures not reported.
avg	0.10	0.64	0.018	0.027	0.23	0.52	0.66	0.68	43,750	68,100	30.0				

[1] Aluminum not reported.
[2] Approximate analysis—actual not reported.

The average composition and tensile properties of the two groups of steel were almost identical. In spite of the differences in heat treatment, these data suggest that some factor not normally measured is important in contributing to low-temperature toughness. The laboratory work indicates that this may be the aluminum content.

Although most of the low-temperature equipment was fabricated without difficulty, there were a few cases in which unexpected problems arose. A description of some of these problems may be of interest in the light of the experimental findings.

A heat exchanger was required for service at −100° C. A heat of chromium-copper-nickel steel was selected which appeared, in the preliminary tests, to have adequate toughness at this temperature. The chemical composition (in percent) of the steel was reported as C, 0.11; Mn, 0.77; P, 0.023; S, 0.019; Si, 0.21; Cu, 0.60; Ni, 0.47; and Cr, 0.74. During the assembly of the equipment, Charpy impact tests showed that the specimens representing one large flange gave unusually low values at −100° C, and even at −80° C, as follows: At −100° C: 3, 3, 2, 5, 4, 3, 3, and 6 ft-lb; at −80° C: 8, 7, 11, 12, 4, 22, and 8 ft-lb.

An investigation was initiated by the fabricator, and a number of different heat treatments were studied. The results are given in table 3.8.

TABLE 3.8. *Effect of heat treatment on low-temperature impact resistance*

Heat treatment			Charpy impact at −100° C
Normalized at—	Renormalized at—	Stress-relieved at—	
° C	*° C*	*° C*	*ft-lb*
925	--------	--------	22, 7, 6
870	--------	--------	28, 22, 18
980	--------	--------	4, 5, 6
925	870	--------	3, 11, 4
980	870	--------	12, 28, 34
925	870	620	29, 2, 3
870	--------	620	13, 36, 24
980	870	620	25, 29, 33

Check chemical analysis revealed that the steel contained only 0.01 percent of aluminum, and the McQuaid-Ehn grain size was from 1 to 3. This example demonstrated several principles that had been suggested by the laboratory work. The composition of the steel was not abnormal, and the low aluminum content, and the associated coarse grain size were probably responsible for the borderline nature of this steel. The improvement shown as the result of the double normalizing treatment has been observed in other tests, but in the interest of simplicity and economy, it is not generally employed.

In this instance, stress-relieving improved the low-temperature toughness of the steel. In many other cases, however, it has been found that stress-relieving, particularly at temperatures above about 620° C, tended to raise the transition temperature. The results given in table 3.9 illustrate an example of extreme sensitivity to heat treatment.

TABLE 3.9. *Effect of heat treatment on low temperature impact resistance*

Chemical composition							
C	Mn	P	S	Si	Cu	Ni	Cr
% 0.10	% 0.73	% 0.028	% 0.029	% 0.26	% 0.50	% 0.66	% 0.80

Heat treatment	Charpy impact (longitudinal) at −100° C
900° C for 1 hr; 650° C for 1 hr	31, 3, 3, 3, 24 ft-lb
870° C for 1/2 hr; 620° C for 1/2 hr	30, 30, 40 ft-lb

In table 3.10 it may also be noted that stress relieving often tended to reduce the impact resistance as compared to that in the normalized condition. In the experimental work, it was observed that stress relieving often raised the transition temperature, especially in the steels with very low transition temperatures in the normalized condition. On the other hand, stress relieving tended to lower the transition of many steels which exhibited nonpearlitic microstructures. As nearly all welded structures for use at low temperatures are stress relieved, it might appear that only these values need be considered. However, there is always the possibility that a part might not be stress-relieved, and it appears advisable that a material of construction for low-temperature service be tough in the normalized as well as in the normalized and stress-relieved condition. The changes taking place on stress-relieving appear to be a fruitful field for further investigation.

Provided welding electrodes and procedures are selected that will produce weld metal capable of giving adequate low-temperature impact resistance, the weldability of the steel is excellent, as shown in earlier studies [6, 7]. In one case, however, the impact strength of specimens notched in the scarf zone gave low values. This difficulty was traced to an improper heat treatment that had developed a fairly coarse grain size, and a renormalizing treatment provided satisfactory properties in the heat-affected zone, as shown in table 3.11.

Although the effect of mass cannot be satisfactorily studied on a laboratory scale, it is, nevertheless, an important problem in the construction of large equipment. The data presented in table 3.10 were obtained from specimens cut from the center of large-sized samples and also from the more conventional ¾- and 1-in. square bars. It is evident that the impact strength decreased markedly in the heavier sections. It is a matter of some controversy as to whether it is possible to subject very large sections to the type of stress involved in the small notched specimen. Although a discussion of this problem is beyond the scope of this paper, it is clear that the future development of low-temperature steel will be modified according to the properties required in heavy sections.

Finally, it may be in order to cite an example of a commercial steel that embodies many of the advantages indicated by the experimental work. The composition was well-balanced with respect to carbide formers and austenite-forming elements, and it was treated with alum-

TABLE 3.10. *Mass effect in low-temperature steel*

Chemical composition								Source of specimens	Heat treatment	Charpy keyhole impact at—							
C	Mn	P	S	Si	Cu	Ni	Cr			+20° C	0° C	−50° C	−75° C	−80° C	−90° C	−100° C	−110° C
								Heat A[a]									
%	%	%	%	%	%	%	%			*ft-lb*	*ft-lb*	*ft-lb*	*ft-lb*	*ft-lb*	*ft-lb*	*ft-lb*	*ft-lb*
0.09	0.71	0.023	0.012	0.21	0.64	0.60	0.78	¾ by ¾-in. bar	Normalized							46, 50 51, 70 67, 57 56, 57 64, 65 56, 56	
								3 by 5-in. bar from 12- by 12-in. bar.	do							50, 49 47, 46 44, 42 52, 49 39, 12 28, 47 47, 47 48	
								Heat B									
0.10	0.74	0.012	0.023	0.28	0.59	0.63	0.76	1- by 1-in. bars	Normalized at 925° C; renormalized at 900° C; stress-relieved at 650° C.							41.6 43.3 36.8	
								5- by 5-in. bars	do	61.2 58.5 ----	62.0 63.9 62.0	50.0 56.8 63.9	45.0 48.2 2.9	31.2 33.5 33.5	33.5 27.2 2.9	33.5 36.8 27.2 2.9 22.0 2.9	6.3 8.2 6.3 2.1 22.0 2.9
								1½- by 1½-in. bars	Normalized at 900° C							63.9 67.3 67.5	
								8-in.-thick plate	do							7.2 24.4 2.9	
								7-in.-thick plate	do							30.2 30.2 24.4	
								do	Normalized at 900° C. Stress-relieved 675° C.							2.9 2.9 2.9	
								do	Normalized at 900° C. Renormalized at 900° C. Stress-relieved at 675° C.							6.3 4.5 6.3	

Heat C

0.08	0.86	0.010	0.022	0.25	0.54	0.62	0.79	1- by 1-in. bar	Normalized at 925° C; renormalized at 900° C; stress-relieved at 650° C.							47.6
																43.3
																43.3
								3- by 3-in. bar	Normalized at 900° C							62.0
																9.2
																48.2
								do	Normalized at 900° C; stress-relieved at 650° C.							1.2
																2.9
																2.9
								do	Normalized at 900° C; stress-relieved at 650° C; and again stress-relieved at 650° C.							40.0
																41.6
																53.3
								4- by 4-in. bar	Normalized at 900° C							24.4
																45.0
																45.0
								5- by 5-in bar	Normalized at 900° C; stress-relieved at 730° C.							4.5
																14.7
																2.9
								do	Normalized at 925° C; renormalized at 900° C; stress-relieved at 650° C.	67.3	67.3	50.0	63.0	6.3	40.0	1.2
										65.5	67.3	40.0	41.6	19.2	18.0	3.7
											70.9	43.3	51.8	2.9	7.2	4.5
														50.0	2.9	2.9
														46.7	2.9	2.9
														40.0	8.2	2.9
															36.8	
															30.2	
															41.6	

[a] McQuaid-Ehn grain size 7.

inum in such a way that a fairly large amount of soluble aluminum was retained. The results of the mechanical tests on this steel are given in table 3.12, and it appears that the transition temperature was below −145° C.

TABLE 3.11. *Effect of renormalizing on Izod impact resistance of welded plates*

Plate thickness	Condition	Chemical composition C	Mn	Si	Cu	Ni	Cr
in.		%	%	%	%	%	%
15/16		0.09	0.68	0.18	0.54	0.62	0.95
1 9/16		.09	.74	.22	.53	.63	.74

Plate thickness	Condition	Average Izod impact values for heat-affected zone at [1]— Room temperature	−50° C	−78° C	−100° C	−120° C
		ft-lb	*ft-lb*	*ft-lb*	*ft-lb*	*ft-lb*
15/16	Before renormalizing	78	53	16	3	3
15/16	After renormalizing	82	75	55	30	22
1 9/16	Before renormalizing	76	27	10	3	3
1 9/16	After renormalizing	82	83	89	72	65

[1] All specimens were stress-relieved for 2 hrs. at 650° C.

TABLE 3.12. *Mechanical properties of commercial chromium-copper-nickel steel*

Chemical composition C	Mn	P	S	Si	Cu	Ni	Cr	Soluble Al	Insoluble Al
%	%	%	%	%	%	%	%	%	%
0.10	0.65	0.013	0.031	0.23	0.48	0.61	0.68	0.050	0.002

Heat treatment	Thickness	Tensile properties (average of two tests) Yield strength	Tensile strength	Elongation in 1½ in.	Reduction of area
	in.	*psi*	*psi*	%	%
As rolled	1/2	48,200	65,900	39.3	75.0
Do	1	41,150	65,400	38.0	70.4
Normalized at 925° C	1/2	45,400	69,050	38.6	73.0
Do	1	41,800	68,600	38.0	69.3

Heat treatment	Thickness	Values of Charpy (keyhole) impact at — −78° C	−100° C	−120° C	−145° C	−180° C
		ft-lb	*ft-lb*	*ft-lb*	*ft-lb*	*ft-lb*
Normalized at 925° C	1/2	53.5	60.0	42.5	39.5	4.5

Summary

The development and application of the chromium-copper-nickel steel has demonstrated that it is possible to obtained adequate impact resistance at low temperatures in low-alloy pearlitic steels. The evaluation of the data presented in this paper indicates that carbon and silicon tend to raise the transition temperature, whereas manganese, chromium, nickel, and copper tend to lower it within the limits of a pearlitic microstructure. However, the most striking result is that a high aluminum content of the steel is one of the most effective means of lowering the transition temperature.

The chromium-copper-nickel steel contains a relatively small amount of strategically critical alloying elements, requires no special manufacturing and fabricating practices, and is consistently serviceable at −80° C. Application at −100° C requires treatment with enough aluminum to ensure a persistently fine grain size. When this condition has been produced effectively, the steel has met the requirements and served successfully at −100° C.

Further lowering of the service temperature without changing the foolproof characteristics of the steel for the application here discussed appears to be possible, if the aluminum content is raised to a level approaching 0.20 percent. Although it has not been common practice to use such high aluminum contents, there do not appear to be any serious difficulties involved, and a significant saving of alloys would be realized.

Discussion

Dr. C. B. Post, Chief Metallurgist, Carpenter Steel Co., Reading, Pa.: Mr. Crafts, why wouldn't SAE 3110 be all right? What is the copper doing that a little bit more nickel would not do in SAE 3110?

Mr. C. M. Offenhauer: The combination of chromium, copper, and nickel appeared, on the basis of our data, to permit the lowest transition temperatures with a minimum alloy content. The copper is a substitute for a part of the nickel which would otherwise be required to obtain the desired transition temperature.

Dr. Post: Well, then you evidently have other data you don't show here. The data do not show it up here. The effect of copper seemed to be the same as nickel so you would suppose that around 1.00 to 1.25 percent nickel and 0.75 percent of chromium would be the same steel. I thought maybe you had the copper in there for some pitting effect on exposure to salt water, or something like that.

Mr. Walter Crafts: Well, in general that is true. The copper has a little edge over the nickel but not enough to make a great deal of difference. You could use nickel for copper. It might take a little more nickel than so-much copper but beyond that you would expect to get about the same results.

Dr. J. M. Hodge, Development Engineer, United States Steel Co., Pittsburgh, Pa.: I would like to ask about your criterion of transition temperature in impact testing. Did you find a scatter band or pick off the point at which the impact temperature curve showed a certain foot pound value?

Mr. Offenhauer: We tested normalized Charpy keyhole specimens at 20-deg C intervals, drew a curve through the results and established the transition temperature at 15 ft-lb on this curve. The transition temperature measurement is no closer than plus or minus 15 deg C.

Dr. H. S. Blumberg, Division Head, Materials Engineering Division, Special Projects Department, M. W. Kellog Co., Jersey City,

N. J.: We have followed the development and application of this steel for nearly 20 years. My recollection is that when this steel was first developed, it was shown in laboratory tests that chromium and copper alone were sufficient to give satisfactory impact values down to −148° F. I believe that the very first heats made of this composition were steels containing only chromium and copper as alloying elements. Subsequently, nickel was introduced for another purpose that has not been mentioned today, and that is to improve the surface of the material. Without the presence of nickel, a scabby surface was found on plates, with a tendency toward segregation of copper at the surface.

The foregoing experience with respect to a composition designed from laboratory experience leads to a note of caution, which can be directed toward the studies of steels, about which we have heard considerable in several of the papers today. Some of these "new" compositions will probably appear very promising, based on laboratory tests, but until large-scale heats are melted and processed, any proposed steel should be considered in a development stage.

When the company, with which the writer is connected, first fabricated pressure vessels from the early Cr-Cu-Ni steel, data were obtained relating the normalizing temperature, microstructure, and longitudinal and transverse Charpy impact values at several temperatures down to −148° F. These data are shown in figure 3.4. These tests represent production material in plate form made from a 100-ton open-hearth heat. When normalizing temperatures are in the range 1,600° to 1,750° F, values of over 25 ft-lb are obtained down to −148° F. As normalizing temperature is raised, lower values are obtained, so that the transition temperature is raised. I offer this data because some people here may not be familiar with this steel. The foregoing refers to Cr-Cu-Ni steel with little or no aluminum additions. The modification proposed by these authors should also serve to widen the temperature range in which normalizing can be done, without lowering the transition temperature.

Mr. Crafts: I can confirm your story about the first heats being chrome-copper. It is true they were very scabby and, in addition, the impact strength and plate sections were not as low as we expected, so we put in nickel and got both surface and low temperature.

Dr. Post: My next question then is why did you get started with the chrome-coppers? I am still mystified about why you have copper in there.

Mr. Crafts: As pointed out in the original publication [8] on this type of steel, copper was shown to have a strong influence on low-temperature toughness, especially in certain combinations with other alloys. In addition, some impact specimens were heat-treated after machining, so that the notch was oxidized, and these specimens of chromium-copper steel gave much higher values than specimens of similar size that were notched after heat treatment. This was not true of some other steel, so that it was felt that the copper-bearing steel offered an extra factor of safety. The significance of this observation has been minimized by later concepts of brittle failure. Although recent work has indicated that nickel might be substituted for copper, experience has demonstrated that such substitution is not required and that copper is a useful component of the steel.

Analysis C 0.10 Mn 0.68 Si 0.21 Cr 0.77 Cu 0.58 Ni 0.63		Charpy Impact Averages, Ft-Lb Temperature of Testing									
Condition	100 × Micrographs Etched in 4% Alcoholic HNO_3	+70°F		-50°F		-75°F		-112°F		-148°F	
		Long.	Tran.	Long.	Tran.	Long.	Tran.	Long.	Tran.	Long.	Tran.
Stress Relieved 1150° F		66.2	57.3	49.6	38.3	44.9	44.9	38.2	31.2	14.4	19.3
Normalized 1600° F, Stress Relieved 1150° F		64.3	56.4	60.3	41.4	47.1	37.1	43.3	36.2	31.7	26.5
Normalized 1650° F, Stress Relieved 1150° F		60.6	52.1	54.2	39.4	51.6	35.7	38.6	33.1	29.9	24.9
Normalized 1700° F, Stress Relieved 1150° F		61.9	50.5	44.8	33.8	41.9	29.7	37.2	28.8	33.4	20.3
Normalized 1750° F, Stress Relieved 1150° F		63.5	52.5	48.2	38.1	42.5	28.9	36.7	30.8	30.8	24.8
Normalized 1800° F, Stress Relieved 1150° F		63.4	50.8	47.5	37.1	38.4	32.7	37.8	24.7	29.4	3.9
Normalized 1850° F, Stress Relieved 1150° F		60.7	54.4	40.5	33.9	38.9	31.4	34.4	14.0	4.7	1.8
Normalized 1900° F, Stress Relieved 1150° F		58.5	47.9	37.7	28.2	33.9	29.4	10.3	28.2	1.9	1.7
Weld Stress Relieved 1150° F,		23.6		19.4		17.3		16.3		10.8	
Zone Adjacent to Weld Stress Relieved 1150° F,		50.7		35.6		48.2		44.4		37.2	

FIGURE 3.4. *Microstructure and Charpy impact values at various subzero temperatures obtained on $1^{15}/_{16}$-in-thick chromium-copper-nickel steel plate after various heat treatments and welding, including data on weld and adjacent zone. The micrographs are reduced to ½ size.*

MR. S. EPSTEIN, Research Engineer, Bethlehem Steel Co., Bethlehem, Pa.: Isn't it a fact that such steels first arose as low-alloy high-strength steels in which atmospheric corrosion resistance was important. In that case the presence of about 0.5 percent of copper is beneficial. But this much copper will cause hot-rolling difficulties in the steel, particularly surface checking. Nickel was found to overcome this difficulty. Thus, when copper was used in these steels, nickel was also used.

DR. BLUMBERG: Perhaps I can answer Dr. Post's question about the presence of nickel in this steel from the viewpoint of the vessel fabricator and user. Pressure vessels usually require a large poundage of plate material, so that even a small addition of an alloying element may result in a considerable increase in cost. Economically, therefore, it would be preferred to use the chromium-copper steel without the addition of nickel. However, as already pointed out, the presence of nickel in the steel is necessary to give satisfactory plate surfaces after rolling. Mr. Crafts has already told us that nickel also improves the notched-bar impact characteristics of the material.

Mr. P. H. Brace, Consulting Metallurgist, Westinghouse Research Laboratories, East Pittsburgh, Pa.: What is the present commercial status of the complex steel carrying the aluminum? Is the aluminum steel a commercial article at the present time?

Mr. Crafts: No, a few heats have been made with higher than normal aluminum, but we have yet to get the high aluminum variation put through the mill processing enough to know whether it can be relied on from heat to heat and that we are positive we will not get into any abnormal trouble.

Mr. Epstein: I would like to ask what the yield strength was of these steels and whether in determining the yield strength a straight stress-strain curve to a drop of the beam was obtained or a rounded stress-strain curve. In these alloys a rounded stress-strain curve is usually accompanied by lower impact resistance. A rounded stress-strain curve thus becomes a good tell-tale—more easily recognizable usually than the microstructure—of the possibility that the impact properties may be lower.

Mr. Crafts: The yield point of the chromium-copper-nickel steel is characterized by a drop of the beam. The yield strength appears to be a significant indicator of the presence of more than an optimum amount of alloys. The appearance of the nonpearlitic constituent results in a marked lowering of yield strength, sometimes to 20,000 psi. As it is very difficult to identify the nonpearlitic constituent positively, the yield strength is a useful criterion.

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4. Tensile Properties of Copper, Nickel, and Some Copper-Nickel Alloys at Low Temperatures

By Glenn W. Geil[1] and Nesbit L. Carwile[1]

Tension tests were made at −196°, −140°, −78°, −30°, +25°, and +100° C on high-conductivity copper, cupro-nickel, 70-percent-nickel–30-percent-copper alloy, and nickel in the annealed condition. True stress-strain curves were determined from numerous simultaneous load and diameter measurements made during the entire course of each test to fracture.

Graphs are presented showing the influence of the testing temperature and chemical composition of the alloy on the mechanical properties and work-hardening characteristics. The decrease in the rates of work-hardening with increase in the strain during the tension tests at the various temperatures is discussed. The effect of the solute content of the alloys on the initial strength, ultimate tensile stress, and on the true stress and true strain at maximum load and at fracture is briefly discussed.

Introduction

In recent studies of the rheological characteristics of metals and alloys, physicists and metallurgists have recognized the importance of the true stress-strain relationship for metals tested in tension under controlled test conditions. In previous papers Ludwik [1],[2] MacGregor [2], Voce [3], and the authors [4, 5] have discussed in detail the procedures for obtaining true stress-strain curves and the advantage of such tests. In these studies the true stress, σ, is the current load divided by the current minimum cross-sectional area of the specimen, and the true strain, δ, is $\log_e A_o/A$, in which A_o and A are the initial and current minimum cross-sectional areas of the specimen, respectively. These true stress and true strain values are the average values of these quantities across the minimum cross-sectional area of the specimen.

If the deformation of a tensile specimen is uniform throughout the gage length and the volume is assumed to remain constant, the true strain also can be represented by $\log_e L/L_o$, in which L_o and L are the initial and current gage lengths. However, as pointed out previously by the authors [4, 5] local contraction can occur in certain metals tested in tension before the maximum load is reached. Consequently, it is the authors' belief that simultaneous measurements of the load and diameter during the entire tension test are essential for determining accurately the true stress-strain relationship.

In a recent paper [6] presenting a mechanism for the deformation of metals, Frederickson and Eyring state:

> In practically all cases the curve of the stress-strain relationship from the maximum load point to the point of fracture is a straight line for low temperatures and rates of strain greater than creep rate. This linearity of stress to strain has been assumed to hold for all temperatures; consequently, many investigators do not attempt to obtain stress measurements much beyond the maximum load point. The stress at fracture is determined

[1] Metallurgists, National Bureau of Standards.

[2] Figures in brackets indicate the literature references on p. 95.

by the intersection of the best straight line through the points beyond the maximum load point and the strain at fracture. For low temperatures and relatively high rates of strain, this approximation is valid for the fracture stress.

Many investigators have assumed such linearity, and their results were discussed from this viewpoint. Moreover, some investigators have presented experimental data that apparently conform to this relationship. However, other reported data, including those of the authors [5], indicate that a linear relationship does not necessarily exist between true stress and true strain for loads ranging from maximum to fracture of metals tested in tension at room and subzero temperatures.

A parabolic true stress-strain relationship for metals and alloys tested in unidirectional tension has been reported by various investigators for deformation between yield and maximum load. This relationship apparently has been almost universally accepted as most nearly conforming to data obtained and can be expressed as

$$\sigma = b\delta^m \tag{1}$$

in which σ is the true stress, δ is the true strain, and b and m are constants. The exponent m is called the work-hardening exponent. According to this equation, graphs of true stress versus true strain plotted on logarithmic coordinates should be linear for the deformation between yield and maximum load with the slope of the line equal to m. However, some data obtained in tension tests with ingot iron [5], aluminum [7], copper [3, 8] and other nonferrous metals [8] do not conform to this relationship.

Recently some formulas [3, 9, 10, 11] have been postulated for the true stress-strain relationship for metals or alloys tested in unidirectional tension or compression. As most of these formulas are relatively complex they cannot be easily verified by experimental data.

The purpose of the present investigation was to determine the influence of the testing temperature on the true stress-strain relationship, work-hardening characteristics, and other mechanical properties of copper, nickel, and two solid-solution alloys of these metals tested in tension in the temperature range of −196° to +100° C. It was also proposed to study the conformance of these data with some of the true stress-strain relationships postulated by other investigators.

Materials and Testing Procedures

The materials used in this investigation were high-purity copper, high-purity nickel, high-purity 70%-Ni–30%-Cu alloy, and 70%-Cu–30%-Ni alloy (cupro-nickel). These bars had been hot-worked and subsequently annealed during manufacture. The chemical compositions of these metals are given in table 4.1. The oxygen-free high-conductivity copper contained 99.99+ percent of copper, as determined by chemical analysis. The arc spectrum of the copper was examined, and lines for Ag, Al, Mg, and Si were identified. There was some indication of the presence of Fe, Ni, and Pb. The chemical analysis of the cupro-nickel indicates that it is comparable to that of commercial cupro-nickel. A special heat was processed for the high-purity 70%-Ni–30%-Cu alloy. Its purity was considerably higher than that of the commercial alloy (Monel) of the same nominal composition.

TABLE 4.1. *Analysis of materials (percentage by weight)*[1]

Material	Chemical composition																	
	Cu	Ni	Fe	Mn	Co	Zn	Si	C	S	P	Pb	Sn	Cr	Ti	Al	Ag	Mg	Ca
OFHC copper	99.99+	FT	FT	----	ND	ND	FT	----	----	----	FT	ND	----	ND	FT	VW	FT	----
Cupro-nickel	68.84	29.89	0.50	0.65	0.04	0.09	0.003	0.023	0.004	ND	VW	T	VW	----	T	VW	VW	FT
70%-Ni-30%-Cu alloy	29.71	70.08	.01	.01	ND	ND	.12	.017	.002	ND	ND	ND	T	VW	FT	T	VW	FT
Nickel	0.009	99.85	.04	.03	ND	ND	.11	.007	.002	ND	ND	ND	VW	ND	FT	ND	VW	FT

[1] Chemical analysis determinations indicated by numerical values. Spectrographic analysis determinations indicated by VW, very weak; T, trace; FT, faint trace; ND, not detected.

The purity of the nickel (99.85 percent of nickel), although somewhat higher than that of typical commercial nickel, was, however, considerably less than that of the copper; principal impurities were Fe, Mn, and Si. The copper was bright annealed at 800° F for 1 hour; the nickel and the 70%-Ni–30%-Cu alloy were annealed 16 hours at 1,100° F, followed by heating 8 hours at 1,000° F. The annealing treatment of the cupro-nickel was not reported by the manufacturer.

Some of the properties of these metals and alloys at room temperature are given in table 4.2. All the materials were of fine to medium grain size; the average grain diameter ranged from 0.02 mm for the 70%-Ni–30%-Cu alloy to 0.045 mm for the nickel. The microstructure of these materials in the initial (annealed) condition is shown in figure 4.1.

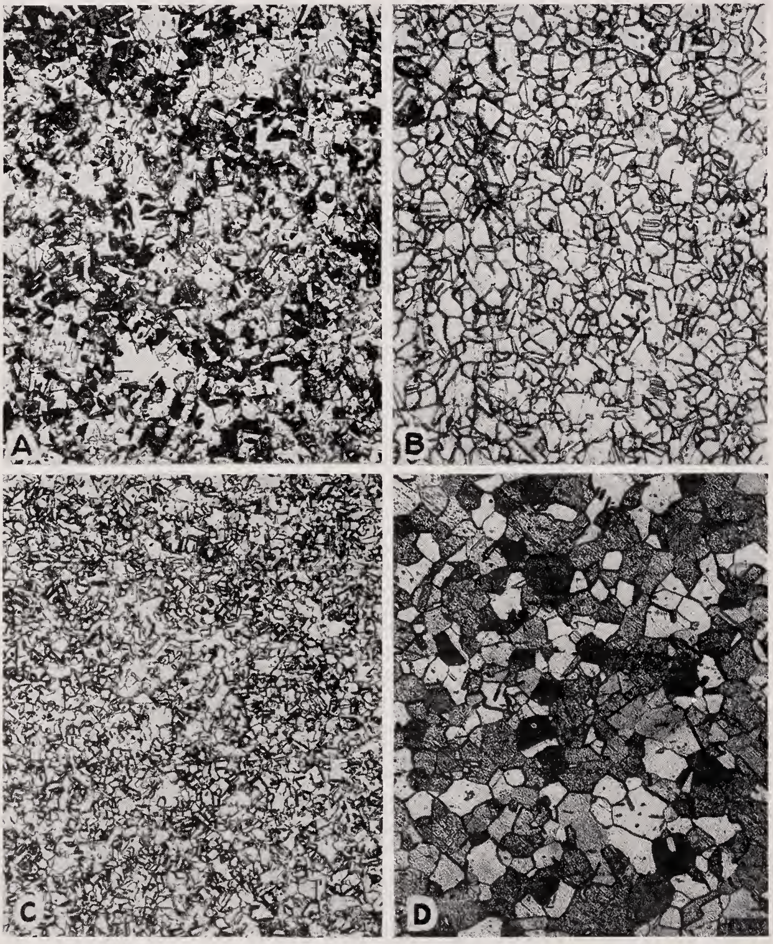

FIGURE 4.1. *Longitudinal structure of annealed specimens,* $\times 75$.

A, Copper etched in equal parts NH_4OH and H_2O_2 (3 percent); B, cupro-nickel, etched electrolytically in 10 percent HNO_3, 5 percent glacial $HC_2H_3O_2$ solution; C, 70-percent Ni-30-percent Cu alloy, etched in equal parts HNO_3 and glacial $HC_2H_3O_2$; D, nickel, etched in equal parts HNO_3 and glacial $HC_2H_3O_2$.

Table 4.2. *Mechanical properties of materials at room temperature*

Material	Average grain diameter	Hardness, Rockwell scale	True stress at a total true strain of 0.0025	Nominal ultimate stress	True stress at maximum load	True breaking stress	Reduction of area at—		
							Maximum load	Initial fracture	Final fracture
	mm		*lb/in.²*	*lb/in.²*	*lb/in.²*	*lb/in.²*	%	%	%
OFHC copper	0.025	34 R_F	7.5×10^3	31.8×10^3	45.6×10^3	81.3×10^3	30	79	88
Cupro-nickel	.035	48 R_B	21.5	54.0	73.2	124.7	26	75	81
70%-Ni-30%-Cu alloy	.020	62 R_B	[b]28.0	67.5	94.0	156.8	28	76	88
Nickel	.045	32 R_B	12.0	51.8	75.1	144.1	30	80	87

[a] At initial fracture.
[b] This also is the value of the true stress at lower yield point.

Cylindrical tensile specimens with a reduced section of 2-in. gage length were used. The reduced section was gradually tapered from each end; the diameter of the midsection was 0.505 in. and was about 0.003 in. less than that of the ends. The specimens were finished to the final dimensions by grinding and polishing in the axial direction, to avoid circumferential tool marks. The ends of the specimens were machined with ¾-in.×10 threads, and the shoulder fillets were machined to a radius of 0.75 in.

The method of testing and maintenance of the specimens at the desired test temperatures was the same as that described in detail in a previous paper [5]. The specimens during the tension test, except those tested at room temperature, were fully immersed in an appropriate liquid contained in an insulated vessel and maintained at the desired temperature. The rate of loading was less than is usual in a tension test to avoid local heating of the specimens during the plastic deformation. The loading was so controlled that the rate of contraction of the specimens beyond the initial yielding was maintained at approximately 1-percent reduction of area per minute.

A reduction of area gage of special design [4], which could function satisfactorily while largely submerged in the bath, was used for determining the change in the minimum diameter of the specimen during the tension tests at subzero and at moderately elevated temperatures. Changes in the diameter of a cylindrical specimen can be measured by this gage with an accuracy of ± 0.0001 in.

True Stress-Strain Curves

The true stress-strain curves obtained for specimens tested in tension at temperatures ranging from $-196°$ to $+100°$ C, are given in figures 4.2 to 4.5. Fracture of an unnotched specimen of a ductile metal tested in ordinary tension at room, subzero, or moderately elevated temperatures is usually initiated at its axis and the crack propagates toward the periphery—the metal at the periphery extends longitudinally and contracts radially during the fracture. The true stress and true strain values at the initial rupture or fracture of the specimen, represented by squares in figures 4.2 to 4.5, will be designated in this paper as the fracture stresses and true strains at fracture, whereas the crosses represent the values usually reported as the true breaking stresses and strains at fracture. These latter values are based on the minimum area of the specimen as determined from diameter measurements made after the specimen has been fractured. The "rim effect" varies with chemical composition, structure, and condition of the metal, as well as with the temperature of the specimen during the tension test. Thus true breaking stress and true strain values determined from diameter measurements made after complete fracture of a specimen of a ductile metal may be of little physical significance and can be used erroneously. For example, the curve is drawn by some investigators as a straight line through the point representing the true stress and true strain values at maximum load and the point representing these values at fracture based on the measurements made after the fracture.

The true stress-strain curves obtained in the present investigation for specimens of copper, nickel, and copper-nickel alloys, tested in tension at temperatures ranging from −196° to +100° C, are not linear between the points representing maximum load and fracture; they are curvilinear from the point of initial plastic flow to fracture. The curvature of the portion of the true stress-strain curve representing deformation beyond that at the maximum load is about the same for all of the tests with these metals and apparently is independent of the temperature.

Results reported in a previous investigation [5] on the true stress-strain relationship for ingot iron tested in tension at temperatures ranging from −196° to +100° C indicated an increasing curvature in the portion of the curves between the points representing maximum load and initial fracture with a decrease in test temperature. The true strain at initial fracture of the ingot iron generally decreased greatly with a decrease in test temperature below room for certain conditions of the iron or below −30° C for other conditions. Furthermore, the strain aging of the iron had an appreciable effect on the shape of the curves. The true strain at initial fracture of the metals studied in the present investigation, however, did not vary greatly with the testing temperature (figs. 4.2 to 4.5), and strain aging during the tension test is not believed to be present to any appreciable extent with perhaps one exception. The tension test at 100° C with nickel (fig. 4.5) indicated that some strain aging may have occurred during the initial part of the deformation; the true stress-strain curve for this test lies above the curve for the specimen tested at 23° C at small strain values, and then lies below the curve for the test at 23° C at true strain values greater than 0.1. All the other true stress-strain curves in these figures exhibit the general trend of an increase in the resistance of the metal to plastic deformation with a decrease in testing temperature.

Influence of Temperature on the Work-Hardening Characteristics

As the term "work-hardening" relates to the increase of the true stress with the increase in the true strain of a metal, the slope of the true stress-strain curve, $d\sigma/d\delta$, at any strain is a measure of the instantaneous rate of work-hardening at that strain. The work-hardening is the combined effect of the ordinary strain-hardening and any aging or recovery during the tension test.

If the parabolic true stress-strain relationship given in eq 1 is assumed, the rate of work-hardening is given by

$$\frac{d\sigma}{d\delta}=m\frac{\sigma}{\delta} \tag{2}$$

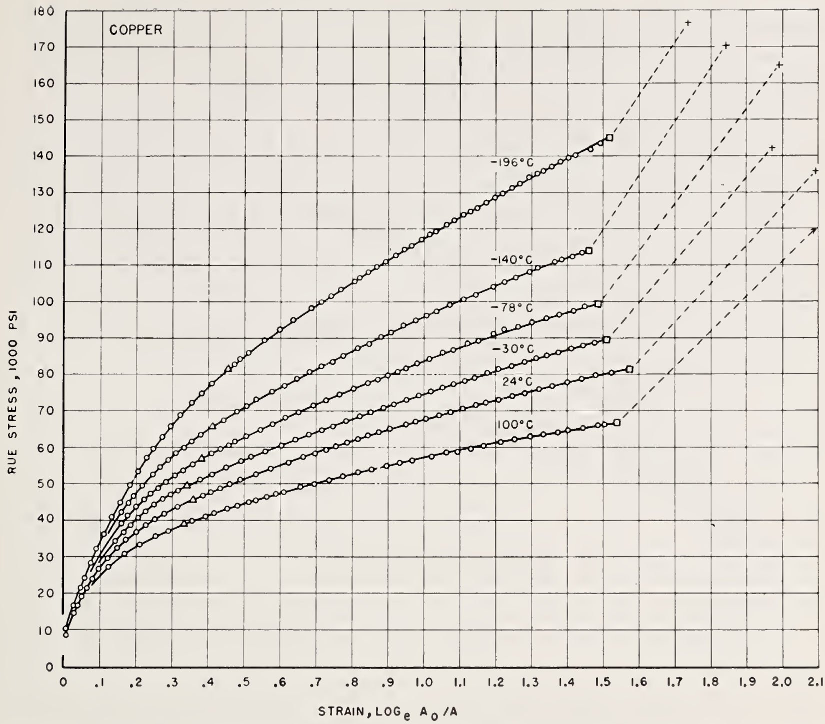

FIGURE 4.2. *True stress-strain curves obtained in tension tests at different temperatures with annealed copper.*

△, Maximum load; □, initial fracture (based upon current minimum diameter at initiation of fracture); +, final fracture (based upon minimum diameter of specimen after fracture).

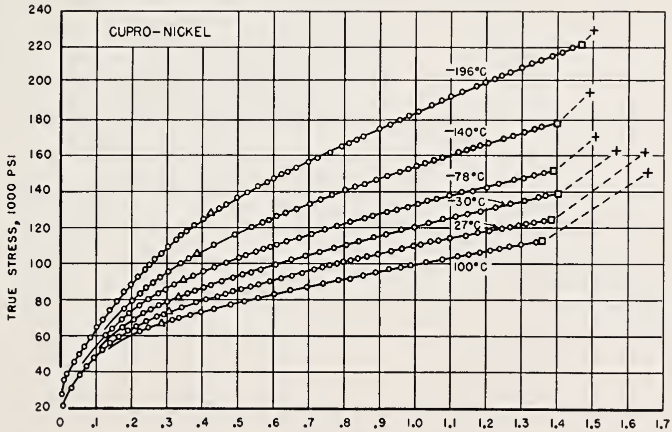

FIGURE 4.3. *True stress-strain curves obtained in tension tests at different temperatures with annealed cupro-nickel.*

△, Maximum load; □, initial fracture; +, final fracture.

According to eq 1, graphs of true stress versus true strain, plotted on logarithmic coordinates, should be linear for deformation between the end of initial yielding and maximum load with the slope of the line equal to *m*. Thus, the rate of work-hardening at any strain δ,

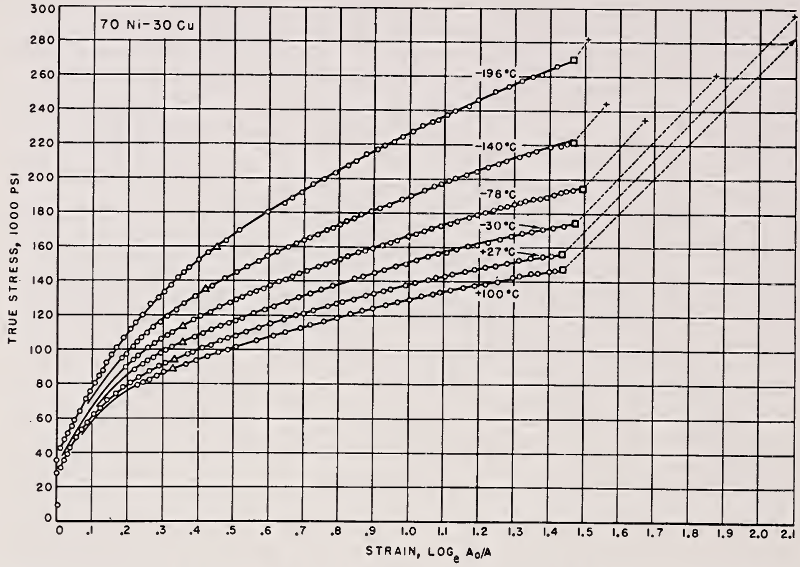

FIGURE 4.4 *True stress-strain curves obtained in tension tests at different temperatures with annealed 70-percent-*Ni*-30-percent-*Cu *alloy.*

△, Maximum load; □, initial fracture; +, final fracture.

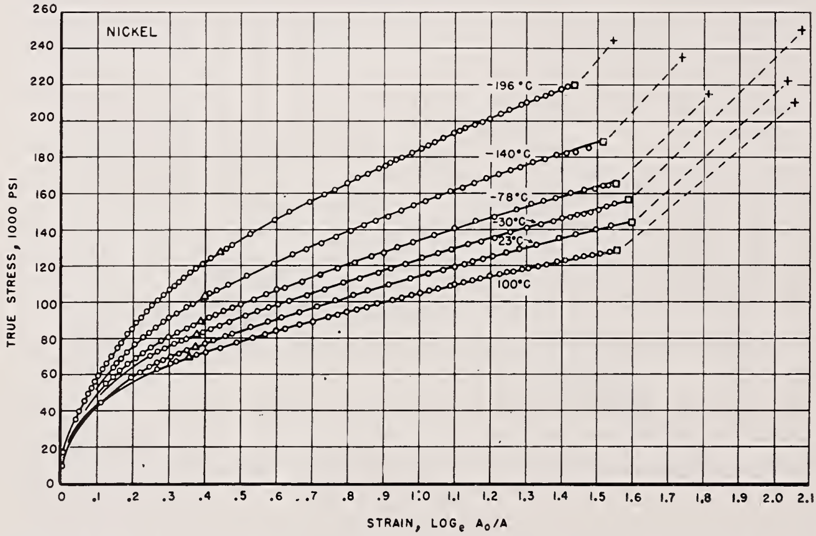

FIGURE 4.5. *True stress-strain curves obtained in tension tests at different temperatures with annealed nickel.*

△, Maximum load; □, initial fracture; +, final fracture.

should be readily determinable from the value of σ at that strain and the slope of the linear logarithmic true stress-strain curve.

The slope of a logarithmic true stress-strain curve at the maximum-load point and also the slopes at the upper and lower yield points, if present, are equal to the respective true strains of the metal at these points. Moreover, the rate of work-hardening of a metal at the upper and lower yield points and at the maximum load is equal to the respective true stresses at these points. These latter two relationships do not depend upon any specified true stress-strain relationhip [5].

The true stress-strain data obtained in the present investigation plotted on logarithmic coordinates, are summarized in figures 4.6 to 4.9. The dashed lines in these figures are drawn through the points representing the true stress-strain values at maximum load, with the slopes of the straight lines equal to the true strains at the maximum load. (It should be noted that the length of a logarithmic cycle on the scale of true stress is twice that for the scale of true strain.)

The true stress-strain values in the immediate vicinity of the maximum load conform very closely to these dashed lines as expected, because the slope of the curve at the maximum-load value must be equal to the true strain at the maximum load. However, the true stress-strain values throughout the remainder of the strain range deviate greatly from this linear relationship, as indicated by the sigmoidal form of the curves. The data obtained in tension tests with these metals at temperatures of $-30°$ and $-140°$ C (not shown in these figures) conformed to logarithmic true stress-strain curves of the same general pattern as those presented in figures 4.6 to 4.9.

Heterogeneous yielding was observed in some of the tension tests with specimens of the 70%-Ni–30%-Cu alloy (fig. 4.8). Upper and lower yield points were obtained with the specimens tested at $+27°$, $-30°$, and $-78°$ C, and some yielding at a constant load was obtained with the specimens tested at 100° C. As the values of the true strain at these yield points were very small, the logarithmic true stress-strain curves are almost horizontal at these positions; the slopes of the logarithmic curves at these points also must be equal to the corresponding values of the true strains.

The data summarized in the logarithmic true stress-strain graphs in figures 4.6 to 4.9, in general, do not conform to the parabolic relationship given in eq 1, and no constant value can be ascribed to the exponent m, usually called the strain-hardening exponent. Thus, for these metals, the strain-hardening exponent, m, has little physical significance and cannot be called a material constant. Moreover, the work-hardening rates cannot be derived readily from these logarithmic true stress-strain graphs.

The reported relationship between the rate of work-hardening and true strain (eq 2), based on the true stress-strain relationship given in eq 1, in which m is equal to δ_m (true strain at maximum load), can be expressed in the form

$$\frac{d\sigma}{d\delta}/\sigma=\delta_m/\delta \tag{3}$$

Thus, graphs of values of $((d\sigma/d\delta)/\sigma)$ versus δ_m/δ should be linear with a slope of unity for true strain values between yield and maximum load. Some of the data obtained in tension tests at room temperature and

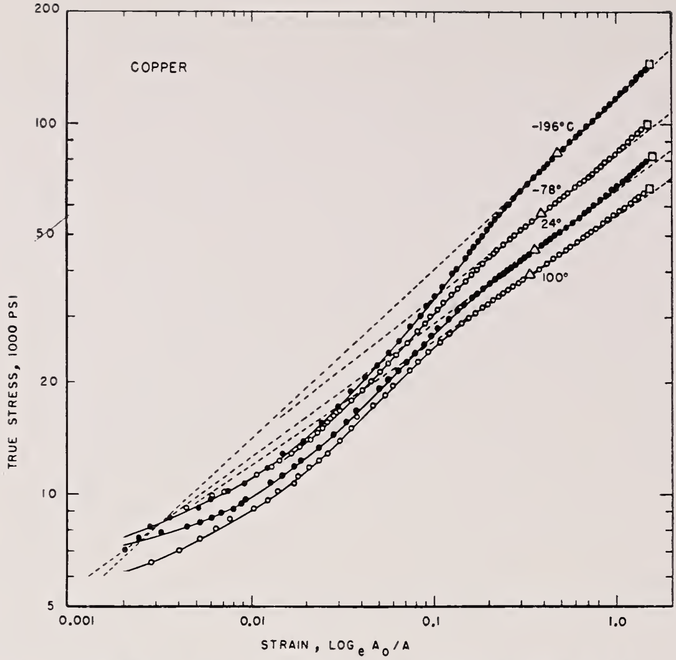

FIGURE 4.6. *Logarithmic true stress-strain curves obtained in tension tests at different temperatures with annealed copper.*

△, Maximum load; □, initial fracture.

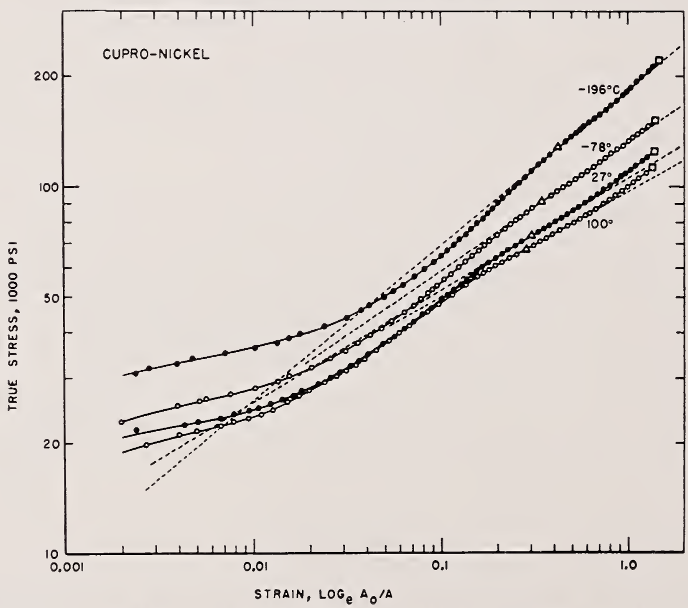

FIGURE 4.7. *Logarithmic true stress-strain curves obtained in tension tests at different temperatures with annealed cupro-nickel.*

△, Maximum load; □, initial fracture.

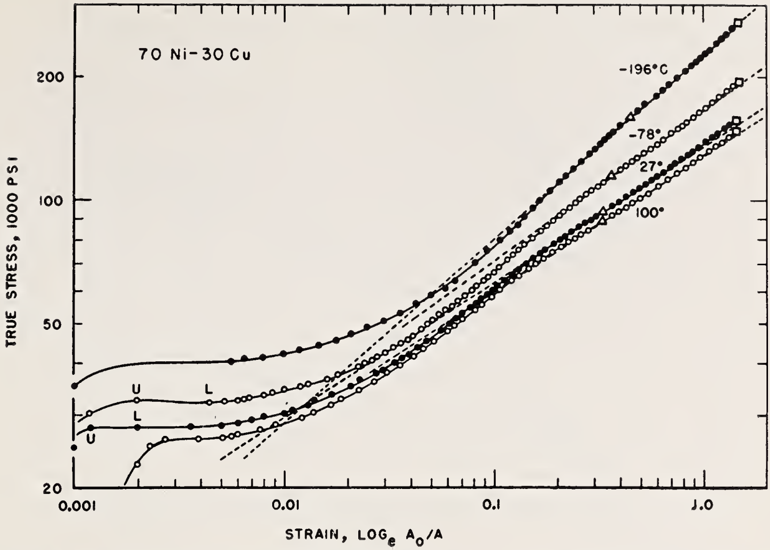

FIGURE 4.8. *Logarithmic true stress-strain curves obtained in tension tests at different temperatures with the 70 percent-Ni–30 percent-Cu alloy.*

U, Upper yield point; L, lower yield point; △, maximum load; □, initial fracture.

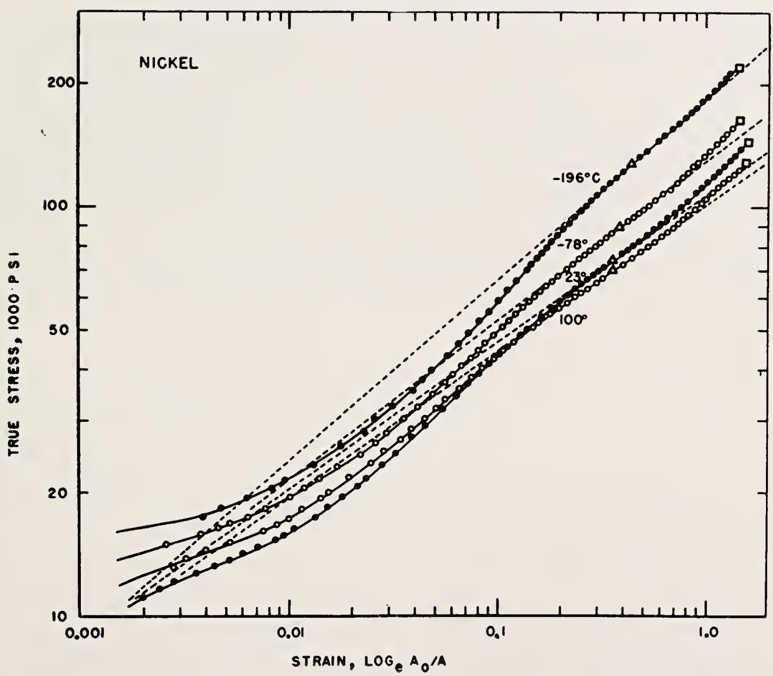

FIGURE 4.9. *Logarithmic true stress-strain curves obtained in tension tests at different temperatures with annealed nickel.*

△, Maximum load; □, initial fracture.

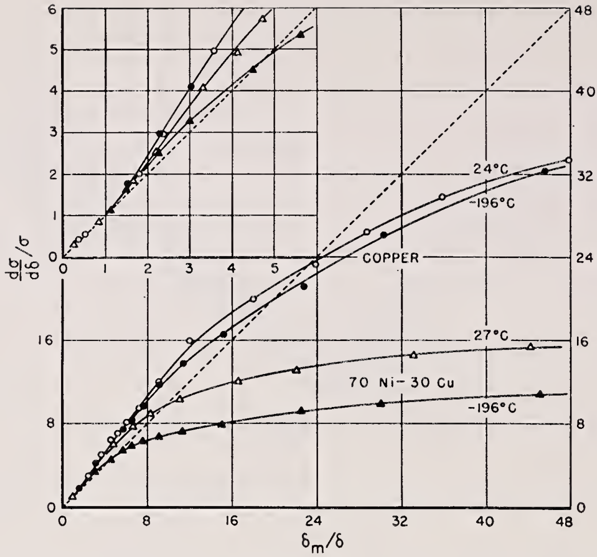

FIGURE 4.10. *Effect of alloy composition and testing temperature on the relationship between $(d\sigma/d\delta)/\sigma$ and δ_m/δ. The insert is an enlarged view of the lower left-hand corner.*

−196° C with the copper and the 70%-Ni–30%-Cu alloy specimens are summarized in figure 4.10. Similar data were obtained in the tension tests at +100°, −30°, −78°, and −140° C with these metals and also with cupro-nickel and nickel specimens. As shown in figure 4.10, these data certainly do not conform to the proposed linear relationship indicated by the broken straight line. Apparently, neither the true stress-strain relationship nor the rate of work-hardening during the tensile deformation of these metals can be represented, even approximately, by relationships of the type given in eq 1 and 2.

Recently a formula has been postulated [3, 9] relating the true stress, σ, and true strain, δ, in uniaxial tension and compression as follows:

$$\sigma=\sigma_\infty-(\sigma_\infty-\sigma_0)e^{-\delta/\delta_c} \tag{4}$$

where σ_o, and $\delta\infty$ and δ_c are constants, or σ_o, the initial, or threshold, stress, corresponds to the stress at which plastic straining starts; $\sigma\infty$, the asymptotic or final stress is the stress to which σ asymptotically approaches at infinite strain; δ_c, the characteristic or specific strain, is at any point of the stress-strain curve, the total strain necessary to raise the stress to the asymptotic stress if the rate of work-hardening, $d\sigma/d\delta$, would remain constant. From eq 4 one obtains

$$\frac{d\sigma}{d\delta}=\frac{\sigma_\infty-\sigma}{\delta_c}. \tag{5}$$

Thus, one should obtain a linear relationship between the rate of work-hardening and the true stress. The data obtained in the present investigation, however, do not conform to this linear relationship. As illustrated in typical examples in figure 4.11, in which the data

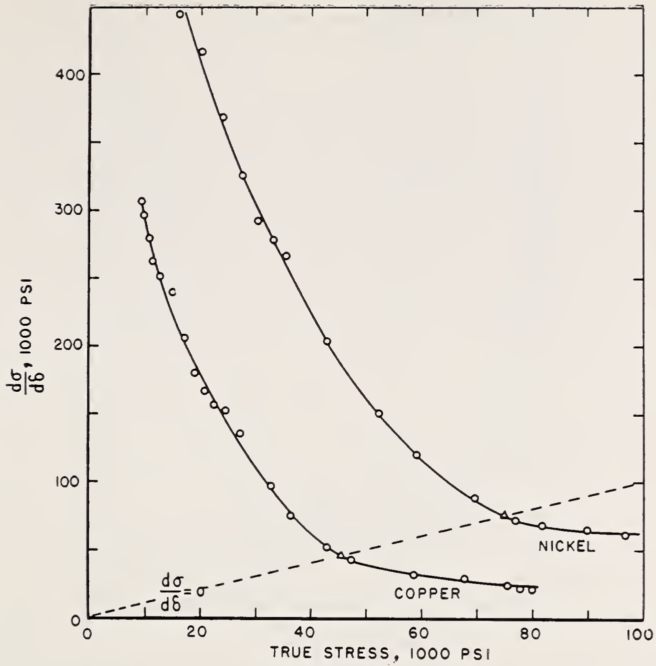

FIGURE 4.11. *Variation of the rate of work-hardening with true stress in tension tests made at room temperature with annealed copper and annealed nickel.*

obtained in the tension tests at room temperature with copper and nickel are summarized, the graphs apparently are curvilinear. The true stress-strain relationship for tension tests with these metals, therefore, cannot be represented closely by the formula proposed in eq 4.

The rate of work-hardening, $d\sigma/d\delta$, of these metals at specified strains can be determined from the slope of the true stress-strain curves (figs. 4.2 to 4.5) at the designated strains. Although tedious, this simple method for determining the rate of work-hardening is the best available, and it has been used in this investigation. Large true stress-strain graphs were prepared and used for these determinations.

The variations in the rate of work-hardening of these metals with deformation during the tension tests at temperatures ranging from $-196°$ to $+100°$ C are summarized in figures 4.12 to 4.15 These graphs present a quantitative representation of the variation of the slope of the true stress-strain curves with the strain, and show clearly the rapid decrease in the rate of work-hardening from relatively high values at very small strains, to low values at maximum load strains and to still smaller values as the strain is further increased. The curves are neither linear nor horizontal beyond the points representing maximum load, and thus neither a constant rate of work-hardening nor a rate of work-hardening proportional to the strain can be ascribed to these metals for this range of deformation. However, as mentioned previously, many investigators have assumed a linear true stress-strain

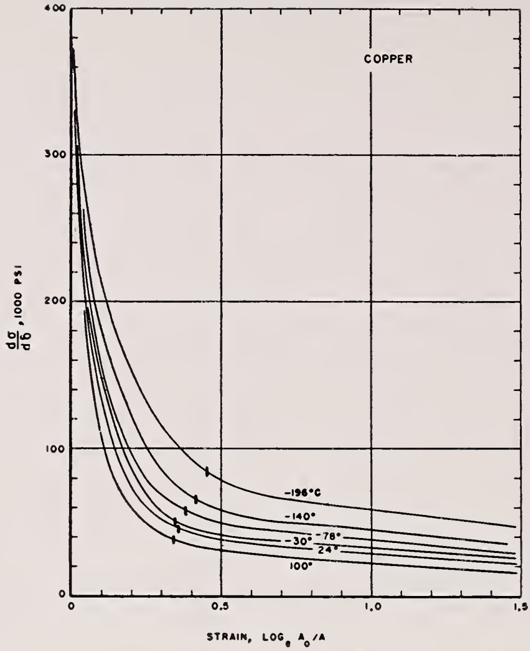

FIGURE 4.12 *Variation in the rate of work-hardening of annealed copper with the strain, in tension tests made at different temperatures.*
| , Maximum load.

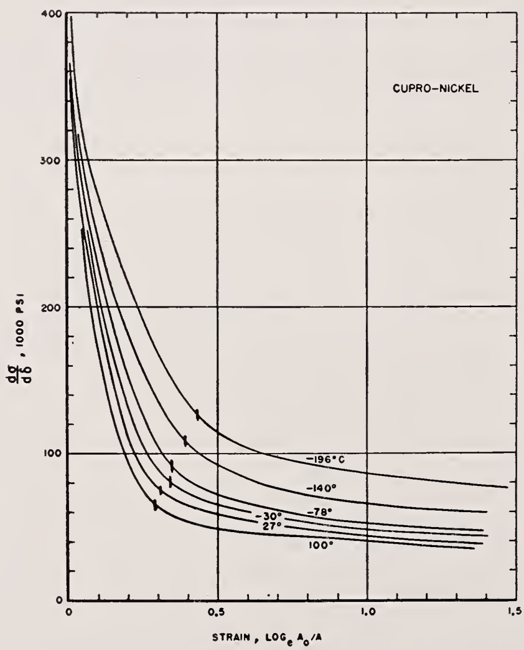

FIGURE 4.13. *Variation in the rate of work-hardening of annealed cupro-nickel with the strain, in tension tests made at different temperatures.*
| , Maximum load.

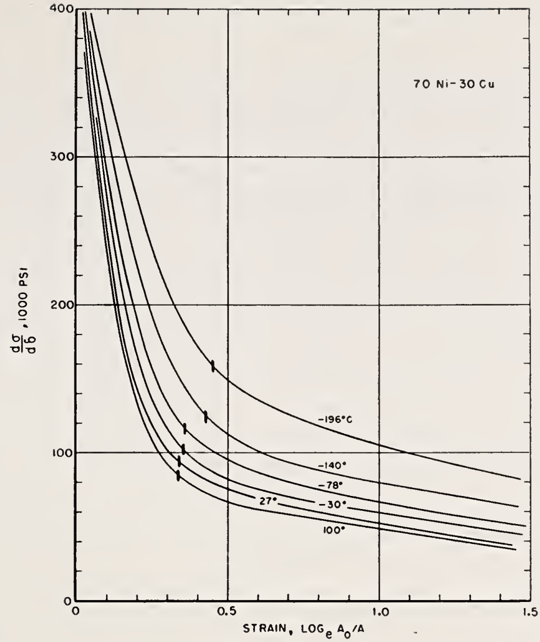

FIGURE 4.14. *Variation of the rate of work-hardening of the annealed 70-percent-Ni–30-percent-Cu alloy with the strain, in tension tests made at different temperatures.*

| , Maximum load.

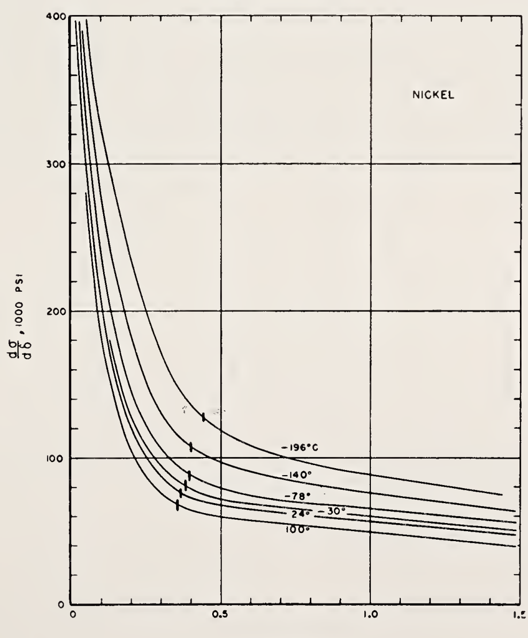

FIGURE 4.15. *Variation of the rate of work-hardening of annealed nickel with the strain, in tension tests made at different temperatures.*

| , Maximum load.

relationship for this range, which implies a constant rate of work-hardening.

Figures 4.12 to 4.15 also show a general trend of an increase in the radius of curvature in the "knee" of the curves with a decrease in the testing temperature. Thus the decrease in the rate of work-hardening of these metals with increasing strain during the tension test is more gradual, the lower the testing temperature. Moreover, the true strain value corresponding to the knee of the curve is greater at the lower temperatures. The influence of these same factors is illustrated in the shape of the true stress-strain curves (figs. 4.2 to 4.5) : the radius of curvature of the knee of the true stress-strain curves increases with decrease in the temperature and the knee of the curve occurs at relatively higher strains as the temperature is lowered, approaching the strain at maximum load. The larger values of the strain at maximum load at the lower temperatures may be attributed in part to these differences in the change of the rate of work-hardening of these metals with increasing strain; the maximum-load condition, $d\sigma/\sigma=-dA/A$, or $d\sigma/d\delta=\sigma$ is reached at larger values of the true strain as the testing temperature is lowered.

The quantitative variation in the rate of work-hardening of these metals at specified constant strains with change in the testing temperature is summarized in figures 4.16 to 4.18. Figure 4.16 is a

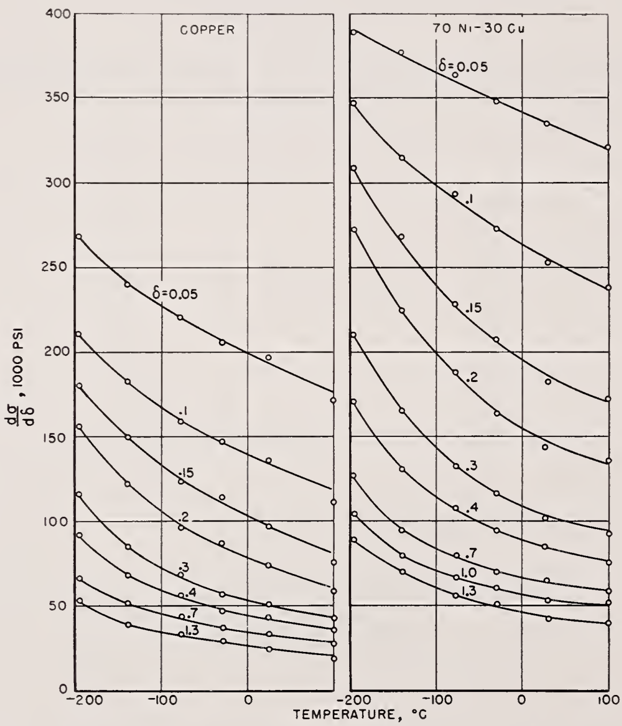

FIGURE 4.16. *Influence of strain on the relationship between temperature and the rate of work-hardening of annealed copper and annealed 70%-Ni–30%-Cu alloy.*

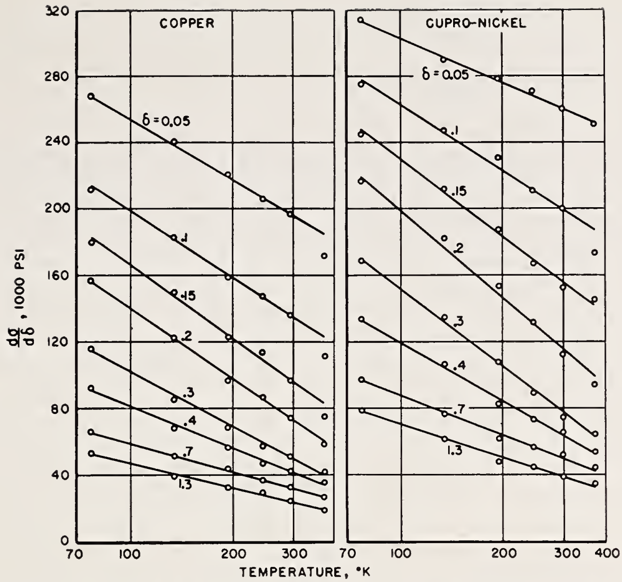

FIGURE 4.17. *Influence of strain on the relationship between temperature and the rate of work-hardening of annealed copper and annealed cupro-nickel.*

representative graph of values of rate of work-hardening versus temperature plotted on linear coordinates. These curves plotted from data obtained in tests with copper and the 70%-Ni–30%-Cu alloy indicate an increase in the rate of work-hardening with decrease in the testing temperature; this temperature effect increases as the temperature is decreased. The data obtained with tension tests on specimens of cupro-nickel and nickel (not shown), indicate the same trends.

The graph of the rates of work-hardening of the metals at specified constant strains versus the logarithm of the temperature (figs. 4.17 and 4.18) indicate a linear relationship which may be expressed

$$\left(\frac{d\sigma}{d\delta}\right)_{\delta=\text{const},} = -k \log_e T + c,$$

where k is the steepness of the straight line and c is a constant. k and c, however, vary both with the metal and the strain of the specimen. The data given in these figures indicate only fair conformity to straight lines as some scatter is exhibited in the plotted values. However, it is believed that for these metals this relationship may be accepted tentatively, until further experimental evidence either definitely establishes or invalidates this relationship.

The values of k (the work-hardening rate-temperature index for a specific strain) vary both with the composition of the metal and with the deformation of the specimen. This relationship is indicated in figure 4.19, in which values of k versus true strain are plotted for each metal. Even though the rates of work-hardening of these metals at small true strains, 0.01 to 0.10, are much greater than those for strains from 0.10 to 0.25 (figs. 4.12 to 4.15), the influence of the temperature on the rate of work-hardening is greatest for strains between

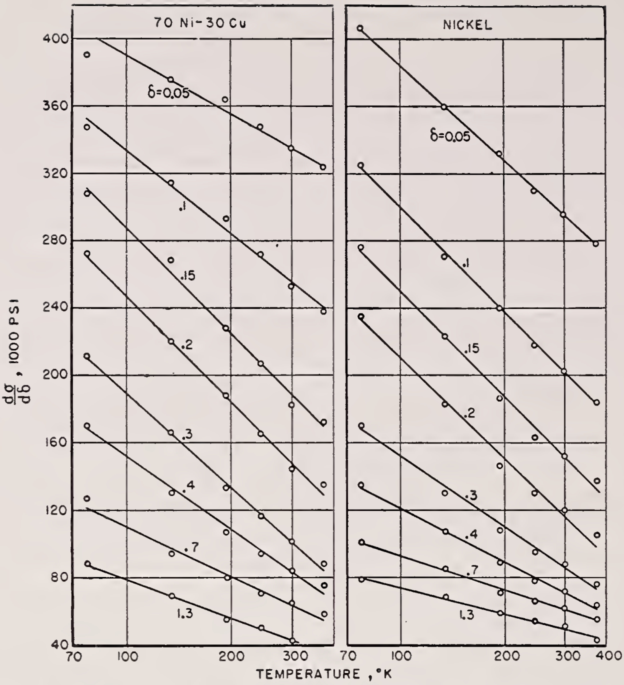

FIGURE 4.18. *Influence of strain on the relationship between temperature and the rate of work-hardening of annealed 70%-Ni-30%-Cu alloy and annealed nickel.*

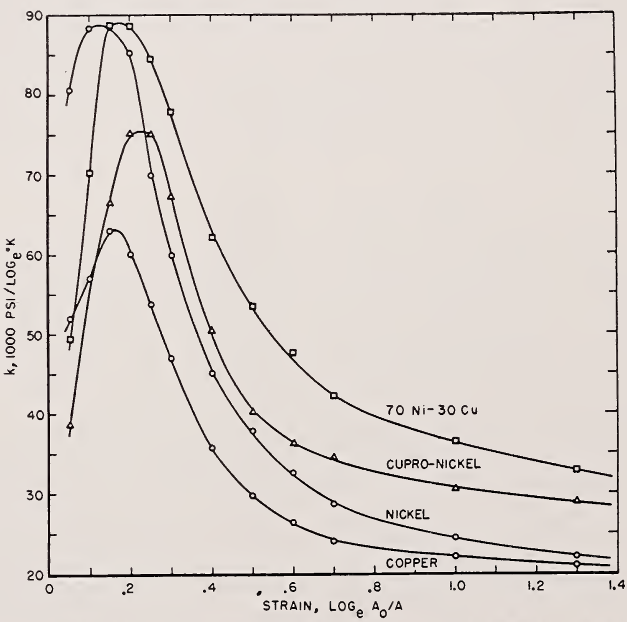

FIGURE 4.19. *Variation of the work-hardening rate-temperature index k, with the tensile strain of the specimen.*

0.10 and 0.25, as indicated by the pronounced maximum in each curve at true strain values between 0.10 and 0.25. The maximum in the curve for cupro-nickel and also that in the curve for the 70%-Ni–30%-Cu alloy occur at slightly greater strains than those for the copper and nickel.

Influence of Temperature on Mechanical Properties

The relationships between the testing temperature and some of the mechanical properties of these metals and alloys as determined in tension tests at temperatures ranging from −196° to +100° C are summarized in figures 4.20 to 4.23. The continuous line curves A, C, and D represent true stress-temperature relationships, and the broken line curves E, F, and G represent true strain-temperature relationships. The continuous line curves, B, represent the variations of the nominal ultimate stress with temperature. As the nominal ultimate stress is based upon the original cross-sectional area of the specimen and not the actual cross-sectional area of the specimen at the maximum load conditions, it does not represent a true stress value. However, it represents a value of interest to structural engineers and is therefore included in these diagrams.

Yield-strength values (0.1 or 0.2% offset from the modulus line) were not determined in these tension tests. However, the true stress values for a very small total true strain (elastic plus plastic) of 0.0025 are included and these values (curve A) can be considered as approximate initial strength values of these metals and alloys in the

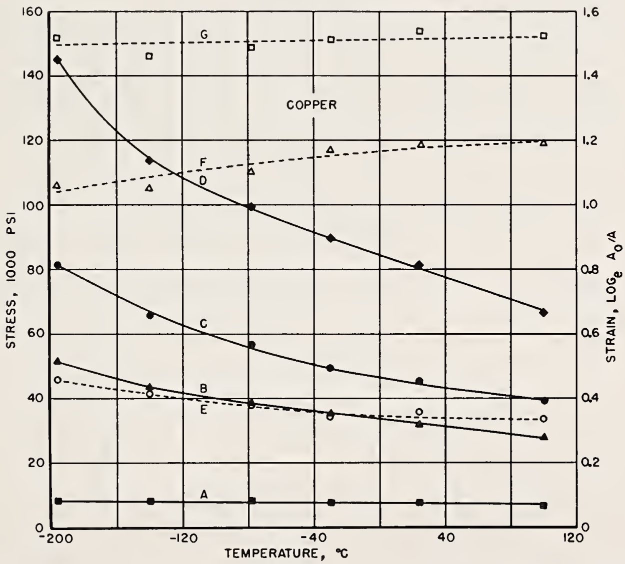

FIGURE 4.20. *Effect of testing temperature on various mechanical properties of annealed copper.*

A, True stress at a true strain of 0.0025; B, nominal ultimate stress; C, true stress at maximum load; D, true stress at initial fracture; E, true strain at maximum load; F, true strain at initial fracture minus true strain at maximum load; G, true strain at initial fracture.

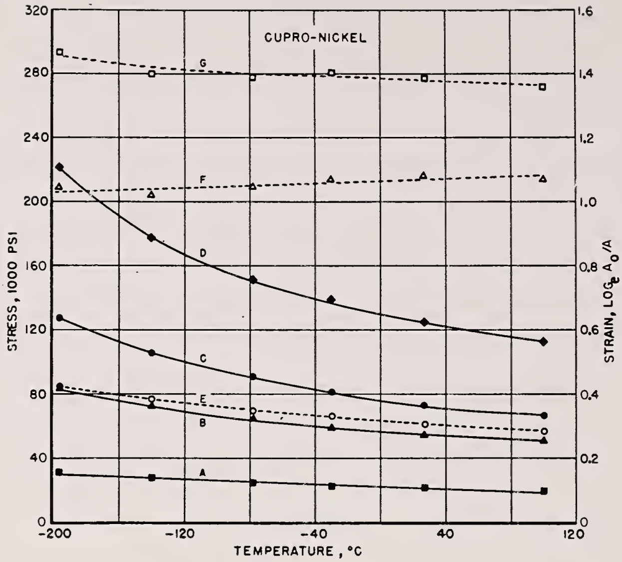

FIGURE 4.21. *Effect of testing temperature on various mechanical properties of annealed cupro-nickel.*

A, True stress at a true strain of 0.0025; B, nominal ultimate stress; C, true stress at maximum load; D, true stress at initial fracture; E, true strain at maximum load; F, true strain at initial fracture minus true strain at maximum load; G, true strain at initial fracture.

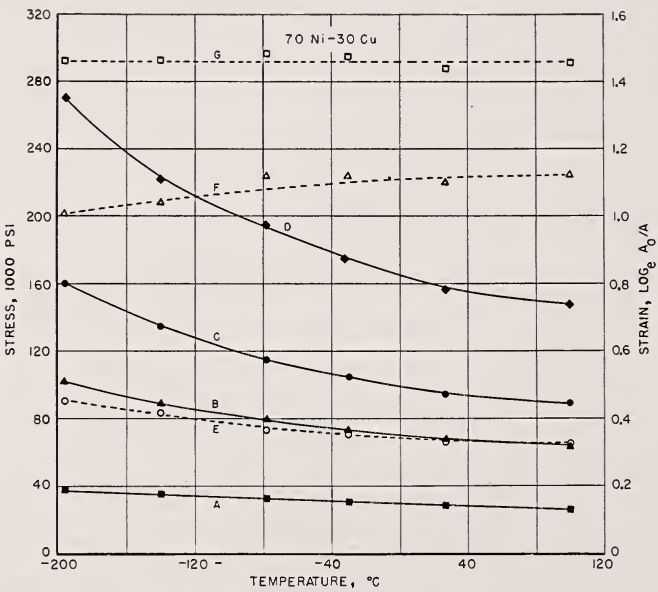

FIGURE 4.22. *Effect of testing temperature on various mechanical properties of annealed 70% Ni-30% Cu alloy.*

A, True stress at a true strain of 0.0025; B, nominal ultimate stress; C, true stress at maximum load; D, true stress at initial fracture; E, true strain at maximum load; F, true strain at initial fracture minus true train at maximum load; G, true strain at initial fracture.

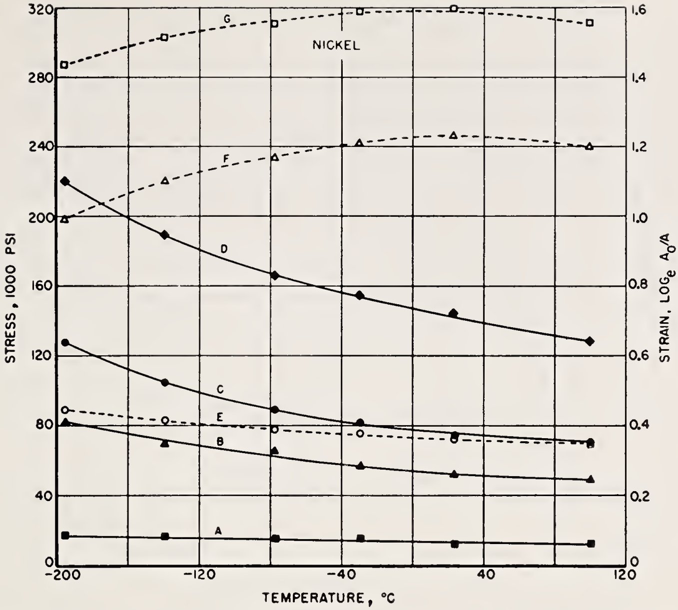

FIGURE 4.23. *Effect of testing temperature on various mechanical properties of annealed nickel.*

A, True stress at a true strain of 0.0025; B, nominal ultimate stress; C, true stress at maximum load; D, true stress at initial fracture; E, true strain at maximum load; F, true strain at initial fracture minus true strain at maximum load; G, true strain at initial fracture.

annealed conditions and at the indicated temperatures. These graphs illustrate the general trend of a continuous increase in the initial strength of these metals and alloys with decrease in the testing temperature from +100° to −196° C.

Curves C in figures 4.20 to 4.23 show a continuous increase in the true stress at maximum load with decrease in temperature from +100° to −196° C. Moreover, this increase is much greater than the corresponding increase in the initial strength with decrease in testing temperature (curves A). This difference is due to the additive effect of two factors: (1) the greatly increased rate of work-hardening with decrease in testing temperature as previously described, and (2) the increase in true strain at maximum load with decrease in temperature (curves E).

Curves D in figures 4.20 to 4.23, representing the variation of the true stress at initial fracture with the testing temperature also show the general trend of a continuous increase in the true stress at fracture with decrease in temperature from +100° to −196° C. As will be noted later, the total strain at fracture did not vary greatly with the temperature (curves G of figs. 4.20 to 4.23), and as the increase in initial strength with decreasing temperature was relatively small, the increase in the true stress at fracture with decrease in temperature may be attributed mainly to the increase in the rate of work-hardening.

All of the curves representing the variation of the true strain at maximum load with testing temperature (curves E of figs. 4.20 to 4.23) show a general trend of a continuous increase in this ductility value with decrease in temperature. On the other hand, curves F (figs. 4.20 to 4.23) representing the variation of the nonuniform strain (true strain at initial fracture minus the true strain at maximum load) show a general trend of a small decrease in this ductility value with decrease in temperature below 25° C. The decrease in this ductility value of the copper and nickel specimens with decrease in temperature was greater than that of the cupro-nickel and the 70%-Ni–30%-Cu alloy specimens.

No general trend is indicated in the variation of the total ductility (true strain at initial fracture, curves G of figs. 4.20 to 4.23) with temperature. The data for nickel (fig. 4.23) indicate a slight decrease in the total ductility with decrease in temperature below 25° C, whereas the data for cupro-nickel (fig. 4.21) indicate a slight increase in the total ductility with decrease in temperature. No significant change in the total ductility with temperature was observed with the specimens of copper (fig. 4.20) and 70%-Ni–30%-Cu alloy (fig. 4.22).

Influence of Solute Content of Copper-Nickel Alloys on the Strength Indices and Work-Hardening Characteristics

The influence of the solute content of these solid-solution alloys of copper and nickel on the strength indices and the rates of work-hardening during deformation in tension is summarized in figures 4.24 to 4.27. As this investigation was limited to tension tests with specimens of copper, nickel, cupro-nickel, and 70%-Ni–30%-Cu alloy no continuous smooth curves are drawn to represent the relationships between these strength indices or the rates of work-hardening and the chemical composition of the alloys. In order to determine these relationships accurately, tension tests with additional alloys of copper and nickel would be required. It is believed, however, that the data presented in these figures exhibit the general trend of the influence of the alloy content on the strength indices and rates of work-hardening.

The addition of the solute element in a substitutional solid-solution alloy may either increase or decrease the lattice parameter. The unit cell usually is contracted by dissolving atoms smaller than the solvent atoms and, to a rough approximation, the lattice parameters vary linearly with the composition expressed in atomic percentage of the solute (Vegard's law). Marked deviations from this rule often occur. However, atoms of very similar size and atomic structure tend to form systems that conform closely to this law. The lattice parameters of face-centered-cubic copper and nickel at room temperature do not vary greatly (3.608 and 3.517 angstroms, respectively). Moreover, the atomic diameters of the copper and nickel atoms are nearly the same, (2.551 and 2.487 angstroms, respectively). Thus only relatively small negative deviations from Vegard's law (smaller lattice parameters) are found in the copper-nickel system [12].

The solution-hardening and the work-hardening of substitutional solid-solution alloys have been reported to vary approximately linearly with the change in the lattice parameters (13, 14). An initial almost linear increase in solid-solution hardness with increase in the atomic percentage of nickel was observed in the copper-nickel system [13].

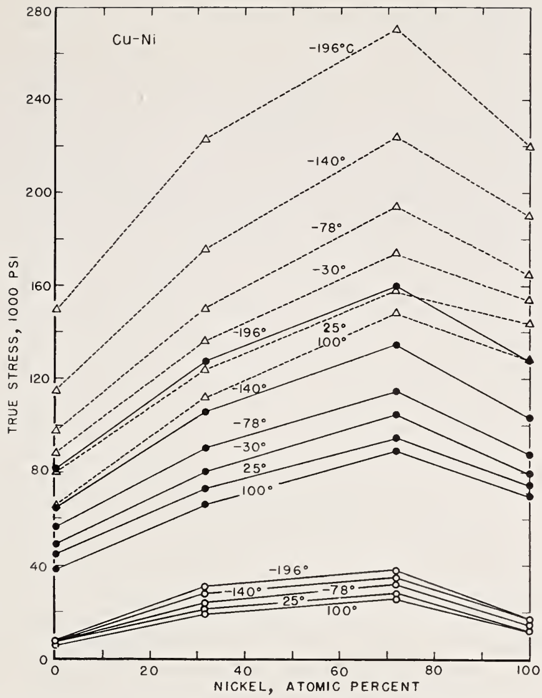

FIGURE 4.24. *Effect of the nickel content of copper-nickel alloys and the testing temperature on various strength indices.*

○, True stress at a true strain of 0.0025; ●, true stress at maximum load; △, true stress at initial fracture.

The rate of the increase in hardness with increasing atomic percentage of nickel, however, diminished at solute contents above 25 atomic percent.

The relatively small difference in the atomic diameters of the two basic components of these copper-nickel alloys is of sufficient magnitude to produce strain and accompanying local residual stresses in the solvent lattice. This factor also has a pronounced effect on the initial strength or solution-hardening and work-hardening characteristics of these alloys; the resistance of these alloys to deformation by slip or movement of dislocations is greater than that of the solvent metal. These effects should increase continuously with increase in the atomic percentage of the solute atoms in the alloy to maximum values for alloy composition of approximately 50 atomic percent. The data summarized in figures 4.24 and 4.25, although limited to the two base metals and two alloy compositions, indicate conformity to the above relationships.

The testing temperature also should have a very appreciable effect in conjunction with the solute content of these solid-solution alloys on the strength indices and rates of work-hardening. As the temperature is lowered, the thermal motion of the atoms becomes less, and thus the effect of the strain and local stress in the solvent lattice induced by the solute atoms is enhanced. This effect of the temperature is

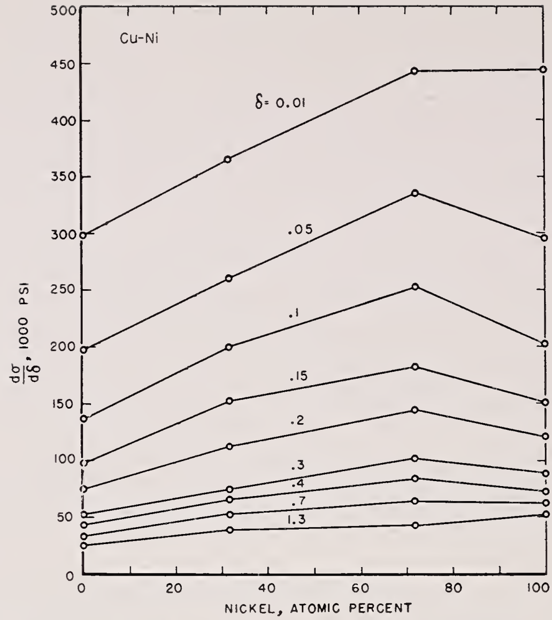

FIGURE 4.25. *Effect of the nickel content of copper-nickel alloys on the rate of work-hardening at room temperature.*

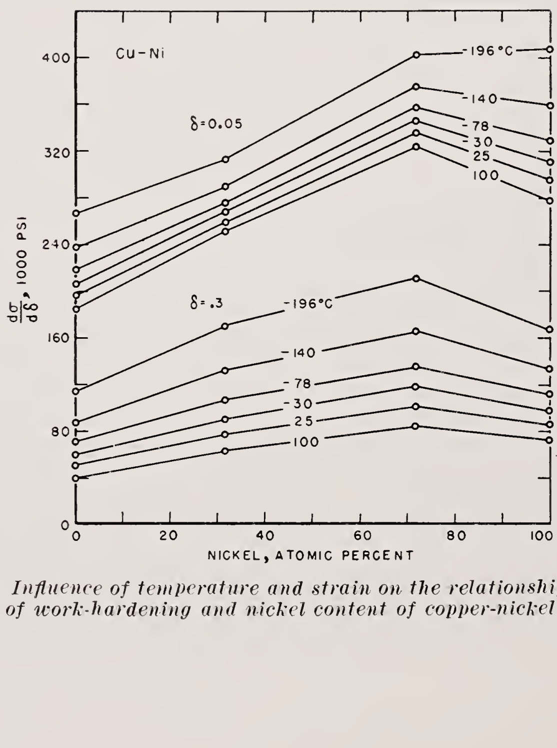

FIGURE 4.26. *Influence of temperature and strain on the relationship between the rates of work-hardening and nickel content of copper-nickel alloys.*

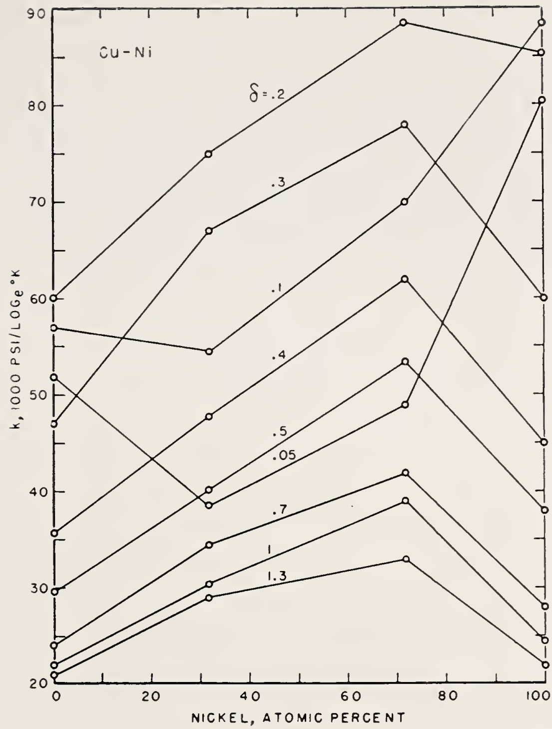

FIGURE 4.27. *Influence of strain on the relationship between the work-hardening rate-temperature index, k, and the nickel content of copper-nickel alloys.*

indicated in figure 4.24 by the relatively large increase in the strength indices of the solid-solution alloys with decrease in temperature as compared to those of the component metals. As illustrated, the increase in the values of the true stress at the small total true strain of 0.0025 for the two copper-nickel alloys, with decrease in temperature from +100° to −196°C is approximately two to three times greater than that for the nickel and copper. A similar, although less pronounced, effect is shown in the data for the true stress at maximum load and the true stress at initial fracture. The latter two sets of values depend upon the rates of work-hardening during the deformation from yielding to the maximum load and to fracture, respectively, and also upon the initial-strength values. The results indicate indirectly that the variation in the rates of work-hardening with test temperature of the solid-solution alloys, as compared to that of the component metals, is not as pronounced as the corresponding variations in the initial-strength values. This effect of the temperature on the variation of the rates of work-hardening of these copper-nickel—solid-solution alloys with the solute content, for constant true strains, is indicated in the data summarized in figure 4.26. At small true strains (0.05) the variation of the rates of work-hardening of the copper-nickel alloys with the temperature, is less than that of the copper or nickel. However, at true strains larger than 0.15 or 0.2, a directly opposite trend is found, as indicated by the data for a true strain of 0.3. These trends are summarized in figure 4.27. In this figure the work-hardening rate-temperature index k for various values of constant true strains, is plotted against the atomic percentage of nickel. The trends shown in this figure could be expected from the graphs of k values versus true strain presented in figure 4.19. As shown in figure 4.19, the curves for the copper and nickel lie above those for the copper-nickel alloys at small strains. However, at true strains of approximately 0.15 to 0.2 the curves for copper and nickel cross over those for the two alloys and remain below the latter at all larger true strain values.

Summary

True stress-strain curves are presented for high-purity copper, high-purity nickel, high-purity 70%-Ni-30%-Cu alloy and cupro-nickel (70%-Cu–30%-Ni) in the annealed conditions and tested in tension at temperatures ranging from $-196°$ to $+100°$ C. Derived graphs also are presented to show the effects of the testing temperature and alloy content of these solid-solution alloys on the mechanical properties and work-hardening characteristics.

The true stress-strain curves for tension tests made at temperatures ranging from $-196°$ to $+100°$ C with these materials are curvilinear throughout the deformation from yield to initial fracture. The degree of curvature of the portion of the curves from maximum load to initial fracture is similar at all testing temperatures.

The "rim effect" obtained during the fracture of these metals and alloys in tension at temperatures ranging from $-196°$ to $+100°$ C was considerable, even at the very low temperatures. Below room temperature the rim effect tended to decrease as the temperature was lowered. The magnitude of this effect also varied with the chemical composition of the metal or alloy. Accurate values of the fracture stresses for these metals and alloys therefore, cannot be determined from diameter measurements made after fracture of the specimen.

The logarithmic graphs of the true stress-strain data for these metals and alloys tested in tension at temperatures ranging from $-196°$ to $+100°$ C are not linear, and thus the true stress-strain relationship cannot be represented accurately by the parabolic equation

$$\sigma = b\delta^m.$$

The strength indices of these metals and alloys, as represented by the initial strength, nominal ultimate stress, true stress at maximum load and true stress at initial fracture, increased continuously with a decrease in testing temperature within the range of $+100°$ to $-196°$ C.

The effect of the solute content of these solid-solution alloys of copper and nickel on the initial strength (true stress at a total true strain of 0.0025) is indicated by the relatively high initial-strength values of the cupro-nickel and the 70%-Ni–30%-Cu alloy specimens as compared to those for the copper and nickel specimens.

The rates of work-hardening of the metal varied greatly with the plastic deformation of the specimen. The rates decreased rapidly from relatively high values at small strains to low values at strains near maximum load and decreased slightly during the deformation from maximum load to fracture.

The rates of work-hardening at specific true strains, in general, decreased continuously with increase in testing temperature from $-196°$ to $+100°$ C; an exception was the slight increase in the rate of work-hardening of nickel at small true strains, with increase in temperature from 23° to 100° C as a result of an appreciable degree of strain aging of the nickel specimen during the test at 100° C. The influence of the temperature on the rates of work-hardening varied greatly with the deformation during the tension test and was a maximum for true strains of approximately 0.15 to 0.2. The solute content of the alloy also affected the work-hardening during the tension test. The increase in the rates of work-hardening of the cupronickel and the 70%-Ni–30%-Cu alloy with decrease in temperature was greater for true strains above 0.2 than that of the two component metals, copper and nickel. However, at very small strains, the increase in the rates of work-hardening of these alloys with decrease in temperature was less than that of the copper and nickel.

The total ductility of copper, nickel, and copper-nickel alloys, as represented by the true strain at initial fracture was affected little by the temperature. However, the uniform deformation of these metals and alloys, as represented by the true strain of the specimens at maximum load, in general, increased with a decrease in temperature. On the other hand, the nonuniform deformation (true strain at fracture minus the true strain at maximum load) usually decreased with a lowering of the temperature.

The authors are indebted to J. D. Grimsley and C. R. Johnson for assistance in this investigation.

Discussion

Mr. J. A. Kies, Metallurgist, Naval Research Laboratory, Washington, D. C.: I believe that there was nothing said about the effect of strain rate. Are you studying that?

Do you find any evidence of more than one micromechanism of deformation? It may be that the size of the flow unit undergoes a change as strain or temperature increases. If the change is sudden, some kind of discontinuity in the expression for the effect of temperature would be required, and one simple analytic expression could not, as you have demonstrated, be valid. By size of flow unit, I do not necessarily mean the length of a Taylor dislocation, as computed by J. E. Dorn.

Mr. T. W. George of Naval Research Laboratory has recently obtained some data for copper at very low strain rates, which seem to indicate definite changes in the flow mechanisms, depending on the strain rate. This will be published soon.

Your paper deals with a most interesting subject and the reasons for departures from simple formulas are the important things to study.

Mr. G. W. Geil: A study of the effect of varying the strain rate upon the true stress-strain relationship was not included in this investigation. These tensile tests were all conducted at an approximately constant strain rate; the deformation of the specimen beyond the region of initial yielding was so controlled that a rate of reduction of area of approximately 1 percent a minute was maintained up to the initial fracture of the specimen. Thus the tests were not conducted at a constant true strain rate. However, for the deformation beyond the maximum load, tests conducted at a constant rate of reduction of area more nearly approximate a constant true strain rate than tests conducted at a constant rate of extension.

It is generally recognized that the deformation properties of metals vary with the strain rate and test temperature, and I believe it is logical to assume that the micromechanism of deformation also varies. The true stress-strain data obtained in this study do not indicate any sudden change in the mechanism of deformation, either with increase in strain or temperature. As mentioned earlier, this investigation did not include tests at different strain rates. No significant variations in microstructure were observed in specimens tested at the different temperatures.

Prof. J. W. Fredrickson, Chief, Division of Metallurgy, Pennsylvania State College, State College, Pa.: There are a couple of points I would like to mention. First, I agree with Mr. Geil in that I do not see why, even though the stress-strain curve is apparently quite simple when observed on paper, how one can expect a simple parabolic equation to fit anything as intricate as the deformation and flow of metals. Second, MacGregor and Fisher published some work in 1946 on true stress-strain diagrams in which they brought up the point that the maximum-load point is short of the point where the derivative of stress-strain becomes a constant value.

I want to bring the following into consideration. You mentioned the effect of atomic radii and its possible effect on strength and hardness. I think that we take "atomic radii" too literally. Possibly we should use "effective radii." These are not billiard balls with which we are working. Their atomic radii can be changed, the valence electrons can be deformed, and their effective radii are very much dependent upon environment. I think that if we think in terms of effective radii rather than the fixed radii of a billiard ball, we might come up with some different ideas.

Mr. Geil: I agree with Prof. Fredrickson's statement that it would be more appropriate to think in terms of the values of effective radii of the atoms instead of the atomic-radii values as published. In either case, however, the differences in radii would be great enough to cause local residual stresses and lattice strains of sufficient magnitude to affect the initial strength and work-hardening characteristics of the alloys.

Dr. G. A. Moore, Assistant Professor, University of Pennsylvania, Philadelphia, Pa.: This seems to be a good time to raise a little theoretical question. Mr. Geil has shown some changes of work-hardening with temperature which have been of great interest to me. In the paper by Frederickson and Eyring [13] I believe there is a generalization that the effective size of a slip unit is directly proportional to the absolute temperature. That would appear to imply that, if the mechanism stayed the same, the rate of work-hardening should effectively double every time you divide the temperature by 2. Now, the data that Mr. Geil has given here apparently show a considerable resemblance to that expected behavior, but not all materials behave that way.

Some materials harden excessively at low temperature, and there is a strong probability that certain other materials do not harden, particularly at very low temperature, and especially when another flow mechanism comes in. I should like to ask Mr. Geil how well he thinks his current data agree with that original suggestion.

Mr. Geil: The data obtained in this investigation were not evaluated in terms of the temperature—effective size of slip unit relationship discussed by Dr. Moore, and thus I do not feel qualified to comment further at the present time on this relationship. Perhaps Professor Fredrickson would like to comment on this factor.

Prof. Fredrickson: Maybe we can talk about it this way. Mr. Kies mentioned the effect of the flow unit, the size of the flow unit, and whether a change in mechanism is present. First of all, we know that a change of mechanism is taking place. It is not an abrupt change but is of a transition type. I have not worked on the problem sufficiently to say just what the change is, but the data we have investigated have shown that it is present. The size of the flow unit is dependent upon the temperature and, as brought out in later investigations, is dependent also upon the stress. Both the stress and temperature will affect the flow unit.

If I recall correctly, small flow units are associated with large stresses and large flow units with small stresses. It is almost an indirect relationship, but not quite. Likewise, when the testing temperature is increased the size of the flow unit increases.

This interpretation is from theoretical investigation. I wish we could see the units. You mentioned microscopic investigation. I do not ever expect to find anything in the microscope so far as flow units are concerned—they are too small. It is like dislocations, one thinks they are there but cannot see them. The flow unit is there and moving, but so small we cannot see it. We wish we could. I do not know whether or not I have answered your question; maybe I have just talked around it.

Mr. Geil: I would like to make one further comment in regard to the mechanism of plastic deformation. Dr. Allen, in the presentation of his paper at this symposium, discussed the recent work of A. F. Brown [14] in which the distribution and form of visible (electron microscope) slip zones in aluminum depended upon the temperature of the aluminum during the deformation. His data certainly indicate a change in the micromechanism of plastic deformation with temperature.

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5. Application of Metals in Aircraft at Low Temperatures

By J. B. Johnson [1] and D. A. Shinn [2]

The lowest temperature that modern aircraft will encounter is approximately —130° F. As the extremes of low temperatures occur at high altitudes, where the vehicle is in a fluid medium, shock loading is not critical unless the metal is subjected to ballistic or explosive impacts. This paper discusses briefly the relationship of the properties of metals as determined by conventional laboratory tests and those selected by the designer. Additional data on test of aircraft metals at temperatures down to —420° F are presented. The metals used in the fabrication of aircraft have good low-temperature properties, and no failures attributable to a reduction in mechanical properties at low temperatures have been encountered. Evaluation of the serviceability of a material at low temperatures should not be based solely on the transition from tough to brittle fracture as represented by results of notched bar impact tests, as this may mean the selection of steel with a higher content of strategic alloying elements than is necessary for the specific application.

Air-borne vehicles, at least for military service, should be designed and constructed to operate under any climatic conditions encountered on the earth's surface, and above the earth's surface to the absolute ceiling of the aircraft or missile. It is probable that, because of the difficulty of maintaining and servicing aircraft at low temperatures, −80° F is the lowest ground-level temperature that need be considered, and actually, −65° F, which has been the standard requirement of the Air Force for several years, is more realistic. Upper-air temperature [1] [3] charts indicate the temperatures at various levels and represent conditions for the entire year. These charts indicate the range of temperatures that would be encountered 80 percent of the time by aircraft flying at the specified levels within the latitude belts. The belt for 0° to 20° north latitude is superimposed over the northern hemisphere tempagram in figure 5.1.

Extreme low temperatures aloft in winter may exceed the limit on the chart by more than 10 percent, but not over the year. Between 50 and 60 thousand feet the low limit approaches −130° F. This is ambient air temperature. Due to aerodynamic heating and thermal gradients within the aircraft structure, it is highly probable that no part of the structure reaches this temperature.

The effects of temperature are indicated by changes in physical and mechanical properties. These concern not only the physicist and metallurgist, but also the designer and fabricator, as they influence the allowable stress levels used in design, the dimensions of mating parts, the timing of forming operations of metal components, and the rate and amount of hardening.

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[3] Figures in brackets indicate the literature references on p. 111.

Design Considerations

The design mechanical properties [2] for the metals used in aircraft construction (table 5.1) are obtained by uniaxial static tests of specimens at approximately 70° F. The ultimate and yield strengths and modulus of elasticity invariably increase at lower temperatures (figs. 5.2 and 5.3, and table 5.2).

These and other numerical data in this paper were obtained at Ohio State University and have been partially reported in a paper by Zambrow and Fontana [3], which gives further details in regard to materials and testing procedures. The tests were made in order to obtain a comparison of the low-temperature properties of the several metals used in aircraft construction which conformed to standard specifications when tested under the same conditions. The values are within the range of published data for similar metals [4, 5, 6].

The higher values assure an increase in the positive margin of safety at lower temperatures with uniaxial static loading. This is less certain for a loading that causes combined (polyaxial) static or cyclic stresses. Experience indicates that the material should have some ductility. Ductility will be defined as the ability of the metal to flow

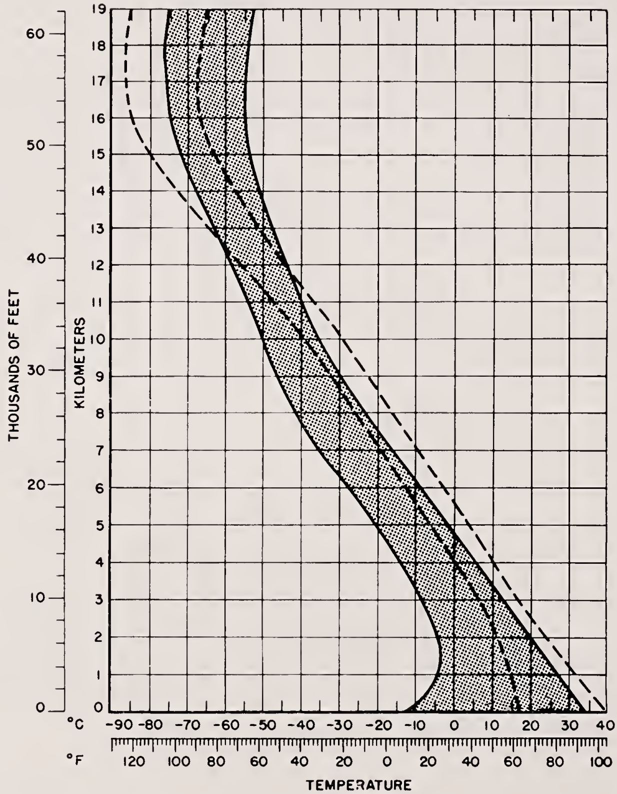

FIGURE 5.1. *Upper-air eighty-percent design tempagram for the Northern Hemisphere and for 0° to 20° N latitude (dashed section).*

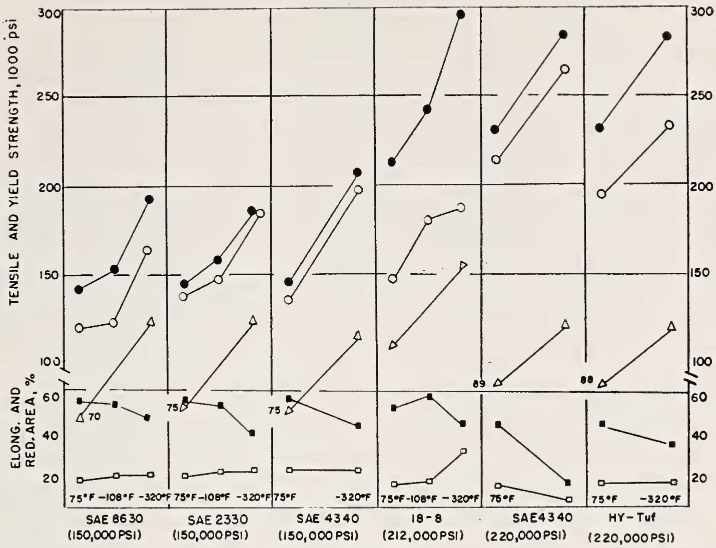

FIGURE 5.2. *Tensile and fatigue properties of ferrous metals at room temperature to —320° F.*

●, Tensile strength; ○, yield strength (0.2 percent); △, fatigue strength at 10^7 cycles; □, elongation (percentage in 2 in.); ■, reduction in area (percent).

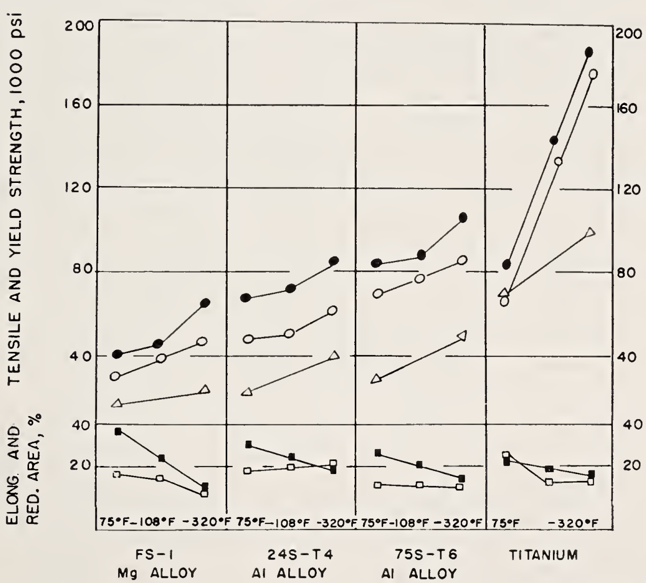

FIGURE 5.3. *Tensile and fatigue properties of nonferrous metals at room temperature to —320° F.*

●, Tensile strength; ○, yield strength (0.2 percent); △, fatigue strength at 10^7 cycles; □, elongation (percentage in 2 in.); ■, reduction in area (percent).

TABLE 5.1. *Chemical composition and heat treatment*

Metal	Specification	Si	Fe	Cu	Mn	Al	Mg	Zn	Cr	Ni	Other
Nonferrous											
		%	%	%	%	%	%	%	%	%	%
24S-T4 aluminum [1]	QQ-A-354	0.16	0.43	3.8 to 4.4	0.6 to .9	Bal.	1.2 to 1.7	0.10	------	<0.05	<0.05 Bi
75S-T6 aluminum [1]	AN-A-9a	.20	.45	1.50	.15	do	2.80	5.70	0.30	------	.05 Ti
FS-1h magnesium [2]	AN-M-27	<.01	<.001	<0.01	.49	3.10	Bal.	1.05	------	<.001	------
Titanium [3]		------	------	------	------	------	------	------	------	------	------
Ferrous											
Steel	**Specification**	**C**	**Mn**	**P**	**S**	**Si**	**Ni**	**Cr**	**Mo**	**Al**	**Ti**
		%	%	%	%	%	%	%	%	%	%
SAE 2330 [4]	AN-QQ-S-589	0.28 to .33	0.60 to .80	<0.026	<0.030	0.20 to .35	3.25 to 3.75	------	------	------	------
SAE 8630 [5]	AN-S-14a	.27 to .33	.70 to .90	<.022	<.033	.20 to .35	0.48 to .67	0.49 to .56	0.18	------	------
Type 304 stainless (18-8) [6]		.054	.49	.019	.015	.42	8.82	18.50	------	------	------
Hy-Tuf [7]		.259	1.26	.016	.017	1.37	1.91	0.09	0.43	------	------
SAE 4340 [8]	AN-QQ-S-756A	.46	0.70	.018	.031	0.288	1.78	.95	.23	------	------

[1] Commercially heat-treated and aged ¾-in. round bars.
[2] Extruded and cold-drawn ¾-in. round bars.
[3] Commercially pure from Remington Arms; received in hot-swaged condition as ⅝-in. rounds with Vickers hardness number of 250.
[4] Received as normalized. Austenitized at 1,500° F., oil-quenched and tempered at 925° F to 150,000-psi tensile strength.
[5] Received as normalized. Austenitized at 1,500° F, oil-quenched and tempered at 850° F to 150,000-psi tensile strength.
[6] Cold-drawn to 210,000-psi tensile strength.
[7] Received as normalized. Austenitized at 1,575° F, oil quenched, and tempered at 550° F (230,000 psi tensile strength).
[8] Received as normalized. Austenitized at 1,550° F, oil-quenched and tempered at (a) 1,200° F (150,000-psi tensile strength) and (b) 800° F (230,000-psi tensile strength).

Table 5.2. *Modulus of elasticity in tension*

Material	Modulus of elasticity			
	−13° F	−108° F	−320° F	Increase +77° to −320° F
	lb/in.²	*lb/in.²*	*lb/in.²*	%
Aluminum alloy, 24S–T4	10.87×10⁶	11.11×10⁶	11.97×10⁶	10.1
Aluminum alloy, 75S–T6	10.27	10.75	11.55	12.4
Magnesium alloy, FS–1	6.36	6.83	7.30	14.7
Titanium, commercially pure	16.7	17.5	18.96	11.3
Steel, SAE 2330 (150,000 psi)	29.18	----------	31.59	8.3
Steel, SAE 8630 (150,000 psi)	29.77	----------	32.38	8.8
Steel, SAE 4340 (150,000 psi)	30.9	----------	31.8	2.9
Steel, SAE 4340 (230,000 psi)	30.7	----------	30.9	0.6
Steel, Hy-Tuf (230,000 psi)	29.8	----------	31.3	5.0
Steel, 18-8 (210,000 psi)	23.57	----------	27.66	17.5

or change dimension permanently before fracture. Ordinarily, parts are not designed to change shape permanently in service, but invariably many of them do, at least locally. A casual examination of any airframe structure will show elongated rivet and bolt holes, and slightly distorted parts that have been forced into alignment or deformed by one or more applications of a load above the elastic range. Innumerable fillets, notches, and welds act as foci for polyaxial stresses.

Ductility is generally evaluated by the elongation and reduction of area in a tension specimen, the distortion and local increase in area of a compression specimen, the angle of twist or number of reversals in a torsion test, the number of reversed bends or angle of bend in a transverse test, and the foot-pounds of energy absorbed in an impact test, with or without notch. These criteria may not measure exactly the same phenomena as variations in structure, strength, and shape of test piece will affect the ratings.

The values obtained by making tests at approximately 70° F are used for the preparation of specifications. Statistical analysis is applied when the number of tests are sufficient. The selection of the minimum values for ductility for acceptance or rejection of material is seldom based on design considerations, but rather to control quality and uniformity, and in some cases, formability and machinability. If the stress level is 200,000 psi for steel, the accepted elongation may be as low as 6 percent, but if the stress level is 100,000 psi, the minimum accepted elongation may be as high as 25 percent. In castings over a wide range of stress levels, 1 to 3 percent is accepted. Acceptance of these values by engineers is based on successful application under operating conditions, i. e., experience. For example, the necessity for a reasonable level of center ductility in forgings for gas turbines was not definitely established until wheel fractures occurred in service, although there was much prior discussion, pro and con.

The ductility of metals changes with a decrease in temperature below 70° F, although the amount varies with the method of measurement. If these changes are used to evaluate metals for engineering applications, the basis should be their effect on the performance in service. The reduction of area (figs. 5.2 and 5.3) decreases with lower temperatures for all the metals investigated, but not appreciably above

−108° F. In fact, the 18-chromium–8-nickel stainless steel (fig. 5.2) has a slightly higher value. It is probable that the slope of the lines connecting the points for SAE–4340 steel are too steep up to −108° F, as the steel with lower nickel content, SAE–8630, shows little change in ductility. The magnesium alloy, FS–1 (fig. 5.3), shows the largest decrease, but no effects of low temperature have been noticed in its serviceability for rather extensive applications in the skin and structure of the B–36 high-altitude bomber. Ductility, as measured by elongation over a length four times the diameter, changes very little for any of these metals down to −108° F.

Static Tensile Tests

The tensile strengths of notched and unnotched specimens of aluminum and magnesium alloys indicate very little effect of the notch at either room temperature or −320° F (figs. 5.4 and 5.5). The specimen was a standard tensile specimen with a circumferential 60-deg. V-notch with a root radius of 0.010 in. and a constant depth of 0.025 in. A notch with a larger root radius would have given an increase in strength [9, 10].

In the case of titanium and steel, except for SAE–4340 at 220,000 psi, where residual stresses may mask the effect, there is an appreciable increase in strength at both temperature levels due to the notch.

The reduction of area at the root of the notch, table 5.3, is a sensitive indicator of temperature effects and rates the materials in practically the same order as the notched impact test. When the ductility is very low at room temperature, as in the case of SAE–4340, the percentage loss in ductility is much less than when the ductility is high.

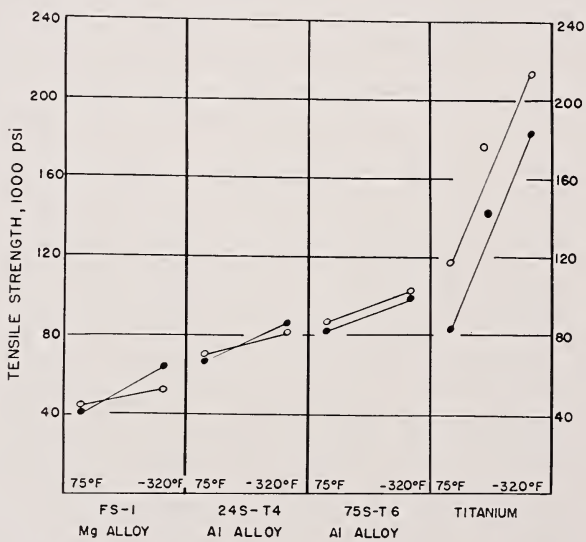

FIGURE 5.4. *Tensile strength of notched and unnotched specimens of nonferrous metals at room temperature to —320° F.*

●, unnotched specimens; ○, notched specimens.

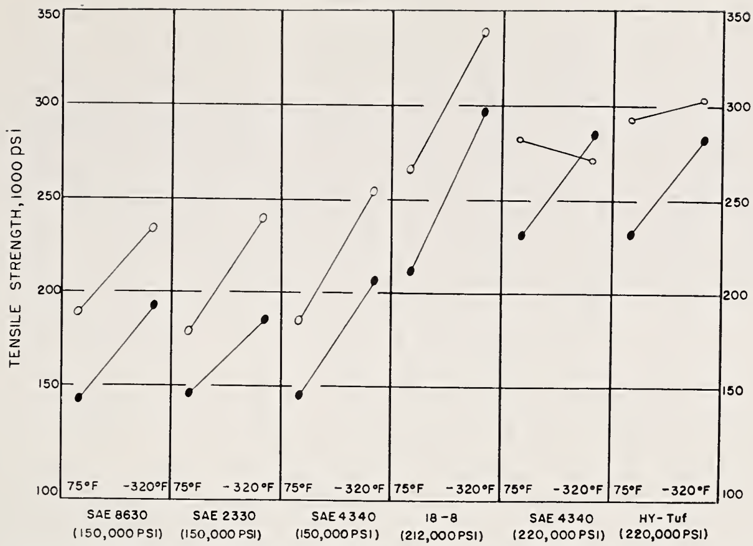

FIGURE 5.5 *Tensile strength of notched and unnotched specimens of ferrous metals at room temperature and —320° F.*

●, unnotched specimens; ○, notched specimens.

TABLE 5.3. *Effect of temperature on ductility factors*

Material	Notched static tensile			Charpy impact		
	Reduction of area at root of notch		Temperature effect	Room temperature	−320° F	Temperature effect
	Room temperature	−320° F				
	%	%	%	*ft-lb*	*ft-lb*	%
75S–T6	2.25	2.25	0	4	6	+50
24S–T4	7	6.5	−7	5.5	6	+10
18.8 stainless	11.5	7.25	−37	24.7	30.3	+25
Titanium	12.5	5.0	−60	14.5	6.6	−54
FS-1h	11.25	4.0	−65	4.2	2.0	−52
SAE 2330 (150,000 psi)	22	6.75	−70	35.8	16.8	−50
SAE 4340 (150,000 psi)	17	6.25	−61	38.7	16.4	−57
SAE 4340 (230,000 psi)	3.5	2.75	−21	19	7.4	−60
Hy-Tuf (230,000 psi)	7.5	2.5	−66	24.8	9.7	−60
SAE 8630 (150,000 psi)	13.25	3.25	−75	39.1	5.0	−87

Several investigators have studied the effects of the shape and size of test specimens, strain rates, and stress distribution over a range of temperatures on the transition from ductile to brittle behavior. It has been shown for ferrous metals that ductile fracture is dependent on the type of test. The change occurs at a higher temperature for notched impact specimens than for tension or torsion [7, 8]. Increasing the sharpness of the notch increases the severity of the test [4].

Notched Impact Tests

The effect of reduction in temperature on the notched-bar impact values for several aircraft metals is shown in figure 5.6.

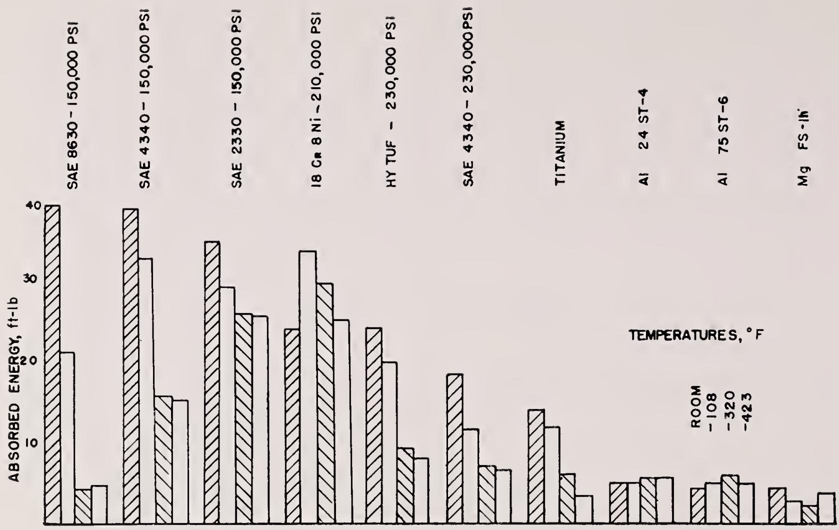

FIGURE 5.6 *Effect of test temperature on the notch toughness of various metals.* (Charpy keyhole specimens.)

The relationship between chemical composition, structure, and mechanical properties of materials such as steel, magnesium, and plastics is not sufficiently well understood to compare them on the basis of notched impact tests. For similar compositions, the impact values indicated the tendency to fail with more or less plastic deformation under the rather complicated stress pattern at the root of the notch. The correlation of impact tests on standard specimens with tests on larger machine components or structures is not too satisfactory with the exception of ballistic tests on armor plate and tear tests on ship plate [12].

Izod notched-bar tests and Stanton impact tests were used extensively at one time for measuring the quality of aircraft-engine forgings, but due to lack of correlation with serviceability, and the fact that standard specimens cut from highly stressed portions of the parts where the grain flow was not perpendicular to the plane of the notch would normally show very low values, discouraged the practice. The reduction in impact values due to temperature are of the same order as those due to grain orientation at room temperature.

Fatigue Tests

Nonrotating flexural fatigue tests were made on specimens having cylindrical test sections with diameters of 0.505, 0.300, and 0.250 in., depending on the material. This variation was necessary because of the limited capacity of the Krouse plate flexure-testing machine. All ferrous specimens were 0.300 in., and the nonferrous, except titanium, were 0.505 in. in diameter. The titanium specimens were 0.250 in. in diameter. Throughout the life of each specimen, adjustments of deflection were made periodically to maintain maximum stresses as nearly constant as possible [3]. The root diameter of notched specimens was the same as the diameter of unnotched specimens. The notch was a 60° V with a 0.010-in. nominal root radius 0.025 in. deep formed by rough machining with a milling tool and finish machining the final few thousandths with a carbide-tipped tool ground to shape. All notches were examined at 75× to determine their conformance to desired shape and smoothness. The theoretical stress concentration factors for the specimens were 3.15, 2.75, and 2.65 for the 0.505-, 0.35-, and 0.30-in. specimens, respectively, computed in the manner described by Neuber [11]. Technical factors, considering the effect of material, were in the same order as above, 1.7, 1.55, and 1.5, using an arbitrary grain size of 0.0189 in. taken from Neuber.

Unnotched Specimens

Figures 5.2 and 5.3 show the fatigue strength of unnotched specimens as compared to the yield and ultimate strengths at room and low temperatures. As previously noted [3], this fatigue strength increases without exception as the temperature is lowered. In most cases the fatigue strength increases a little faster in percentage than does the ultimate, but for the magnesium and titanium materials (close packed hexagonal) the increases in ultimate strength are 57 and 108 percent, respectively, whereas the fatigue strength increases only 47 and 37 percent. No explanation is offered except that such a behavior may be due to the different structure of these two materials as compared to the others. These comparisons are all based on fatigue strengths at 10^7 cycles.

Figure 5.7 shows the effect of temperature on the fatigue strength of unnotched specimens at three different lifetimes. It will be noted here again that magnesium alloy and titanium react differently to temperature than do any of the other materials, with the possible exception of 18–8 stainless steel. A decrease in temperature causes, for these materials, an increase in fatigue strength gain as the lifetime is reduced, whereas for the remainder of the materials the reverse is true. For example, the fatigue strength of the magnesium alloy (FS–1) is increase by low temperature only 30 percent at 10^7 cycles but in increased 50 percent at 10^4 cycles. This condition might be expected as the static strength is more rapidly increased by decreasing the temperature, and 10^4 cycles is nearer the static ¼-cycle test than is 10^7 cycles. To take an extreme example of the reverse situation, the fatigue strength of the Hy-Tuf steel (220,000 psi) is increased by low temperature about 35 percent at 10^7 cycles, but the change at 10^4 cycles is slightly negative. This phenomenon might be important in those designs where high allowable cyclic stresses are encountered.

Notched Specimens

Figure 5.8 shows the same comparisons for notched specimens between materials and effect of temperature as did figure 5.7 for unnotched specimens. The materials are arranged in the same order as in figure 5.7, and it will be noted that the order would be completely changed in Figure 5.8 if the sequence was for notched specimens: 75S–T6 would be fourth from the left rather than at the extreme right.

There are several points of interest on this figure, the first of which is the high percentage increases in fatigue strength at 10^7 cycles at -320 F for 24S–T4 aluminum alloy and 18–8 steel. These increases are even higher than for unnotched specimens. Otherwise, increases were less for the notched specimens than for unnotched specimens, with the exception of titanium, which reacted similarly, notched and unnotched.

In several cases, notching caused a decrease in fatigue strength, notably for the heat-treated steels, which is large enough to be considered in designs for applications at very low temperatures below those of significance in the operation of aircraft.

The same general trend as noted for unnotched specimens relative to effect of low temperatures at various numbers of cycles was observed also for notched specimens. The magnesium alloy and titanium in this case followed the trend of the other materials in having their fatigue curves start close together and diverge as the number of cycles increased.

Figure 5.9 shows the relation of the experimentally determined fatigue-stress concentration factors to the theoretical factor calculated by Neuber's [11] method and formula.

$$\frac{K_f-1}{K_t-1}=q,$$

where

$$K_f=\frac{\textit{fatigue strength (unnotched)}}{\textit{fatigue strength (notched)}}$$

K_t = *theoretical stress concentration factor.*

The three sizes of specimens are thus on one basis. Any point falling above the value $q=1$, indicates that the actual experimental stress-concentration factor is greater than the theoretical factor. It will be noted that one point at room temperature and several at low tempera-

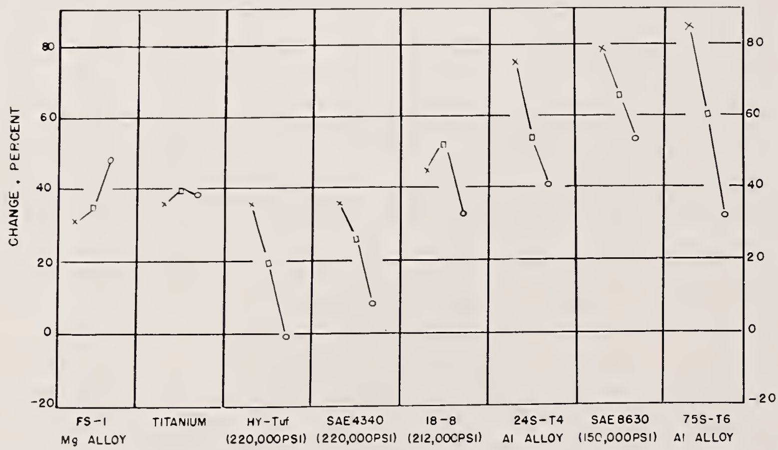

FIGURE 5.7. *Effect of change in temperature from room temperature to —320° F on the fatigue strengths at 10^4, 10^5, and 10^7 cycles of unnotched specimens of various metals.*

The metals are arranged from left to right in the order of increasing effect at 10^7 cycles. ×, 10^7 cycles; □, 10^5 cycles; ○, 10^4 cycles.

ture do have q values greater than 1. (This phenomenon has also been observed at room temperature by other investigators.) No explanation is readily available for the occurrence of an actual factor greater than theoretical. In fact, there is considerable evidence pointing in the opposite direction. Neuber has developed along this line a technical stress-concentration factor based on materials charac-

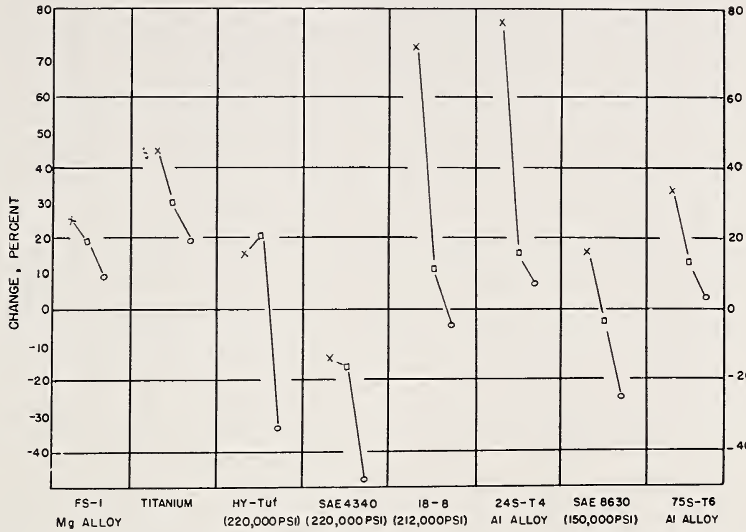

FIGURE 5.8. *Effect of change in temperature from room to —320° F on the fatigue strengths at 10^4, 10^5, and 10^7 cycles of notched specimens of various metals.*

The metals are arranged in the same order as given in figure 5.7. ×, 10^7 cycles; □, 10^5 cycles; ○, 10^4 cycles.

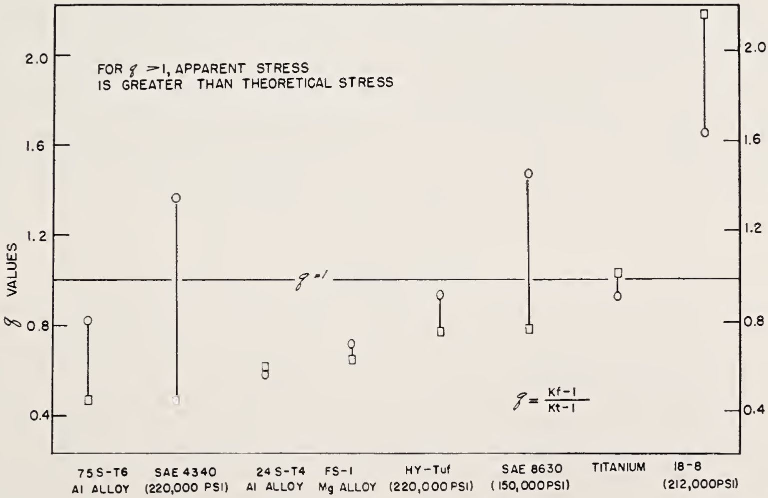

FIGURE 5.9. *Effect of temperature on the notch sensitivity (q) for fatigue strength at 10^7 cycles of various metals.*

□, room temperature; ○, -320°F.

teristics, and flank angle of the notch that reduces the theoretical factor as much as 50 percent or more, and some agreement, especially in ferrous materials, has been found between this factor and the experimental factor obtained for long lifetime fatigue tests. In view of the results shown in figure 5.9, however, it should not be assumed, especially for low temperatures, that stress concentrations will be no larger than those calculated theoretically.

Tests at Liquid-Hydrogen Temperatures (−420°F)

A few fatigue tests have been made on notched specimens of 24S–T4, 75S–T6, aluminum alloys, and SAE 8630 steel (150,000 psi) at −420° F. The test results for the aluminum alloys fall within the scatter band for the temperature range +75° F to −320° F. However, there are indications of a knee appearing in the curves for both materials at about 10^5 cycles, whereas no such indications occurred at the other temperatures. Tests were not carried out past 10^6 cycles because of the large quantities of hydrogen consumed, but it appeared that the fatigue limit at −420° F, due to the leveling out of the curve past the knee, would be slightly higher than for the other temperatures.

In the case of SAE 8630, the curves at +75° F and −108° F were about parallel, with the −108° F curve a little higher. At −320° F the curve was much lower than the other two for a low number of cycles, crossing them at about 200,000 cycles and then falling into the −108° F curve. The −420° F curve was still lower at a lower number of cycles, crossing the others at about 200,000 cycles and terminating at 10^6 cycles, considerably above the others (58,000 psi compared to around 40,000 psi). The −420° F curve was almost flat, however, from 10^4 to 10^6 cycles, failing in 11,000 cycles at 65,000 psi. Failure occurred in 1,100 cycles at 85,000 psi compared to about 10,000 cycles at 125,000 psi at room temperature. This effect in the low cycle range again emphasizes the care that should be taken in the design of applications for high stresses as only a few applications of load may cause fatigue failures.

Fatigue tests on small specimens, with or without stress raisers such as notches, are useful for indicating the effects of changes in environment. The actual values are not too significant to designers. The stress patterns in many machine components cannot be completely resolved and analyzed to the point where such basic data as that obtained in conventional fatigue tests are applicable.

Conclusions

The laboratory tests on standard and notched specimens at low temperatures indicate that the allowable design values determined at room temperature for the standard aircraft metals are safe for low-temperature operations. This is confirmed by the fact that no failures attributable to a change in mechanical properties of the material have been reported from operating stations in cold climates. Most failures due to low temperatures are caused by unequal contraction of unlike materials producing galling or binding, with subsequent overstress and distortion or failure.

The effect of crystal structure, chemical composition, grain size, and microstructure on the mechanical properties of metals at low temperatures has been investigated and considerable data are published, especially for the iron-base alloys. The low-alloy steels normally contain two or more of the elements nickel, chromium, molybdenum, aluminum, and vanadium in sufficient proportions to lower the transition temperature below that of steel with equal carbon content, but no other alloying element. With a large demand for low-alloy steels, the supply of alloying elements becomes a problem in conservation, and any savings that will not be detrimental to the end use of the product are desirable. The use of alloy to reduce the transition temperature should be approached by making a diagnosis of the effect of lower ductility on the serviceability of the part. The use of the notched impact test as the sole criterion of ductility is questionable, unless service failures of components can be correlated with the impact values.

Impact tests are satisfactory for demonstrating a transition from tough to brittle fracture. The designer can eliminate the sharp notch, and an impact load is seldom transmitted to the metal parts in flight or landing. The evaluation of the metal for low-temperature serviceability requires more extensive data. The change in ductility in a tensile test is more informative and subject to better interpretation by an engineer. The best criterion is a simulated service test on the actual machine part or structure at the lowest temperature expected in service, using the leanest alloy that satisfies the requirements for strength and hardenability at room temperature.

Discussion

Mr. F. M. Reinhart, Metallurgist, National Bureau of Standards, Washington, D. C.: Mr. Shinn, do you know how much carbon your titanium contains?

Mr. D. A. Shinn: Approximately ½ percent.

Dr. Peter B. Kosting, Metallurgist, Watertown Arsenal, Watertown, Mass.: I should like to make a few comments, which apply not only to this paper but also to others. The value of the impact test to the engineer lies in the correlation between temperature and strain rate, which may be capable of being utilized in design procedures, considering conditions for brittle failure. This correlation incorporates a constant that has been referred to sometimes as a "metal constant" or a "material constant." Perhaps not enough thought and experimental work has been given to the study of such constants.

The engineer is responsible for the functioning of the components that he designs, and he has to be sure that the components will function at the service temperature. It is convenient to distinguish between functioning on the first application of the load without failure and functioning after repeated applications of load without failure. These go hand in hand with considerations of the strength of the component and the strength, ductility, and toughness of the material going into the component. The work of MacGregor and of Lessels demonstrates that the repeated application of load to mild steel does affect the toughness and low-temperature performance. The effect might simply be a cold-work effect. However, I think their observations are of considerable importance and take this opportunity to bring them to your attention.

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Dr. Alexander L. Feild, Associate Director of Research Laboratories, Armco Steel Corp., Baltimore, Md.: I have listened to the paper by Johnson and Shinn with particular interest. I attended a meeting some years ago at the Bureau at which the subject of the yield strength of 18–8 sheet was under discussion. Today I have heard a great deal about yield strength in tension but not a word about yield in compression.

Now, in most cases we have to consider them both because the stressed member is usually a beam in flexure. Therefore, the lowest yield strength is the one on which the designer must depend. When we speak of yield strength in aircraft members, we have to consider the yield in compression, as well as in tension.

The excellent properties of the austenitic chromium-nickel steels at low temperatures have attracted much attention from design engineers over a period of years. It seems timely, however, to point out that the tensile properties of such an austenitic steel as 18–8 when processed by cold-rolling to high tensile strengths are decidedly anisotropic. So far as I am aware, systematic measurements of such tensile properties in the case of the high tensile 18–8 material has been confined to room temperature or thereabouts.

For previous work with respect to tensile measurements at room temperature on high tensile 18–8, reference is made to published bibliographies [13, 14]. Reference might also be made to the commercial bulletin published in 1944 by the American Rolling Mill Co. (now Armco Steel Corp.), entitled Design data on high tensile stainless-steel sheets for structural purposes. In the case of flat-rolled products the design engineer must consider both tensile and compressive properties because, in general, the stresses on such products are those of a beam due to bending, and the design must be based on the lower value, whether of compression or tension. About 12 years ago certain industries became very much interested in the use of full-hard cold-rolled 18–8 as a possible substitute for other flat-rolled products in a number of applications. The importance of this problem was realized by the National Advisory Committee for Aeronautics, and as a result the Committee requested the National Bureau of Standards to determine the compressive properties of 18–8 sheet or strip. Much work was carried on to perfect a suitable compressive-test method for sheet. Since that time other improved test methods have been developed, as disclosed in the bibliographies referred to above. As an indication of the anisotropy of 18–8 at room temperature, reference is made to page 8 of the above-mentioned bulletin, where the yield strength (0.2 percent offset) in longitudinal tension of full-hard cold-rolled 18–8 was 153,000 psi, whereas the same property in longitudinal compression amounted only to 102,000 psi.

It seems to me that it would be of interest to know whether this anisotropy, observed to such a marked degree at room temperature, is diminished at subzero temperatures or is magnified.

Mr. Shinn: I might point out that the yield strength in tension and compression both increase with a decrease in test temperature.

Dr. Feild: Well, that is interesting, Mr. Shinn. The point I wanted to make is that in the prior work the longitudinal compressive yield is far lower at room temperature than the tensile yield.

Dr. V. N. Krivobok, Charge Stainless Steel Section, International Nickel Co., New York, N. Y.: Do you have any data?

Mr. Shinn: We have a little, I am not sure we have much.

Dr. Krivobok: I agree with Dr. Feild on the importance of data in compression for the reason that he stated. I was asked if data were available on compression at subzero temperature. We have not secured any data of our own. There are no data in the literature, as far as I know. I wrote to several possible sources of information, but was not successful in getting any, so I assumed that experimental data on yield in compression at subzero temperatures are not available. It is for this good reason that I omitted its discussion in my paper.

Mr. Shinn: Compressive yield strength data, longitudinally, are available on five of the materials described in the paper. One of these was cold-rolled 18–8 rod. These results showed that the longitudinal compressive yield strength for 18–8 increases from 131,100 to 150,700 to 175,500 psi for room temperature, −108° and −320° C, respectively. No data are available for transverse properties, but it would be expected that they would also increase. As an increase occurs, not too much emphasis was placed on obtaining these properties, as those which decrease, or may decrease, are of most importance. The compressive data are listed in Air Force Technical Report 5662, part II.

The comments and additional information presented by the various discussers are appreciated.

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6. Properties of Austenitic Stainless Steels at Low Temperatures

By V. N. Krivobok [1]

This paper presents a study and summary of published and unpublished information on the strength characteristics of austenitic stainless steels at various subzero temperatures. The effect of composition on the mechanical properties at room and subzero temperatures, the calculations of true strength from test results, and the influence of heat treatments, especially those that might result either in carbide precipitation or sigma formation, on the mechanical properties at low temperatures are discussed in some detail. The important beneficial influence of extra low carbon content is presented.

Data on modulus of elasticity, fatigue, impact, tensile impact, effect of notches, distribution of ductility at the notch and the beneficial effects of low temperature on ductility characteristics are summarized and appraised.

Introduction

This discussion is a summary of available knowledge on the engineering properties of austenitic stainless steels at subzero temperatures, augmented by the results of several investigations currently in progress. Preliminary data are presented in connection with some specific studies, such as those of impact properties in which the concept of local plastic deformation is supported by experimental results. In the majority of cases, the inherent properties of metals are not permanently changed by exposure to low temperatures; upon return to room temperature, they resume their usual values.

Many intriguing observations are made while carrying out experiments on metals and alloys at subzero temperatures. Some observations made in testing austenitic stainless steels in tension, impact, and in fatigue are described.

Effect of Subzero Temperatures on Tensile Properties

Stainless steels of the 18-percent-chromium–8-percent-nickel class (with and without other additions) are of the metastable austenitic type. Working, that is, plastic deformation, of the metastable austenitic steels causes a preferential, as to location, phase change. This phase change, confined at first to a few and finally to numerous slip planes, results in the hardening and strengthening of the metal. During mechanical working, such as stretching, bending, and testing in tension, the resulting phase change proceeds at a greater rate at subzero temperatures. McAdam and coworkers [1] [2] in their comprehensive and detailed study of the shape of the stress-strain curves used the

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[2] Figures in brackets indicate the literature references on p. 135.

above concepts to explain the presence of two maxima observed in stress-strain curves of 18-percent-chromium–8-percent-nickel steels initially as annealed and as cold-worked.

Tensile Strength, Annealed Steels

In agreement with many other metals, the tensile properties of austenitic stainless steels, or, more precisely, the tensile strength and true breaking strength, are found to be markedly increased as the temperature of the test decreases. Tests have been performed at temperatures as low as $-320°$ F, and no exception to the above rule has ben found. The effect of low temperatures on yield strength is much less pronounced. An attempt has been made to arrive at some general conclusion regarding the "probable" magnitude of such an increase. As the rate of hardening of austenitic stainless steels by cold-working is profoundly influenced by the composition, it is to be expected that the relative increase in strength at subzero temperatures also would be similarly affected by the actual analysis of the steel. A study of numerous test results shows that in straight 18-percent-chromium–8-percent-nickel varieties the influence of composition is not too great (see fig. 6.1). Each separate curve is for a steel of the general 18-percent-chromium–8-percent-nickel type, but with variations in chromium content from 16.5 to 20.8 percent, nickel from 8.2 to 11.4 percent, and carbon from 0.06 to 0.18 percent.

Stabilized grades 321 and 347, to which should also be added type 316, are somewhat less responsive to the effect of low testing temperatures than type 302, as is shown in figure 6.2, in which various types of stainless steels are represented by their average strength values at room and at below room temperatures.

Tensile Strength, Cold-Worked Steel

As in many instances the austenitic stainless steels are used in the cold-rolled condition, i. e., hardened (or strengthened), it is of interest to know what can be expected from these steels when tested in tension at subzero temperatures. To permit the comparison between annealed and cold-rolled steels, the former are included. The results given in figure 6.3 were obtained from steels processed at the mill to standard specifications. As the temperature of the test is lowered from room temperature to 32° F, then to $-60°$ F, and finally to $-105°$ F, the increase in strength is considerably greater in annealed than in cold-worked steels. The results also show that the analysis of the steel has an appreciable effect upon the actual amount of increase, although the rate appears to be somewhat near constant. Such a profound increase in strength with decrease in test temperature is not accompanied by a sharp drop in ductility as judged by elongation and reduction of area. The latter especially is little affected, with the result that the true breaking strength of stainless steels at subzero temperatures is unusually high. An illustration of the magnitude of the expected values is given in table 6.1. The steel is type 302, tested in two conditions, annealed and cold-drawn, to "full hard" temper.

A recent publication [2] supplies interesting data on the true breaking strength and elongation at $-320°$ F of cold-worked stainless steel and other materials (table 6.2).

Table 6.1. *Influence of test temperature on the tensile properties of type 302 steel*

Condition	Temperature	Yield strength (0.2% offset)	Tensile strength	Elongation in 2 in.	Reduction of area	Breaking strength
	° F	*psi*	*psi*	*%*	*%*	*psi*
Annealed	75	59, 000	108, 000	51. 0	72. 0	285, 000
Do	0	62, 000	146, 000	47. 0	70. 0	330, 000
Do	−60	56, 000	162, 000	39. 0	68. 0	340, 000
Do	−105	50, 000	175, 000	33. 0	65. 0	337, 000
Full-hard	75	150, 000	180, 000	15. 0	69. 0	358, 000
Do	0	----------	185, 000	26. 0	67. 0	375, 000
Do	−60	162, 000	195, 000	37. 0	66. 0	384, 000
Do	−105	175, 000	197, 000	36. 0	65. 0	390, 000

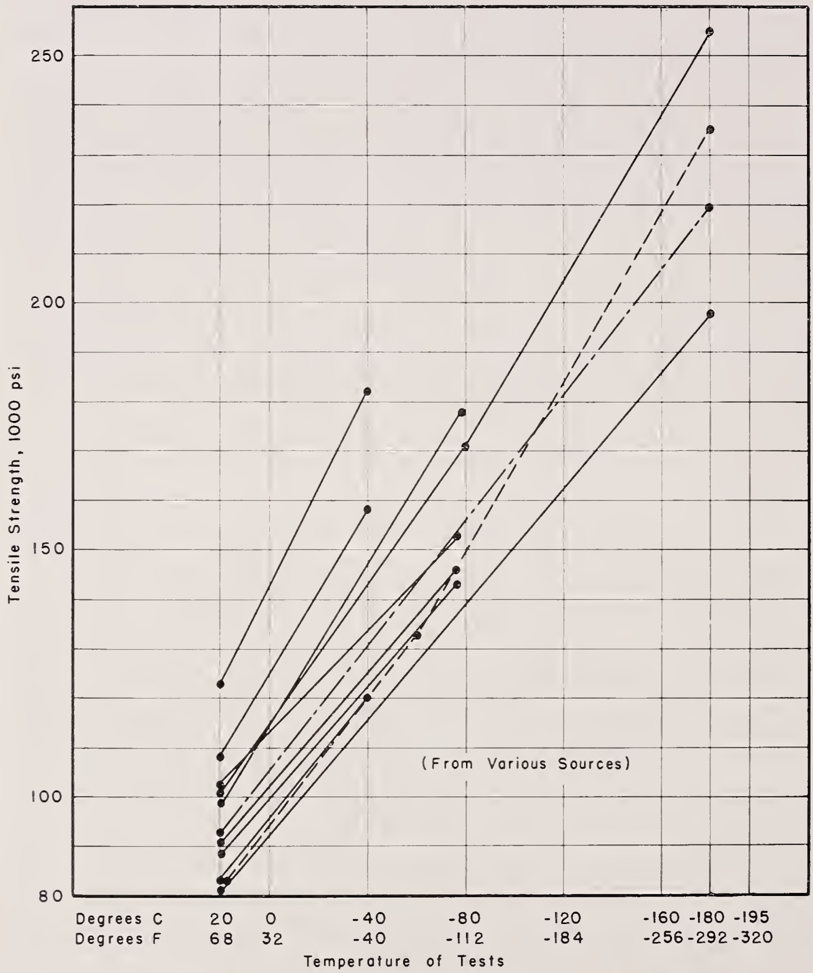

Figure 6.1. *Influence of test temperature on the tensile strength of austenitic stainless steels of 18-percent Cr–8-percent Ni type composition*

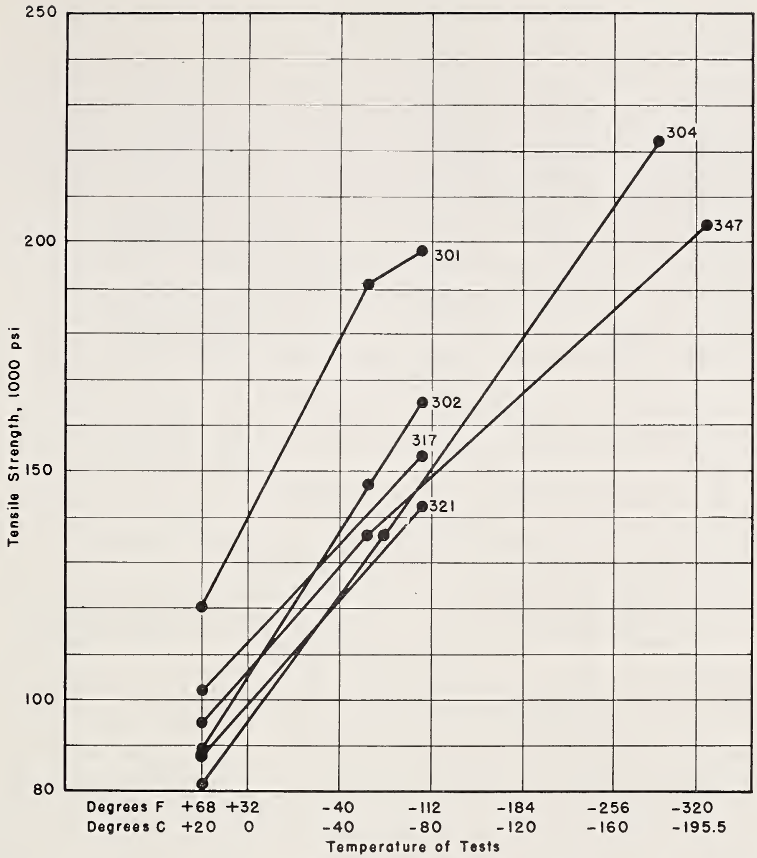

FIGURE 6.2. *Influence of test temperature on the tensile strength of different types of commercial austenitic stainless steels.*

TABLE 6.2. *Tensile properties at room temperature and —320° F of different materials*

Material	Tensile strength at room temperature	Elongation at room temperature	Tensile strength at —320° F	Fracture stress at —320° F	Elongation at —320° F
	psi	%	*psi*	*psi*	%
24S–T aluminum	69, 900	20. 0	87, 300	107, 000	21. 5
FS–1 magnesium	40, 100	16. 5	62, 800	69, 300	6. 5
Titanium	89, 600	27. 5	186, 500	221, 500	14. 0
18–8 stainless steel	210, 000	12. 0	296, 200	538, 700	29. 5

It seems appropriate to point out a fundamental difference in the effects of subzero temperatures on the mechanical properties of austenitic chromium-nickel and ferritic (or martensitic) chromium stainless steels; the beneficial effect of low temperature on the latter is much less than on the former. As an illustration of the importance of this observation, figure 6.4 is of interest. The rate at which the strength of chromium-nickel steel is increased is so much greater

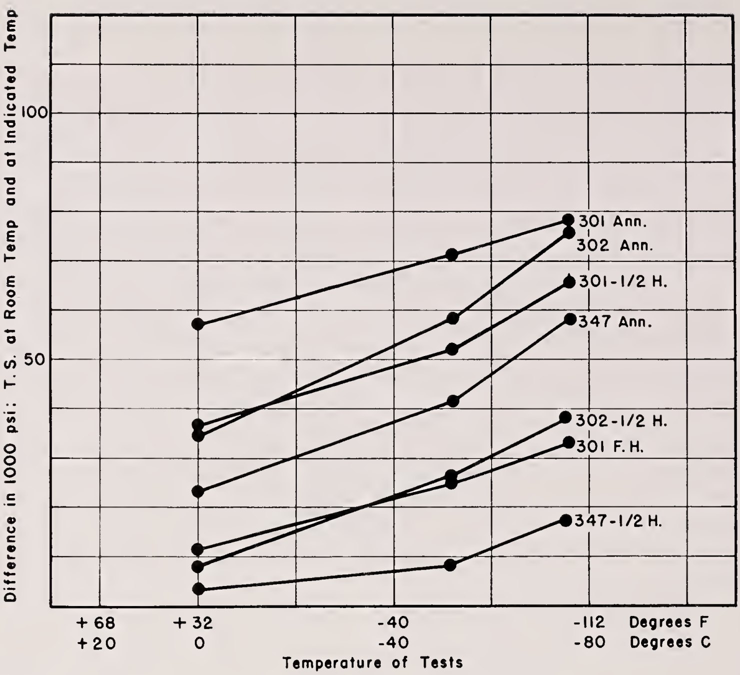

FIGURE. 6.3. *Influence of test temperature on the tensile strength of annealed and cold-rolled austenitic stainless steels.*

than that of chromium steels of either the 410 or 430 type, that in either the annealed or hardened tempers the chromium-nickel steel will show at subzero temperatures a marked superiority over the straight chromium steel. In fact, the strength of half-hard type 302 at room temperature may actually be lower than that of hardened and stress-relieved type 410, yet at −300° F or thereabouts the strength of the chromium-nickel steel is greater than that of the chromium steel. This superior strength of the austenitic steel at low temperature in the hardened or in the annealed condition is accompanied by pronounced ductility and toughness, the two qualities in which, as is well known, straight chromium steels are lacking.

The effect of low temperature on the yield strength does not appear to be as clear cut and proved as its effect on the ultimate strength. In fact, numerous data show that in some steels, either annealed or cold-worked (i. e., in the hard tempers), the yield strength is lowered as the temperature goes down. The work of the author and his colleagues at the Research Laboratories of The International Nickel Co., Inc., has often shown a small, gradual, but definite lowering of experimental values for yield strength (or proof stress) as the temperature of the tests was carried down to −105° F. Other workers have observed and reported the same tendency, which persisted even at temperatures lower than −105° F. According to recent results reported by Hoke, et al. [3], however, both the proof stress and yield strength are not noticeably affected by lowering the temperature of the tests down to −300° F. In fact, a slight increase is given for an-

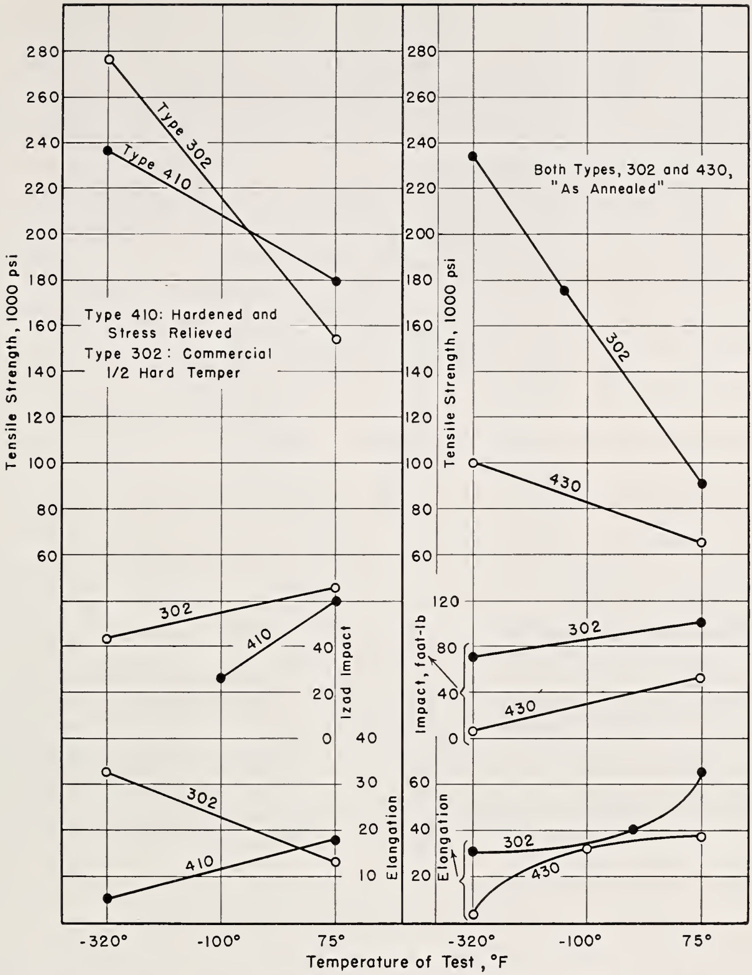

FIGURE 6.4. *Effect of test temperature on the tensile and impact properties of annealed and cold-worked ferritic and austenitic stainless steels.*

(Data for types 410 and 430 by Research Laboratories, Armco Steel Corp.)

nealed types 302 and 316. It is believed that these two steels should be grouped with the others, i.e., to say the slight increases in yield and proof strengths may not be real as the testing technique at such low temperatures is indeed difficult.

For the actual values for yield strength and proof stress, the reader is advised to study the excellent paper by Hoke, et al., as well as other references mentioned in the appended bibliography.

Ductility of Annealed and Cold-Worked Steels

The annealed austenitic stainless steels, similar to other ferrous alloys, lose some ductility when tested in tension at extremely low temperatures. However, the ductility as measured by the elongation in 2 in. and reduction of area are still sufficiently high to justify the statement that all types of austenitic steels remain ductile down to −320° F. The ductility of these steels increases slightly as the temperature is lowered immediately below that of room temperature. The values for elongation expected at room temperature and at −320° F are presented in table 6.3.

Table 6.3. *Effect of test temperature on the elongation of annealed austenitic stainless steels*

Steel	Elongation in 2 in. at—	
	Room temperature	−320° F.
	%	%
301	65 to 60	40 to 30
302	65 to 60	48 to 40
303	60 to 50	30 to 20
304	65 to 60	45 tc 35
321	55 to 50	40 to 30
347	60 to 50	45 to 35
310	50 to 45	45 to 35
316	55 to 50	50 to 35

A few individual tests performed at about −423° F showed that specimens of both types 302 and 304 had an elongation of approximately 25 to 28 percent.

The values given in the above table are based on strain rates within the limits of 0.01 to 0.05 (in./in.)/min; the rate of straining may have a pronounced effect on the ductility.

Reduction of area of the annealed steels is even less affected by the lowering of the temperature of testing than is the elongation. Numerous data show that the average reduction of area at about −300° F is approximately two-thirds of its value at room temperature. This characteristic has been pointed out in the discussion of true breaking strength.

With cold-rolled austenitic stainless steel, the elongation increases as the temperature of deforming decreases. This is an important fact that has not received the attention it merits. Furthermore, the increase in ductility can be of very considerable magnitude, being governed by the analysis of the steel, by the temper (i. e., by the degree of cold-work), by the temperature, and possibly by the rate at which deformation (in tension) is carried out.

An illustration of the above fact is shown in figure 6.5, which summarizes the findings of several investigations on cold-rolled types 302 and 347 stainless steels. It is clear that with a decrease in the temperature of testing, the average increase in elongation over the 2-in. gage length is very appreciable. The word "average" is emphasized because the distribution of elongation (degree of necking) is also affected by the test temperature. This will be discussed presently.

In figure 6.6 are summarized the data collected on commercially produced steels but rolled to various tempers in the laboratory and then tested at room temperature and at −320° F. There seems to be an indication of a trend, namely, that improvement in ductility at subzero temperature is about the same, regardless of the initial tensile

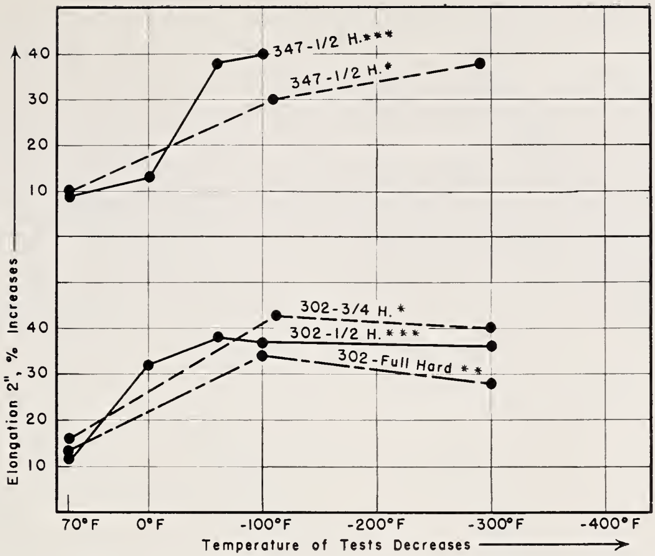

FIGURE 6.5. *Effect of test temperature on the elongation of cold-worked austenitic stainless steels.*

*National Bureau of Standards; **Ohio State University, Research Foundation; ***Research Laboratory, International Nickel Co.

strengths to which steels 302, 304, and 347 have been cold-rolled. Note that the solid lines (for each alloy) are roughly parallel to each other. One notable exception on this graph is that of steel 301. It is assumed that the instability of austenite of this composition, and its rate of work-hardening, are responsible for the observed behavior, which differs from all other types, including such relatively stable ones as 310.

Distribution of Elongation

The origin of the selection of 2 in. as the length by the stretch of which the ductility of various materials is judged is an enigma. With due and profound respect for the tradition (and vast accumulated data), it is submitted that it is a simplified, or rather over-simplified, criterion for ductility. It does call for further diagnosis and sometimes with interesting results. Percentage elongation in 2-in. gage length measured after the fracture of the test piece gives us an average stretch, including the "break", and thus obscuring the true stretching characteristics of the metal, *per se*. An observed fact is that the percentage elongation in 2 in. is not necessarily a dependable measuring stick for the behavior of a given metal, or of a metal in a given state of heat treatment, in forming or other mill operations. Studies during and after World War II, brought forth the significant differences in distribution of elongation, which are observed and measured in samples plastically stretched to the beginning of "necking" or in the samples tested to destruction. The pattern of distribution of elongation is obtained by dividing the standard gage length of 2 in.

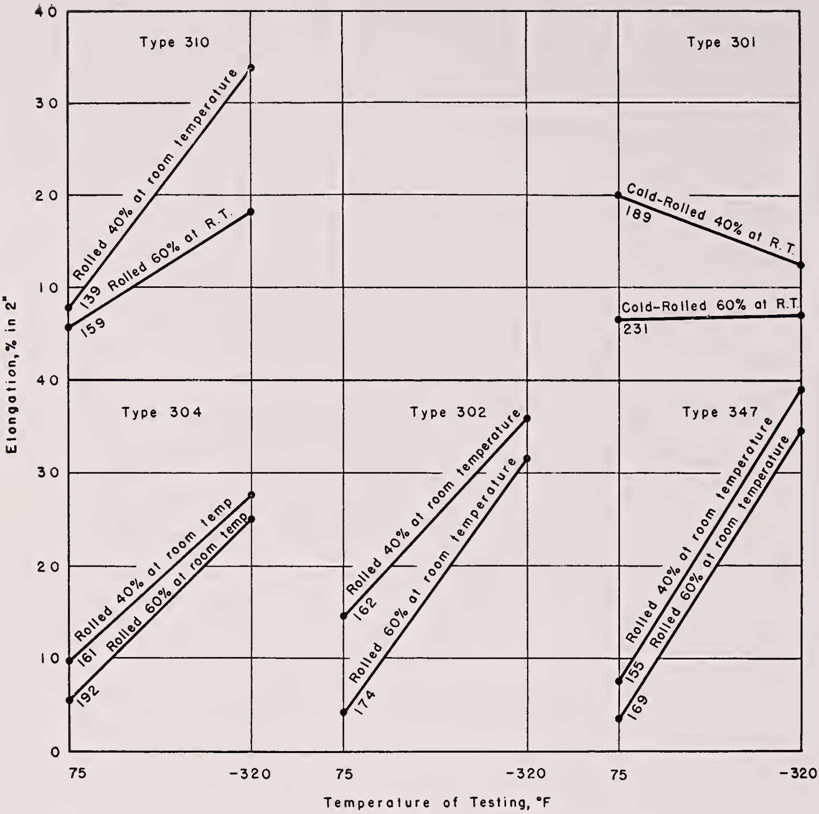

FIGURE 6.6. *Effect of test temperature on the elongation of cold-rolled austenitic stainless steels.*

Figures at the left of each curve indicate tensile strength in 1,000 psi.

into smaller components, that is, using as a unit of length values of 0.25, 0.10, and even 0.010 in. The significant differences can readily be obtained by using 0.25-in. gage length, and this will be used throughout the discussion that follows.

Figure 6.7 illustrates this difference in the distribution of elongation of annealed and cold-rolled type 302 stainless steel when tested in tension at room temperature. For the annealed condition, the pattern of the distribution of elongation over the whole 2-in. gage length is essentially the same, whether it is tested at room temperature or at −105° F. The pattern is that of a sharp maximum at or close to the final fracture, this maximum gradually diminishing toward each end of the gage length. The ductility is quite high, although it is somewhat lower at −105° F than at room temperature.

An altogether different picture is presented by the same steel when tested in the cold-rolled condition (approximately 155,000- to 161,000-psi tensile strength). Obviously, the pattern of distribution of elongation is rather profoundly altered by the temperature of the test. At

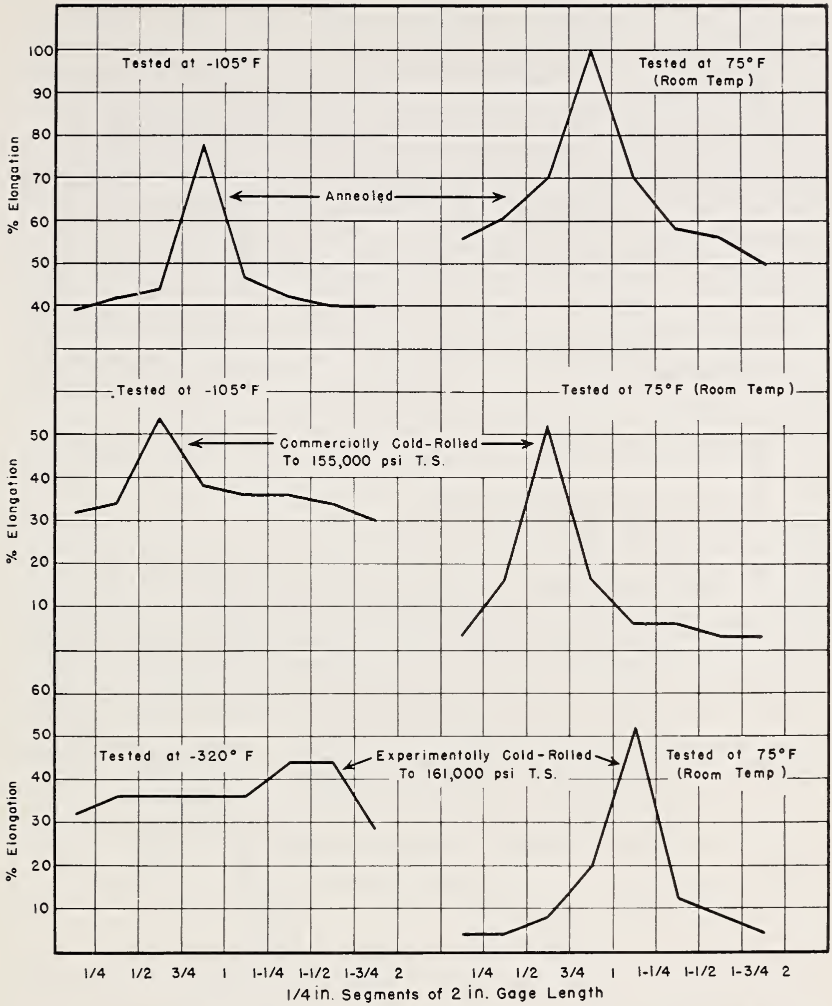

FIGURE 6.7. *Effect of test temperature on distribution of elongation in 2 inches of annealed and cold-rolled 302 stainless steel.*

room temperature practically all the elongation measured on 2-in. gage length is confined to that relatively small part of the test piece that contains the ultimate fracture. A short distance away from the break, the elongation is down to relatively low values of about 15 percent and continues to drop with further increase in distance from the point of fracture (peak in curves). This obviously suggests that the location of the initial plastic stretch, followed by "necking," determines the location of the final fracture. At a test temperature of −105° F maximum elongation is attained at or near the final break, and the value is the same (about 51 percent) as that of the sample tested at room temperature. Even at a very considerable distance away from the break the ductility is about 30 percent. When the temperature of the test is further lowered to −320° F, the pattern of the distribution of elongation assumes almost a flat shape. This is a sharp contrast to the pattern for the same steel when tested at room temperature, as is illustrated by the two lower curves.

Similar differences in the patterns for distribution of elongation have been found in the case of other types of stainless steels, cold-rolled and subsequently tested at subzero temperatures; 304, 347, 321, 316, and even 310, show the same general picture, differing only in detail. Extensive experimental data, the result of work done by the Research Laboratories of The International Nickel Co., Inc., have been found useful in applying the deductions from the observed factual data to various problems of forming and fabrication.

A rather good illustration of the changes in the pattern of distribution of elongation is furnished by the studies of type 347 stainless steel originally rolled at room temperature to a tensile strength of 154,000 psi and then tested at a gradually decreasing temperature, starting at room temperature, then at 0°, −60°, −105°, and finally at −320° F. These data are summarized in figure 6.8, which shows how the distribution of elongation of cold-worked stainless steels is affected by gradually lowered temperatures. For comparison, the data on steel 302 are given in the same figure. In the case of 302, it is evident that lowering the testing temperature only to 0° F materially alters the stretching characteristics and materially improves the ductility. Further lowering of the test temperature to −60° F merely tends to distribute the elongation somewhat more evenly, but still

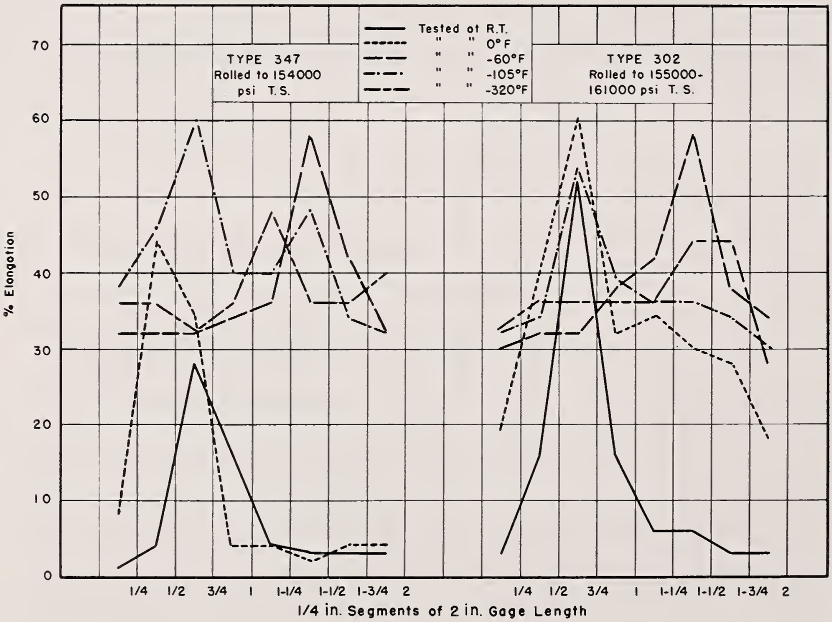

FIGURE 6.8. *Variation in elongation in 2-inch gage length of cold-worked 302 and 347 stainless steels tested in tension at different temperatures.*

with a pronounced peak. Reference to the original records will permit an interesting observation, namely, that the average percentage elongation as measured on 2-in. gage length when tested at −60°, −105°, and −320° F is, for all practical purposes, the same; 38, 37, and 36 percent, respectively. Yet one will see at once the differences as are presented by the distribution pattern at, for example, —60° and at −320° F.

In the case of type 347, further lowering of the test temperature to 0°F produces practically no difference in the distribution pattern; this pattern is similar to the one obtained during the test at room temperature. Further lowering to −60° F brought a most marked change, as can well be seen from figure 6.8.

Thus it must be surmised that the optimum conditions for ductility, and especially for its distribution, are governed by the analysis of the steel, as well as by the temperature. Undoubtedly, too, the amount of cold-working to which the steel has been subjected prior to testing or prior to stretching would also influence the behavior of the steel ductility-wise at various levels of subzero temperature.

Modulus of Elasticity

The work that was completed in the Research Laboratories of The International Nickel Co., Inc. (table 6.4), and also a few published results, seem to prove beyond reasonable doubt that lowering of the temperature of the test raises the moduli of elasticity in tension. It is difficult, however, to go beyond this general statement as the technique of determining a true stress-strain curve at very low temperatures is a rather difficult experimental procedure.

TABLE 6.4. *Effect of test temperature on Young's modulus of elasticity of cold-worked austenitic stainless steels*

Steel	Tensile strength at room temperature	Modulus of elasticity in tension at—	
		Room temperature	−320° F
	psi	*psi*	*psi*
301	189, 000	25. 4×10⁶	28. 6×10⁶
301	231, 000	25. 8	27. 8
302	162, 000	26. 2	27. 2
304	192, 000	24. 4	26. 2
347	155, 000	25. 0	27. 0
347	169, 000	23. 3	27. 6

A rather early investigation [4] suggests numerical values for the effect of low temperatures on the properties of cold-worked and annealed stainless steels of the general 18-percent-chromium–8-percent-nickel composition. This investigation proposes between 10,000 and 12,000 psi as an increase in Young's modulus per degree centigrade fall in temperature. The values calculated from the results given in table 6.4 are in fair agreement with these suggested figures.

Recording of the above values was incidental to other work at low temperatures. The experimental difficulties associated with determination of the stress-strain curve at quite low temperatures, the shape of such curves, and the influence of such external factors as the rate of strain, do not permit us to claim for the above results the accuracy that would entitle them to be used as design criteria.

McAdam [1] in his work, already frequently quoted, clearly shows that the modulus is increased with falling temperature, and that this observation applies to annealed as well as cold-worked steels. In view of the continually changing slopes obtained on conventional stress-strain diagrams, it is obvious that the values for modulus depend on the level of stress. A better evaluation of these constants would be the study of tangent or secant modulus at any given level of stress.

Effect of Notches

The earlier investigation by McAdam, et al. [1], and the recent one by Spretnak, et al. [2], present limited data on the effect of notches. According to the work of these investigators, the effect of notches is beneficial, that is, the ultimate tensile strength of the notched sample at room or subzero temperatures is greater than that of the unnotched sample at the same temperature. McAdam also comments on the high ductility, 26 percent, at the fracture of a cold-worked 18–8 steel tested at −306° F. It is obvious that the ultimate strength of the notched samples is appreciably raised by the phase change taking place during plastic deformation. As the latter is affected by composition, temperature, and possibly the rate of plastic strain, much more additional data are needed before any further generalizations are made.

Effect of Subzero Temperatures on Impact Properties

Tension Impact

Very little data are available concerning the tension-impact properties of austenitic stainless steels, even at room temperature. The importance of this property is perhaps overlooked, or its value for design purposes has not been evaluated. At any rate, the investigation of tensile-impact characteristics of sheet material was encouraged by several leading design engineers and is now being studied by The International Nickel Co., Inc. Only preliminary results are available, but they are of sufficient importance to warrant brief mention.[3]

The purpose of the investigation is as follows: (a) To evaluate the ability of austenitic stainless steels, in various tempers, to withstand shock loading, by measuring the absorbed energy. (b) To study the type of fracture that accompanies the rupture by tension impact stresses. (c) To study the effects of notches. (d) To study the effects of subzero temperatures.

To carry out the above program, testing of a standard tension impact sample for sheet material is to be implemented by the evaluation of plastic flow, preceding the fracture, the distribution of plastic strains throughout the length of the test sample, and by testing additional samples, so designed as to simulate the effects of notches and stress concentrations.

The fact that became obvious at once is that rupture under tension impact at the average velocity of about 18 ft/sec is definitely of the

[3] Comprehensive program of tests is carried out with the cooperation of Prof. Carl Muhlenbruch of Northwestern University. The results of these studies will soon be made a subject of a technical paper.

"ductile" or "tough" type. Measuring the elongation over the ¼-in. gage length, the smallest practical distance to contain the fracture, yields figures of between 70 and 90 percent, with corresponding values for tension impact of 150 to 225 ft-lb in the case of annealed steels. Testing of cold-worked grades shows that as the steels are worked to harder tempers, the tension-impact toughness is gradually lowered, but is nevertheless accompanied by considerable ductility, judged or measured by the plastic straining that accompanies the final fracture.

An attempt to evaluate the effects of the notches shows that the tension-impact toughness is very appreciably changed by the configuration of the test sample. Figure 6.9 shows the experimental values, expressed in foot-pounds of absorbed energy, when the standard sample (a) was changed to one marked (b) and then to one marked (c). Samples (b) and (c) required only 17.0 and 21.0 ft-lb, respectively, to effect fracture. The estimate of the required energy (to break the samples) was accompanied by the measure of the permanent plastic deformation accompanying fracture. It is found that irrespective of the shape and configuration of the sample, the average plastic strain over the ¼-in. distance is of the same high order of magnitude. This is shown by values of 76, 84, and 76 percent of plastic strain, as is indicated for the samples in figure 6.9.

As already mentioned, similar studies are to be carried out at subzero temperatures. Results that have just become available show that toughness in tension impact at subzero temperatures is a firmly established characteristic of austenitic stainless steels. Type 301 annealed shows the ability to "stretch", which is greater at −40° F than at room temperature. Moreover, type 301 half-hard showed an increase rather than a decrease in tension-impact toughness as the temperature of the test was lowered from +70° to −40° F.

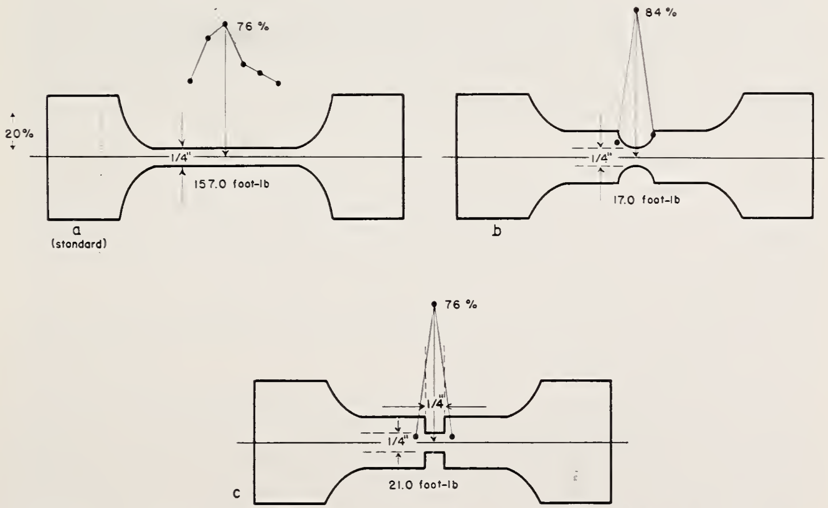

Figure 6.9. *Effect of shape of notch on energy absorbed and plastic deformation in fracturing annealed 302 stainless steel in tension impact at room temperature.*

Standard (Bending) Impact

Austenitic stainless steels have long been noted for their ability to maintain toughness, i.e., to exhibit high impact resistance at temperatures below normal. Although annealed austenitic stainless steels, when tested at progressively lower temperatures, down to −300° F, do show a tendency toward lowered values under notched impact, the residual toughness is still more than adequate. This is well illustrated by the data in table 6.5, which are based on findings of the National Bureau of Standards and others [5, 6, 7].

TABLE 6.5. *Impact resistance of annealed stainless steels at low temperatures*

(From various sources [5, 6, 7])

Steel	Charpy (V-notch) impact at—				
	+32° F	−4° F	−40° F	−112° F	−300° F
	ft-lb	*ft-lb*	*ft-lb*	*ft-lb*	*ft-lb*
302	[a] 119	------	------	------	[a] 117
304	------	157	143	124	------
321	182	181	178	169	------
347	120	122	120	123	[a] 104

[a] Izod value.

A few tests on modified test pieces [4] have been carried out at −297° and −423° F for stainless steels corresponding in composition to types 304, 309, and 321. The results of each steel showed substantially the same values at both temperatures. More recently [8], impact testing of 304 (in the form of standard Charpy impact tests) has also shown that its remarkable resistance to impact is maintained down to −423° F.

From scattered published results it appears that retention of toughness at low temperatures is also a characteristic of types 316, 309, and 310 in the annealed condition. Thus it may be concluded that all commercial types of austenitic stainless steels when properly annealed are well suited for applications where maximum toughness is required at low temperatures.

Recent studies of types 310 and 316 in the Research Laboratories of The International Nickel Co., Inc., not only fully agree with the above general statement, but furnish additional valuable information (table 6.6). The latter is in connection with the effect of prestressing (while at subzero temperatures) on impact toughness of annealed 310 and 316 stainless steels.

The experimental values clearly show that holding steel 310 at subzero temperatures and under stress produces no detrimental effects within a limited time (14 days).

The presence of ferrite in the structure of steel 316, 309, or 310 is undesirable. Alloys of "mixed" structure will show lower impact values at subzero temperatures. The only data available on impact properties at low temperatures is for stainless "W".[5] Fontana [8]

[4] So-called DVM sample measuring 55 mm (2.17 in.) in length, 10 mm (0.394 in.) in width, and 10 mm (0.394 in.) in depth, with a keyhole notch of 3 mm (0.118 in.) in depth, including a 2-mm (0.0787 in.) hole at the bottom of the slot.

[5] Proprietary name of the United States Steel Co. Unofficial, AISI number 322.

TABLE 6.6. *Effect of test temperature on the notch toughness of annealed austenitic stainless steels*

[Data from Research Laboratories International Nickel Co.]

Steel	Chemical composition						Charpy (keyhole) notch toughness at—			
							Room temperature		−320° F	
	C	Mn	Si	Ni	Cr	Mo	Annealed	Stress-aged [a]	Annealed	Stress-aged [a]
	%	%	%	%	%	%			*ft-lb*	*ft-lb*
310	0.11	1.51	0.42	21.64	27.22	----	65, 66	63	49, 50, 48	50, 50, 45
316	.08	1.54	.56	12.01	17.65	2.74	----	----	64, 65	58, 62

[a] Stressed in a special jig to 30,000 psi at −320° F. Held at −320° F. for 14 days.

reports that this steel in the "age (or precipitation) hardened" condition has very low impact values, namely, 4 ft-lb at −108° F and 2 ft-lb at −314° F.

Effect of Sensitization on Annealed Steels

As sensitization produces different effects at room and subzero temperatures, these effects should be described. Sensitization, i. e., the effect of precipitated carbides, is a function of carbon content. This was established some time ago by Tindula [9]. More recent work by E. H. Schmidt [10] partly substantiates the previous findings, as will be seen from the data on annealed samples of commercial steels 302 and 304. Figure 6.10 shows that the effect of sensitization in these two steels on impact properties at room temperature is insignificant. In fact, the data for 302 (actual analysis: C, 0.14; Cr, 18.43; and Ni, 8.90%) seem to be in disagreement with the above-cited work of Tindula. The explanation appears to be as follows: Tindula's steel was sensitized for 100 hours, and that of Schmidt only 2 hours. The spread in the values reported by Schmidt was from 70 to 100 ft-lb for sensitized steel. The lower values of this spread are not too far from values published by Tindula. Taking into consideration the difference in sensitizing treatment, one may conclude that at least the trends are in substantial agreement.

At temperatures below zero the effect of sensitization is altogether different. The impact properties of sensitized steels may be profoundly affected, depending on the carbon content, as is clearly shown in figure 6.10. Sensitized type 302 (0.14% carbon) is quite brittle at −300° F, whereas 304 (0.07% carbon) is still tough at −300° F. This and other results would suggest that recently developed "extra-low-carbon" stainless steels may have their properties unaffected by sensitization.

In our tests extra-low-carbon 304 steel, containing less than 0.03 percent of carbon, showed no detrimental effects of sensitization (2 hr at 1,200° F) at temperatures as low as −320° F. Interesting results by Dr. F. K. Bloom, Research Laboratories, Armco Steel Corp., showing the effect of sensitizing or stress-relieving treatment of welded samples on the impact properties of stainless steel at low temperatures are given in table 6.7.

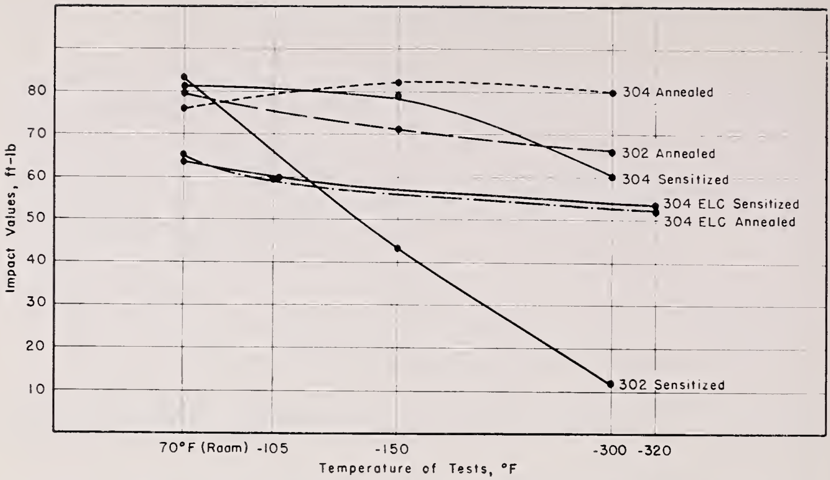

FIG. 6.10. *Effect of test temperature on the impact properties of annealed and sensitized 302 and 304 stainless steels.*

Charpy keyhole notch specimens. (Schmidt [10] and Research Laboratories, International Nickel Co).

TABLE 6.7. *Notched-bar impact test on welded specimens* [a]

[From Research Laboratories, Armco Steel Corp.]

Base metal	Condition prior to Testing [b]	Energy absorbed					
		Unaffected base metal			Heat-affected zone		
		+75° F	−60° F	−320° F	+75° F	−60° F	−320° F
	ft-lb	*ft-lb*	*ft-lb*	*ft-lb*	*ft-lb*	*ft-lb*	*ft-lb*
304 ELC	AW	76	90	71	70	76	65
Do	S	58	63	52	63	75	36
Do	SR	75	79	70	58	52	38

[a] Charpy samples, keyhole notch. Tests conducted in triplicate.
[b] AW, as welded; S, sensitized at 1,200° F for 2 hr; air-cooled; SR, stress-relieved at 1,600° F for 2 hr; air-cooled.

It will be noted that sensitization or stress-relieving treatments when imposed upon the heat-affected zone produce a drop in impact at very low temperatures. Even though the lowering of impact is noticeable, the material will still be classified as tough.

Steels 347 and 316 behave much in the manner of type 304, already described. With the advent of "extra-low carbon", columbium, or molybdenum bearing steels, inquiries were made concerning their impact behavior, in the sensitized condition, at room and low temperatures. This inquiry is natural as the low-carbon content and the presence of columbium or molybdenum, or both, might result in a mixed, that is, austenitic and ferritic structure. The tests, shown in table 6.8, tell us that no embrittlement, at room or at subzero temperatures (down to −320° F) due to sensitization, may be expected.

While sensitization is commonly associated with the well-known phenomenon of carbide precipitation, the harmful effects of heating within the range of "sigma" formation must not overlooked. This

TABLE 6.8. *Effect of sensitizing treatment on the notch toughness at room and low temperatures of different austenitic stainless steels*

Chemical composition					Condition	Charpy (keyhole) impact at—		
C	Cr	Ni	Cb	Mo		Room temperature	−105° F	−320° F
%	%	%	%	%		*ft-lb*	*ft-lb*	*ft-lb*
0.020	17.01	11.00	0.36	------	Annealed	63	72	65
					Sensitized	66	64	67
.030	18.52	12.25	.45	------	Annealed	66	63	71
					Sensitized	56	58	64
.020	17.71	11.17	------	2.44	Annealed	65	63	65
					Sensitized	68	68	69
.020	17.25	13.10	------	2.10	Annealed	63	57	58
					Sensitized	67	56	58
.020	17.80	13.10	------	2.76	Annealed	69	61	61
					Sensitized	69	64	62
.030	17.85	11.34	------	2.25	Annealed	76	65	53
					Sensitized	71	65	53

Sensitizing treatment: 2 hr at 1,200° F; air-cooled.

range and the rate of sigma formation are somewhat dependent upon the chemical composition of the steels. The maximum rate is usually associated with the temperature range between 1,350° and 1,650° F. The harmful effect consists, besides other changes, in a much lowered impact toughness at room and subroom temperatures.

The loss of impact caused by the presence of sigma phase is quite insignificant in types 302, 304, 304 ELC, and other straight chromium-nickel alloys, with the possible exception of type 309.

In types 347, 316, 317, and 318, especially of low-carbon variety, the heat treatment that would be expected to favor the formation of sigma phase is likely to affect unfavorably the subzero impact toughness of the above steels.

This general and rather guarded statement is advisable because of the well-known difficulty of positive identification of sigma phase. At any rate, the data in table 6.9 are of interest.

A study of the data reveals that, as is well known, the effect of "sigma" formation has a pronounced detrimental effect on the impact toughness of types 316 and 318, even at room temperature. This effect

TABLE 6.9. *Effect of sigma phase on the notch toughness at room and low temperatures of different austenitic stainless steels*

Chemical composition					Condition	Charpy (keyhole) impact at—		
C	Cr	Ni	Mo	Cb		Room temperature	−105° F	−320° F
%	%	%	%	%		*ft-lb*	*ft-lb*	*ft-lb*
0.020	17.71	11.17	2.44	------	Annealed	65	63	65
					Sigmatized	29	21	25
.023	17.80	13.10	2.76	------	Annealed	69	61	61
					Sigmatized	43	31	22
.021	17.30	14.10	2.94	------	Annealed	69	63	60
					Sigmatized	47	37	29
.022	18.12	12.68	2.12	0.48	Annealed	54	50	53
					Sigmatized	12	10	8
.022	18.62	14.02	2.15	.32	Annealed	70	61	61
					Sigmatized	23	17	12
.020	17.01	11.0	------	.36	Annealed	63	72	65
					Sigmatized	50	52	43

is further aggravated by lowering the temperature of the test to subzero, but the relative degree of lowering is not as great as might have been expected.

A set of data kindly furnished to us by Dr. F. K. Bloom, Research Laboratories, Armco Steel Corp. shows that "sigmatizing" treatment may not necessarily cause detrimental effects on impact toughness at subzero temperatures. These data are reproduced in table 6.10. It is to be noted that the nickel content of the tested steel was unusually high. Whether or not this circumstance affected the findings, we are not prepared to argue.

TABLE 6.10. *Notched bar (Charpy keyhole) impact test results on welded specimens*

[From Research Laboratories, Armco Steel Corp.]

Type 316 ELC: C, 0.028%; Mn, 1.28%; Si, 0.49%; Cr, 17.32%; Ni, 13.78%; Mo, 2.20%

Condition prior to testing [a]	Energy absorbed at—					
	Unaffected base metal			Heat-affected zone		
	+75° F	−60° F	−320° F	+75° F	−60° F	−320° F
	ft-lb	*ft-lb*	*ft-lb*	*ft-lb*	*ft-lb*	*ft-lb*
AW	54, 54, 55	59, 64, 53	50, 52, 50	47, 46, 43	45, 45, 46	33, 36, 45
S	56, 58, 54	50, 73, 60	47, 55, 50	51, 46, 43	50, 57, 48	42, 40, 45
SR	65, 56, 53	60, 51, 62	45, 46, 41	68, 49, 51	45, 74, 54	43, 45, 45

[a] AW, As welded; S, sensitized at 1,200° F for 2 hr; air-cooled; SR, stress relieved at 1,600° F for 2 hr; air cooled.

Effect of Cold-Work

Cold-working of austenitic stainless steels has a definite detrimental effect on impact toughness. The extent of the effect is governed by the analysis of the steel and by the amount of cold-work. One would expect that the effect of superimposed low temperatures of testing would further lower the impact toughness of the cold-worked steel, but the experimental data do not support this belief. According to W. B. Pierce, Allegheny Ludlum Steel Corp [11]: "cold-working does not cause austenitic stainless steels to become dangerously sensitive to temperature."

It should be noted that the notch toughness of stainless steel 347 is particularly sensitive to cold-working. However, cold-worked type 347 does not show a further pronounced lowering of impact strength at temperatures as low as −320° F.

Effect of Sensitization on Cold-Worked Steels

At room temperature, the effect of sensitization of either annealed or cold-worked steels (302 and 304) is practically nil (see fig. 6.11).

At subzero temperatures the sensitization of cold-worked steels has a profound detrimental effect on impact strength, much in the same manner as for annealed steels. Data supporting the above statement will be found in the portion of the curve on the right side of figure 6.11. Similar to the annealed steels, the composition, i. e., primarily carbon content, influences the extent of damage to the impact strength.

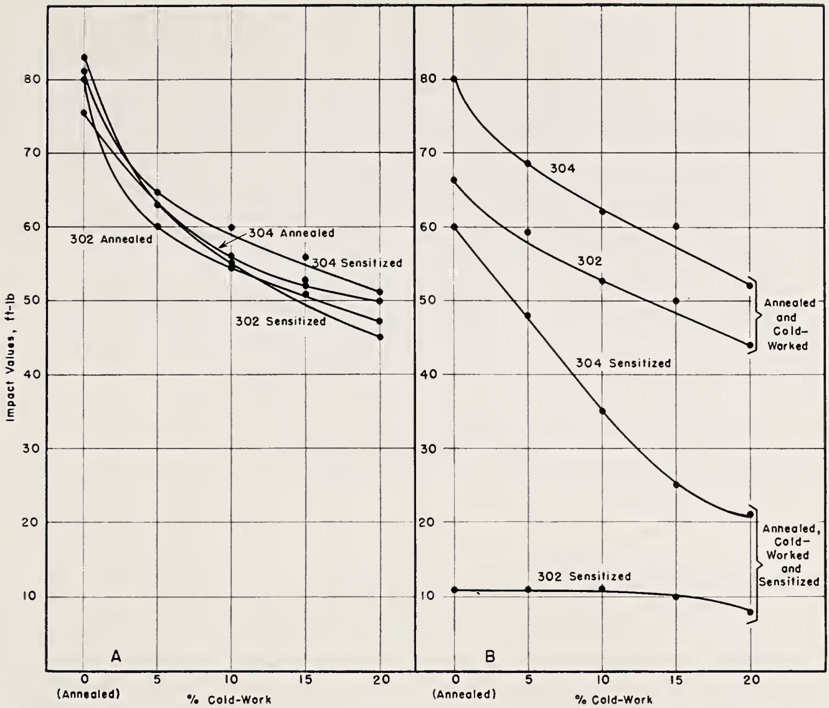

FIGURE 6.11. *Effect of sensitization on the impact properties at room temperature and —300° F. of annealed and cold worked 302 and 304 stainless steels.*

Charpy keyhole notch specimens. A, tested at room temperature; B, tested at —300° F. (Schmidt [10] and Research Laboratories, International Nickel Co.)

Effect of Long Exposure to Low Temperatures on the Properties of Stainless Steels

Exposure to low temperatures has no effect whatsoever on the mechanical properties of steels tested at room temperature subsequent to exposure. The experimental evidence has been obtained at temperatures as low as —320° F and for the duration of as long as 100 hr or in a special test for 1 year. This time was used because of the logical assumption that if any change is likely to take place, it certainly would do so in this period of time. Our experimental results were obtained on 301 annealed, half-hard, and full-hard; 302 annealed and half-hard; and 347 annealed on half-hard.

By way of illustration, 301 half-hard showed the same yield strength of 115,000 psi, the same tensile strength of 160,000 psi, and the same elongation of 30 percent before and after being exposed at —100° F for 100 hr. (fig. 6.12).

The possibility of structural changes that might occur on prolonged exposures to extremely low temperatures and affect the mechanical properties has not been too extensively studied. In the laboratories

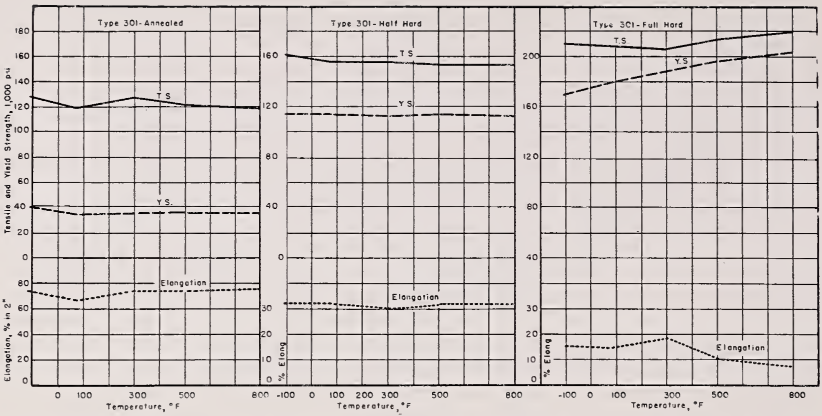

FIGURE 6.12. *Effect of exposure for 100 hours at different temperatures on the tensile properties of annealed and cold-worked 301 stainless steel tested at room temperature.*

of our company we studied the effect of the holding time at −320° F on impact, believing that the impact properties would be most sensitive to change. Tests made on commercial type 304 steel are reported by Armstrong and Miller [12], and the results show that as long as 12 months of exposure at −320° F did not affect the room-temperature impact strength of the steel. It should also be stated that exposure to liquid nitrogen for a year had no adverse effect on impact strength at the temperature of liquid nitrogen. Although these data (table 6.11) are confined to only one steel, it seems probable that tests on other types would yield similar results.

TABLE 6.11. *Effect of prolonged exposure to low temperature (−320° F) on the impact properties at room temeprature of annealed 304 stainless steel [12].*

Time of exposure at −320° F	Charpy (keyhole) impact at—	
	Room temperature	−320° F
	ft-lb	*ft-lb*
None	80 to 90	----------
30 minutes	90 to 91	72 to 73
6 months	83 to 91	72 to 75
12 months	83 to 89	73 to 77

The results of earlier work of Colbeck, MacGillivray, and Manning [13] are given in table 6.12. In their work, type 302 stainless steel was subjected to the influence of −292° F for 192 hours and tested again at room temperature. Thus the results reported by Armstrong and Miller confirm those of the above authors.

Very recent work by Kulin and Cohen [14] suggests that cooling stainless steel 304 (annealed) to extremely low temperatures of 1° K (−272° C or −458° F) produced no phase transformation; admit-

TABLE 6.12. *Effect of exposure at —292° F on the Izod impact properties at room temperature and —292° F of annealed 302 stainless steel* [13]

Condition	Izod impact at—	
	Room temperature	−292° F
	ft-lb	*ft-lb*
Prior to exposure at −292° F	114	----------
After exposure at −292° F for 192 hr	117	100

tedly the holding time was not long, but one is encouraged to think that if a phase change could take place (in the absence of stress), it would do so at this low temperature, despite the shortness of holding time.

A stress, applied either dynamically or at a slow rate, produces a rapid phase change; this change occurs during deformation. The authors refer to this change as "martensitic transformation", the amount of which progressively increases as the temperature is lowered. The martensitic transformation is accompanied by the appearance of magnetism, as should be anticipated.

Although it is established with a considerable degree of certainty that at low temperatures prevailing at stratospheric or even interstellar space, austenitic chromium-nickel steels (not only of the most common 18-percent-chromium–8-percent-nickel analysis, but any analysis in which the percentage of chromium and nickel insure an austenitic matrix) in the absence of stress, undergo no transformation, it is not to be assumed that all austenitic steels behave in a similar manner. A classic example is the Hadfield austenitic manganese steels, which are known to undergo transformation and become extremely brittle at sufficiently low temperatures.

Interestingly enough, partial substitution of manganese for nickel in chromium-nickel austenitic stainless steels, or additions of other elements in substantial amounts, do not affect detrimentally the low-temperature toughness, provided such steels remain austenitic, and further that these steels do not have a high carbon content.

Fatigue

A most careful search of the literature will demonstrate that data on the endurance properties of austenitic stainless steels are indeed very meager. If any data are available in the files of various research laboratories, this may be an appropriate occasion to plead for the publication of research dealing with this subject.

It is generally held that the endurance properties of either annealed or cold-worked steels under discussion are greater at low temperatures than at room temperature. As the work of Russell [15], and Johnson and Oberg [16] has been extensively quoted in many publications dealing with the subject, there is no need to reproduce their data.

The more recent work of Spretnak, et al [2] substantiates the former findings and suggests that the fatigue strength at −320° F. of severely cold-worked steel (tensile strength 230,000 psi) at room temperature may be improved by between 32 and 53 percent, depending on the magnitude of the stress. Unfortunately for our discussion, the careful work of the authors does not furnish as yet additional experimental results vital for the over-all appraisal of the endurance properties of steels and the effect of temperatures between room temperature and −320° F, the level and nature of stress, etc. This suggestion seems to be in order, because other (private) sources of information do not agree with the above-quoted figures for the magnitude of the increase. It may be prudent, in view of very limited experimental data, to accept the more conservative values for increase in endurance limit with decrease in temperature for both annealed and cold-worked steels of this class.

Of considerable interest would be a comprehensive picture of the effect of notches on the fatigue strength at subzero temperatures, especially in comparison with the same at room temperature. The reason for this interest is based on intriguing possibilities that can be speculated upon on the basis of partial data, recently secured by The International Nickel Co., Inc., in cooperation with Battelle Memorial Institute [17]. Direct-tension fatigue tests revealed a striking feature of stress-cycles diagram, namely, long lifetime (large number of cycles) at high stresses, giving rise to a very flat S–N curve. The same observations were made by other investigators, including Zamrow and Fontana [8]. The explanation might well lie in the fact that this result is produced by work-hardening of the test section, especially if the conditions of testing consist of a steady stress plus superimposed alternating stress. Obviously, each work-hardening would be influenced by the magnitude of plastic strain, produced if the level of stress is sufficiently high. It is submitted that through the very nature of notches the stress concentration that occurs there may result in a sufficiently high rate of work hardening to offset the effect of the notch, *per se*.

Thus, lowered temperatures, which have been found to increase the rate of work-hardening by mechanical means [18] should exert a beneficial influence on the fatigue properties of notched steels. This is what has been found in the already quoted work of Spretnak, et al. [2].

Very little published information is available on the fatigue properties at low temperatures of the precipitation hardening type of stainless steels. Zambrow and Fontana [8] report that as the temperature is lowered to −320° F, the endurance limit of stainless "W" is practically doubled. For the details, the reader is referred to the original work.

Discussion

Dr. C. B. Post, Chief Metallurgist, Carpenter Steel Co., Reading, Pa.: Dr. Krivobok, have you done any testing on the substitution of manganese for nickel? Do those types of austenitic steels follow the same type you have been showing at low temperature, so far as the effect of hardening? Let's say 6 to 8 percent of manganese and 2 to 3 percent of nickel, somewhere along in there.

Dr. V. N. Krivobok: In preparation for the presented summary, I searched the literature for data on the influence of substantial amounts of manganese as a substitute for nickel. It is know that Hadfield austenitic steels do not maintain subzero toughness in all conditions of heat treatment, and hence the substitution of manganese for nickel might conceivably result in altered properties at low temperatures. While the effect of manganese substitution on properties at room temperature has been studied, I found no data on impact at subzero temperatures. My own private files contain limited information on an alloy of 18 percent of chromium, 0.09 percent of carbon, 6.50 percent of manganese and 4.50 percent of nickel. After a quench annealing treatment from 2,050° F this steel possessed a toughness in excess of 60 foot pounds at −78° F.

Dr. N. P. Allen, Superintendent of Metallurgy Division, National Physical Laboratory, Teddington, England: Well, since it is mentioned, among the slides I set aside yesterday afternoon—because I was in the same difficulty as you were—I did deal with the question of brittleness in austenitic manganese steel. There are some well-known austenitic steels which have been proved to be fully austenitic, that is to say, they contain no martensite, which are nevertheless brittle at low temperatures. There is one with about 45 percent of manganese and 2 percent of carbon mentioned in the classical paper by Sir Robert Hadfield. The people who make manganese steel in Sheffield have been very much concerned about this very brittleness at low temperatures of the 12 percent of manganese, 1 percent of carbon steel.

One of my slides showed how they have been dealing with the situation by progressively lowering the carbon, taking out manganese, and putting in nickel and chromium. By these changes they make steels with very satisfactory impact values at lower temperatures.

Dr. Post: The question is: suppose you take one of these manganese-nickel stainless steels in the form of a bar or a sheet down to a cold temperature and then roll or work it to get the hardness and toughness you are talking about. That is my question.

Dr. Krivobok: An additional question of Dr. Post, however, is in connection with rolling or working at subzero temperatures of austenitic stainless with manganese substituted for nickel. A few preliminary tests which were carried out on steel of 18 percent of chromium, 5 percent of nickel, 5 percent of manganese and 0.07 percent of carbon show that the rates of work hardening at 70°, −105° and −320° F are slightly higher than those previously found for type 302, but not as high as those for type 301 stainless steels. The actual data can be obtained upon request.

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7. Dimensional Effects in Fracture

By C. W. MacGregor [1] and N. Grossman [2]

Three phases of the general problem of the effects of dimensional changes of test specimens on fracture characteristics are reported. These are (1) the influence of size on the transition temperature from ductile to brittle fracture, (2) the effect of various ratios of combined stresses on the brittle transition temperature, and (3) the effect of combined stresses on fracture strength.

In the first problem, flat circular steel risks containing 0.95-percent of carbon, simply supported around the circumference and loaded by a concentrated force at the center, were tested at constant deflection rates and at various constant temperatures in the MIT slow-bend testing device described previously. Sizes were changed in the ratios, 6, 2, 1. When the sizes were altered, all dimensions, including the disks, supports, and loading members, were changed in the same proportions. Comparisons were made of the transition temperatures for brittle fracture at the same effective strain rates rather than at equal deflection rates in order to maintain mechanical similitude. It was found that the largest disk of 6-inch diameter had a transition temperature only 8 deg F higher than the disk of 1-inch diameter. Thus the size effect for this material is of a trivial nature in ratio ranges of 6 to 1, as far as the brittle transition temperature is concerned.

To study the influence of biaxiality of the stresses on the brittle transition temperature, rectangular steel plates containing 0.95 percent of carbon were simply supported along two sides, free along the other two sides, and loaded by a central concentrated force. The thickness of all these plates was equal to ⅛ inch. The ratios of supported lengths to unsupported lengths varied from 0.5 to 2.667, which produced a variation of the biaxiality of stresses from 0.483 to 0.855. As the circular disks previously described had a biaxiality ratio of 1.0, the present tests included a study of the effect of this combined stress ratio between 0.483 and 1.0. It was found that if the biaxiality ratio varied from 0.483 to 1.0, the brittle transition temperature at a given effective strain rate increased on the average 62 deg F.

The data obtained in these tests also permitted some conclusions to be drawn regarding the effect of combined stresses on fracture strength. It is shown that as the contraint effect increases, i. e., as the ratio of the biaxial stresses increases from 0.483 to 1.0, the brittle fracture strength also increases markedly. A definite increase in fracture strength is also shown to accompany a decrease in size.

Introduction

The influence of size and shape of test specimen on mechanical properties has occupied the attention of investigators for many years, especially since the early work of Barba [1] [3] and Kick [2]. For a homogeneous and isotropic elastic material, mechanical similitude in two geometrically similar bodies of the same material tested at the same temperature and subjected to similarly applied external loads in equilibrium exists when the states of stress (or of strain) are proportional to each other at corresponding points [3]. As far as plastic flow is concerned, Barba and Kick have suggested that to maintain in two

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[3] Figures in brackets indicate the literature references on p. 151.

geometrically similar bodies of the same homogeneous material similar states of plastic strain requires that they be deformed under equal stress at corresponding locations. This assumes the strain rate has no effect. It is in general necessary to add the condition that at any instant the strain rate is the same in both bodies, as well as the strains and stresses for complete mechanical similitude. Even the latter assumes the path of loading or prior strain history is the same for both bodies, the temperature conditions are the same, no phase changes or other metallurgical disturbances take place during loading, and that the grain size relative to specimen size has no effect. If brittle fracture takes place with no macroscopic plastic flow, statistical considerations may also influence the mechanical properties.

In many previous experiments purporting to show a size effect one or more of the above conditions have frequently not been met, such as in the experiments of Kuntze [4] where true mechanical similitude was not maintained [3]. Another condition all too prevalent in many experiments indicating a size effect is a lack of control of the strength and ductility properties of the specimens of different sizes at the position in the body where fracture is initiated. When specimens of various sizes are machined from a billet or plate, great care has to be exercised that there exists no difference in basic strength and ductility levels between specimens of different sizes at the origin of failure, or else the true size effect is clouded over by nonhomogeneities of this sort.

A very important technical problem today is the possible vulnerability of a normally ductile steel to brittle failure. It has been discussed previously [5] that there are many mechanical and metallurgical factors that tend to increase the temperature at which a metal may transform from ductile to brittle behavior. A few of these are increasing triaxiality of the stresses, strain rate, prestrain, prior fatigue cycles, strain-aging, grain size, carbon content, pearlite size, etc. When this temperature equals the service or testing temperature, brittle failure may ensue with little or no energy absorption. It is the conditioning of the material for such brittle behavior that is the most important pragmatic question rather than the strength condition or law followed when brittle behavior takes place. Another possible influence on the brittle transition temperature is the size effect. Davidenkov, Shevandin, and Wittman [6] have reported increases in the brittle transition temperature of as much as 38 deg C for a 0.25-percent-carbon steel when sizes of unnotched tension specimens were increased from 2 to 10 mm. They made careful impact tension tests, but the specimens of different sizes apparently had some variation in basic strength and ductility levels, which showed up more prominently at the lower temperatures. The reductions in area at fracture from room-temperature tests showed a slight decrease as the specimen size was reduced. The properties probably varied somewhat over the section of the bar from which the specimens were machined. The scatter band of the transition temperatures was also greater for each specimen size than the increase in transition temperature found. There appeared to be a limiting size, however, above which little size effect was found. One of the objects of the present study was to investigate under well-controlled conditions the possible size effect as regards the brittle transition temperature. If such were present to any appreciable degree, great care would have to be observed in applying the results of small-scale laboratory tests to full-scale structures or machine parts in service.

If geometric similarity of test specimens is not preserved, it has been well known that considerable changes in brittle transition temperatures may result. Early tests [7] have shown the marked influence of the greater constraint produced by widening a notched-beam impact specimen. As the exact state of stress in the bottom of the notch for such specimens is practically unknown, it has been difficult to express the degree of constraint produced in notched specimens or to disassociate the pure constraint effect of the state of combined stresses from the effect of the stress concentration. By utilizing plate specimens of different proportions without the presence of a notch, the authors some time ago [8] showed quantitatively for an SAE 1020 steel how increased triaxiality increased the transition temperature. Three cases were tested having ratios of lateral to longitudinal stresses of 1, 0.56, and 0. At least 100 deg F difference in transition temperature was found between the uniaxial and equal biaxial stress conditions at the same effective strain rate. Moreover, equivalent notched bars were found that would give the same transition temperature at the same effective strain rate as the simple plate specimens, thus establishing a structure-notched bar equivalence. Tests showed this correlation to be a purely formal one independent of the material, although different materials had, of course, different base transition temperatures. It is data of this type that would be directly useful in design, as discussed previously [8]. The investigation reported herein carries this phase of the problem further and considers the effect of combined stresses on the transition temperature of a 0.95-percent-carbon steel for additional ratios of lateral to longitudinal stresses between 1 and 0.

In addition, the tests will allow certain conclusions to be drawn regarding the effect of combined stresses on fracture strength.

Velocity-Temperature Relation

Experiments have shown that the testing temperature and the deflection velocity, or strain rate, can be closely correlated. When the entire temperature-velocity range is considered, use may be made of the velocity-modified temperature concept [9]. For experiments involving brittle fracture by little or no plastic flow and conducted well below room temperature, as in the present case, the relationship is

$$\log_e \dot{\delta} = C - \frac{Q}{RT},$$

where $\dot{\delta}$ is the deflection velocity, C a material constant, Q the heat of activation of the material, R the molal gas constant, and T the absolute temperature of testing [10]. With a given stable material and system of constraint, a straight line is obtained if the logarithm of the deflection velocity is plotted versus the reciprocal of the absolute temperature of testing. This property is utilized in the present investigation.

Material Tested and Specimens

The material used for this study was Ryerson VD (0.95% C) tool steel obtained in the form of a 6½-in.-diameter bar. The chemical analysis is listed in table 7.1. Photomicrographs are included in figure 7.1. The metal was tested in the annealed condition. Figure 7.2 shows the dimensions of the plate specimens used. The size-effect tests were made on circular disks simply supported along the circumference and bent as a plate by a concentrated load at the center. The diameters were 6 in., 2 in., and 1 in., with all other dimensions, including thickness, radii of knife edges, and supports, changed in the

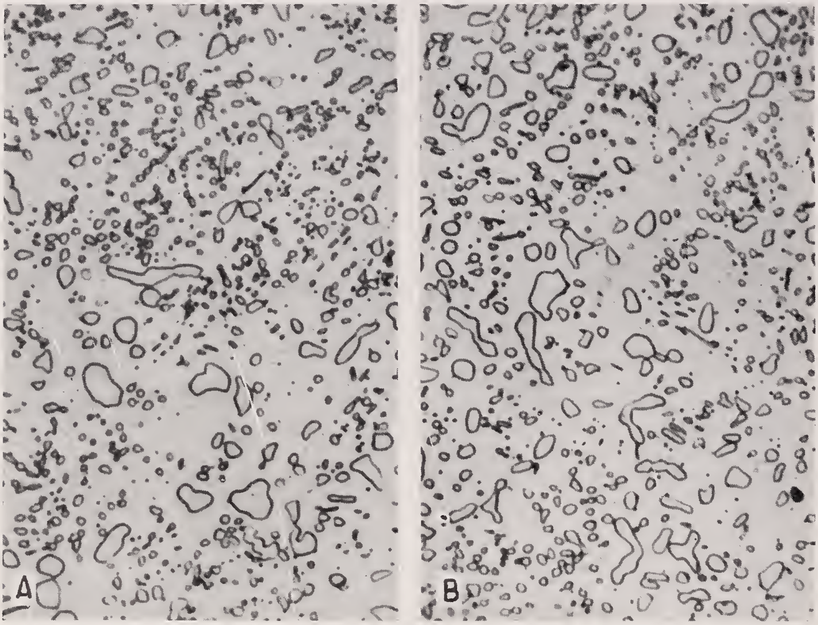

FIGURE 7.1. *Microstructure of the annealed steel used,* ×*1,000.*

A, Longitudinal section; B, transverse section.

same ratio. These are shown in figure 7.3. The specimens used to study the effect of combined stresses on the brittle transition temperature are also shown in figure 7.2 and the supports in figure 7.3. The thickness of the latter was kept constant, and the ratios of supported to unsupported lengths varied from 0.5 to 2.667. These plates were thus simply supported along two edges, free along the other two and loaded by a concentrated force at the center.

TABLE 7.1. *Chemical composition of steels used*

Element	Percent	Element	Percent
Carbon	1.03	Chromium	0.15
Manganese	.14	Vanadium	.19
Phosphorus	.008	Molybdenum	.02
Sulfur	.024	Copper	.08
Silicon	.20	Aluminum	.01
Nickel	None		

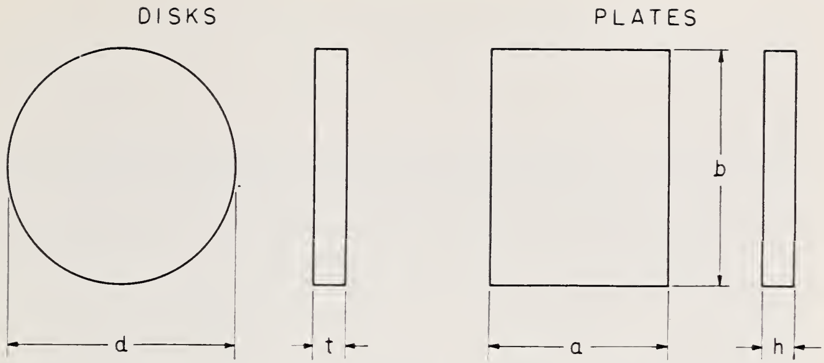

FIGURE 7.2. *Forms and dimensions of the specimens used.*

Disks		Plates		
d	*t*	*a*	*b*	*h*
in.	*in.*	*in.*	*in.*	*in.*
6	3/16	4	4	1/8
2	1/16	1½	4	1/8
1	1/32	1	½	1/8

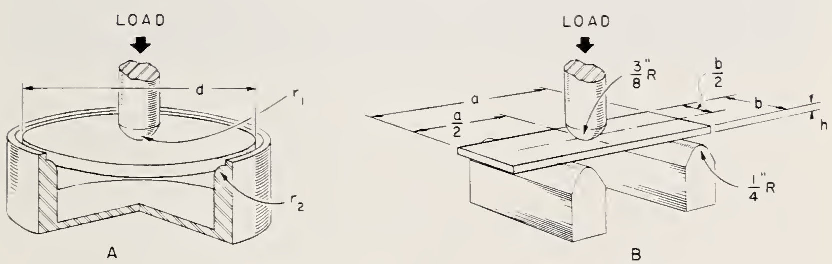

FIGURE 7.3. *Specimen supports and loading devices.*

A. Disks			B. Plates		
d	r_1	r_2	*a*	*b*	*h*
in.	*in.*	*in.*	*in.*	*in.*	*in.*
6	3/8	9/16	4	4	1/8
2	1/8	3/16	1½	4	1/8
1	1/16	3/32	1	½	1/8

In order to insure that the portions of the test specimens where fracture originates had the same strength and ductility level for all tests, the specimens were machined in such a way that the center of the specimen under the load coincided with the original axis of the 6½-in.-diameter bar supplied.

Apparatus and Testing Technique

The testing equipment is designed to load a specimen in simple bending at a given uniform speed and temperature and to supply a load-deflection record of the test. AB–7, SR–4 electric strain gages connected in a bridge circuit are used to measure both the load and deflection. The integral parts of the apparatus include the loading machine with strain gages, the thermocouple and millivoltmeter, stop watch, electronic recorder, and a still camera.

The loading device is a lever system which transmits the load to the specimen resting on a suitable support. The support, specimen, and electric strain gages are housed in a subzero test cabinet. The cabinet can be cooled to $-100°F$ by forced circulation of dry ice, and the specimen can be further cooled by pouring liquid nitrogen around it.

The time of testing is measured either by the axis of the electronic recorder, which places a "dot" on the load-deflection record at regular time intervals, or by a stop watch at the slower speeds.

An a-c bridge system is used whereby a 5,500-c/s voltage is fed into the load bridge and the deflection bridge, with the resulting unbalance from loading or deflecting detected, amplified and sent to a 3FPI tube. A record is made with a still camera. The load bridge and the deflection bridge contain an AB–7, SR–4 strain gage in each of the four arms. The detailed description of the apparatus has been given previously [8].

The disks and plates were freely supported and centrally loaded with a concentrated load. The specimen was cooled to a temperature slightly above the expected transition temperature and then loaded at a given uniform rate. If a slight yielding took place the test was stopped, the temperature was lowered and the loading process repeated until failure took place with a slight amount of yielding. A new specimen was then tested at a lower temperature to account for the cold working of the trial sample. This procedure was repeated until brittle fracture was experienced with a minimum amount of yielding at the first trial. This same technique was then repeated at some other deflection rate to obtain the strain-velocity–brittle temperature relation for the disk under investigation. Thus the results listed in tables 7.2 and 7.3 do not represent one single test but rather the limit of a number of successive approximations, with the finally satisfactory test value repeated two to three times. The accuracy of the temperature determination is ± 3 deg F that of the deflection rate ± 5 percent. The amount of macroscopic flow may be judged from a previous study [11] when for a given notched bar a permanent angular bend of $1/2°$ corresponded to a rise of about 100 deg F above the respective transition temperature, as determined by the MIT slow bend test.

Figure 7.4 includes a view of the apparatus.

Analytical Considerations

As discussed previously [8], due to the biaxial nature of the problem it is necessary to utilize composite expressions for stress and strain that depict conditions during the short twilight zone between the cessation of plastic flow and the initiation of brittle fracture. While this is imperfectly understood, it is considered reasonable to utilize the composite values of stress and strain that are valid during plastic flow, namely,

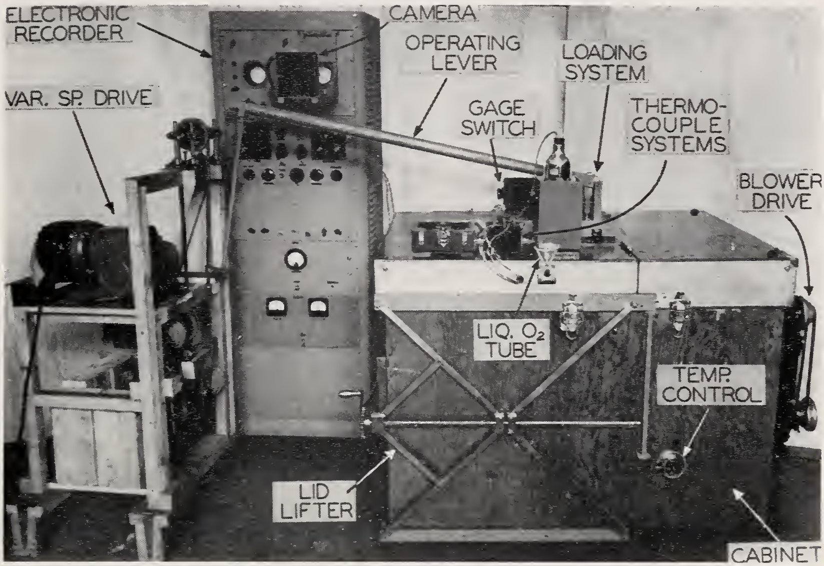

FIGURE 7.4 *Load-deflection apparatus*

$$\left.\begin{aligned} \sigma &= \frac{1}{\sqrt{2}} \sqrt{(\sigma_1-\sigma_2)^2+(\sigma_2-\sigma_3)^2+(\sigma_3-\sigma_1)^2} \\ \epsilon &= \frac{\sqrt{2}}{3} \sqrt{(\epsilon_1-\epsilon_2)^2+(\epsilon_2-\epsilon_3)^2+(\epsilon_3-\epsilon_1)^2} \end{aligned}\right\}, \qquad (1)$$

where σ, σ_1, σ_2, σ_3 are the effective stress, and the three principal stresses, respectively, and ϵ, ϵ_1, ϵ_2, ϵ_3 are the effective strain and the three principal strains, respectively.

TABLE 7.2. *Effect of size of test specimen on some low-temperature properties of the steel*

Diameter of disk	Brittle transition temperature	$T°$ (absolute)	$1/T$	$\dot{\delta}$	$\dot{\epsilon}$
in.	° *F*	° *R*		*in./sec.*	*(in./in.)/sec.*
6	−283	177	0.00565	0.0020	0.00019
6	−238	222	.00450	.037	.0035
6	−210	250	.0040	.214	.0203
2	−280	180	.00556	.0014	.0004
2	−230	230	.00435	.024	.00683
2	−180	280	.00357	.206	.0587
1	−270	190	.00527	.0014	.0008
1	−230	230	.00435	.015	.00855
1	−160	300	.00333	.57	.325

Table 7.3. *Effect of combined stresses on some low-temperature properties of the steel*

Plate size	Brittle transition temperature	$T°$ (absolute)	$1/T$	$\dot{\delta}$	$\dot{\epsilon}$
in.	° *F.*	° *R*		*in./sec*	*(in./in.)/sec*
4 by 4	−310	150	0.00667	0.0022	0.000140
4 by 4	−280	180	.00556	.026	.00165
1½ by 4	−279	181	.00552	.0012	.000565
1½ by 4	−239	221	.00452	.011	.00518
1 by ½	−320	140	.00715	.00157	.00118
1 by ½	−265	195	.00513	.0060	.00451
1 by ½	−245	215	.00465	.108	.0811

It is then considered that for a stable metal under these conditions there exists some limited functional relation connecting σ, ϵ, the absolute transition temperature T, and the effective strain rate $\dot{\epsilon}$, which is the time derivative of ϵ of the form

$$\sigma = f(\epsilon, \dot{\epsilon}, T). \tag{2}$$

For the size-effect tests with the circular disks, the value of $\dot{\epsilon}$ was determined as follows. The deflection at the center was computed by classical formulas [12] in terms of the central load P. Incidentally, this agreed very well with measured deflections. The maximum stresses were next calculated in terms of P on the bottom side of the disk under the load. The principal strains were then determined from Hooke's generalized law and substituted in eq (1) to obtain the effective strain in terms of P. Thus the deflection-effective strain relation was determined, and consequently the relation between effective strain rate and deflection rate. Hence the deflection rates would be converted to effective strain rates. These showed that to maintain equal effective strain rates in the disks of increasing size it was necessary for comparison to increase the deflection rates proportionately. This is discussed further under Test Results, where comparisons are made of the transition temperatures for disks of different sizes.

For the plate tests where the effects of combined stresses on the transition temperature were determined, the procedure was similar. The only difference was that corresponding rectangular plate formulas were utilized [13].

To represent the effects of constraint, the same index is used as originally proposed by Jackson [14] and utilized formerly [8], namely,

$$C = \frac{|\sigma_1 + \sigma_2 + \sigma_3|}{3|\sigma_{\max}|}. \tag{3}$$

As discussed earlier [8], this factor is 0 for pure sheer, 0.333 for pure tension or compression, 0.666 for equal biaxial tensions, and 1.00 for triaxial tension or compression.

Test Results

The effect of size on the transition temperature for the disks (equal biaxial tensions) is shown in figure 7.5, where the logarithm of the deflection rates, δ, are plotted versus the reciprocal of the absolute testing temperature $1/T$. In order to compare the results at the same effective strain rates, $\dot{\epsilon}$, it is necessary to choose deflection rates that increase in direct proportion to the size. Hence auxiliary construction lines are indicated in figure 7.5, which accomplish this. It is thus seen that an almost trivial effect is indicated, namely, that the 6-in. disk had a transition temperature of about 8 deg F higher than that for the 1-in. disk at the same effective strain rate. Thus while indicating a small increase in transition temperature for this material in the size ranges studied as the size is increased, it is still very small.

Figure 7.6 shows the effect of combined stresses on the transition temperature. An increase in the ratio of the supported to the unsupported length increased the ratio α of the transverse stress to the axial bending stress from 0.483 to 1.00, while the constraint index, C, increased from 0.494 to 0.666. It may be seen from this figure that as the constraint index increased from 0.494 to 0.666, the brittle transition temperature increase on the average 62 deg F for this material at a given effective strain rate. The test results for $C=0.520$ were taken from the previous study [8] for SAE 1020 steel, making due allowance for the different base transition temperatures of the two materials.

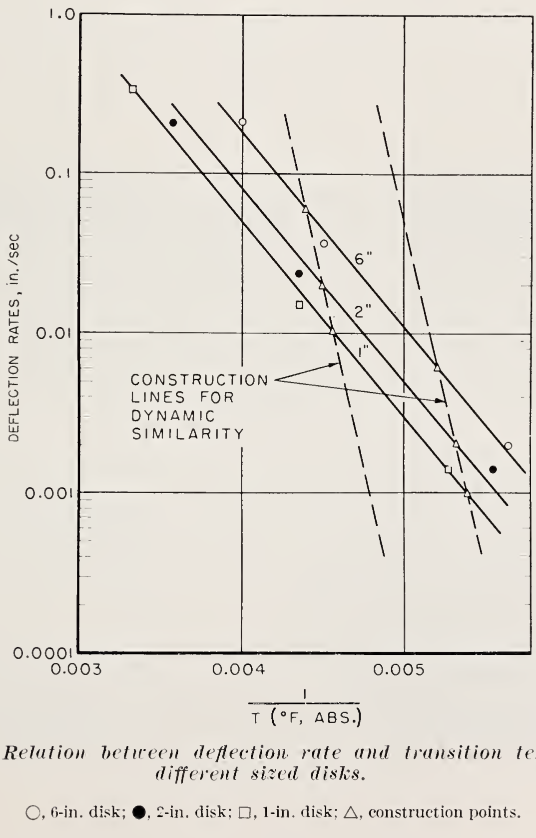

FIGURE 7.5. *Relation between deflection rate and transition temperature for different sized disks.*

○, 6-in. disk; ●, 2-in. disk; □, 1-in. disk; △, construction points.

Figure 7.7 includes views of the fractures received.

The data obtained also permit a study of the effect of combined stresses on the fracture strength. Table 7.4 records calculations of fracture stresses, strain rates, and strains for the various cases considered. It will be noted that in certain cases checked the fracture deflection calculated agreed quite well with experimental values. Particularly illuminating is the relation between the effective stress and effective strain, ε, (calculated from eq 1) at fracture. This relation is plotted in figure 7.8 for all cases listed in table 7.4. It will be noted that all points fall on a straight line, the slope of which is

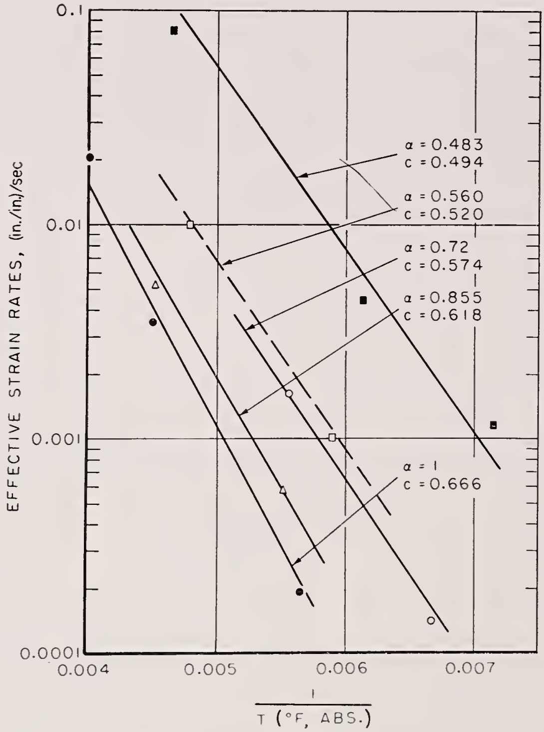

FIGURE 7.6. *Effect of combined stresses on transition temperatures.*

Plates

	a	*b*
	in.	*in.*
○	4	4
△	1½	4
[a] □	3	1⅞
■	1	½
● 6 in. disk		

[a] Estimated from previous data.

34.5×10^6 psi. As shown previously [15], the theoretical relation between the effective stress, σ, and the effective strain, ϵ, for an ideally elastic body is

$$\frac{\sigma}{\epsilon} = \frac{3E}{2(1+\gamma)}, \tag{4}$$

where E is Young's modulus, and γ is Poisson's ratio. Using $E = 30 \times 10^6$ psi and $\gamma = 0.3$ for steel, eq (4) shows that

$$\frac{\sigma}{\epsilon} = 34.6 \times 10^6 \text{ psi}, \tag{5}$$

which agrees well with the slope of figure 7.8. The values for figure 7.8 are for various cases of disks and plates with different ratios of supported to unsupported lengths and were calculated from elastic theory, using the experimental fracture loads. Thus figure 7.8 is the elastic σ-ϵ relation for this material. It is worth while to note the position on the curve corresponding to various ratios of combined stresses or to values of the constraint index, C, of eq (3). It was found possible to utilize the velocity-modified temperature concept [9] and to thus account, for a given constraint factor, for the various strength levels produced by different values of strain rate and temperature. Some overlapping is present, between cases representing various values of C. This is interpreted as reflecting a size effect as indicated by the different positions on the curve for the 2-in. and 6-in. disks. In spite of this, however, it is unmistakably clear that as the

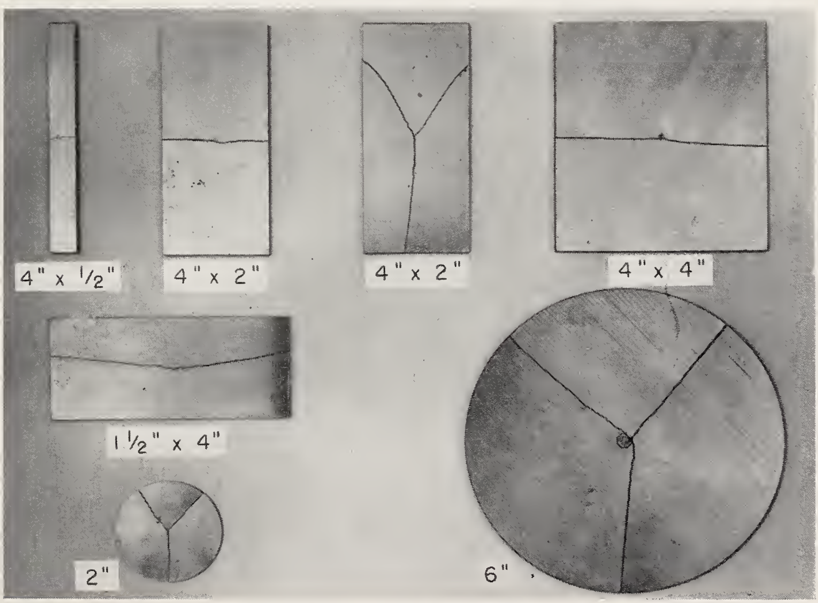

FIGURE 7.7. *Typical fractures.*

biaxial stress ratio increases, the fracture stress and strain increase. This is further made clear from table 7.5, in which the average values of effective fracture strengths are listed for each value of C and α, whereby the average is meant the mean point of the range of values of effective fracture strength for each C and α.

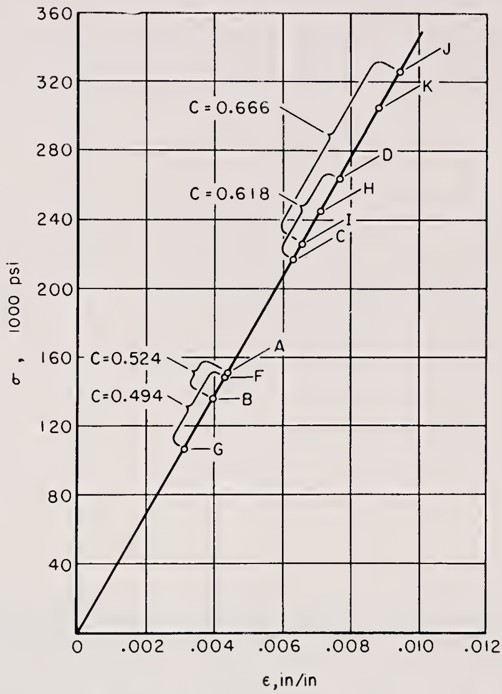

FIGURE 7.8. *Effect of combined stresses on effective fracture stresses and strains.*

That the fracture strength increases with triaxiality, as shown here, agrees well with the concepts of McAdam, et al. [16], and Sachs and Lubahn [17] and is contrary to the prediction of Fisher and Hollomon [18]. The latter disagrees with McAdam, Sachs, and Lubahn, contending that the increase in fracture strength reported by them as due to increased triaxiality was in reality a size effect and criticized the use of notched specimens to show this. In the present case, however, the use of notches was avoided, thus separating out the complicated effects of constraint and stress concentration, and still the test results show an unmistakable influence of triaxiality in increasing the fracture strength. More in agreement with Fisher and Hollomon is the size effect on fracture strength shown particularly by the relation between the strength of the 2-in. and 6-in. disks. Size effect alone cannot, however, explain the increases shown for all the cases reported herein.

TABLE 7.4. *Effect of combined stresses and size of test specimen on the fracture strength*

Test piece: Designation	Test piece: Size	Brittle transition temperature	$\dot{\delta}$ (experimental)	$\dot{\epsilon}$ (experimental)	Fracture load, P	Calculated fracture stress	δ (experimental)	δ (calculated)	ϵ (calculated)	C	α	Effective stress
	in.	° *F.*	*in./sec.*	*(in./in.)/sec.*	*lb.*	*psi*	*in.*	*in.*	*in./in.*			*lb./in.²*
A	4 by 4	−310	0.0022	0.000140	1,000	σx = 169,000; σy = 121,700	0.0795	0.0686	0.00436	0.574	0.72	151,000
B	4 by 4	−280	.026	.00165	900	σx = 152,000; σy = 109,000		.0619	.00392	.571	72	135,600
C	1½ by 4	−279	.0012	.000565	1,700	σx = 232,000; σy = 199,000	.020	.0118	.00630	.618	.855	217,000
D	1½ by 4	−239	.011	.00518	2,660	σx = 282,000; σy = 241,000		.0143	.00765	.618	.855	264,000
E	1 by ½	−320	.00157	.00118								
F	1 by ½	−265	.0060	.00451	708	σx = 171,500; σy = 82,800		.00569	.00430	.494	.483	148,500
G	1 by ½	−245	.108	.0811	510	σx = 123,500; σy = 59,800	.097	.00412	.00310	.494	.483	107,000
H	[a] 6	−283	.0020	.00019	2,980	σx = 245,000; σy = 245,000	.072	.0745	.00710	.666	1.00	245,000
I	[a] 6	−238	.037	.0035	2,750	σx = 226,000; σy = 226,000	.0675	.0689	.00655	.666	1.00	226,000
J	[a] 2	−280	.0014	.0004	440	σx = 326,000; σy = 326,000	.026	.0330	.00940	.666	1.00	326,000
K	[a] 2	−230	.024	.00683	410	σx = 304,000; σy = 304,000	.026	.0309	.00875	.666	1.00	304,000

[a] Diameter of disk.

TABLE 7.5. *Effect of constraint on the average effective fracture strength*

C	α	Average σ at fracture
		psi
0.494	0.483	127,700
.574	.72	143,300
.618	.855	240,000
.666	1.000	275,200

Conclusion and Summary

Three phases of the general problem of the effects of dimensional changes of test specimens on fracture characteristics are reported.

(1) The influence of size on the transition temperature from ductile to brittle fracture, (2) the effect of various ratios of combined stresses on the brittle transition temperature, and (3) the effect of combined stresses on fracture strength.

In the first problem, flat circular steel disks containing 0.95 percent of carbon, simply supported around the circumference and loaded by a concentrated force at the center, were tested at constant deflection rates and at various constant temperatures in the MIT slow-bend testing device described previously. Sizes were changed in the ratios 6, 2, 1. When the sizes were altered, all dimensions, including the disks, supports, and loading members, were changed in the same proportions. Comparisons were made of the transition temperatures for brittle fracture at the same effective strain rates rather than at equal deflection rates in order to maintain mechanical similitude. It was found that the largest disk of 6-in. diameter had a transition temperature only 8 deg. F higher than the disk of 1-in. diameter. Thus the size effect for this material is of a trivial nature in ratio ranges of 6 to 1, as far as the brittle transition temperature is concerned.

To study the influence of biaxiality of the stresses on the brittle transition temperature, rectangular steel plates containing 0.95 percent of carbon were simply supported along two sides, free along the other two sides, and loaded by a central concentrated force. The thickness of all these plates was equal to 1/8 in. The ratios of supported lengths to unsupported lengths varied from 0.5 to 2.667, which produced a variation of the biaxiality of stresses from 0.483 to 0.855. As the circular disks described previously had a biaxiality ratio of 1.0, the present tests studied the effect of this combined stress ratio between 0.483 and 1.0. It was found if the biaxiality ratio varied from 0.483 to 1.0 that the brittle transition temperature at a given effective strain rate increased on the average 62 deg. F.

The data obtained in these tables also permitted some conclusions to be drawn regarding the effect of combined stresses on fracture strength. It is shown that as the constraint effect increases, i. e., as the ratio of the biaxial stresses increases from 0.483 to 1.0, the brittle fracture strength also increases markedly. A definite increase in fracture strength is also shown to accompany a decrease in size.

Discussion

DR. M. J. LETICH, American Bureau of Shipping, New York, N. Y.: If in the interpretation of the test results, some thought was given to abide by the laws of similitude, could predictions be attempted on the behavior of larger size specimens than the ones tested?

Dr. C. W. MacGregor: If the laws of mechanical similitude are strictly observed and the material is in the same metallurgical state in both small and large sizes, it is the author's opinion that as far as transition temperature is concerned the behavior of even larger size specimens than the ones tested would show similar results, that is, no appreciable effect of size on transition temperature. As mentioned in the paper, extreme care has to be observed to maintain true mechanical similitude and often where size effects have been claimed in the literature true mechanical similitude did not exist.

Dr. J. W. Fredrickson, Chief, Division of Metallurgy, Pennsylvania State College, State College, Pa.: I want to make one comment, Professor MacGregor. You indicated that the size effect was not an important one in transition temperature. I wish to state that, based on our work at Penn State, I agree with you. We have made impact tests using subsize Charpy specimens to try to determine if there were a size effect. We could find none. I have a question.

I would like to know if you have made a comparison between the transition temperatures obtained by the slow bend test with those from the impact test, and what might be the result.

Dr. C. W. MacGregor: The relationship between the transition temperatures determined by our special slow-bend test and those obtained from the Charpy test is not as easy to find as one might at first assume. The Charpy test is a valuable one and we also use it in many cases. It does not, however, yield any quantitative information of direct use in design. Such considerations as variable and uncontrolled velocity of deformation during the Charpy test (the final velocity being different from the striking velocity), unavoidable temperature changes from the bath, adiabatic rather than isothermal deformation, etc., render it difficult to correlate with the special slow-bend test used in which all of these are well controlled. In order to attempt some correlation, however, we conducted experiments some time ago using the regular Charpy specimen for both tests [19]. A comparison of the transition temperatures was made which indicated that due to the inherently basic differences in both forms of test, the resulting transition temperatures (predicted by extrapolation from the slow-bend test results) were in all cases higher than those determined from the Charpy test.

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8. Mechanical Properties of High-Purity Iron-Carbon Alloys at Low Temperatures

By R. L. Smith[1], G. A. Moore[2], and R. M. Brick[3]

Natural stress-strain data have been obtained over a range of temperatures from +23° to −185° C on four high-purity iron-carbon alloys, 0.05 to 0.49 percent carbon, having microstructures with ASTM No. 4–5 ferrite grain size and pearlite of spacing equivalent to normalized steels. Analyses of these data lead to the following conclusions: (1) Yield points and flow stresses increase with increase of carbon content and with decrease in temperature. (2) Fracture stresses also increase with increase in carbon content and with decrease in temperature, until at around −150° C, loss of ductility becomes pronounced and the fracture stress drops to a considerably lower "unworked" value at −185° C. (3) The transition temperature, based on one half the maximum energy to fracture of tensile specimens, is at about −153° to −170° C for all carbon contents investigated. The transition temperature, based on a selected value of energy to fracture, increases with carbon content but so slightly as not to be significant. (4) Increasing carbon content results in a marked decrease in total strain to fracture and, relatedly, of total energy absorbed to fracture at all test temperatures used here. (5) Comparison of data for these iron-carbon alloys with those for low-aluminum high-purity ferrites indicates that carbon reduces the ductility of iron at all temperatures in the range investigated.

Introduction

A wealth of notched-bar test data on commercial steels have been accumulated in the last 10 years in the course of investigations on the general problem of low-temperature brittle failures of pearlitic steels. These data have led to certain useful conclusions, e. g., that aluminum-killed steels are superior (i. e., have lower transition temperatures), that normalizing frequently has a comparable beneficial effect, that fine grain size is desirable, that less than 11.4 ft-lb energy absorbed in V-notch Charpy tests at the service temperatures is conducive to brittle failure, that nickel improves low-temperature ductility, etc.

Useful as these conclusions are, the energy absorbed in notched-bar tests is not a numerical value susceptible of analysis nor it is useful in design. As practically all these data have been obtained on commercial steels, too many variables are present to isolate the cause of the beneficial or detrimental effects that are observed. For example, is the beneficial effect of aluminum deoxidation attributable to a solid-solution effect of excess aluminum, to the reduction or near absence of FeO, to fixing nitrogen as aluminum nitride, to a grain refining effect of residual Al_2O_3, or to an effect of dissolved aluminum on the structure of Fe_3C in pearlite?

The investigation reported here is a part of a general program designed to answer such questions.[4] The low-temperature properties

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[4] The present work has been supported by the Ship Structure Committee as Project SR-109 (contract NObs-50062, Index No. NS011-078).

measured are uniaxial tensile properties employing customary natural stress-strain curves obtained with somewhat novel equipment. The alloys tested were made in 6- to 8-pound ingots, which, apart from the added element, represent 99.9 percent of Fe. Analyses for impurities in quantities of 0.00X percent are troublesome, to say the least. To support the use of the term "high-purity" with reference to the alloys being studied, it is preferable to state that some of these low-alloy ferrites (not iron-carbon alloys, however) have shown yield strengths at room temperature of 11,000 psi, which is as low as any values for high-purity iron of comparable grain size reported in the literature. At the moment, this seems to be a more useful indicator of effective purity than chemical analyses where numerical percentages are in the vicinity of the limit of sensitivity of the analytical methods.

Preparation of Iron-Carbon Alloys

The general approach was to melt electrolytic iron in air, thereby oxidizing some of the impurities and removing them as much as possible by a slagging process. Carbon was originally added by dropping into the melt pieces of a master alloy of high carbon content. This method proved unsatisfactory because of the violent explosive boil that resulted. It was found that the air-melted alloys could be produced by dropping sugar charcoal directly on the surface of the oxidized melt. The violence of this reaction could be controlled by adding the carbon slowly.

The air-melting procedure used was as follows:

1. Electrolytic iron was melted in air in a magnesia crucible.
2. The heavily oxidized melt was slagged with pure dried calcium oxide, the slag was removed and the operation repeated several times.
3. Carbon was added to the surface of the melt in the form of sugar charcoal. The amount of carbon for a desired alloy had to be estimated by the boiling action in the crucible.
4. The melt was cast into a dry graphite mold.
5. The resulting ingot was sampled, and if of satisfactory carbon content and cleanliness, it was freed from surface scale in preparation for vacuum melting.

The air-melted alloys were remelted in a 6-in.-diameter high-frequency vacuum melting furnace. This included a standard 6-in. quartz tube, a brass head, and 1-in. glass tubing connected through a flexible bellows to a liquid-air trap to catch undesired products distilled from the furnace. The high-vacuum portion could be isolated from the system by a valve to estimate the equilibrium pressure over a melt, or to preserve the vacuum during the final cooling of the ingots. For continuously indicating the working vacuum, a thermocouple gage was permanently sealed into the high-vacuum manifold. The indications of this gage could be read either on a meter on the panel board or switched to a recording potentiometer to give a plot of the change of pressure with time. The absolute pressure was determined at intervals by means of the McLeod gage, and these readings were used to maintain a corrected calibration of the thermocouple gage.

For the alloys studied in this report (table 8.1), the units of the pumping system were a two-stage jet-diffusion mercury pump backed by a high-vacuum mechanical oil pump. In order to flush the system before melting the alloys, or to melt them in hydrogen or other controlled atmospheres, stopcocks were provided for connection to gas tanks and a purifying train.

TABLE 8.1. *Chemical composition of the high-purity iron-carbon alloys used*

Element	Alloy—			
	49v	51v	43v	53v
	%	%	%	%
Carbon	0.05	0.12	0.25	0.49
Manganese	.010	.016	.007	.011
Phosphorous	.006	.005	.006	.005
Sulfur	.004	.003	.001	.004
Silicon	.001	.005	.001	.005
Nickel	----------	Nil	----------	----------
Chromium	----------	Nil	----------	----------
Beryllium	.005	----------	----------	----------
Oxygen (vacuum fusion)	.0005	.0004	.0008	.0009
Nitrogen (vacuum fusion)	.0016	.0013	.0008	.0010

[1] Analyses were made by W. B. Coleman Co., except that determinations for S, O, and N, were made at the National Bureau of Standards. Cu, Mo, and Sn were reported at "limit of sensitivity."

The vacuum-melting procedure posed a problem for some time because of the violent boiling experienced when the previously air-melted ingot became molten under vacuum. The FeO: C reaction was finally controlled by conducting the actual melting of the solid ingot under nearly an atmosphere of evolved gases and then evacuating very slowly. With stabilized zirconia crucibles, a suitable vacuum was not obtained because of a ZrO_2: C reaction and solid ingots were not produced.

Solid ingots were produced, using both beryllia and magnesia crucibles but a suitable vacuum was not obtainable even with degassing periods of 25 hours. Experiments finally led to the use of beryllia crucibles backed with beryllia powder contained by a graphite sleeve. It was found that if this type of crucible assembly was prefired in vacuum at about 1,900° C and then degassed at 1,550° C, satisfactory melts could be subsequently realized.

The final vacuum-melting procedure used was as follows:

1. A 2½-in. beryllia crucible was packed with beryllia powder in a graphite sleeve. This assembly was in turn packed with beryllia powder in a quartz sleeve.

2. The crucible assembly was prefired at 1,900° C in vacuum for about ½ hour and then degassed at 1,550° C to less than 10 microns.

3. The air-melted ingot with a small amount of dry calcium oxide was placed in the above assembly, melted and degassed, then allowed to solidify directionally in order to minimize pipe formation. The alloys reported here were held molten from 2½ to 7 hours and were solidified under pressures ranging from 15 to 200 microns, figure 8.1.[5]

Vacuum-melted alloys were prepared for testing as follows:

1. The 2½-in. diameter ingots were hot-forged to 1-in. rounds, hot rolled at 1,100° C to ⅝-in. square bars, cooled and surface cleaned.

2. The bars were vacuum annealed at 1,000° C for at least 100 hours to eliminate gas absorbed during hot-working.

[5] The vacuum system has since been modified so as to maintain pumping pressures of less than 1 micron over the melt. Two-in. stainless-steel tubing leads from the melting chamber past a liquid-air trap to an oil diffusion pump, thence to a high-vacuum fore pump.

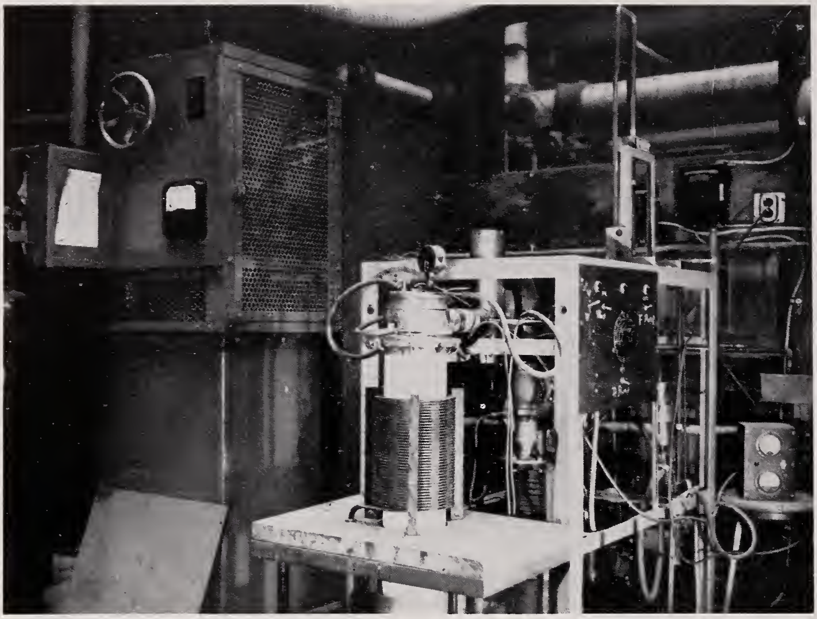

FIGURE 8.1 *High-frequency induction vacuum furnace.*

3. The annealed stock was then cold-rolled 60 percent to $\frac{7}{16}$-in. rounds and heat treated to an equiaxed No. 4 to 5 ASTM grain size with a medium-fine pearlite spacing. A strong Widmanstätten structure resulted from simple normalizing treatments. A normal structure, as shown in figure 8.2 was obtained by furnace-cooling in helium from austenite temperatures into the two phase $\alpha+\gamma$ region, followed by air-cooling. The relative amount of pearlite resolved at 500× was approximately the same for all heats.

4. The heat-treated bars were then turned to 4-in. long tensile specimens having a 1-in. gage length and a 0.252-in. diameter with a standard central taper (0.001 in.). The bars were given a hand polish with No. 600 polishing paper.

5. For most alloys, tensile tests were started 10 days after heat treatment and final tests completed within 5 weeks. Two bars of the 0.12-percent-carbon alloy were tested 3 months after heat treatment and gave results similar to those obtained earlier.

Mechanical Testing at Low Temperatures

The mechanical testing of the tensile specimen was done in a hydraulic testing machine of 60,000 pounds capacity, provided with special equipment for low-temperature work. This equipment was built previously for a similar research project, and is described elsewhere.[6] [7] However, because of some recent modification, a brief description is given here.

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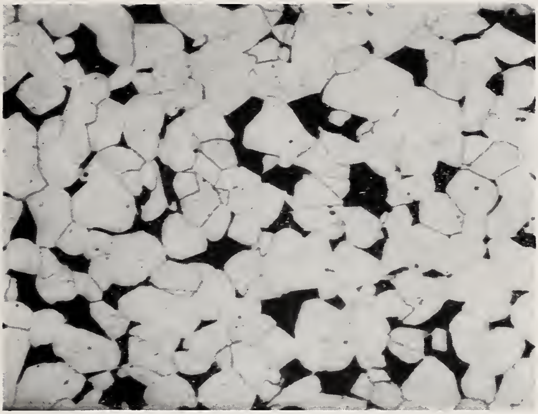

FIGURE 8.2 *Structure of iron-carbon alloy containing 0.12 percent of carbon.* Specimen was heated at 900° C for 15 minutes, cooled in furnace to 780° C, held 10 minutes and then cooled in air—etched with nital, ×100.

The low-temperature testing is conducted by submerging the specimen in a suitable refrigerant that is kept at a specific constant low temperature. The cooling fluid is contained in a Dewar vessel, supported around the test bar without interfering with the movements of the specimen grips and their alinement. Such a setup is shown in figures 8.3 and 8.4.

The metal Dewar, or test vessel, is supported by the movable crosshead of the machine, and therefore can, prior to testing, be moved vertically along with this cross bar into any desired position. The stem of the lower specimen grip enters the test vessel through a wide opening in the bottom with the ring-shaped space between stem and bottom being closed by light flexible bellows (fig. 8.3). Thus alinement of the grips during the test can take place freely without being hindered by the rather weighty test vessel. The test bar is screwed into the bottom grip from above and then the upper grip screwed downward over the threaded top part of the specimen. To provide for practically ideal self-alinement during the test, case-hardened chains are used to transfer the tensile force to the specimen, that is, from the loading ram to the upper specimen grip, and from the lower grip to the movable crosshead.

The cooling media used in this investigation were liquid air for −185° C and Freon No. 12 for the range −150° to −40° C. To cool the Freon, a copper coil carrying liquid air is fitted in the test vessel, the temperature being regulated by an automatic control circuit. To check the temperature independently at any time, an extra copper-constantan thermocouple is placed in the vessel close to the gage of the specimen, and wired to a potentiometer. An air-bubbler stirring arrangement is used to keep the temperature as uniform as possible.

Natural stress-strain data were obtained at low temperatures, using a microformer type of diameter gage immersed with the test specimens in the heat-transfer fluid, see figure 8.5. The diameter gage was calibrated against accurately machined test sections of a control bar.[8] By

[8] All dimensional data are reported in terms of inches measured at room temperature.

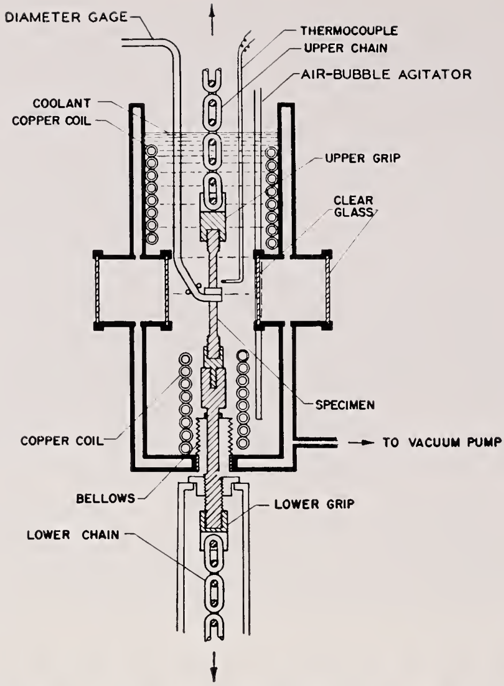

FIGURE 8.3. *Schematic diagram of tensile-test apparatus.*

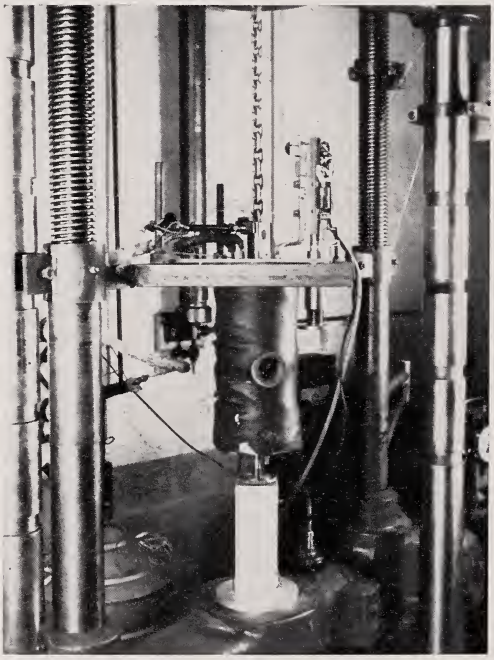

FIGURE 8.4. *Low-temperature tensile-test apparatus.*

traversing the mircrometer knife edges of the gage along the length of the tensile specimen during the test, the instantaneous diameter could be recorded directly with the corresponding load on an autographic recorder. A strain pacer was connected in series with the diameter gage and the autographic recorder. The investigator could raise or lower the jaws of the diameter gage. While the specimen is decreasing in diameter, the diameter gage causes the strain pacer needle to rotate clockwise. If the spacing of the jaws of the diameter gage increases, the pacer needle reverses its direction of rotation. Thus, by traversing the specimen, it is readily possible to find the region of necking and then to remain in the position of minimum diameter.

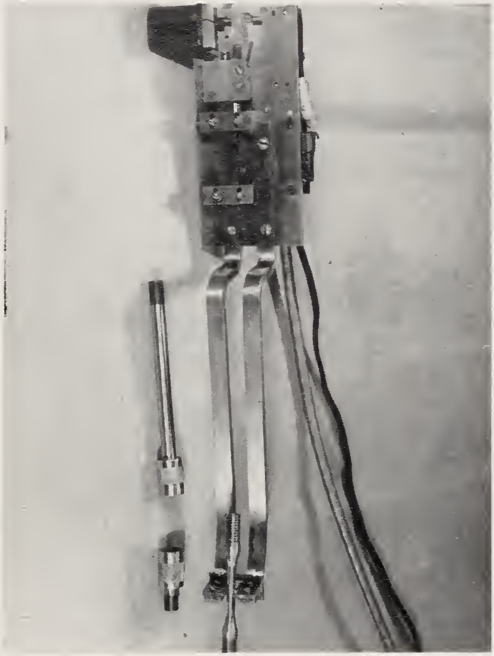

FIGURE 8.5. *Diameter gage used in tensile testing.*

Results

Tensile data for four alloys are tabulated in tables 8.2 to 8.5, and are graphically illustrated in figures 8.6 to 8.9. This information was determined from the natural stress-strain diagrams calculated in the conventional manner, where the average stress is taken as $\sigma = L/A$ and $\epsilon = \ln A_0/A$. (σ=average instantaneous stress; L=load; ϵ=natural strain; A_0=original area; and A=instantaneous area.)

A strain rate of 7×10^{-4} sec^{-1} was used throughout, with the exception of the 0.12 percent carbon alloys, for which the strain rate varied from 3×10^{-4} sec^{-1} to 15×10^{-4} sec^{-1}, as is shown in figure 8.8.

A minimum of two specimens was tested at each temperature investigated, extra tests being conducted when duplicate runs gave unsatisfactory correspondence.

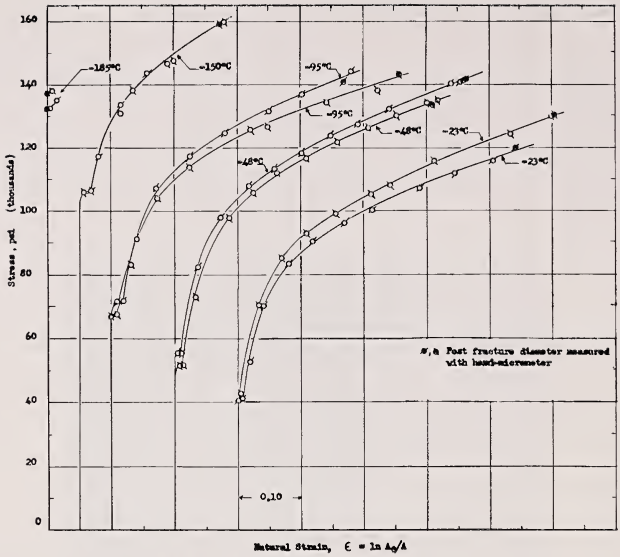

FIGURE 8.6. *Stress-strain curves for iron-carbon alloy (53V) containing 0.49 percent of carbon.*

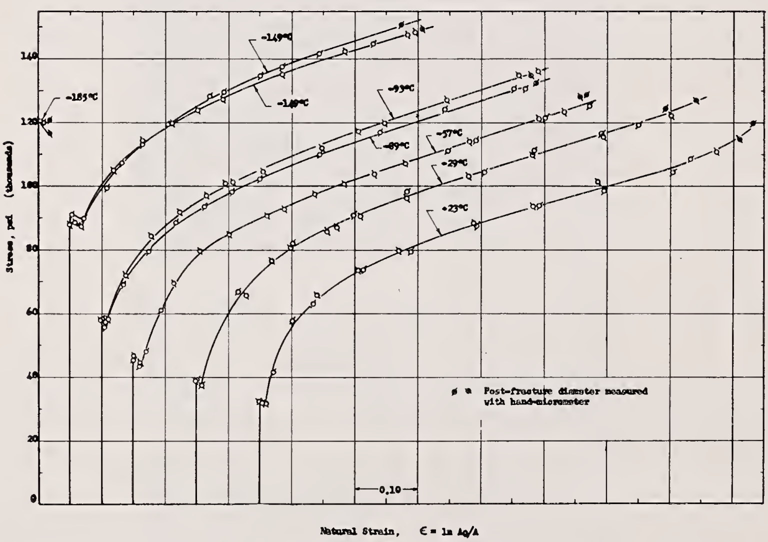

FIGURE 8.7. *Stress-strain curves for iron-carbon alloy (43V) containing 0.25 percent of carbon.*

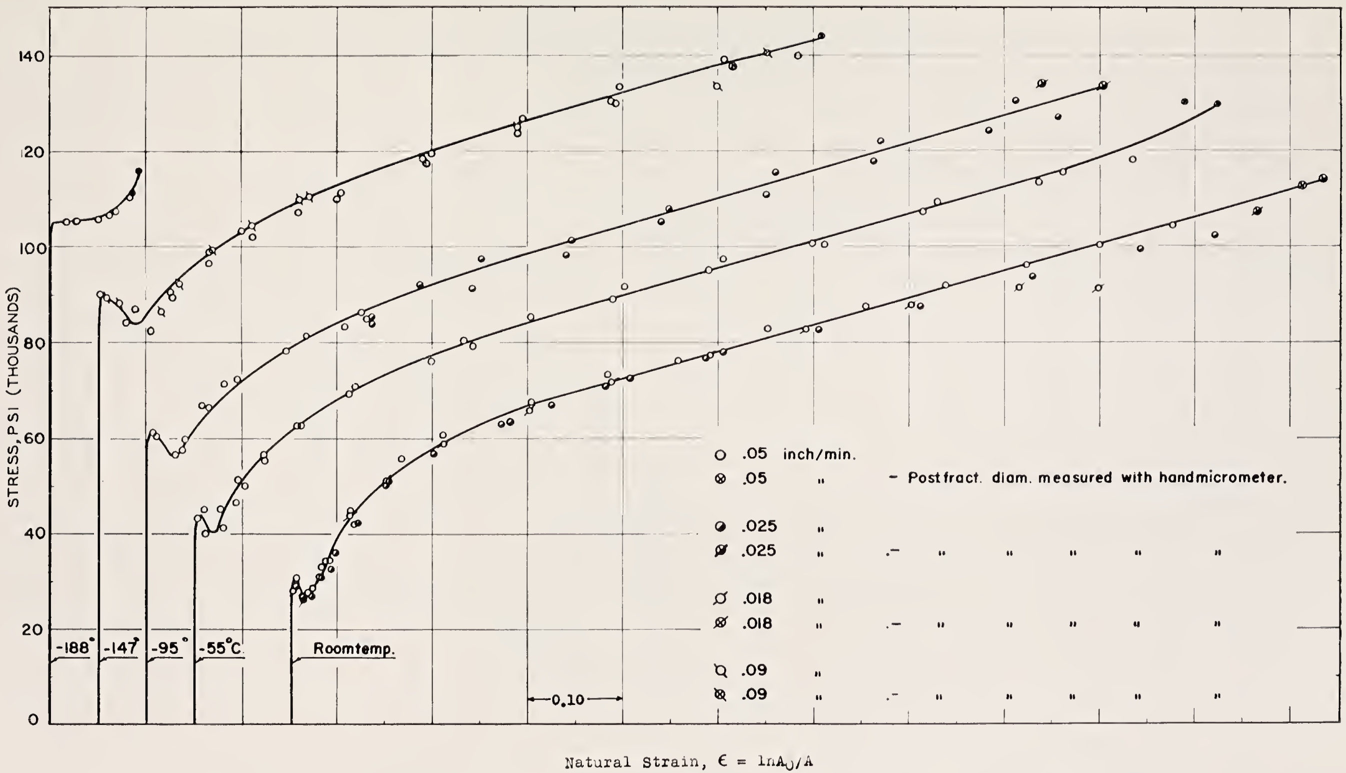

FIGURE 8.8. *Stress-strain curves for iron-carbon alloy (51V) containing 0.12 percent of carbon.*

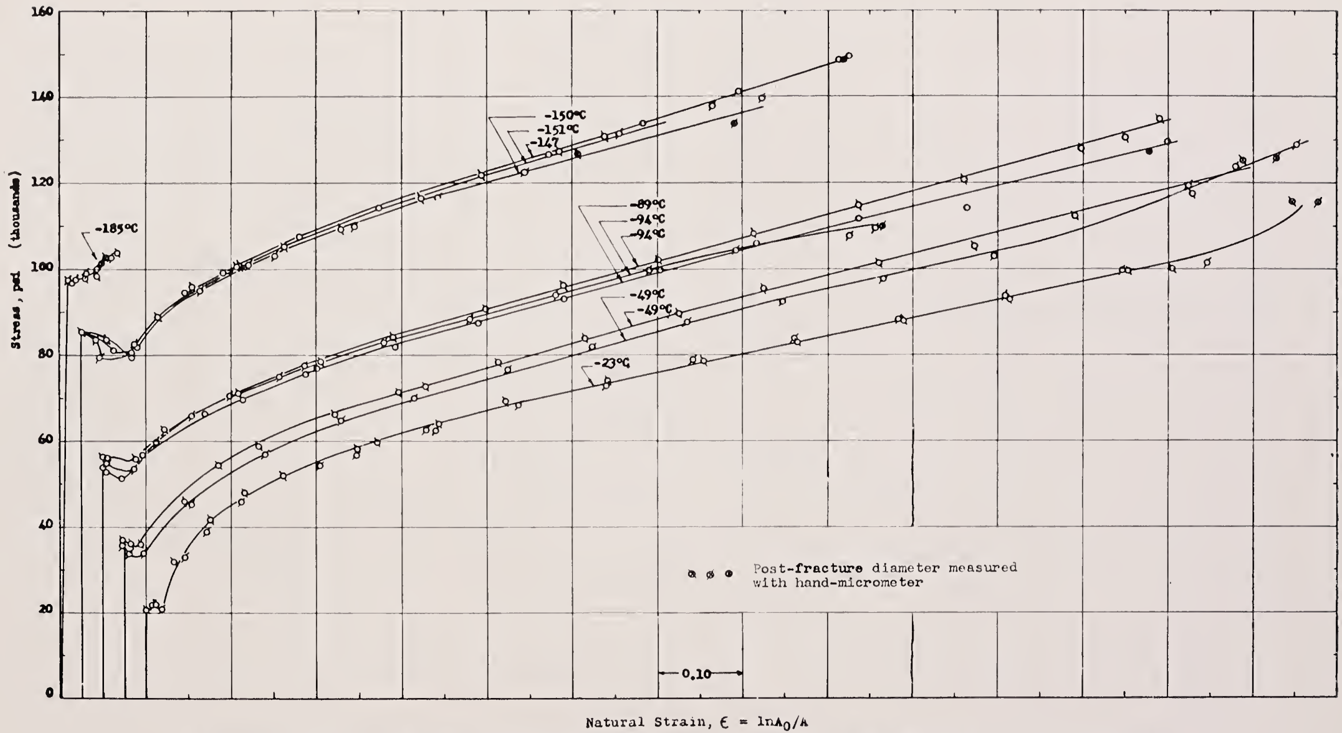

FIGURE 8.9. *Stress-strain curves for iron-carbon alloy (49V) containing 0.05 percent of carbon.*

Table 8.2. *Tensile data for iron-carbon alloy containing 0.45 percent of carbon*

[Strain rate for all bars was 0.04 in./min]

Number of bar	Test temperature	Upper yield stress	Lower yield stress	Worked fracture stress	Flow stress at natural strain of 0.2	Uniform strain	Total strain	Ultimate strength	Elongation in 1 in.[a]	Reduction in area [a]
	° *C.*	*psi*	*psi*	*psi*	*psi*			*psi*	%	%
7	+23	40,800	41,100	120,600	98,500	0.213	0.443	81,300	27	36
23	+23	42,600	None	130,200	105,000	.217	.504	84,800	27	40
15	−46	55,600	None	------	118,000	------	------	97,000	------	------
15	−48	------	------	141,800	------	.201	.463	------	25	38
3	−48	51,500	51,400	------	116,700	------	------	------	------	------
3	−49	------	------	135,000	------	.208	.417	95,000	27	33
27	−94	67,000	67,500	------	122,500	------	------	------	------	------
27	−95	------	------	143,500	------	.223	.457	101,100	28	37
12	−95	71,300	71,800	144,200	127,000	.180	.379	104,100	27	30
24	−151	106,600	(b)	(b)	------	------	------	------	------	------
4	−150	106,000	107,000	160,000	156,000	.138	.229	128,000	9	20
16	−185	------	------	138,700	------	.008	.008	137,700	0	0
11	−185	133,000	------	135,200	------	.016	.016	133,000	0	0

[a] Elongation and reduction of area values calculated from post-fracture data to nearest 1 percent.
[b] Broke on shoulder.

TABLE 8.3. *Tensile data for iron-carbon alloy containing 0.25 percent of carbon*

[Strain rate for all bars was 0.04 in./min]

Number of bar	Test temperature	Upper yield stress	Lower yield stress	Worked fracture stress	Flow stress at natural strain of 0.2	Uniform strain	Total strain	Ultimate strength	Elongation in 1 in.	Reduction in area
	° C	psi	psi	psi	psi			psi	%	%
19	+23	31, 950	31, 400	119, 300	76, 200	0. 239	0. 782	62, 400	36	59. 2
7	+23	32, 000	31, 900	114, 700	78, 000	. 221	. 761	63, 500	34	53. 3
26	−29	39, 300	37, 400	124, 700	86, 000	. 252	. 742	70, 500	32	52. 5
14	−29	38, 700	37, 400	127, 200	85, 000	. 258	. 790	69, 800	37	54. 7
23	−52	45, 400	43, 600							
23	−54			124, 200			. 715		38	51
15	−57	46, 100	44, 600		89, 000	. 210		73, 100		
15	−56			128, 500			. 706		43	51. 5
18	−90	57, 700	55, 700	132, 000	97, 500					
18	−88					. 207	. 686	79, 500	30	49. 6
6	−92	57, 500								
6	−94		57, 500	134, 500	100, 000	. 195	. 678	82, 700		49. 2
11	−149	90, 700	89, 300		125, 000					
	−148			150, 700		. 220	. 524	102, 500	22	40. 8
27	−149	87, 800		149, 500	123, 200	. 160	. 557	101, 400	25	42. 7
10	−185	120, 200		121, 000		. 019	. 019	119, 200	2	2
22	−185	120, 800		116, 600		. 019	. 019	120, 000	2	2

Table 8.4. *Tensile data for iron-carbon alloy containing 0.12 percent of carbon*

Number of bar	Test temperature	Rate of strain	Upper yield stress	Lower yield stress	Worked fracture stress	Flow stress at natural strain of 0.2	Uniform strain	Total strain	Ultimate strength	Elongation in 1 in.	Reduction in area
	° *C*	*in./min*	*psi*	*psi*	*psi*	*psi*			*psi*	%	%
35	+23	0.025	28,000	26,100	----	61,300	0.350	0.967	51,000	38	63
24	+23	.025	28,800	26,600	114,800	61,000	.329	1.083	51,000	39	66
45	+23	.05	30,100	27,100	104,200	63,000	.332	.925	52,000	39	63
19	+23	.018	27,400	25,400	107,300	61,500	.336	1.012	51,000	38	64
23	−55	.05	43,600	40,100	----	73,000	.291	----	59,000	42	66
23	−51	----	----	----	129,800	----	----	1.077	----	----	----
44	−55	.05	45,100	41,900	----	73,700	.281	----	61,000	45	60
44	−50	----	----	----	130,300	----	----	1.044	----	----	----
36	−93	.05 to 0.025	61,100	57,400	134,000	83,300	.207	1.008	70,000	42	63
41	−92	.05 to 0.025	61,300	56,600	----	84,600	.275	----	69,000	42	61
41	−98	----	----	----	134,000	----	----	.940	----	----	----
20	−95	.04	58,500	----	----	----	----	----	----	----	----
20	−90	----	----	55,000	135,200	80,200	.220	.993	65,500	----	63
49	−147	.05 to 0.025	90,100	86,600	139,200	107,300	.228	.657	89,000	38	54
32	−148	.05 to 0.025	89,700	87,000	----	109,000	.211	----	89,000	39	50
32	−147	----	----	----	140,400	----	----	.702	----	----	----
28	−147	.04	87,900	82,600	143,200	105,000	.212	.767	86,500	32	53
31	−188	.05	105,300	106,000	111,000	----	.070	.070	104,000	3	5
27	−188	.05	105,800	106,800	116,200	----	.094	.094	109,000	6	8

TABLE 8.5. *Tensile data for iron-carbon alloy containing 0.5 percent of carbon*

[Strain rate for all bars was 0.04 in./min]

Number of bar	Test temperature	Upper yield stress	Lower yield stress	Worked fracture stress	Flow stress at natural strain of 0.2	Uniform strain	Total strain	Ultimate strength	Elongation in 1 in.	Reduction in area
	° *C.*	*psi*	*psi*	*psi*	*psi*			*psi*	%	%
4	+23	21,200	20,600	115,200	53,500	0.330	1.378	44,700	42	75
10	+23	20,500	None	115.800	54,500	.274	1.348	45,200	41	75
1	−49	35,900	------------	------------	60,100	------------	------------	------------	------------	------------
1	−47	------------	33,600	------------	------------	------------	------------	------------	------------	------------
1	−49	------------	------------	126,000	------------	.339	1.376	49,800	52	74
14	−49	36,600	35,600	125,100	62,500	.322	1.313	51,600	52	73
2	−89	53,700	51,500	127,000	72,500	.239	1.230	59,300	------------	71
5	−97	54,600	------------	------------	74,000	------------	------------	------------	------------	------------
5	−92	------------	53,200	109,800	------------	.208	.911	60,600	47	60
88	−96	55,800	55,200	------------	73,500	------------	------------	------------	------------	------------
88	−93	------------	------------	134,900	------------	.256	1.240	60,200	48	71
6	−150	83,600	79,200	133,900	101,000	.196	.765	82,500	25	54
9	−152	85,100	79,000	126,400	103,000	.185	.551	83,700	22	43
13	−147	85,200	80,700	148,100	------------	.187	.895	83,700	------------	58
11	−183	97,400	99,600	102,300	------------	.049	.049	96,300	6	5
7	−183	97,100	98,600	103,700	------------	.054	.054	97,200	9	5

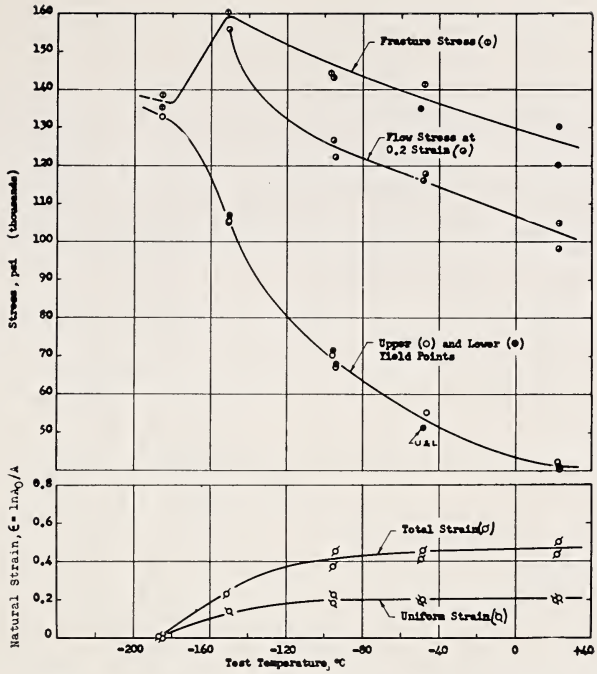

FIGURE 8.10. *Effect of temperature on the tensile properties of iron-carbon alloy containing 0.49 percent of carbon.*

In the property plots versus temperature all observation points are shown and smooth curves drawn through them, figures 8.10 to 8.13. Insufficient data are available at this time to establish the precise curves for total strain and fracture stress. Figures 8.14 to 8.17 show properties plotted versus carbon content. With the exception of the yield-point curves, the points shown are obtained from the curves drawn through the observation points of the property versus temperature plots.

Discussion of Results

Yield stresses, figures 8.10 to 8.13, increase rapidly on **S**-shaped curves as the temperature decreases. Double yields were observed for every temperature and every carbon content tested, although the difference between upper and lower yield points became very small for both the lowest and highest carbon alloys tested. It can be seen that in some cases, lower yield stresses are higher than upper yield stresses. This occurs when the amount of yield-point strain is great in comparison to the drop in load, thus resulting in an apparent loss of a lower yield in the stress calculation.

At room temperature the yield stresses increased nearly linearly with carbon content, a 0.5 percent increase in carbon having roughly the same effect as a 40°C decrease in temperature. At lower temperatures, there is a tendency for yield points to remain constant in the 0.1- to 0.25-percent carbon-range. Observation points in the graph

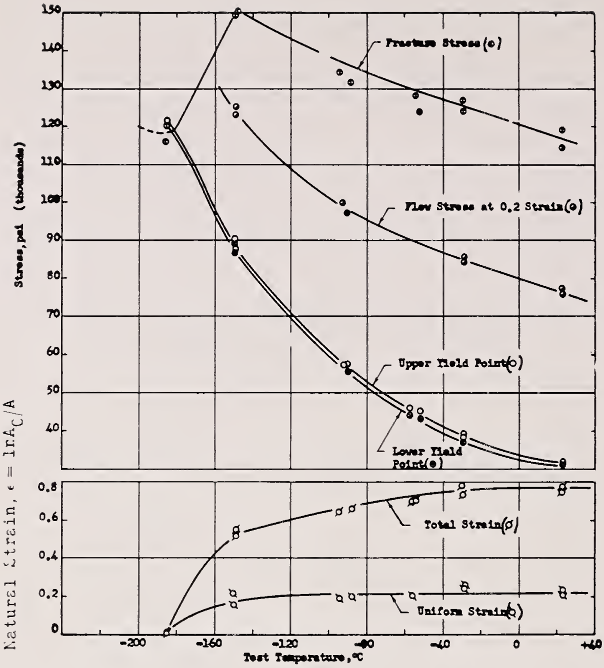

FIGURE 8.11. *Effect of temperature on the tensile properties of iron-carbon alloy containing 0.25 percent of carbon.*

(figs 8.16, and 8.17) illustrate this trend. Apparently heterogeneous yielding is insensitive to carbon content in this carbon range at low temperature.

Flow stress curves for a natural strain of 0.2 are shown in figures 8.10 to 8.13. These increase on upwardly curving lines as the temperature decreases and are similar in shape to ultimate strength curves versus temperature shown in figure 8.18. This similarity is to be expected as maximum load was generally found in the vicinity of 0.2 strain. Flow-stress values were obtained at various strains for several temperatures and flow-stress versus carbon-content curves were constructed, as is shown in figures 8.14 to 8.17. The flow stress for various constant strains increases linearly with increasing carbon content up to about 0.3 percent of carbon, then the curves tend to decrease slightly in slope from 0.3 to 0.5 percent of carbon. The slope of the flow-stress curves increases as the strain increases but remains essentially unchanged as temperature decreases.

Fracture-stress values are uncorrected in any manner and were calculated from the load at fracture divided by the area at fracture as determined by post-fracture measurements made with a micrometer. Figures 8.10 to 8.13 show that the fracture stress increases as the temperature decreases to about −150° C and then falls rapidly with further decrease in temperature. This drop corresponds to a change from ductile to brittle behavior and the loss of strain hardening effects on the fracture stress.

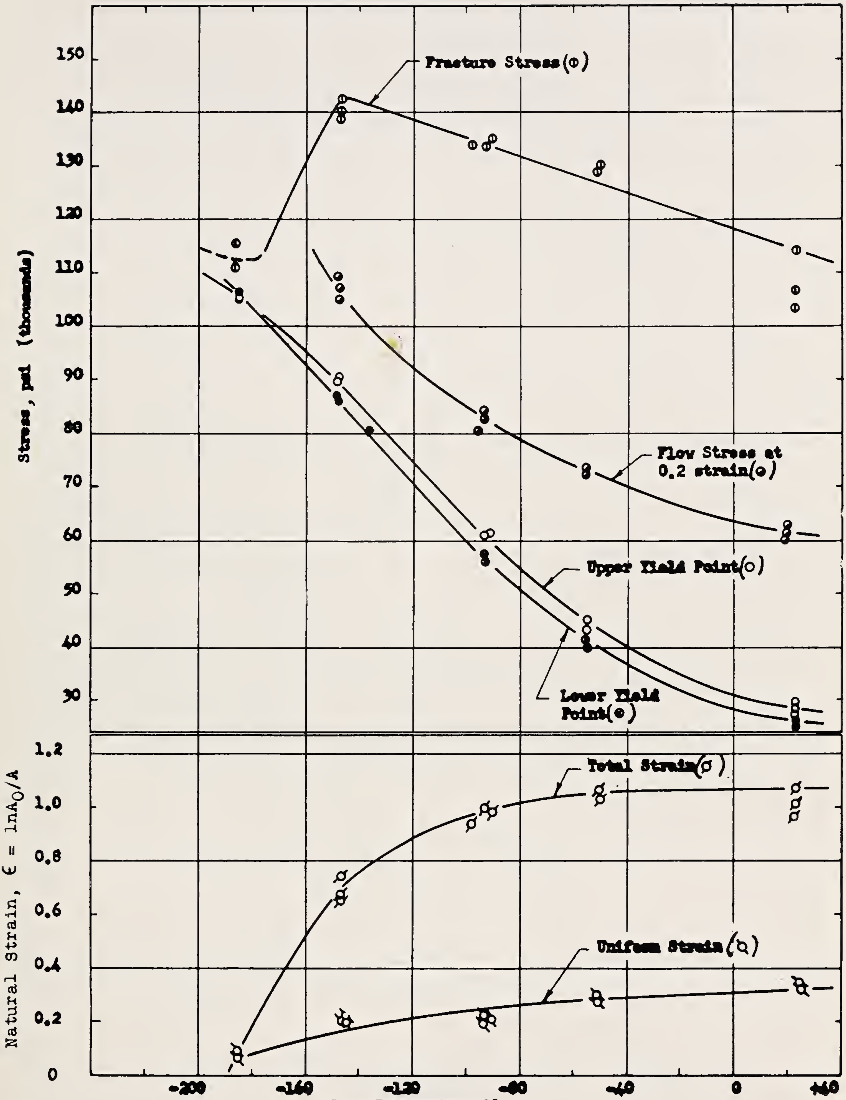

FIGURE 8.12. *Effect of temperature on the tensile properties of iron-carbon alloy containing 0.12 percent of carbon.*

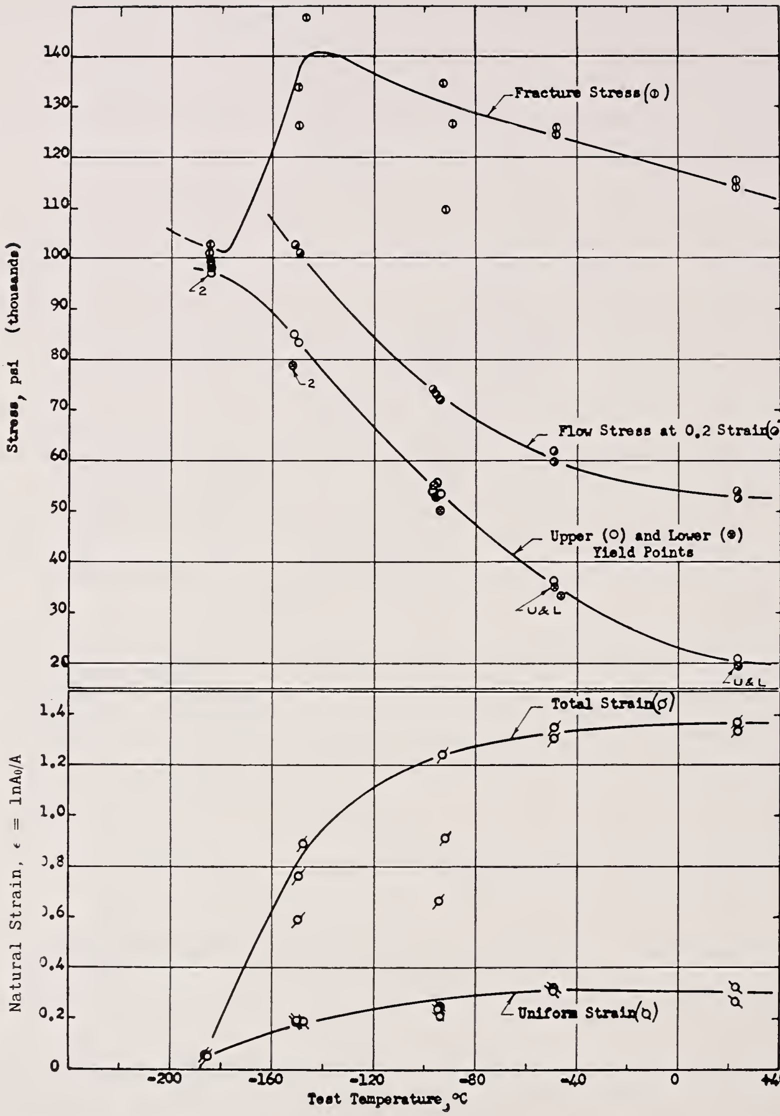

FIGURE 8.13. *Effect of temperature on the tensile properties of iron-carbon alloy containing 0.05 percent of carbon.*

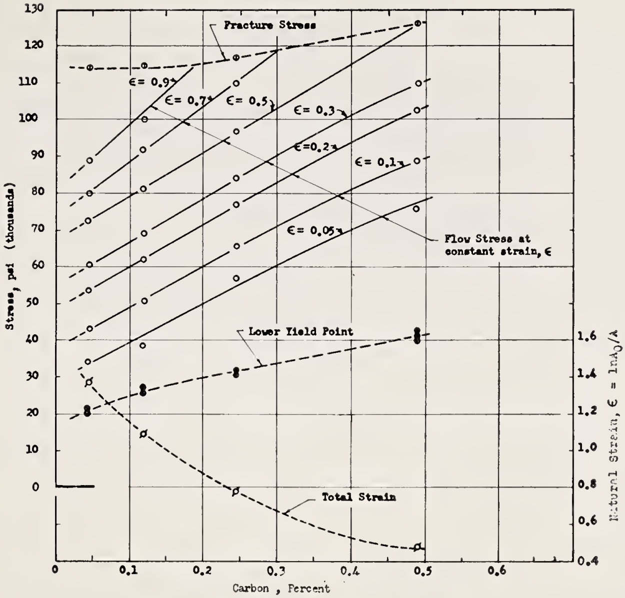

FIGURE 8.14. *Effect of carbon on the tensile properties of iron-carbon alloys at +23°C.*

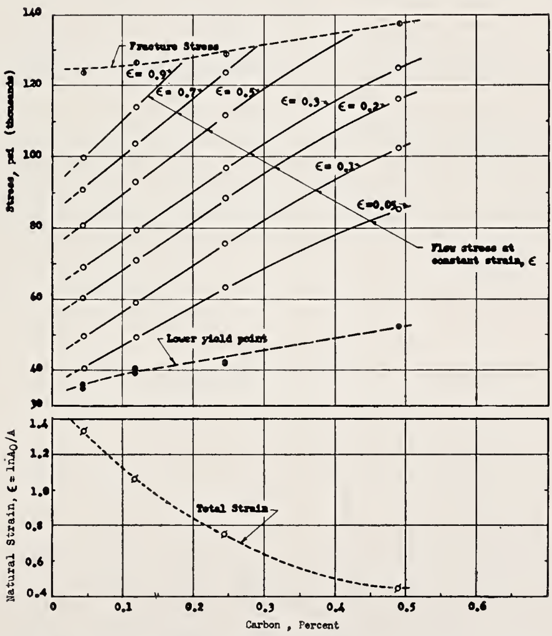

FIGURE 8.15. *Effect of carbon on the tensile properties of iron-carbon alloys at —50°C.*

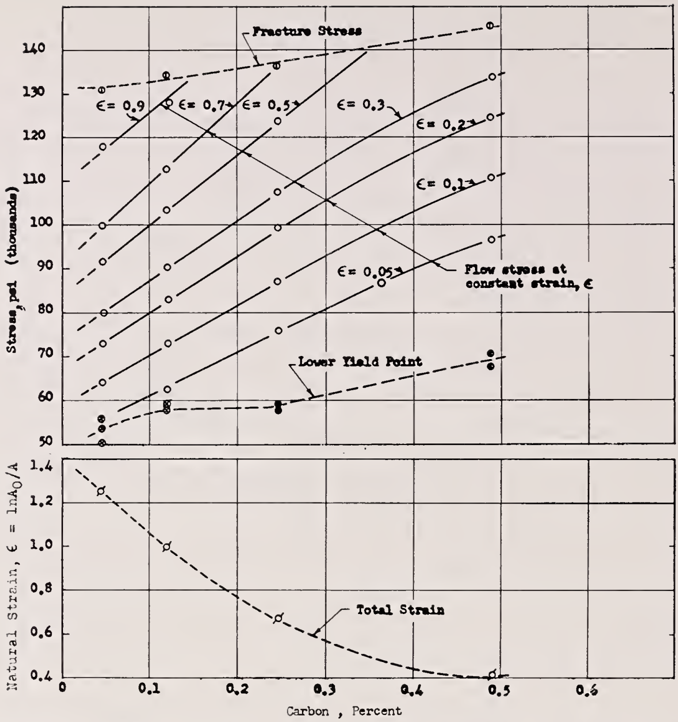

FIGURE 8.16. *Effect of carbon on the tensile properties of iron-carbon alloys at —95°C.*

There is a definite increase of fracture stress from the 0.05-percent-carbon alloy to the 0.49-percent-carbon alloy; however, because of the scatter in results it is not possible to specify with certainty the shape of the fracture-stress curve versus carbon content. It appears that at room temperature, fracture stress is relatively insensitive to changes in carbon content in the range 0.05 to 0.2 percent of carbon, whereas it increases more nearly linearly throughout the carbon range at lower temperatures. The fracture stress at liquid-air temperatures may be taken as the fracture stress essentially unaffected by work-hardening, triaxiality, changing strain rate, and structural preferment resulting from prior strain. This "unworked" fracture stress varies nearly linearly with increasing carbon content, a 0.5-percent-carbon increase causing a 35,000-psi increase in fracture stress.

Total strain, uniform strain, percentage elongation, and percentage reduction of area (figs. 8.10 to 8.13 and 8.18) generally decrease as the temperature decreases. They decrease slowly in the higher temperature range and more rapidly in the lower temperature range.

Total strain at all test temperatures decreases as the carbon content increases, figures 8.14 to 8.17. Some of the strain may be attributed to twinning (figs. 8.19, 8.20) for tests conducted at −140° C and lower. At −140° C both twins and deformation bands were observed, but at liquid air temperatures, only twins were observed. These occurred

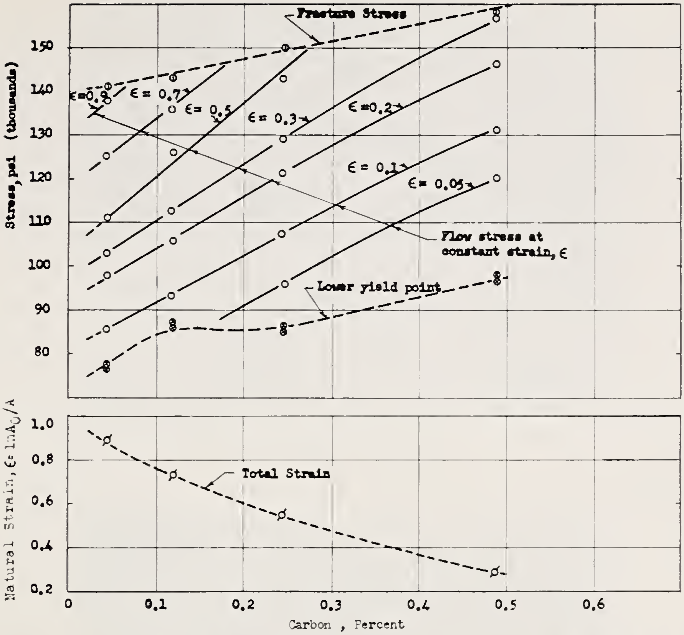

FIGURE 8.17. *Effect of carbon on the tensile properties of iron-carbon alloys at —145° C.*

FIGURE 8.18. *Effect of temperature on the engineering tensile properties of iron-carbon alloy containing 0.12 percent of carbon.*

FIGURE 8.19. *Structure of iron-carbon alloy containing 0.12 percent carbon after testing in liquid air.*

Considerable twinning is apparent over the entire gage section, etched with nital, ×250.

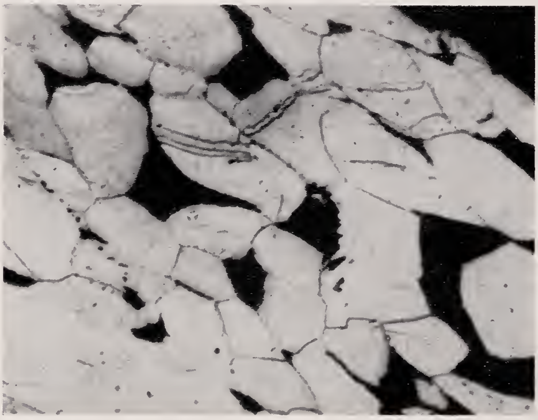

FIGURE 8.20. *Structure of iron-carbon alloy containing 0.12 percent of carbon after testing at —147° C.*

Section, about ¼ in. from the fracture, shows twinning and some deformation bands—etched with nital, ×250.

all along the gage section and it is probable that a large part of the strain at this temperature can be ascribed to twinning.

Uniform strain is defined as the strain taking place up to the onset of necking, and here was taken as the strain at which the maximum load occurred. Uniform strain decreases slowly with decreasing temperatures and decreases slightly with increasing carbon content.

From the plots of figures 8.14 to 8.17, stress-strain diagrams may be reconstructed for iron-carbon alloys from the range of solid solubility to 0.5 percent of carbon and for temperatures from +23° to −145° C. Such curves apply to similar microstructures, and could prove useful in determining the effect of alloying elements as differentiated from the effects of carbon alone.

In discussing properties derived from the portion of the stress-strain diagram after necking begins, it should be recognized that the effects of triaxiality, varying strain rate, plastic deformation, and varying degrees of structural preferment are acting on the test specimen. It should be further emphasized that the stress-strain diagrams as usually constructed are plots of the average axial stress versus natural strain at the minimum diameter of the test bar.

The method used for the tests described in this report for controlling the strain rate was to maintain a constant rate of head motion of the testing machine, thus keeping the rate of longitudinal extension constant. Since natural strain is defined as ln A_0/A, it can be seen that a constant rate of head motion results in a much greater rate of natural strain as the severity of necking increases. Both increasing the strain rate and setting up a condition of triaxiality have the same general effect as decreasing the temperature in that stress levels for particular strains are increased and total strains are decreased or uncorrected fracture stresses increased.

Some experiments have been performed to adjust for the change in natural strain rate during necking. When the strain rate is decreased to the same value as that employed at the point of maximum load, the flow stress drops abruptly. Increasing the strain rate to the original constant head motion raises the flow stress to the appropriate position on the original flow curve.

Attempts have been made to correct the natural stress-strain curves for triaxiality, using both the Bridgman correction [9] and a correction based on the observation that the average stress at a section of the neck away from the minimum diameter is not on the stress-strain curve for the indicated lesser strain but has a lower value. The corrected stress-strain curves are not yet sufficiently complete to permit any conclusions to be included in this report.

Both corrected and uncorrected natural stress-strain curves have been analyzed for the so-called modulus of strain hardening and the strain-hardening index n. These analyses have not yet revealed a consistent trend for strain hardening of the iron-carbon alloys as affected by temperatures in the range +23° to −185° C.

The drop in fracture stresses of figures 8.10 to 8.13 was previously stated to be caused by a change from ductile to brittle behavior. The temperature of this change, or the transition temperature of an iron-carbon alloy, may be determined in several ways. The usual basis for stipulating transition temperatures is the energy to fracture. This was determined by graphical integration of the area under the uncorrected natural stress-strain curves. The plots E_f or energy to fracture versus temperature gave curves that are approximately parallel to the total strain versus temperature curves. The E_f curves are nearly horizontal in the range +23° to −140° C with a slight maximum within this range; then they decrease sharply to a low value at −185° C. The transition temperature was chosen as that at which E_f was half of the maximum value. Figure 8.21 shows a plot of transition temperature versus percentage of carbon. A slightly sloped line showing at most a small effect of carbon content (in the range 0.05 to 0.5) on the transition temperature satisfactorily repre-

[9] P. W. Bridgman, Stress distribution at the neck of a tension specimen, Trans. Am. Soc. Metals **32**, 553 (1944).

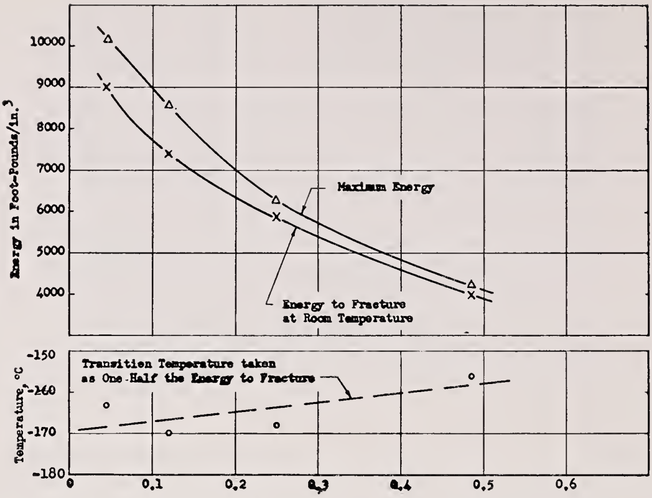

FIGURE 8.21. *Variation of transition temperature and energy parameters with carbon content.*

sents the data, considering the error in individual points. Separate analyses for transition temperature based on maximum fracture stress or total strain lead to the same conclusion as to the absence of a carbon effect on transition temperature. However, it is noted that the energy to fracture at room temperature and the maximum energy decrease strongly with the increase in carbon content.[10]

It is desirable to compare these data on high-purity iron-carbon alloys with the room-temperature properties recently reported by Raring, Rinebolt, and Harris (see footnote 10) for experimental steels containing approximately 1.0 percent of Mn and 0.25 percent of Si. Figure 8.22 shows that their reported yield points are consistently about 15,000 psi above those for the high-purity alloys. They obtained greater total strains and relatedly higher fracture stresses. It is unexpected to find both lower strengths and lower ductilities for the purer alloys. Unfortunately, this difference cannot be ascribed solely to compositional differences, particularly manganese, since the purer alloys reported here were coarser grained, ASTM No. 4 to 5, as compared to ASTM No. 8 for alloys tested by Raring, Rinebolt, and Harris. Apart from the differences in absolute values for stresses and strains, both investigations show similar effects of carbon on room-temperature properties of the perlitic structures tested.

Finally, properties of the lowest carbon content iron-carbon alloy may be compared with those of some high-purity binary ferrites prepared and tested on a companion program in the same laboratory, figures 8.23 and 8.24. None of the iron-carbon alloys show ductility comparable to high-purity aluminum ferrites containing 0.02 to 0.2 percent of Al. Furthermore, if the property versus percentage of carbon graphs of figures 8.14 to 8.17 are extrapolated to zero percent of carbon, the extrapolated high-purity iron data show too high strength and too low ductility values in comparison with values for

[10] Raring, Rinebolt, and Harris, J. Metals, p. 395 (May 1951).

the aluminum ferrites. Of course, the extrapolation is not valid anyway, since within the range of these tests, 0.05 to 0.49 percent of carbon, there is a structural similarity of carbon-saturated ferrite plus pearlite, while below 0.02 percent of carbon, a different type of structure would be found.

It apparently is demonstrable from the data of figures 8.23 and 8.24 and other as yet unpublished data on aluminum and silicon ferrites that both silicon and carbon adversely effect low-temperature ductility of iron.

Conclusions

Natural stress-strain data have been obtained over the range +23° to −185° C on four high-purity iron-carbon alloys, 0.05 to 0.49 percent of carbon, having microstructures with ASTM No. 4–5 ferrite grain size and pearlite spacing equivalent to normalized steels. Analyses of these data lead to the following conclusions:

1. Yield points and flow stresses increase with increase of carbon content and with decrease in temperature.

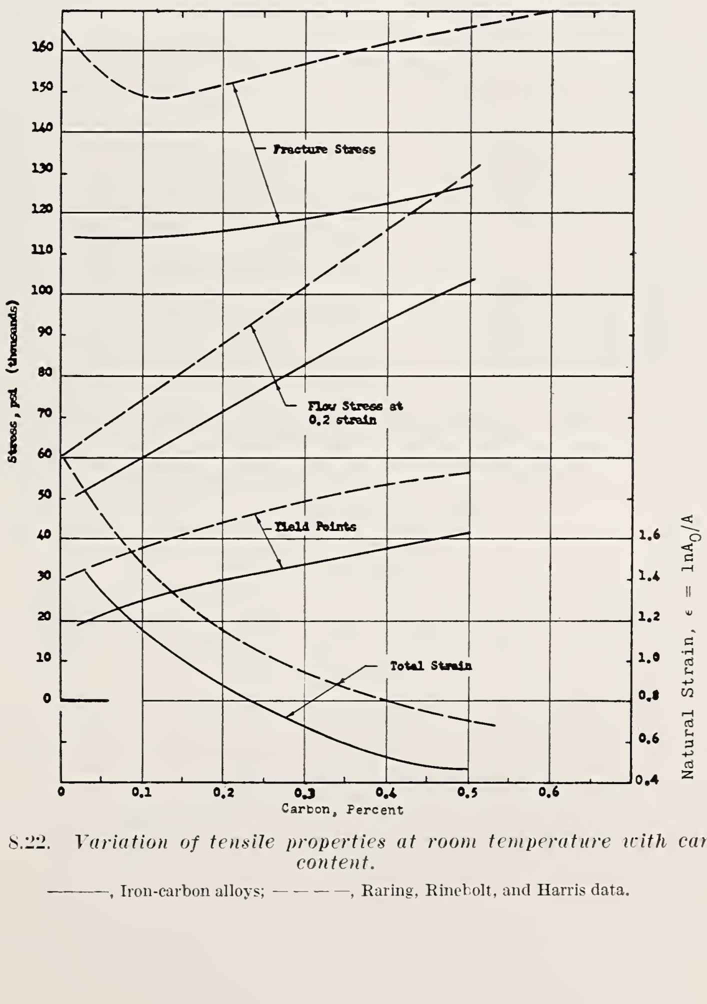

FIGURE 8.22. *Variation of tensile properties at room temperature with carbon content.*

———, Iron-carbon alloys; — — — —, Raring, Rinebolt, and Harris data.

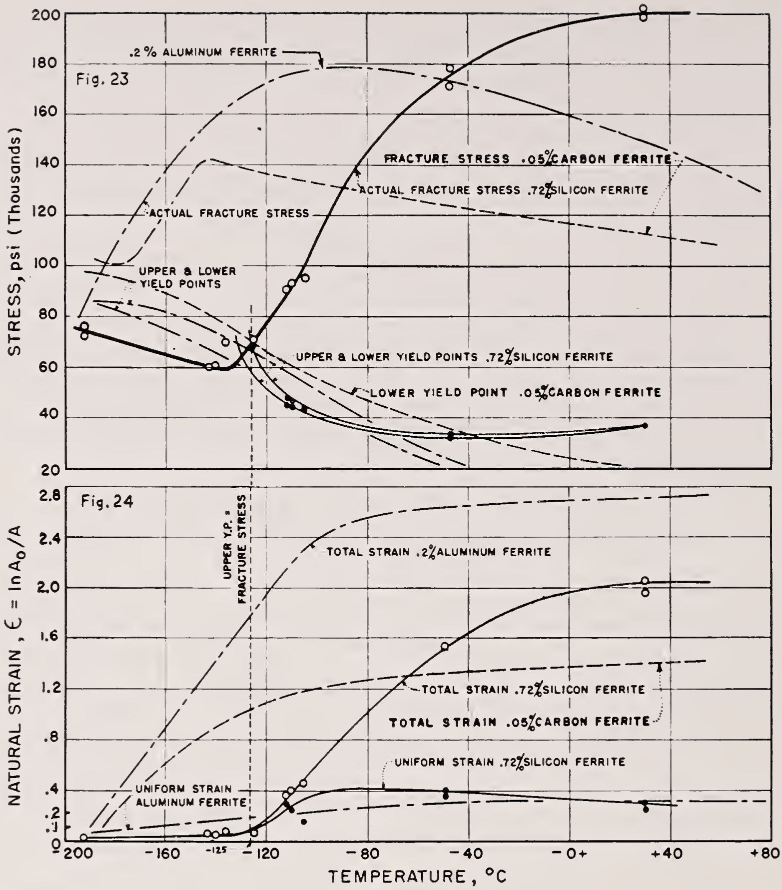

FIGURE 8.23. (Top) *Effect of temperature on the actual fracture stress and yield stress of some iron alloys.*

———, alloy X3VR, 0.72-percent-silicon ferrite (footnote 7, p. 156); —— – ——, 0.2-percent-aluminum ferrite (footnote 6, p. 156); — — — —, 0.05-percent-carbon ferrite.

FIGURE 8.24. (Bottom) *Effect of temperature on the natural strain of some iron alloys.*

2. Fracture stresses also increase with increase in carbon content and with decrease in temperature until, at around −150° C, loss of ductility becomes pronounced and the fracture stress drops to a considerably lower "unworked" value at −185° C.

3. The transition temperature, based on one-half the maximum energy to fracture of tensile specimens, is at about −156° to −170° C for all carbon contents investigated. The transition temperature, based on a selected value of energy to fracture, increases with carbon content but so slightly as not to be significant.

4. Increasing carbon content results in a marked decrease in total strain to fracture and, relatedly, of total energy absorbed to fracture at all test temperatures used here.

5. Comparison of data for these iron-carbon alloys with those for low-aluminum high-purity ferrites indicates that carbon reduces the ductility of iron at all temperatures in the range investigated.

Discussion

Mr. R. A. Lindberg, Metallurgist, Naval Research Laboratory, Washington, D. C.: It is perhaps time to consider or recognize a definition for "high-purity metal or metals", in order to avoid being misled by the literature or other fostered claims. To date, the term "high-purity" has been used rather loosely to describe a metal of varying degrees of purity. Other claims are similarly expressed as pure, ultrapure, ultrahigh purity or similar expressions to denote a metal of claimed purity that may range from 99.0 to 99.99 percent. In my opinion, iron of 99.9 percent purity is too low to be classed as a high-purity iron. It would seem desirable, therefore, to establish a standard nomenclature for metals of various degrees of purity.

Dr. R. M. Brick: That might well be a problem for the National Bureau of Standards to solve. The statement that these alloys were of 99.9 percent Fe, apart from the added element, is probably conservative. The electrolytic iron is reportedly 99.91 percent Fe with Mn, Si, Ni, Cr, Cu, Sn, etc. present in the range 0.001 to 0.005 percent. Our hydrogen-treated stock (Al and Si ferrites produced from thin sheet heated 6 weeks at 1,100° C in dry hydrogen) has been purified by the substantial removal of carbon, oxygen, and sulfur, and must be appreciably better than 99.9 percent Fe. The alloys of this report undoubtedly have higher oxygen and sulfur contents but still appear to be of "three nines" grade.

Dr. J. G. Thompson, Chief, Metallurgy Division, National Bureau of Standards: There is no such thing as pure iron. The more you analyze it and examine it, the more impurities you will find. I sympathize with Dr. Brick in his analytical difficulties. If he has iron that is definitely established as 99.9, I think he is entitled to call it high-purity iron.

Dr. E. E. Schumacher, Chief Metallurgist, Bell Telephone Laboratories, Murray Hill, N. J.: Have you used thoria crucibles in preparing your alloys? Have you tried melting your alloys in thoria-lined alumina crucibles?

Dr. Brick: No we have not. I believe that Dr. Allen has had some experience with them.

Dr. N. P. Allen, Superintendent of Metallurgy Division, National Physical Laboratory, Teddington, Middlesex, England: We have used sintered alumina crucibles in the preparation of our high-purity alloys. We have found these crucibles to be suitable, provided no carbon is present and they are not used at excessively high temperatures.

9. Brittle Fractures in Ship Plates

By Morgan L. Williams [1]

The principal factors contributing to structural failures of welded ships and preventive measures that have been applied are discussed. Improvements of design details and welding workmanship have materially decreased the incidence of failures, but notch sensitivity of the steel is also an important factor, as geometrical or metallurgical notches resulting from operation or maintenance, as well as from unavoidable structural details or welding defects in the original construction, can never be entirely eliminated.

Statistical interpretation of the data obtained from laboratory investigations of 113 plates removed from fractured ships indicates that the probability of a fracture starting in a plate increases markedly with increasing notch sensitivity, as measured by the 15 foot-pound transition temperature of Charpy V-notch specimens. Within the range of the chemical compositions of the ship plates, the notch sensitivity is increased with increasing carbon or phosphorus content and decreased with finer grain size or with increasing silicon or manganese content.

Introduction

The properties of metals at low temperatures are not merely an academic problem of importance only at the very low temperatures obtainable in the laboratory, but a very practical and serious problem at ordinary outdoor temperatures. Structural failures of welded merchant ships alone have cost the Nation almost 50 million dollars in the past 9 years [1],[2] and there have also been costly failures in other structures, such as bridges, storage tanks, and pressure vessels. Most of these failures occurred at low temperatures.

Approximately 5,000 merchant ships were built during World War II, and more than one-fifth of them had developed cracks in varying degrees before April 1946, when most of these ships were less than 3 years old [2]. From November 1942 to date, about 200 ships have sustained fractures that were classified as serious [3, 4]. This number does not include casualties resulting from war damage or from external causes, such as grounding or collision. Some of these failures involved the loss of lives and of ships.

Figure 9.1 shows the forward section of a tanker that broke in two at sea. Several people were removed from this section before it capsized and was sunk by gunfire.

Eight tankers and three Liberty type cargo vessels have broken completely in two as a result of the propagation of fractures that originated at some point in the hull structure or its appendages. About 25 other vessels have suffered complete fractures of the strength deck or the bottom.

Fractures have occurred in riveted ships and structures, but in a riveted ship the fractures usually ended at the first joint, and did not

[1] Metallurgist, National Bureau of Standards.
[2] Figures in brackets indicate the literature references on p. 206.

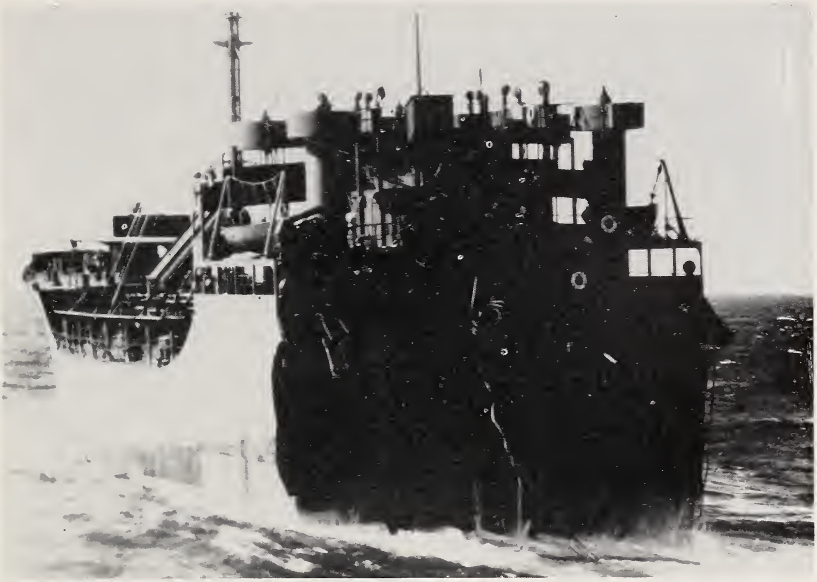

FIGURE 9.1. *Forward section of a T2 tanker that broke in two at sea.*

propagate into adjoining plates. A welded structure, however, is continuous, and a crack may propagate across the welds from one plate to the next, resulting in a complete structural failure. Therefore, the problem of fracture origin and propagation has received more attention in the past few years, because welding has largely replaced riveting in the fabrication of ships and other structures.

In April 1943 the Secretary of the Navy established a Board to Investigate the Design and Methods of Construction of Welded Steel Merchant Vessels. That Board and its successor, the Ship Structure Committee, conducted extensive technical and statistical studies of the casualties, and initiated research projects in several laboratories for the study of some of the factors involved. The Metallurgy Division of the National Bureau of Standards had already assisted in investigations of some of the earlier failures and was assigned to the investigation of the plates removed from the fractured ships. To obtain material for this investigation, the U. S. Coast Guard in November 1943 issued instructions [5] to the effect that, if steel is removed in repairing a fractured ship, samples of the fractured plates, properly identified, should be sent to the National Bureau of Standards.

Samples from some 80 ships in which fractures occurred have been received and examined, and tests have been completed on 113 plates selected from 58 of these ships. When sufficient and suitable material was available, the laboratory investigation included examination of the fracture and of welds, microscopic examination, chemical analyses of the plates, Charpy V-notched bar tests over a range of temperatures, and tension tests. Measurements of the reduction of thickness at the fracture edge were made on plates that were not too badly corroded or battered, and gas analyses by the vacuum-fusion method were completed on about half of the plates. In some instances other tests were made, but the main effort has been to obtain as much information as possible from tests that were more or less standard when this investigation was started.

To coordinate the results of the laboratory investigations with service experience, information was obtained from the Coast Guard and other cooperating agencies regarding the circumstances of the casualties, structural features of the ships involved, location and extent of the fractures, and other pertinent details.

Factors Involved in Structural Failures

The failures of the welded ships resulted largely from the lack of adequate knowledge of the distribution of stresses in ships under different operating conditions and of the significance of certain properties of structural steel, especially their relation to the performance of structures containing notches or similar discontinuities. The Board of Investigation, the Ship Structure Committee and its member agencies, and a number of cooperating laboratories have completed investigations to secure information on these phases of the subject.

There are three principal factors involved in the prevention of failures of welded structures. First, the material, mostly steel plates and weld metal, should possess, uniformly, the properties anticipated by the designer. Furthermore, these properties should not be affected adversely by the effects of fabricating operations or operating conditions. Second, the design should aim to produce a structure functionally suitable for the purpose intended, with adequate strength in all parts of the structure, under all possible operating conditions. Third, workmanship or fabrication—the work of the fitters, welders, supervisors, and inspectors—must carry out the ideas of the designer, and must produce joints of the strength that he intended.

Of course, these three factors are somewhat interrelated, and there may be some others involved, such as cost and availability of materials. For example, a design suitable for one material might be entirely unsuitable for a different material. The designer must consider the properties of the materials that are to be used, in relation to the design strength required and to the effects of any conditions that might be expected in fabrication or in service. Workmanship is also related to design; it is easier to make a good weld downhand than in a vertical or overhead position, or in a location that the welder cannot reach easily.

The failures in the ships usually resulted from a combination of these three factors, but each of them will be considered separately, with a brief discussion of what has been learned from the studies of the failures, and what has been done and what more can be done in the way of improvement.

Workmanship

Many of the earlier failures originated in welds which contained obvious defects, such as the following examples:

Figure 9.2 shows a fracture source in a butt weld in the bulwark cap rail of a cargo vessel. There was no penetration of weld metal in the center of the joint, resulting in a mechanical notch, as well as a

FIGURE 9.2. *Fracture source in defective weld of bulwark cap rail,* $\times 2/3$.

reduction in the effective cross section, and resultant strength, of the weld. The herringbone markings, or chevrons, in the fracture of the bulwark plate at the right are characteristic of the brittle type of fractures that were found in most of the ship plates examined. This type of fracture was observed and reproduced at the National Bureau of Standards about 15 years ago, in tests conducted to determine the source of a brittle failure of an aircraft part. These tests, which were discussed in the first NBS report on the examination of steels from a fractured ship[6], showed that a brittle failure may be produced in a normally ductile metal by applying a tensile stress to a notched plate specimen, and also established the fact that the chevrons always point back toward the origin of the fracture, which made it possible to locate the sources of the fractures in the ships.

The bulwark plate and cap rail are not ordinarily considered as strength members in the design calculations. Actually, however, these parts are at a critical location—the extreme fibers, considering the entire ship as a beam—and they are joined to the hull structure by welding. A number of other fractures started in appendages not considered as strength members, and propagated through welds into the hull structure. This illustrates one important point. As cracks originating in any part of a ship may propagate into the hull structure through the welds, the standards of quality of materials, design, and workmanship, even for nonstrength members attached to the hull by welding, should be equal to those required in the hull structure itself.

Figure 9.3 shows another butt weld, in a hatch side longitudinal girder. Again, the weld metal did not penetrate to the full thickness of the joint—the original flame-cut edge may be seen in the center of the picture. The upper side of the weld had cracked at some previous time, probably during or shortly after the welding, and the crack was covered over when the part was painted. The light area near the top is red lead paint, which had penetrated the crack. After this vessel was launched, but before it was completed, there was a sudden complete failure of the strength deck starting at this joint and a similar faulty weld on the opposite side. Repairs cost $40,000. This

FIGURE 9.3. *Fractured weld showing paint penetration in previous crack.*

costly failure might have been averted if, by better supervision and inspection, the initial crack in one side of the weld had been found and repaired before being painted.

There are a number of other causes leading to cracks in welds, such as gas pockets, or porosity; slag and dirt inclusions in the weld metal; poor fusion; undercutting of the plate at the edge of the weld; shrinkage cracks; improper edge preparation or poor fitup of the plates; or the use of slugs to fill up space that should be welded.

In figure 9.4 the chevrons show two fracture sources in the plate near a lap-seam weld. Both sources are at small craters where apparently the welder had struck his arc on the plate, away from the area of the weld. This creates a small mechanical notch, combined with a "metallurgical notch", in which the properties are decidedly different from the properties of the parent metal. A small volume of metal is heated by the arc, but because of the short time and the small total heat input, the surrounding base metal is not heated appreciably. Consequently, the heated zone is cooled very rapidly by conduction to the surrounding mass of cold metal, which amounts, in effect, to a very drastic quench. The result is a hard, extremely notch-sensitive region, probably combined with residual tensile stresses resulting from thermal contraction. A metallurgical notch may also result from a small weld

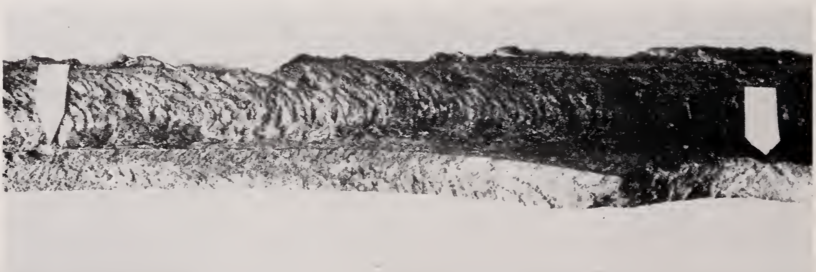

FIGURE 9.4. *Fracture sources at craters caused by touching welding electrode to plate.*

on heavy plate; a number of fractures have originated at light tack welds or at shallow welds joining small clips or brackets to heavier structural members.

As soon as it was realized that welding practices and defects of welding workmanship were a factor in the ship failures, the Board of Investigation initiated studies of welding procedures in the shipyards and in various laboratories, and started a campaign of education of welders and inspectors. A part of this effort was a small illustrated booklet [7] showing a few typical examples of serious failures resulting from welding defects, with a discussion of causes and cures of defective workmanship. The incidence of fractures resulting from welding defects has decreased noticeably in ships built since the issuance of this pamphlet in 1945.

Design

As several of the early failures originated in defective welds, and there had been few such spectacular failures in riveted ships, all the failures were at first attributed to the effects of welding. If no obvious defects could be found, the failures were explained as a result of underbead cracking or of residual or locked-in stress resulting from weld contraction, and studies of these phenomena were initiated.

It was soon noted, however, that many of the fractures originated at structural details such as hatch corners, ladder cutouts, or other openings, or at the abrupt ending of stiffeners, such as a bilge keel or a deck superstructure. The origin of the fractures at these points may be attributed primarily to the stress concentration resulting from a geometrical or structural notch, although in every case the metallurgical effects of welding, flame cutting, or mechanical working were also present, and these effects are not separable. It is unfortunate that the nature of ship construction required large amounts of welding at such points of geometrical discontinuity, as in many cases the welding was blamed for a failure that was primarily a result of stress concentration at a structural notch.

Some structural discontinuities are necessary to the function and operation of a vessel, but the design details could be improved, and they were improved. Alterations were made on existing ships and in new construction. Reinforcements were added at critical locations, hatch corners and other openings were rounded, and the ends of some stiffeners were tapered, to reduce stress concentrations. Butt welds of appendages such as the bulwark plates or bilge keels were isolated from the hull structure, by cutting serrations or holes to break the continuity between these welds and the hull plates. Even with these improvements, cracks may start at unsuspected locations, so in many types of vessels barriers were installed to arrest such cracks before extensive damage resulted. The observation that in riveted ships the cracks usually stopped at a joint led to the installation of some riveted seams or angles in new construction. In existing all-welded vessels, longitudinal slots were cut in the deck plating, the sides, or the bottom, and the plates were then rejoined with a riveted crack-arrestor strap, similar to a riveted butt-strap joint.

The effectiveness of these improvements in workmanship and design details may be seen in the record of service performance [4]. The original all-welded Liberty ships had 88 class I, or major, casualties in 2,100 ship-years of service, or 4.18 serious casualties per 100 ship-years. The all-welded Liberty ships with improved details, with and without riveted crack arrestors, have accumulated more than 7,000 ship-years of service to January 1, 1950, with only 37 major casualties, or 0.52 per 100 ship-years. Thus, the number of major casualties has been reduced by a factor of 8, largely as a result of design modifications in these ships. The design of the Victory ships incorporated many of the lessons learned from the earlier failures, and no major structural casualties have been reported in this type of vessel as originally built, with a service record of nearly 1,800 ship-years.

The stresses and stress concentrations in operating ships have been measured, both in this country and in England. The effects of reinforcements and of modifications of design details are being investigated, in service applications and in a number of laboratories. The designer now has available a better knowledge of the fundamental details on which to base his calculations, and we may expect further improvements in design.

Materials

Tension tests indicated that the plates that had fractured in service were of normal ductility, and that the steels would meet the specification requirements under which they were purchased. This type of steel, in the usual tensile test, elongates under load more than 20 percent in 8 inches, and the reduction of cross-sectional area is about 50 percent, indicating that the material is capable of considerable plastic deformation before it will fracture.

The fractures in the ships, however, showed very little evidence of plastic deformation, or ductility. Nearly all these fractures were of a brittle type, characterized by a break nearly perpendicular to the plate surfaces, and a coarse granular, or crystalline, appearance of the fracture. The reduction of thickness at the fracture edge was very small, usually less than 2 or 3 percent, and the paint or scale on the plate surfaces was not cracked, even very near to the fracture. This showed that the fractures had propagated with very little plastic deformation of the steel, and probably at low stress levels. As energy, or the capacity for doing work, is proportional to the integral of stress (or force) times deformation (or distance through which the force acts), it is evident that much less energy was absorbed in the propagation of these fractures than in the normal ductile fracture of this material.

The lack of ductility and the brittle nature of these fractures indicated that when the steel was incorporated in the structure of the ship, the mechanism of fracture, or the mechanical properties of the steel, were not the same as when determined by the usual tensile test, using relatively small specimens. This phenomenon is similar to that observed in tests of notched specimens in tension or bending, particularly at low temperatures. The similarity is also evident in the facts that the fractures in ships occurred more frequently at the lower temperatures, and that the starting points of the fractures could be traced, invariably, to geometrical or metallurgical notches resulting from structural or design details, fabrication processes, or welding defects. This phenomenon, called notch brittleness, is not peculiar to ship plate alone, and is not confined to metals. The scoring of glass for cutting and the notching of cellophane wrappers are familiar examples in which notch sensitivity is utilized to control the location or direction of a tear or fracture.

The importance of notch sensitivity and brittle fractures of the steel as a factor in the failures was recognized early in the investigation, and notched-bar tests on V-notch Charpy specimens were included in the examinations of the fractured plates. At the same time, investigations of the notch sensitivity of selected steels were conducted in other laboratories, and several new testing techniques and test specimens were developed. However, pending the results of these investigations, very little was known about the possibility of improving the notch toughness of steel in tonnage quantities. Consequently, during the war, preventive measures were directed largely to reduction of stresses by careful loading and ballasting, elimination of notches by improvements of design details and welding workmanship, and prevention of the spread of fractures by the installation of riveted crack arrestors. The effectiveness of these crack arrestors shows quite conclusively that the propagation of the fractures is not a direct result of the loss of the strength and support of the structural members initially fractured, but is due primarily to notch sensitivity. In the presence of the sharp notch formed by the initial crack, the fractures could propagate at relatively low stress and energy levels until the continuity of metal was broken by the crack-arrestor slot. However, even in cases where the entire deck or bottom was fractured, the remaining structure, in the absence of a notch, absorbed sufficient energy to prevent further spread of the fracture.

The failure of the tanker shown in figure 9.5 placed further emphasis on the importance of the steel itself in the prevention of structural failures. This ship broke almost completely in two, while tied up at a dock for conversion after the war. This conversion was to include the installation of crack arrestors. The positions assumed by the two parts of the vessel are an indication of the bending moments that contributed to the origin of the fracture, but the calculated nominal tensile stress in the deck was only 12,000 lb/in.^{2}, or roughly one-third the yield point of the steel.

The starting point of this fracture (fig. 9.6) was at the toe of a small fillet weld joining a clip to the deck stringer plate. The clip was evidently attached after the ship was built, and was placed very near to the end of a chock base, which was also welded to the deck. Two small arc craters were formed at the toe of the weld, probably because of the difficult welding position between the clip and the chock base. These craters, combined with the overlapping welds, one of which was small and shallow, formed a metallurgical notch in addition to the mechanical notch between the welds, and aggravated the stress concentration resulting from the stiffening effect of the chock base. The deck plate itself was not abnormally notch sensitive compared to other ship plates, but in the vicinity of the welds and the arc craters the notch sensitivity was probably greatly increased, as has been observed in a number of other ship plates.

The significant point illustrated by the casualty mentioned above is that even the greatest refinements of design and workmanship in the

FIGURE 9.5 *Tanker that broke in two at dock.*

original construction cannot prevent failures that may result from notches introduced after the vessel is in service. A mechanical or metallurgical notch may be formed at any time during the life of a structure as a result of minor accidents, alterations, temporary attachments, or even accidental touching of the steel by a welding electrode or an electrical conductor. Such notches, resulting from conditions of operation and maintenance, as well as notches resulting from design details or defective welds in the original construction, can never be entirely eliminated. Therefore, it is evident that the material itself must be capable of resisting the initiation and propagation of fractures. In other words, the steel must be notch tough, and furthermore, this toughness must not be seriously impaired by fabricating operations, such as welding, flame cutting, or cold-forming, or by operating conditions, such as low temperature.

In tension, slow-bend, or impact tests of notched specimens, most steels are notch tough at higher temperatures, but as the temperature is lowered they become, suddenly or gradually, more notch sensitive. The fracture changes from ductile to brittle, and the energy required to produce fracture is reduced to a small fraction of that which can be absorbed at higher temperatures. The range of temperatures within which these changes take place, for a given steel, is called the transition range of that steel. Usually, for convenience, a specific point in this range is referred to as the transition temperature, defined by such criteria as the temperature at which brittle fractures first appear, the thickness reduction at some point in the specimen, the temperature at

FIGURE 9.6. *Source of fracture shown in figure 9.5 at arc craters in weld of small clip to deck.*

which the energy absorbed is 50 percent of the maximum, or the temperature at which the energy absorption is at a definite level, say 15 ft-lb for a V-notch Charpy specimen.

The transition range of a steel is dependent on the test method and on the geometry of the test specimen, but, in general, a series of steels will be rated in approximately the same order of merit by the various tests and by the different criteria of transition temperature. It should be noted, however, that the transition temperatures determined in laboratory tests are not necessarily the same as the transition temperatures of the same steels in a structure such as a ship, where the stress concentration, rate of stress application, notch geometry, etc., may differ widely.

It has been realized only recently that the usual tests for notch sensitivity measure, in varying proportions, a combination of two factors that may vary independently in their relations to stress concentration, temperature, and rate of stress application. The ductility transition is related to the resistance of a material to the initiation of a fracture, and is sensitive to specimen design, that is, the degree of restraint and the sharpness of the notch. The fracture transition is related to the resistance of the material to propagation of the fracture, and is relatively less sensitive to specimen geometry. Which of these transitions is of greater importance in relation to the problem of brittle fractures in ships is still a matter of debate [8, 9].

Relation of Notch Sensitivity to the Failures of the Plates in Service

At the start of the investigation of the plates removed from fractured ships, the Charpy notched-bar impact test on V-notch specimens was chosen as a measure of the notch sensitivity of the steels. The amount of material available in the fractured plates was often limited, and this test, which was in fairly wide use at that time, requires a relatively small amount of material. It also has the advantages that tests can be conducted easily and quickly over a wide range of temperatures, and that the tests provide a numerical value of the energy required to fracture a specimen at a given temperature, which is independent of the personal judgment involved in determinations based on the appearance of the fractures.

After notched-bar tests were completed on only a few plates, it was observed that the plates in which the ship fractures originated showed lower energy absorption (or higher notch sensitivity) than plates that did not contain a fracture source. Therefore, the fractured plates for which definite information was available were classified into three categories: those which contained a fracture source (in the ship failures), those which were fractured through, and those which contained the end of a fracture. The plates were also classified in groups on the basis of plate thickness, to permit comparisons of data for similar material, and plates that were not strictly "hull plates of welded ships fractured under normal operating conditions" were placed in a separate miscellaneous group.

Figure 9.7 shows the notched-bar test curves in the transition range for plates in the "fracture-source" and "fracture-end" categories, for one thickness group. For each plate, points representing the average energy absorbed in tests at different temperatures are connected by straight lines ,with no attempt to smooth the curves. "T" on some of the curves indicates the temperature, if known, at the time of the ship failure. The 15 ft-lb transition temperature of each plate, defined as the temperature at which the notched-bar test curve crosses the line of 15 ft-lb energy absorption, is used to compare the plates at the same energy level. The horizontal bars superimposed on each set of curves represent the range of 15 ft-lb transition temperatures of the plates in each category, and the vertical bars represent the range of average values of energy absorbed in tests at 30° and at 70°F. It may be seen that the fracture-source plates (above) had higher transition temperatures, and lower values of energy absorption at corresponding temperatures than the plates in the fracture end category (below). That is, the plates in which fractures originated were considerably more notch sensitive than the plates in which the fractures ended. Similar sets of curves were drawn for the plates in the "fracture-through" category and for the other thickness groups, and the results are summarized in the following figures.

Figure 9.8 shows the energy absorbed by Charpy V-notch specimens at the temperatures of the ship failures in which the plates were involved. The horizontal bars represent the range of values of energy absorption at the failure temperatures, for each group of plates. The vertical lines in the bars show the values for individual plates, and

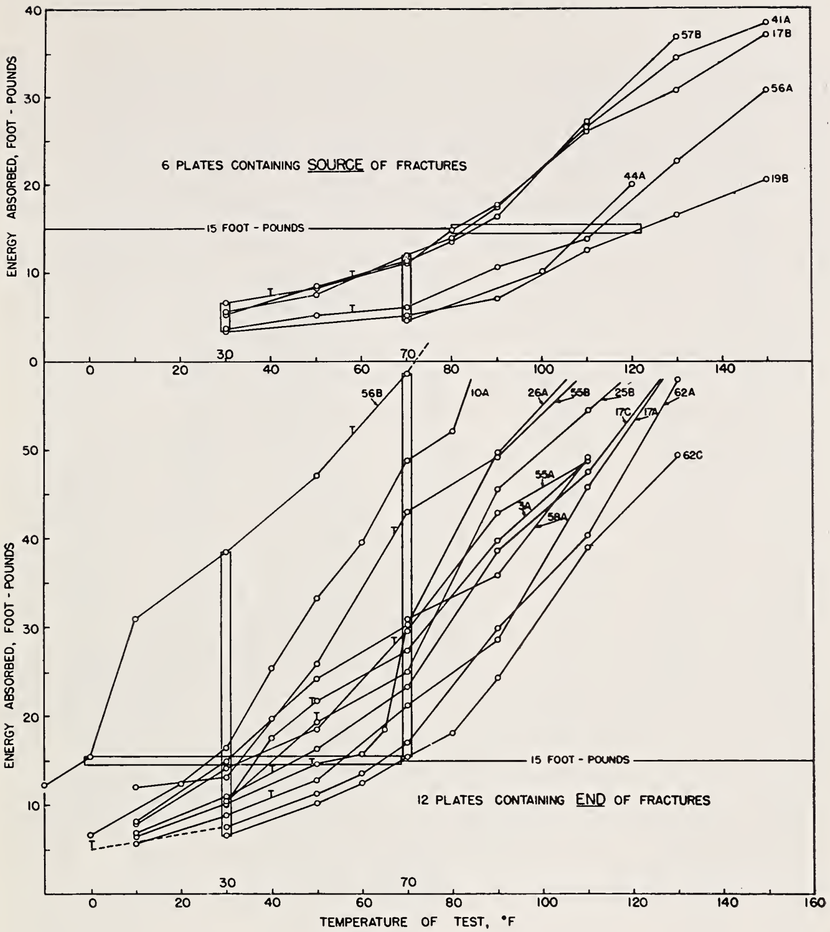

FIGURE 9.7. *Notched-bar test curves for plates containing sources and ends of fractures in the ships.*

the circles above the bars indicate the average value for each thickness group.

The 15 ft-lb transition temperatures of the plates are compared in figure 9.9. In addition to the average, individual values, and range of observed values for the plates in each category, the arrows show the range of plus or minus twice the standard deviation of the observed transition temperatures, which should include about 95 percent of the total observations. The upper three sections of the figure show the transition temperatures of fractured hull plates in the source, through, and end categories for each of the three plate-thickness groups, and the fourth section shows the data for all of the above thickness groups combined. In the lower section of the figure the data for the miscellaneous plates are shown. These are plates from riveted ships or from

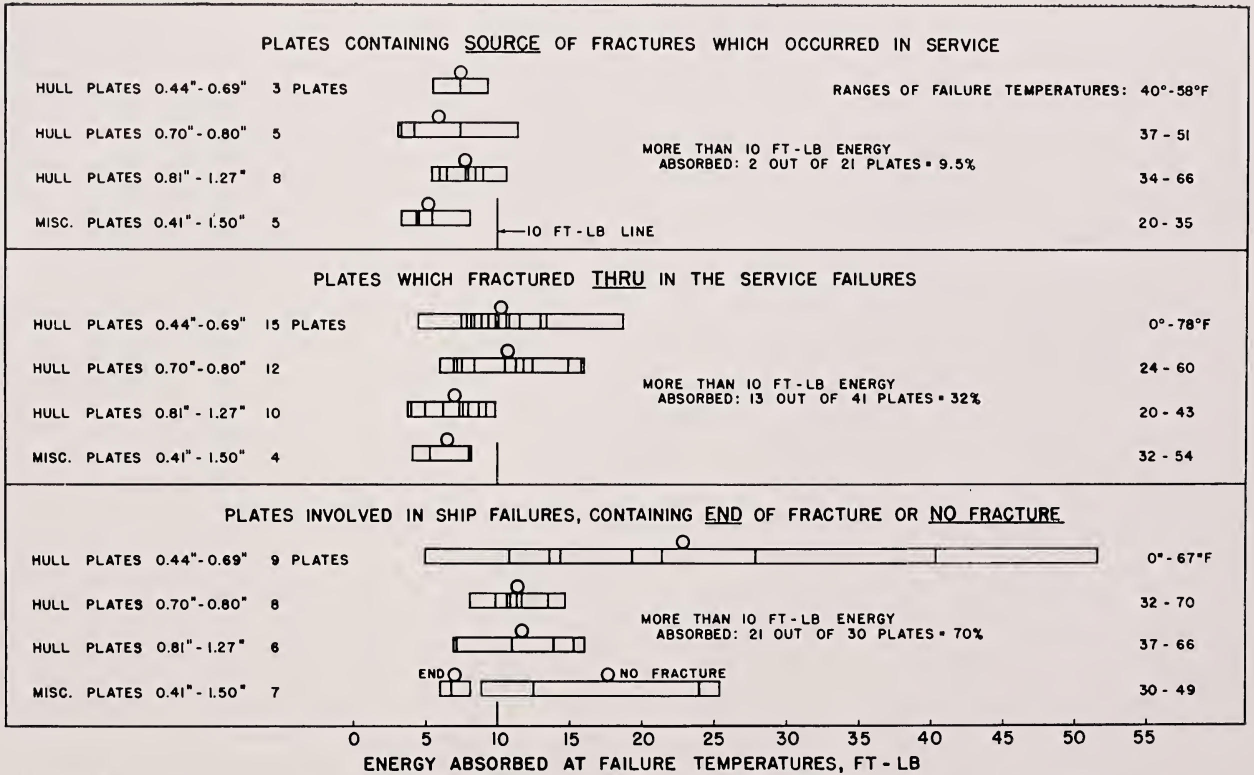

FIGURE 9.8. *Relation of energy absorbed by Charpy V-notch specimens at the temperatures of the ship failures to the nature of the fractures that occurred in service.*

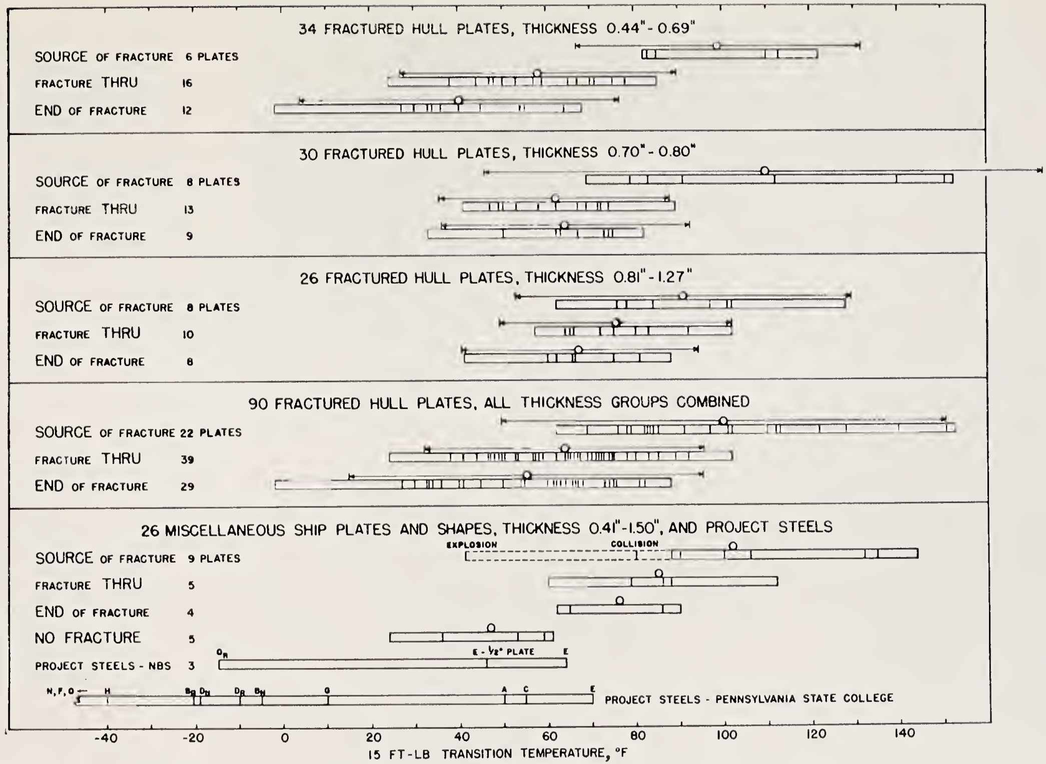

FIGURE 9.9. *Relation of 15 foot-pound transition temperature to the nature of the fractures in ship plates.*

parts other than the hull structure, plates in which the damage resulted from collision or explosion, plates in which the nature of the fracture was doubtful, or plates that contained no fracture, although fractures occurred in adjoining plates or welds. The 15 ft-lb transition temperatures of some of the Ship Structure Committee "Project Steels," are also shown for comparison.

A relation between the plate thickness and the transition temperatures of the plates is shown in the upper three sections of the figure. The thinner plates (top) show generally lower transition temperatures, and a greater spread of values than the thicker plates. This indicates that the notch toughness of a steel may be improved by the lower finishing temperature or the greater amount of hot-working, or other factors involved in rolling to thinner dimensions. However, the greater spread of values for the thinner plates suggests that the notch sensitivity is influenced also by other variables in the rolling process, such as reduction per pass or rate of cooling, since these variations would naturally be greater for thinner plates.

In each thickness group, and in the miscellaneous group, the average transition temperature of the plates in the "fracture source" category is higher than the average values for plates which did not contain the source of a fracture in the ship failures. Plates in which the ship fractures ended show (with one exception) lower average transition temperatures than plates in the fracture-through category, although this difference is not as great. The same consistent relationship is indicated also by both the upper and lower limits of the ranges of observed values of the transition temperatures, and the ranges of the scatter bands defined by the average value plus or minus twice the standard deviation.

Among the fractured hull plates, for the three thickness groups combined, only 2 out of 22, or 9 percent, of the plates in which fractures originated had transition temperatures below 70° F, the lowest being 62° F. In both of these plates, the fracture originated at an arc crater—a metallurgical notch. The transition temperatures of 64 percent of the fracture-through plates and of 76 percent of the fracture-end plates were below 70° F. In the miscellaneous group, the trend is similar. None of the fracture-source plates, except one damaged by an explosion, 20 percent of the fracture-through category, and 78 percent of the plates containing the end of a fracture or no fracture had transition temperatures below 70° F.

This evidence shows a very definite difference of the notch sensitivity of the plates in the different categories, which may be interpreted as follows: The plates in which fractures originated were, so to speak, "selected" because of lower than average notch toughness at critical locations in the structure. The fact that the majority of the ships of the same design and construction did not fracture, and that the fractures originated in different locations in the ships, indicates that this low notch toughness was a borderline deficiency of the steel. The plates in the "through" and "end" categories were "run-of-the-mill" plates that happened to be in the path of the fracture as it propagated. In some cases, higher than average notch toughness of the plates (that is, greater ability to absorb energy) contributed to the halting of the fractures, although there were many other contributing factors, such as dissipation of elastic energy, changes in stress levels due to transient loads, readjustment of stress distribution, or the effect of nearby structural reinforcements.

It appears reasonable to believe that plates of the quality used in ship construction would show a distribution of notch sensitivity similar to the familiar Gaussian probability curve, that is, the great majority of the plates would have notch-sensitivity values near the normal or average for the type of material under consideration, and relatively few would have abnormally high or low values. Thus there is little probability that a small sample taken at random would include any plates with abnormally high or low values. This is shown by the ship plates which did not contain fracture sources and by samples taken from steel mills and from shipyard stocks [2]—these relatively small random samples did not include any steels with transition temperatures as high as those found for some "fracture-source" plates. However, if a much larger sample were taken, it is probable that such steels would be found. The fracture-source plates were not taken at random, but were "selected" because they contained the origin of a fracture. This selection was from a large sample (all of the plates in service at numerous critical locations in some 5,000 ships), and consequently the probability of finding plates with the abnormally high transition temperatures was much greater.

In figure 9.10 this interpretation is tested by analyzing the frequency distributions of the observed 15 ft-lb transition temperatures of the plates in each of the categories. The vertical bars indicate the number of plates (left-hand scale) or the percentage of the total number of plates in each category (right-hand scale) in each 15° F interval of transition temperatures. Fractured hull plates are represented by the center-lined portion of the bars, and the open parts of the bars represent the additional plates of the miscellaneous group. The average transition temperature of all plates in each category is indicated by the circles, and the arrows show the range of plus or minus one, two, or three times the standard deviation. The symbol × indicates the average for the hull plates alone. The essential data are given in table 9.1.

TABLE 9.1. *Comparison of average transition temperatures of all plates in the source, through, and end categories*

Nature of fracture	Number of plates = N	Average 15 ft-lb transition temp = $\bar{X}$	Standard deviation $\sigma = \sqrt{\Sigma(\bar{X}-X)^2/N}$
		° *F*	° *F*
Source	30	102.5	25.0
Through	45	65.7	17.5
End	38	56.3	20.1

The distribution curve for the fracture-through plates is nearly symmetrical, and closely approximates the normal curve (dashed line) computed from the values of the average and standard deviation. The curve for the fracture-source plates (top) is decidedly unsymmetrical, and the average transition temperature is 37 deg F higher than the average for the fracture-through plates. The distribution of the plates containing the end of a fracture or no fracture is also

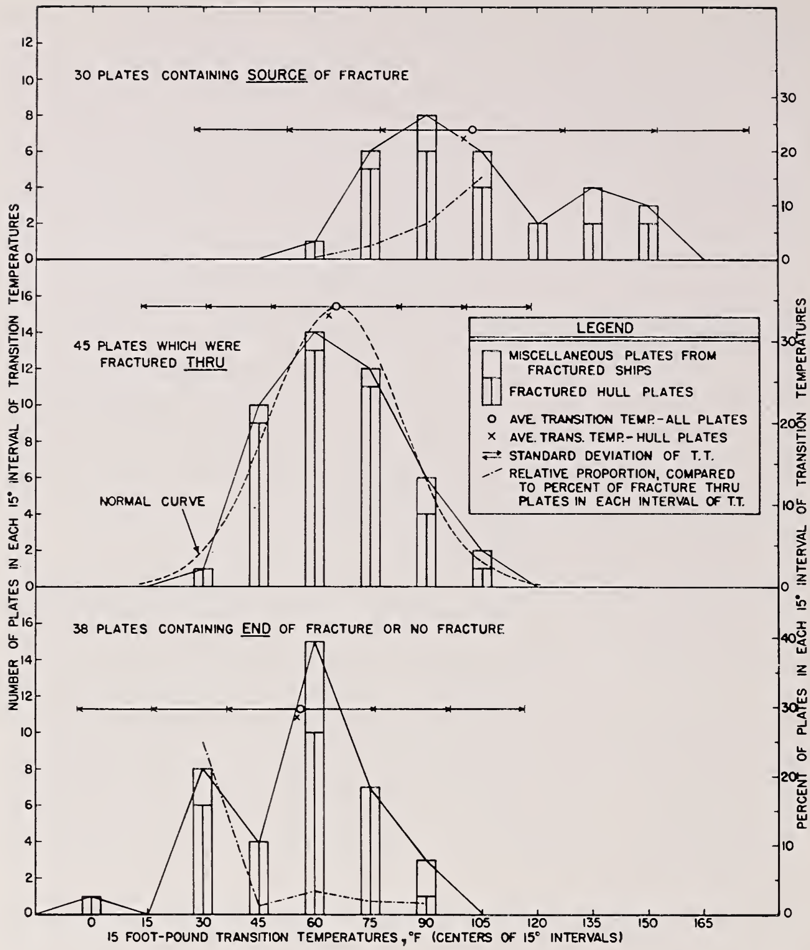

FIGURE 9.10 *Frequency distribution of 15 foot-pound transition temperatures of plates in the fracture source, fracture-through, and fracture-end categories.*

somewhat unsymmetrical, and the average transition temperature is 9 deg F lower than the average for the fracture-through plates.

Twenty percent of the plates in the fracture-source category had transition temperatures in the 75-deg interval, which is only a few degrees above the average transition temperature of the fracture-through plates. This might appear to indicate that there is no sharp demarcation between the transition temperatures of the run-of-the-mill plates and those that contained a fracture source. However, we must consider that these fracture-source plates were selected from a population in which the distribution of transition temperatures was not uniform. For example, in the general population (represented approximately by the sample of fracture-through plates) the 75-deg interval included about six times as many plates as the 105-deg interval, but these two intervals included equal numbers of fracture-source plates.

The dot-dash line in the upper part of the figure shows the relative proportion of fracture-source plates to fracture-through plates in each interval of transition temperatures. This curve indicates that the probability of a fracture starting in a plate, under the conditions existing in a structure such as a ship, increases markedly with increasing notch sensitivity of the plate, as determined by the 15 ft-lb transition temperature of V-notch Charpy specimens.

The frequency-distribution curves also show that the plates with abnormally high transition temperatures, in which fractures are most likely to originate, represent the relatively few plates whose notch sensitivity falls in the tail of the probability curve for steels of the quality used when these ships were built—the borderline deficiency previously mentioned. This suggests two possible remedies (1) improvement of the average quality of the steel with respect to notch sensitivity. (In effect, moving the entire probability curve in the direction of increased notch toughness, so that a much smaller proportion of the plates in the tail of the curve would fall beyond acceptable limits of notch sensitivity); (2) determination of the notch sensitivity of every heat of steel by inspection tests, and rejection of all heats that fail to meet suitable prescribed standards. (In effect, "cutting off the tail" of the probability curve.)

In the application of these remedies, however, due consideration must be given to the possibility that the notch sensitivity of the steel as fabricated in the ship may be considerably greater than the notch sensitivity as determined by mill tests of the prime plate. It has been noted that all of the fractures originating in plates were associated with the effects of welding, flame cutting, or previous deformation of the plate. Notched-bar tests on a number of the fractured plates, and investigations in other laboratories, have indicated that the notch sensitivity is increased locally as a result of these operations. There is evidence also that this effect might be greater in some steels than in others, but more knowledge is needed in this field.

Relation of Notch Sensitivity to Chemical Composition

To determine the effect of chemical composition on the notch sensitivity of the steels, scatter diagrams similar to figures 9.11 and 9.12 were plotted for each chemical element. The content of the element under consideration was plotted against the 15 ft-lb transition temperature, for the individual plates in each thickness group, using different symbols to indicate the nature of the fractures that occurred in the plates in service. The vertical bars at the left represent the range of element content for plates in the source, through, and end categories, respectively, and the symbols within the bars indicate the average values.

Carbon (figure 9.11) has a pronounced effect on both the nature of the fracture and the transition temperature. In each thickness group, the range and the average values of carbon content of plates in the fracture-source category are higher than for plates that did not contain a fracture source. Only one fracture-source plate (in all thickness groups) had a carbon content as low as 0.20 percent, whereas the carbon was at least this low in 28 percent of the fracture-through

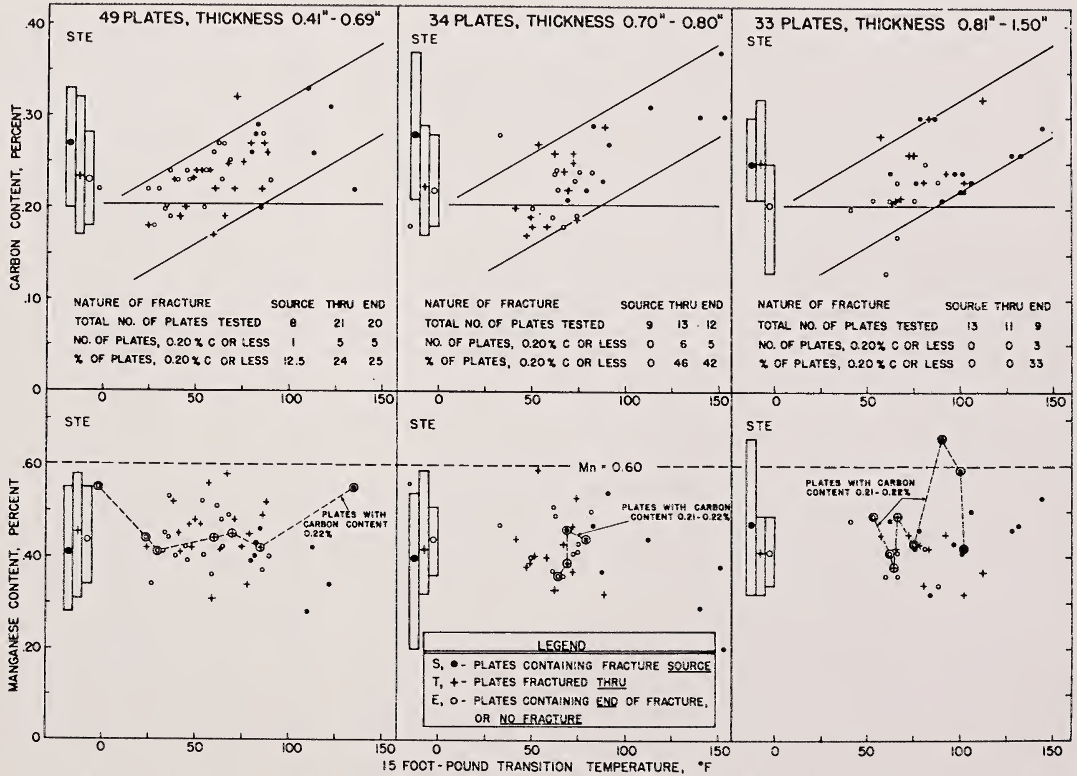

FIGURE 9.11. *Relation of carbon and manganese content to nature of fractures and to 15 foot-pound transition temperatures.*

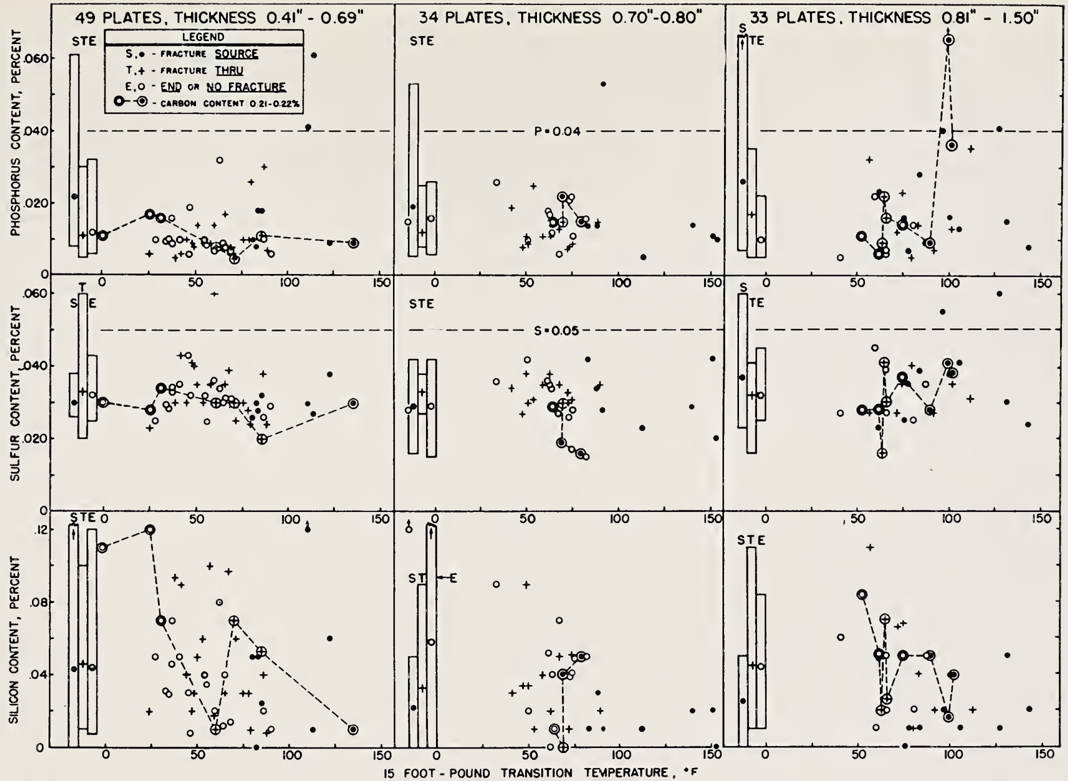

FIGURE 9.12. *Relation of phosphorus, sulfur, and silicon content to nature of fractures and to 15 foot-pound transition temperatures.*

and fracture-end plates. In the scatter diagrams, the majority of the points fall within a broad band (sloping lines), showing a general increase of the transition temperature with increasing carbon content.

Manganese content does not show any consistent relationship to either the nature of the fractures or the transition temperatures of the plates, even when plates of the same carbon content are compared, as indicated by the dashed lines.

Figure 9.12 shows the effects of phosphorus, sulfur, and silicon. Plates with high phosphorus generally had higher transition temperatures, although the trend is not very evident, as high transition temperatures were also found for a number of plates with low phosphorus. The relation to the nature of the fractures is more apparent, as in all thickness groups the plates with high phosphorus were in the fracture source category. This is probably related to the increase of notch sensitivity in the vicinity of welds, which other investigators [10] have shown is greater in steels with high phosphorus content.

Most of the plates had a normal sulfur content of 0.02 to 0.04 percent, and in this range sulfur apparently has little or no effect.

Silicon content, in the range zero to 0.12 percent, affects the transition temperature and the nature of fracture in the opposite direction to the effect of carbon. The transition temperatures were generally lower for the plates with higher silicon, and with the exception of one plate, the range and average values of silicon content were lower for the fracture source plates than for the plates which did not contain a fracture source.

Similar plots were made for each of the other elements determined in the chemical analysis, and for the gases. No definite relations could be observed between the content of any of these elements, considered individually, and the notch sensitivity as indicated by either the transition temperature or the nature of the fractures. This suggests that the effects of carbon, phosphorus, and silicon are considerably greater than the effects of other elements, in the amounts in steels of this type. However, many of these other elements were present only in small amounts, or, as in the case of manganese, in limited ranges, in the group of steels under consideration. Thus, it is possible that the effects of the relatively small variations of these other elements might be obscured by the effects of carbon, phosphorus, and silicon.

An attempt was made to find the effects of these other elements by a method of successive elimination of variables, assuming, provisionally, that the effects of the individual elements might be additive. Tentative correction factors for the effects of carbon and silicon, estimated from the scatter diagrams for these elements, were added to the observed transition temperatures of the plates in each thickness group, and the content of each element was plotted against the corrected transition temperature. With the variations due to carbon and silicon at least partially eliminated, the effects of other elements could be seen, and tentative correction factors for these elements were also added. This process was repeated several times, and at each step the estimated correction factors for each element were checked by the slope of the scatter diagrams, and modified if necessary. As it is well known that the transition temperature of a steel is affected by the grain size, a correction factor for the fracture grain size was determined in a similar manner.

The corrections for the content of the chemical elements reduced the spread of the corrected transition temperatures of the plates in each thickness group, so that most of them fell within a band of plus or minus 30 deg. F, but the positions of these bands were different for the three thickness groups. The addition of the correction for grain size practically eliminated the differences between the thickness groups, indicating that the generally higher transition temperatures of the thicker plates were due primarily to larger grain size.

The corrected transition temperatures of about 95 percent of the ship plates tested, in all thickness groups, now fell within the same plus or minus 30-deg band, or stated in mathematical terms: 15 ft-lb transition temperature (°F) + correction factors for chemical composition and grain size = constant ± 30° F (for 95 percent of the plates tested). Most of the plates that had corrected transition temperatures higher than this band had been severely deformed in fabrication or incident to the casualty, and the high transition temperatures of these plates may probably be attributed to the deformation or to subsequent strain aging.

This looks like a formula for predicting the transition temperature of a steel, but the assumed additive relation on which the method was based had not been proved, and the results were applicable only in the limited range of compositions of the ship plates. The correction factors could not be determined accurately for some of the elements that were present only in small amounts, but the results obtained appeared to indicate that the transition temperatures of the steels were increased with increasing content of carbon, phosphorus, molybdenum, and arsenic, and decreased with finer grain size and with increasing amounts of manganese, silicon, copper, and nickel.

While this work was still in progress, results of an investigation at the Naval Research Laboratory became available. The investigators [11] determined the effects of various elements on the transition temperatures of controlled laboratory heats of steel in which only one element was varied, and also showed that the effects of carbon, manganese, and nickel are very nearly additive.

In the range of compositions covered by both investigations, the results obtained by "working backward" from the compositions of the fractured ship plates were in good agreement with the results for the 15 ft-lb transition temperature obtained at NRL by the more straightforward laboratory method. This appears to validate the assumption of an additive relation for the effects of the other elements. However, the two investigations did not agree on the effects of copper and silicon, which illustrates the dangers of optimistic extrapolation. In the low ranges of copper and silicon content found in the ship plates, these elements appear to lower the transition temperature, but in the higher ranges covered by the NRL investigation, the transition temperature is raised by further additions of these elements.

Further investigations at NRL, and similar research at another laboratory [12], using different specimen types, have shown that the notch sensitivity is first lowered, and then raised, as the silicon content is increased. The effects of some other elements also change, either in magnitude or direction, as the content of the element is increased, and there may also be interactions between certain elements that do not show up when only a single element is varied. The work at NRL also showed that some elements, such as carbon, change the shape of the transition curve for Charpy V-notch specimens, so that the effects of these elements are not the same for different criteria of the transition temperature. All of these factors must be considered in evaluating the effects of composition on the notch sensitivity of hot-rolled steels.

Using the published data from the NRL investigation, combined with the ship-plate data, better values for the correction factors were obtained. A tentative formula including 11 correction factors, for 8 chemical elements and the grain size, was tried out on more than 200 steels, covering a 400-deg range of observed 15 ft-lb transition temperatures. The calculated transition temperatures of more than 90 percent of these steels were within ±30 deg of the observed transition temperatures, but it is hoped that this formula can be improved, and it will be necessary to try it on a larger variety of steels to determine its limitations.

A simpler short form of the formula would probably be more useful in the range of compositions of the usual hot-rolled structural or ship-plate steels. Nine-five percent of the 113 ship plates, and a number of other steels of similar composition, had transition temperatures less than the maximum indicated by the formula

$$\text{(Max) 15 ft-lb transition temperature } (^\circ\text{F}) = 100 + 300 \times \%\ \text{C} + 1000 \times \%\ \text{P} - 100 \times \%\ \text{Mn} - 300 \times \%\ \text{Si} - 5 \times \text{fracture grain-size number}$$

This formula, however, is not applicable for compositions containing more than about 0.35 percent of carbon, 0.10 percent of phosphorus, 0.25 percent of silicon, 0.20 percent of copper, or 0.2 percent of molybdenum, chromium, and arsenic combined, inasmuch as such amounts of these elements may raise the transition temperature above the limit indicated. Residual copper in amounts less than 0.20 percent, and nickel in any amount, are beneficial, and need not be considered in this calculation of the maximum transition temperature to be expected. Other elements generally are present in such small amounts in steels of this type that their effects appear to be negligible.

This formula shows that the notch toughness of hot-rolled steels may be improved by reducing the carbon or phosphorus content, by increasing the manganese or silicon content, or by mill practices that produce a finer grain size. It also shows that silicon, in amounts up to about 0.25 percent, is three times as effective as manganese, which may be important if the manganese situation gets worse.

The metallurgical effects of welding and flame cutting, and the increase of notch sensitivity resulting from deformation or strain aging, must also be considered, and we need to know more about the relation of these phenomena to the composition of the steels. However, experience shows that these effects will be less serious in steels of lower notch sensitivity than in steels that are more notch sensitive to begin with.

Summary

Structural failures of welded merchant ships have cost the Nation almost 50 million dollars in the past 9 years, and there have been other costly failures in such structures as bridges and storage tanks. The failures usually occurred at low temperatures, and the origin of

the fractures could be traced, invariably, to a point of stress concentration at a geometrical or metallurgical notch resulting from design details or welding defects. The fractures in the ships generally were of a brittle type, showing little evidence of ductility, although the steels showed normal ductility in the usual type of tensile test. The principal factors contributing to these failures—workmanship, design, and materials—and the preventive measures that have been applied are discussed. The incidence of failures in ships has been reduced materially by improvements of design details and of welding workmanship, but it is evident that stress concentrations resulting from unavoidable structural notches, welding defects, or conditions incident to operation and maintenance can never be entirely eliminated. Therefore, the quality of the steel, especially with regard to notch sensitivity, must be regarded as an important factor in the prevention of failures.

Charpy V-notch bar tests of 113 plates removed from fractured ships show that the plates in which the fractures originated were generally more notch sensitive than plates that did not contain a fracture source. Statistical interpretation of the data indicates that under the conditions existing in a structure such as a ship, the probability that a fracture will originate in a plate increases markedly with increasing notch sensitivity of the steel, as measured by the 15 ft-lb transition temperature of Charpy V-notch specimens. Relations between the 15 ft-lb transition temperatures of the plates and their chemical composition show that the notch sensitivity is increased with increasing amounts of carbon and phosphorus, and decreased with finer grain size and with increasing amounts of silicon and manganese, within the range of the chemical compositions of the ship plates.

Investigations of failures in welded ships have been conducted at the National Bureau of Standards over a period of several years, starting early in 1943. Until July 1, 1946, the examinations of fractured ship plates were carried on as a part of an extended research program sponsored by the Bureau of Ships, Navy Department. When that program was terminated, the work was continued by the National Bureau of Standards as a fundamental part of a study of the nature of fracture and of fracture propagation in metals. Since July 1947 the Ship Structure Committee has sponsored the investigations of failures in welded ships, and this work has been conducted by the same personnel and closely coordinated with the work on the nature of fracture in metals.

The early part of this investigation was conducted under the supervision of George A. Ellinger, who has continued to show an active interest in the project and has given valuable counsel throughout the investigation. Other present and former members of the Metallurgy Division of the National Bureau of Standards who have contributed to the work are Melvin R. Meyerson, Gordon L. Kluge, Leo R. Dale, James T. Sterling, H. G. MacKerrow, and Lura F. Roehl.

Much credit is due also the U. S. Coast Guard inspectors, who were responsible for the selection and marking of the samples of fractured plates, and to members of the Ship Structure Committee and of the Merchant Marine Technical Division, U. S. Coast Guard Headquarters,

who supplied information regarding the structural features of the ships and the circumstances of the failures, and assisted in many other ways.

Discussion

Mr. J. J. Kanter, Directing Engineer, Engineering Division, Crane Co., Chicago, Ill.: Has any attempt been made to correlate the transition temperature with the aluminum content of the plates? Did you make any attempt to distinguish between soluble and total aluminum?

Dr. M. L. Williams: Aluminum was plotted in the same manner as the other elements, but we did not find any consistent relation between the transition temperature and the aluminum content. The chemical analyses of the plates did not distinguish between soluble and total aluminum.

Mr. S. Tour, General Manager, Sam Tour & Co., Inc., New York, N. Y.: Dr. Williams has chosen 15 foot-pounds as the line of demarcation between a brittle condition and a tough condition for ship plates. This is an arbitrary line and not a transition line. In fact, it has no relationship to transition temperatures. It is possible to have steel that will have this 15 foot-pound notched-bar Charpy impact strength below its transition temperature. On the other hand it is possible to have steel that, even above its transition temperature, will have less than 15 foot-pound Charpy notched-bar impact strength. What justification is there for this arbitrary selection of 15 foot-pounds as a criterion of anything? Is crack propagation dependent solely on notched bar impact strength and independent of whether the fracture is of the brittle or nonbrittle type?

Dr. Williams has referred to defective welds as being the point of initiation of certain of the brittle fractures that have occurred in ship plates. That brittle fractures of great length can develop in relatively soft steel has been demonstrated in the case of certain gas pipelines. That brittle type fractures can travel through several plates for a distance of 100 or 200 feet has been demonstrated in ship plates. The longest crack of this nature that has come to the writer's knowledge did not occur in a ship. It developed in a 30-inch-diameter pipeline laid in a trench and traveled a total of 3,300 feet. It did not originate in a defective weld, it traveled longitudinally of the pipe, through more than 100 circumferential welds and was not deflected by such welds. There is no indication that a steel with a 15 foot-pound notched-bar Charpy impact strength would be satisfactory for a gas pipe or for a ship plate.

Designs should be based on stress levels and on length of steel involved. In the case of pressure pipelines the lengths involved are the circumferences involved. At a given stress level when failure is initiated in a 30-inch-diameter pipe, the crack will progress farther than it will in a 15-inch-diameter pipe stressed to the same degree. The larger the length of steel in which elastic energy may be stored the greater the tendency to brittle fractures. In other words, the notched bar impact strength necessary to stop or retard the propagation of brittle fractures must be higher for large structures than for small structures. No arbitrary figure such as 15 foot-pounds may be selected for all structures.

Dr. Williams: The subjects of notch sensitivity and transition temperatures are discussed more fully in the written paper, but in the condensation of this paper for oral presentation it was necessary to omit part of this discussion.

It might be well to emphasize here that the 15 foot-pound value was chosen as a means of comparing all plates at the same energy level, and not as a line of demarcation between tough and brittle behavior. It is true that the temperature at which the energy absorption is 15 foot-pounds (for V-notch Charpy specimens) is not strictly a transition temperature for all types of steels, and perhaps it might have been better to call it by some other name. However, for steels of ship plate quality, the 15 foot-pound level lies within the transition range, and the temperature at which the notched-bar test curve crosses the 15 foot-pound line provides a convenient and easily determined numerical value which is one of the many possible indices of notch sensitivity. The value of 15 foot-pounds was chosen arbitrarily because it was the minimum acceptable value specified for notched bar tests under certain of the ASME pressure vessel codes, and because it appeared to be a reasonable value based on experience in the field.

The justification of this tentative selection of the 15 foot-pound transition temperature as an index of notch sensitivity lies in the fact that it does provide a reasonably good correlation between laboratory tests and service experience. Similar correlations are shown also for the other indices of notch sensitivity which were used: the energy absorbed by Charpy V-notch specimens at the failure temperatures of the respective plates or at fixed temperatures of 30° or 70° F. Possibly the correlation might be even better at a different fixed temperature, say 40° F, or at a different energy level, for example, the 10 foot-pound transition temperature, and further work in this direction is contemplated.

From these correlations with service experience, using any one of these indices, acceptable limits of notch sensitivity may be determined for steels intended for service under similar conditions. For other service conditions, such as in pipelines or pressure vessels of different sizes, it would be necessary to establish similar correlations with service experience, and the acceptable limits of notch sensitivity would probably be different. For example, the accumulated data for ship plates indicate that steels having a 15 foot-pound transition temperature somewhat higher than the lowest operating temperatures (that is, steels capable of absorbing somewhat less than 15 foot-pounds at the operating temperatures) would be acceptable. To maintain the same energy absorption at lower operating temperatures, a correspondingly lower transition temperature would be required. For more severe service conditions, more than 15 foot-pound energy absorption at the operating temperature might be found necessary, which would require a 15 foot-pound transition temperature lower than the operating temperature.

Mr. Tour's comments on the failures in gas pipelines are appreciated. It would be interesting and instructive to compare the operating conditions, circumstances of the failures, and the properties and chemical compositions of the steels involved with the similar data that have been obtained in connection with the ship failures.

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