Creep of Annealed Nickel, Copper, and Two Nickel-Copper Alloys

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Creep tests were made in tension under constant loads at temperatures of 300°, 700°, and 900° F on initially annealed specimens of nickel, copper, and 70-percent-nickel–30-percent-copper and 30-percent-nickel–70-percent-copper alloys. Tests at 1,200° F were also made on the nickel and the two alloys, but not on the annealed copper as the resistance to creep of copper is relatively low at this temperature. The investigation included a study of the influence of rate of loading on the creep stress and of prior thermal-mechanical history on the creep behavior of the alloys at several selected temperatures. Contour and hardness surveys and metallographic examinations were also carried out on some of the fractured specimens to ascertain the effect of creep on the necking characteristics, hardnesses, and structures of the metals.

1. Introduction

The effects of temperature on the short-time tensile and some other properties of the same lots of nickel, copper, and nickel-copper alloys initially in the annealed and cold-drawn condition were discussed in previous publications [1 to 7].[1] Parker and Hazlett [8] made a comprehensive survey of the literature on the principles of solution hardening in relation to the tensile properties of nickel-copper alloys. Hazlett and Hansen [9] showed that the shape of the creep curves for nickel extended in tension was markedly altered by either prestraining or alloying. French and Hibbard [10] discussed the changes in lattice parameters and mechanical properties of copper introduced by alloying with nickel.

According to Barrett [11] nickel and copper form a continuous series of substitutional solid solutions. All alloys of the system consist of a face-centered-cubic lattice with copper and nickel atoms distributed at random on the lattice points. Nickel has nearly the same atomic size as copper and the presence of nickel atoms alter only slightly the copper lattice. However, Averbach [12] pointed out that recent researches show that solid solutions are not random; that the atoms are displaced from their lattice sites; and that the atoms neither retain their pure metal sizes nor assume the average size calculated from the alloy lattice parameter. There is considerable uncertainty concerning the role of strain energy in solutions containing atoms of disparate sizes; the solutions, however, may be randomized by plastic flow.

The present study was made of the creep characteristics of nickel, copper, and two alloys of these metals as a part of a comprehensive investigation designed to evaluate the rheological properties of samples of these metals at temperatures ranging from −320° to +1,700° F. Special attention was directed towards an evaluation of the resistance to creep of the alloys in relation to that of the component metals.

[1] Figures in brackets indicate the literature reference at the end of this paper.

2. Metals, Apparatus, and Procedure

The chemical composition and average grain size of the metals used in this investigation are shown in table 1. All the bars of each metal were processed from a single heat. The creep specimens of each metal were machined from one annealed bar to an 0.505-in. diameter over a 2-in. gage length. The preparation of the specimens and the apparatus used were described previously [1]. Essentially, each creep specimen was heated in air to the desired temperature and held at temperature 48 hr before loading. Load increments, each equivalent to a stress value of 5,333 lb/in.^{2}, were applied at 1-hr intervals (hereafter designated as the "standard" rate of loading) until the desired stress was attained. This standard loading procedure was modified in those tests made to determine the influence of prior-strain history on the subsequent creep behavior. The prior history and procedures used in this series of experiments are described in tables 5 and 6.

TABLE 1. *Chemical composition (percentage by weight) of the metals and alloys used as determined by chemical, spectrochemical, and vacuum fusion analyses*

Metal	C	Cu	Ni	Co	Fe	Mn	Si	S	Zn	O_2	N_2	H_2	Average [c] grain diameter
													mm
Copper [a] (OFHC)	------	99.99+	------	----	----	----	------	------	----	------	------	------	0.025
30%-Ni–70%-Cu	0.023	68.84	29.89	0.04	0.50	0.65	0.003	0.004	0.09	0.001	0.001	0.0002	.040
70%-Ni–30%-Cu	.017	29.71	70.08	([b])	.01	.01	.12	.002	----	.001	.0015	.0003	.025
Nickel	.007	0.009	99.85	([b])	.04	.03	.11	.002	----	.002	.001	.0002	.045

[a] The arc spectrum of the copper was examined for the sensitive lines of Ag, Al, B, Be, Co, Fe, In, Ir, Mg, Mo, Na, Ni, Pb, Sb, Si, Sn, Ti, V, and Zn. The lines for Ag, Al, Mg, and Si were identified and there was some indication of the presence of Fe, Ni, and Pb.
[b] Not detected.
[c] Values obtained on specimens prepared from the annealed bars.

The temperatures of the creep furnaces were controlled within ±1° F of the desired temperatures over the specimen length and the probable error in measuring the extensions was less than 0.00002 in. Specimen contours were determined by measuring the diameter of the specimens at various distances from a fractured end. Rockwell hardness determinations were made along the longitudinal axis of specimens prepared by the usual procedures for making these measurements [1].

3. Results and Discussion

The results of creep tests made on specimens loaded at the standard rate are summarized in figures 1 to 20, and in tables 2, 3, and 4. Data obtained on specimens tested to determine the effect of prior-strain history on the creep properties are presented in figures 21 to 27 and in tables 5 and 6. Values for the copper at 110° and 300° F and for the nickel are not tabulated, as they were presented previously [1, 2, and 5].

3.1. Influence of Temperature and Stress on Creep Behavior of Specimens Loaded at the Standard Rate

a. Stress-Strain Relations During Loading

The effect of temperature on the stress-strain relations during loading is shown in figure 1. At each temperature, the general shape of the family of curves show that, after the start of deformation, the incremental changes of strain increased with each successive increment of stress. These values also increased with an increase in temperature. Thus, it is apparent that the strain hardening becomes less pronounced as the temperature is raised.

At each of the temperatures and at equal strains the strengths of the alloys were greater than those of the component metals. Furthermore, the differences in the rate of strain hardening between the component metals and the alloys were strongly dependent on temperature. A strengthening effect due to small strains associated with alloying the component materials has been described by Fisher [13].

b. Effect of Temperature and Composition on Strain-Time and Creep Rate-Strain Relations

The shape of the creep curves for all the specimens used in this investigation was, with a few exceptions, similar to those shown previously for copper [1, 2] and nickel [5].

Some of the strain-time curves of specimens of different compositions tested under identical conditions at 700°, 900°, and 1,200° F are shown in figures 2, 3, and 4. Due to differences in strengths of the metals, no identical tests were run at 300° F. It is apparent that the resistance to deformation (fig. 2) of the alloys was greater than that of the pure nickel. Furthermore, the alloy containing 30 percent of copper was more creep resistant at all values of time than the alloy containing 70 percent of copper. Other examples of the effect of alloying the nickel with 70 percent of copper are shown in figure 3 for specimens tested at 700° and 900° F. These data indicate that the strengthening effects due to alloying are more pronounced as the time at stress is

Table 2. *Summary of conditions used and results of creep tests on 70-percent-nickel–30-percent-copper alloy, initially as annealed*

Test		Creep stress	Plastic strain		Average creep rate, second stage	Beginning of third stage			End of test					Remarks
Number	Temperature		1 hr after application of load	Intercept		Time	Plastic strain	True stress	Time	Plastic strain or elongation	Contraction of area or reduction of area	$\log_e (A_0/A)$ at fracture	True stress	
	°F	*lb/in.²*	*Percent*	*Percent*	*%/1,000 hr.*	*hr*	*Percent*	*lb/in.²*	*hr*	*Percent*	*Percent*		*lb/in.²*	
B–11	300	56,030	15.73	16.04	0	------	----	----	2,202	16.04	13.9	----	65,000	Test stopped in second stage.
B–12	300	58,670	20.28	20.37	0	------	----	----	1,699	20.37	17.0	----	70,600	Do.
B–15	300	61,000	27.03	27.1	0.006	------	----	----	2,036	27.11	21.3	----	77,500	Do.
B–17	300	61,840	27.59	27.61	0	------	----	----	1,871	27.61	21.6	----	78,900	Do.
B–18	300	62,500	----	30.75	220,000	0.03	39.5	87,200	0.043	55.5	85.4	1.925	427,000	Tested to complete fracture.
B–1	700	37,330	3.97	5.83	0.12	------	----	----	5,034	6.45	6.1	----	39,800	Test stopped in second stage.
B–6	700	42,670	6.74	8.27	.52	------	----	----	3,357	10.39	9.4	----	47,100	Do.
B–9	700	48,000	10.18	16.22	1.09	------	----	----	3,051	19.61	16.4	----	57,400	Do.
B–13	700	56,030	21.20	21.14	59.0	5.1	21.6	68,100	243	60.5	74.2	1.356	217,300	Tested to complete fracture.
B–14	900	18,670	0.26	0.523	0.61	------	----	----	2,468	2.23	2.15	----	19,100	Test stopped in second stage.
B–4	900	21,330	.31	1.32	1.4	2,200	4.4	22,300	9,927	53	48.6	0.666	41,500	Tested to complete fracture.
B–2	900	26,670	2.11	3.90	11.1	------	----	----	408	8.42	7.8	----	28,900	Test stopped in second stage.
B–7	900	28,000	3.02	5.75	14.3	650	15.0	32,200	1,696	54.5	45.3	.604	51,200	Tested to complete fracture.
B–10	900	32,000	5.03	8.5	51.8	315	24.5	39,800	480	56.5	47.3	.641	60,800	Do.
B–16	1,200	2,665	0	0.70	0.34	------	----	----	1,414.5	1.18	1.16	----	2,700	Test stopped in second stage.
B–8	1,200	5,330	0.09	5.07	6.6	2,200	19.6	5,800	4,048	43.5	31.7	.381	7,800	Tested to complete fracture.
B–5	1,200	8,000	.186	0.18	41	68	2.9	8,200	643	61.5	47.9	.651	15,300	Do.
B–3	1,200	10,670	.236	.12	125	23	3.0	11,000	165	61.3	54.6	.790	23,500	Do.

TABLE 3. *Summary of conditions used and results of creep tests on 30-percent-nickel–70-percent-copper alloy, initially as annealed*

Test		Creep stress	Plastic strain		Average creep rate, second stage	Beginning of third stage			End of test					Remarks
Number	Temperature		1 hr after application of load	Intercept		Time	Plastic strain	True stress	Time	Plastic strain or elongation	Contraction of area or reduction of area	$\log_e (A_0/A)$ at fracture	True stress	
	° F	*lb/in.²*	*Percent*	*Percent*	*%/1,000 hr.*	*hr*	*Percent*	*lb/in.²*	*hr*	*Percent*	*Percent*		*lb/in.²*	
C–24	300	37,330	7.71	7.74	0	--------	--------	--------	1,507	7.74	7.2	--------	40,200	Test stopped in second stage.
C–6	300	42,670	13.07	13.16	0	--------	--------	--------	2,370	13.16	11.7	--------	48,300	Do.
C–11	300	46,000	19.69	19.84	0	--------	--------	--------	2,182	19.84	16.5	--------	55,200	Do.
C–15	300	47,000	20.18	20.42	0	--------	--------	--------	2,340	20.42	16.9	--------	56,600	Do.
C–23	300	47,200	20.6	20.7	0	--------	--------	--------	1,484	20.7	17.2	--------	57,000	Do.
C–22	300	47,350	23.57	23.45	510	--------	--------	--------	13	30.5	23.4	--------	61,800	Do.
C–26	300	47,650	47.0	23.1	6000	0.67	27.1	60,600	1	47.0	80.1	1.612	238,900	Tested to complete fracture.
C–19	700	26,670	3.58	6.43	0.27	--------	--------	--------	3,330	7.73	6.8	--------	28,600	Test stopped in second stage.
C–4	700	32,000	6.39	11.35	3.63	4,020	26	40,300	5,346	46.0	51.7	0.728	66,300	Tested to complete fracture.
C–20	700	34,670	8.49	15.2	33.3	490	31.5	45,600	692	56.0	65.0	1.045	98,600	Do.
C–10	700	37,330	11.92	16.0	83.0	200	32.6	49,500	250	53.0	66.5	1.092	111,300	Do.
C–16	700	40,000	15.13	18.0	290	35	28.2	51,300	56	59.5	71.5	1.253	140,000	Do.
C–21	900	10,670	0.034	0.11	0.19	--------	--------	--------	2,345	0.57	0.57	--------	10,700	Test stopped in second stage.
C–17	900	13,335	.21	.8	2.1	1,800	4.6	13,900	3,860	10.12	9.2	--------	14,700	Test stopped in third stage.
C–14	900	16,000	.91	2.6	2.6	1,310	6.0	17,000	6,864	49.0	40.2	0.513	26,700	Tested to complete fracture.
C–8	900	18,670	1.03	2.3	9.5	1,300	14.7	21,400	2,752	49.0	44.5	0.590	33,700	Do.
C–7	900	21,330	3.58	6.6	73.6	220	22.8	26,200	344	61.0	52.5	0.744	44,900	Do.
C–18	900	24,000	6.32	9.2	286	55	26.0	30,200	97	72.0	66.5	1.088	71,300	Do.
C–25	1,200	1,450	0.03	0.15	1.54	--------	--------	--------	1,439	2.37	2.3	--------	1,500	Test stopped in second stage.
C–13	1,200	1,700	0.13	.29	2.2	795	2	1,750	5.206	10.22	9.3	--------	1,900	Test stopped in third stage.
C–9	1,200	2,665	.07	.34	4.9	860	4.6	2,800	944	5.3	5.0	--------	2,800	Do.
C–5	1,200	5,330	.10	.03	58	45	2.65	5,470	215	23	23.2	0.265	6,950	Tested to complete fracture.
C–12	1,200	8,000	.66	.22	425	1.2	0.75	8,060	29	28	28.6	0.336	11,200	Do.

TABLE 4. *Summary of conditions used and results of creep tests on copper, initially as annealed*

Test		Creep stress	Plastic strain		Average creep rate, second stage	Beginning of third stage			End of test					Remarks
Number	Temperature		1 hr after application of load	Intercept at zero time		Time	Plastic strain	True stress	Time	Elongation	Reduction of area	$\log e\ (A_0/A)$ at fracture	True stress	
	° F	*lb/in.²*	*Percent*	*Percent*	*%/1000 hr*	*hr*	*Percent*	*lb/in.²*	*hr*	*Percent*	*Percent*		*lb/in.²*	
25	700	2,100	0.02	0.06	2.43	450	1.15	2,100	1,485	11	6.0	0.06	2,250	Tested to complete fracture.
34	700	2,765	.07	.14	5.04	400	2.23	2,850	797	11	6.9	0.07	2,950	Do.
31	700	5,330	1.08	1.11	77	15	2.25	5,450	62.5	11	8.8	0.09	5,850	Do.
30	700	10,670	10.52	6.32	4,200	1.5	12.7	12,000	2.4	25.5	21.3	0.24	13,600	Do.
27	900	725	0.02	0.02	2.4	610	1.5	750	2,806	21	13.1	0.14	850	Do.
33	900	1,450	.04	.05	9	140	1.3	1,500	452	15	8.9	0.09	1,600	Do.
35	900	2,100	.20	.21	39.6	30	1.4	2,150	119	12.5	9.5	0.10	2,300	Do.
32	900	2,755	.41	.6	80	6.5	1.17	2,800	39	14	10.3	0.11	3,100	Do.

increased and as the temperature is decreased. These observations agree with the results obtained on specimens of the two alloys tested at 700° or 1,200° F (fig. 4). If the strengthening effects associated with alloying were due wholly to the lattice distortion or short-order arrangements, then it appears that the magnitude of these differences would not be as great. Apparently, some additional mechanism is partly responsible for this behavior. Cottrell [14] describes a process of this type and relates the properties found in substitutional solid solution alloys to the relative degree of interaction of moving dislocations with moving solute atoms. The solute atoms may act as barriers to the motion of dislocations. The Cottrell theory accounts for the alternating of dislocations between slow and fast motions and may be applied to the serrated creep rate-strain curves shown in figures 5 and 6. An additional prediction in this theory based on the creation of vacancies during straining at constant load, is that these curves (figs. 5 and 6) should be at first rather smooth, should become serrated as the strain is increased, and, at large strains, should again be comparatively smooth. An inspection of these data reveals that these effects do occur. Moreover, it appears that increasing the temperature from 900° to 1,200° F, increasing the stress at constant temperature, or increasing the copper content of the alloys tends to decrease the magnitude of the serrations.

It was not possible to deduce a simple mathe-

TABLE 5. *Effect of prior-strain history on creep of 70-percent-nickel–30-percent-copper alloy at 700°, 900°, and 1,200° F*

Test			Creep stress	Plastic strain		Average creep rate, second stage	Beginning of third stage			End of test				Remarks
Number	Temperature	Prior history		1 hr. after application of load	Intercept		Time	Plastic strain	True stress	Time	Plasic strain or elongation	Contraction of area or reduction of area	True stress	
	° F		*lb/in.²*	*% in 2 in.*	*% in 2 in.*	*%/1,000 hr*	*hr*	*% in 2 in.*	*lb/in.²*	*hr*	*% in 2 in.*	*Percent*	*lb/in.²*	
B–1	700	Loaded 5,333 lb/in.² hr to 37,330 lb/in.² at 700° F; stress changed to 42,670 lb/in.² after 5,034 hr.	42, 670	8. 44	8. 9	0. 52	-----	-----	------	1, 511	9. 74	8. 9	46, 800	Test stopped in second stage.
B–1	700	Same specimen as B–1 (above). Stress changed to 48,000 lb/in.² after 1,511 hr.	48, 000	12. 47	15	. 99	-----	-----	------	5, 796	20. 76	17. 2	58, 000	Do.
B–6	700	Loaded 5,333 lb/in.² hr at 700° F to 42,670 lb/in.². Stress changed to 48,000 after 3,357 hr.	48, 000	12. 54	12. 9	1. 75	-----	-----	------	1, 700	16. 25	13. 9	55, 800	Do.
B–1	700	Same specimen as B–1 (above). Stress changed to 56,030 lb/in.² after 5,796 hr at 48,000 lb/in.² and 1 hr at 53,330 lb/in.²	56, 030	27. 26	27. 2	52	35	29	72, 300	165	56	61. 8	146, 600	Tested to complete fracture.
B–6	700	Same specimen as B–6 (above). Stress changed to 56,030 lb/in.² after 1,700 hr at 48,000 lb/in.² and 1 hr at 53,330 lb/in.²	56, 030	22. 61	22. 0	74. 5	122	31. 1	73, 500	235. 5	66	65. 8	163, 800	Do.
B–14	900	Loaded 5,333 lb/in.² hr at 900° F to 18, 670 lb/in.² Unloaded after 2,468 hr and temperature held at 900° F. Reloaded (after 216 hr) 5,333 lb/in.² hr to 21,330 lb/in.²	21, 330	2. 26	2. 46	1. 64	-----	-----	------	1, 259. 3	4. 47	4. 3	22, 300	Test stopped in second stage.
B–14	900	Same specimen as B–14 (above). Stress changed to 28,000 lb/in.² after 1,259 hr at 21,330 lb/in.² and 1 hr at 26,670 lb/in.²	28, 000	5. 46	6. 8	11. 8	1, 150	20. 6	33, 800	2, 133	58	43. 2	49, 300	Tested to complete fracture.
B–9	900	Loaded 5,333 lb/in.² hr to 48,000 lb/in.² at 700° F. Unloaded after 3,051 hr. Temperature raised to 900° F; loaded 5,333 lb/in.² hr to 28,000 lb/in.²	28, 000	19. 48	18. 6	19	540	28. 8	36, 100	847. 5	48	38. 1	45, 200	Do.
B–16	900	Loaded instantaneously to 2,665 lb/in.² at 1,200° F; load removed and temperature changed to 900° F after 1,414.5 hr; loaded 5,333 lb/in.² hr to 28,000 lb/in.²	28, 000	9. 32	14. 1	42. 6	220	23. 5	34, 600	499	63	70. 9	96, 100	Do.
B–17	900	Loaded 5,333 lb/in.² hr to 61,840 lb/in.² at 300° F; load removed and temperature changed to 900° F after 1,871 hr; loaded 5,333 lb/in.² hr to 28,000 lb/in.²	28, 000	27. 65	27. 6	17	250	32	37, 000	762. 5	56. 5	52. 4	58, 900	Do.
B–2	1, 200	Loaded 5,333 lb/in.² hr to 26,670 lb/in.² at 900° F; load removed; temperature raised to 1,200° F; reloaded in 2 increments (5,333 and 2,667 lb/in.²) to 8,000 lb/in.²	8, 000	10. 8	11. 15	27. 2	210	16. 9	9, 350	375. 3	27	21. 3	10, 200	Do.

matical formula for describing the complete strain-time curves. This observation is attributed to the fact that adequate parameters have not been determined for describing the complex structural changes that occur during the different stages of creep. Therefore, it was considered to be more fruitful to tabulate the data, as in tables 2, 3, and 4, and then to analyze separately the data as applied to each of the various stages.

c. First Stage of Creep

The nature of the strain-time curve during the first stage (flow at a decelerating rate) has resulted in a number of mathematical analyses of this stage of the creep process. In general, a period of decreasing creep rate with increasing time was obtained in specimens used in this investigation. Several of the specimens used indicated the existence of an incubation period similar to those shown previously for some specimens of initially annealed nickel at 300° F [5]. The existence of this period in the alloys is also partly attributed to the magnitude of the final applied stress.

Probably the most complete analysis of the creep curves has been made by Andrade [15] who indicated that his data could be described by the equation:

$$l=l_0(1+Bt^{1/3})e^{Kt}$$

TABLE 6. *Effect of prior-strain history on creep of 30-percent-nickel–70-percent-copper alloy, initially as annealed*

Test			Creep stress	Plastic strain		Average creep rate, second stage	Beginning of third stage			End of test			
Number	Temperature	Prior-strain history		1 hr after application of load	Intercept zero time		Time	Plastic strain	True stress	Time	Elongation	Reduction of area	True stress
	°F		*lb/in^2*	*Percent*	*Percent*	*%/1,000 hr*	*hr*	*Percent*	*lb/in^2*	*hr*	*Percent*	*Percent*	*lb/in^2*
C–6	300	Loaded 5,333 lb/in.^{2} hr to 42,670 lb/in.^{2} Stress changed to 46,000 lb/in.^{2} after 2,370 hr; to 47,000 lb/in.^{2} after 3,499 hr; to 47,900 lb/in.^{2} after 4,318 hr at 300° F.	47, 900	20. 54	--------	--------	--------	--------	--------	10	49. 5	80. 6	246, 500
C–11	300	Loaded 5,333 lb/in.^{2} hr to 46,000 lb/in.^{2} Stress changed to 47,900 lb/in.^{2} after 2,182 hr at 300° F.	47, 900	19. 82	--------	--------	--------	--------	--------	10	46. 5	80. 7	248, 100
C–15	300	Loaded 5,333 lb/in.^{2} hr to 47,000 lb/in.^{2} Stress changed to 47,900 lb/in.^{2} after 2,340 hr; to 48,440 lb/in.^{2} after 3,373 hr; to 48,980 lb/in.^{2} after 4,693 hr at 300° F.	48, 980	20. 42	--------	--------	--------	--------	--------	1. 5	47. 5	80. 2	247, 500
C–24	900	Loaded 5,333 lb/in.^{2} hr to 37,330 lb/in.^{2} at 300° F; unloaded after 1,507 hr; temperature changed to 900° F; loaded 5,333 lb/in.^{2} hr to 18,670 lb/in.^{2} at 900° F.	18, 670	7. 80	7. 6	18	600	18. 2	22, 100	1, 123	52	56	42, 600
C–23	900	Loaded 5,333 lb/in.^{2} hr to 47,200 lb/in.^{2} at 300° F.; unloaded after 1,484 hr; temperature changed to 900° F, loaded 5,333 lb/in.^{2} hr to 18,670 lb/in.^{2} at 900° F.	18, 670	21. 04	20. 9	19. 2	260	26	23, 500	696	78	63	49, 800
C–19	900	Loaded 5,333 lb/in.^{2} hr to 26,670 lb/in.^{2} at 700° F, unloaded after 3,330 hr; temperature changed to 900° F; loaded 5,333 lb/in.^{2} hr to 18,670 lb/in.^{2} at 900° F.	18, 670	7. 36	7. 1	21. 5	600	20	22, 400	1, 065	51	49. 5	37, 000
C–21	900	Loaded 5,333 lb/in.^{2} hr to 10,670 lb/in.^{2} at 900° F; stress changed to 18,670 lb/in.^{2} (2 increments) after 2,345 hr at temperature under stress.	18, 670	1. 56	4. 5	11. 6	730	12	20, 900	1, 214	40	45	34, 000
C–17	900	Loaded 5,333 lb/in.^{2} hr to 13,335 lb/in.^{2} at 900° F; stress changed to 18,670 lb/in.^{2} (1 increment) after 3,860 hr at temperature under stress.	18, 670	10. 6	11	57	210	23	23, 000	342	48	51	38, 000
C–25	900	Load corresponding to a stress of 1,450 lb/in.^{2} applied instantaneously at 1,200° F; unloaded after 1,439 hr; temperature changed to 900° F; loaded 5,333 lb/in.^{2} hr to 18,670 lb/in.^{2} at 900° F.	18, 670	6. 4	9. 7	60	267	25. 5	23, 400	380	55	50. 5	37, 800
C–13	900	Load corresponding to a stress of 1,700 lb/in.^{2} applied instantaneously at 1,200° F.; unloaded after 5,205 hr; temperature changed to 900° F; loaded 5,333 lb/in.^{2} hr to 18,670 lb/in.^{2} at 900° F.	18, 670	18. 3	18. 15	905	13. 5	30. 3	24, 300	18	41	43	33, 000

Where l is the length of the specimen at time, t, and l_0, B, and K are constants. The term, $Bt^{1/3}$, was considered as representative of transient creep (essentially first stage) and the term, Kt, represented the steady-state component (second stage of creep). If no steady-state component exists then K would equal 0 and

$$l=l_0(1+Bt^{1/3})$$

or

$$\frac{dl}{dt}=1/3 l_0 B t^{-2/3}$$

whence

$$\log \dot{e}=-2/3 \log t+C_1$$

where

$$\dot{e}=\frac{1}{l_0}\frac{dl}{dt}=\text{creep rate.}$$

With the exception of the data for several specimens, previously mentioned, a linear relation was obtained when the logarithm of the creep rate was plotted against the logarithm of the time during the first stage of creep. The relations between the slopes of the curves and the initially applied creep stresses are shown in figures 7A and 8. It is apparent that transient creep, as described by Andrade, did not exist as the slopes are dependent on stress, temperature, and nickel content. Obviously, the slopes are displaced to higher values (fig. 7A) as the temperatures are increased. No general trend was observed to describe the influence of alloying the nickel with copper (fig. 8).

The modified exhaustion theory of Davis and Thompson [16] is described by the equation

$$\log(t\dot{e})=A+B\log e$$

where t=time, $\dot{e}$=creep rate, and e=strain. Attempts were made to apply this formula to the data obtained for the 70-percent-Ni–30-percent-Cu alloy. However, the agreement was not as good as that previously recorded for nickel [5] as shown by the curves in figures 7B and 9. The values for the slopes in this analysis should be between +1 and $-\frac{1}{2}$. The effect of stress on the values of the slopes, for the 70-percent-Ni–30-percent-Cu alloy is shown in figure 7B, and the influence of alloying the nickel

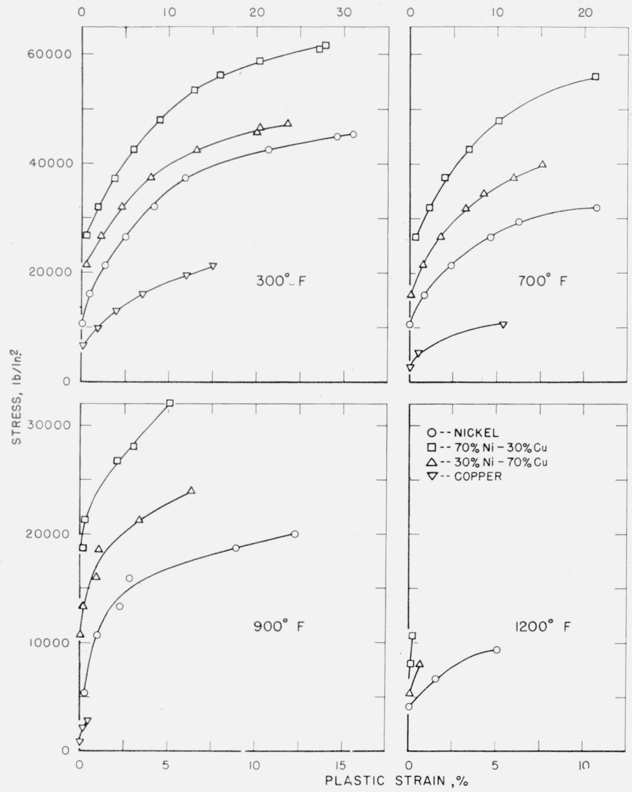


Figure 1. *Stress-strain relations of specimens tested at different temperatures.*

The strain values shown were those measured 1 hr after application of the corresponding stress.

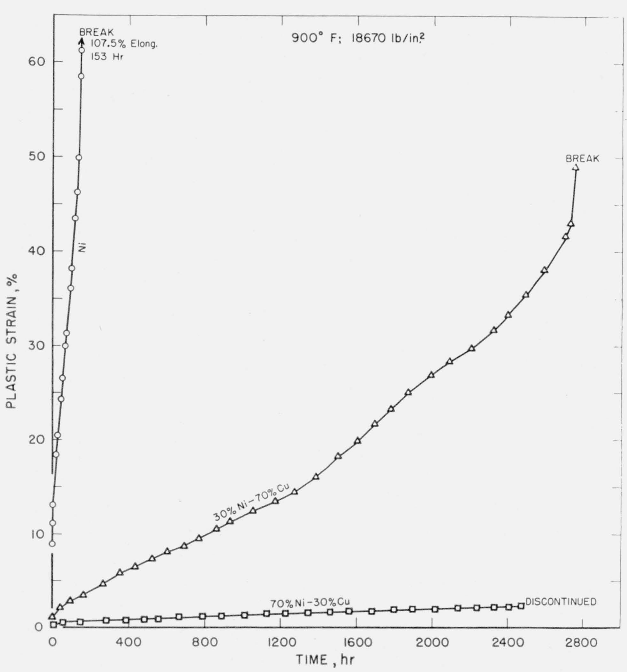


Figure 2. *Strain-time relations of specimens tested at 900° F with a stress of 18,670 lb/in².*

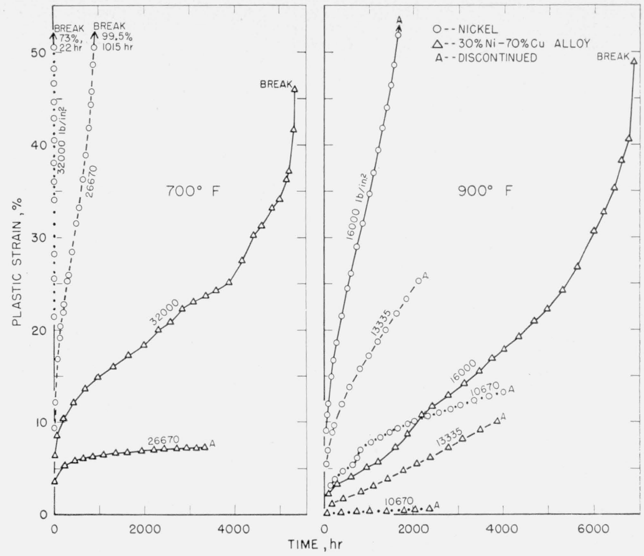


Figure 3. *Strain-time relations of specimens tested at 700° or 900° F.*

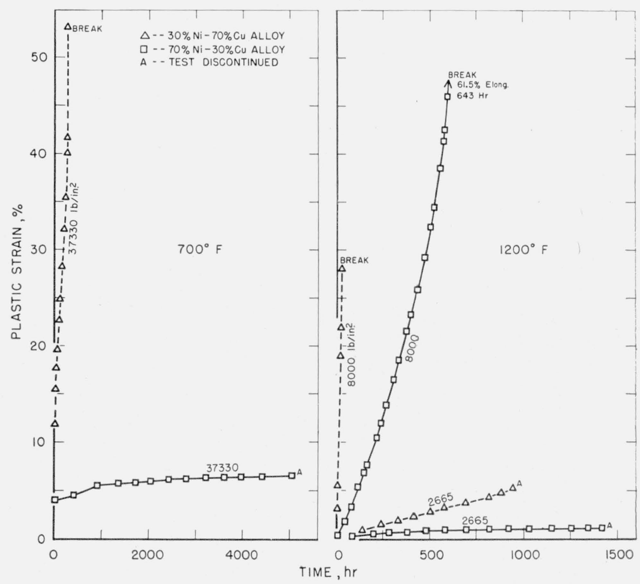


Figure 4. *Strain-time relations for two specimens of* Cu-Ni *alloys tested at 700° or 1,200° F with different stresses.*

with 30 percent of copper is shown by the data reproduced in figure 9. For the alloy (fig. 7B), it appears that the slopes increase with increasing temperature (stress constant) or with increasing stress (temperature constant).

The shapes of the curves appear to indicate that the deformation process changes somewhat with variations in temperature and stress. With the exception of the data obtained at 1,200° F (fig. 9), alloying the nickel displaced the values of the slopes to lower values. Apparently, the solute atoms that cause a strengthening effect in the alloy at low temperatures are less effective strengtheners as the temperature is raised.

The stress dependence of creep rate in the first stage of the 70-percent-Ni–30-percent-Cu alloy is shown in figure 10. Nadai and McVetty [17] indicated that this relation could be expressed by an equation of the form

$$\dot{e}=\dot{e}_0 \sinh\left(\frac{\sigma}{\sigma_0}\right)$$

where $\dot{e}$=the creep rate, σ=stress, and $\dot{e}_0$ and σ_0 are material constants. At low stresses this equation becomes

$$\dot{e}=\dot{e}_0\left(\frac{\sigma}{\sigma_0}\right).$$

This suggests that at low stresses the material behaves like a Newtonian fluid. At high stresses, however, the expression may be shown to exist in the form

$$\dot{e}=\left(\frac{\dot{e}_0}{2}\right)\exp\left(\frac{\sigma}{\sigma_0}\right)$$

or

$$\log \dot{e}=K\left(\frac{\sigma}{\sigma_0}\right)+k.$$

The data were plotted on semilog paper and reproduced in figure 10. The curves at 300° F indicate an approach to linearity as the time at stress is increased. At 900° F, the shape of the curves indicates that the flow mechanisms, although temperature-sensitive, are less affected by the time at which the creep rate was measured for any stress value employed. The positions of the curves, however, indicate that the short-time (5 to 20 min) strain hardening occurring at 700° F is equal to or greater than that occurring at 300° F. This phenomenon is in agreement with another prediction of the theory of the effect of solute atoms as proposed by Cottrell [14]. In the short-time tensile tests of this alloy [7], the magnitude of the serrations in the stress-strain curve attained a maximum in the temperature range 600° to 800° F.

d. Second Stage of Creep

It was previously pointed out in Andrade's creep equation [15] that one of the components described a steady state of flow. This would indicate that when $B=0$,

$$l=l_0e^{Kt}$$

or

$$\frac{1}{l}\frac{dl}{dt}=K.$$

The region to which this formula applies is generally called the second stage of creep. The apparent balance between the competing mechanisms of strain hardening and recovery of metals during this stage has caused a number of theories to be proposed to describe this steady state. Generally, the analyses of the relations between stress, temperature, and second stage creep rate are based on Eyring's chemical-rate theory [18] or are a result of curve fitting.

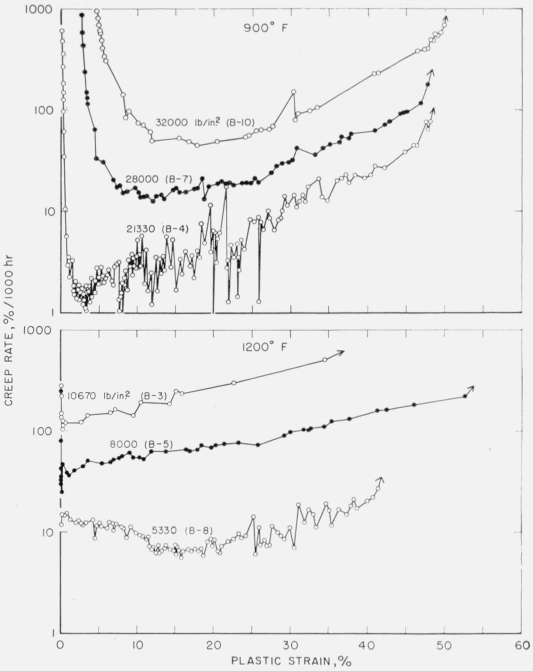


FIGURE 5. *Creep rate-strain characteristics of 70 percent* Ni-*30-percent*-Cu *alloy at 900° and 1,200° F.*

Specimen numbers correspond to those found in the tables.

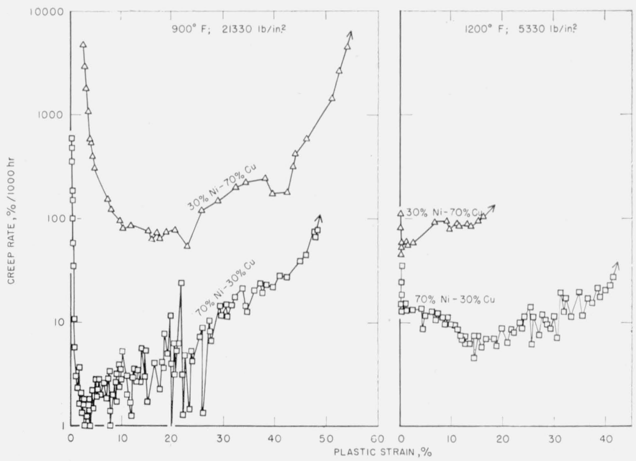


FIGURE 6. *Creep rate-strain relations for specimens of two nickel-copper alloys tested at 900° or 1,200° F.*

Essential features of each of these proposals have been incorporated into a logarithmic rate law in which the second stage creep rate is considered to be proportional to a power function of the stress. The relations between stress and creep rate of the specimens used in the present investigation are shown in figures 11 and 12. At 300°, 700°, and 900° F the curves for the alloys are at higher stress levels than those for the component metals although at 300° F the creep strength of the nickel was only slightly

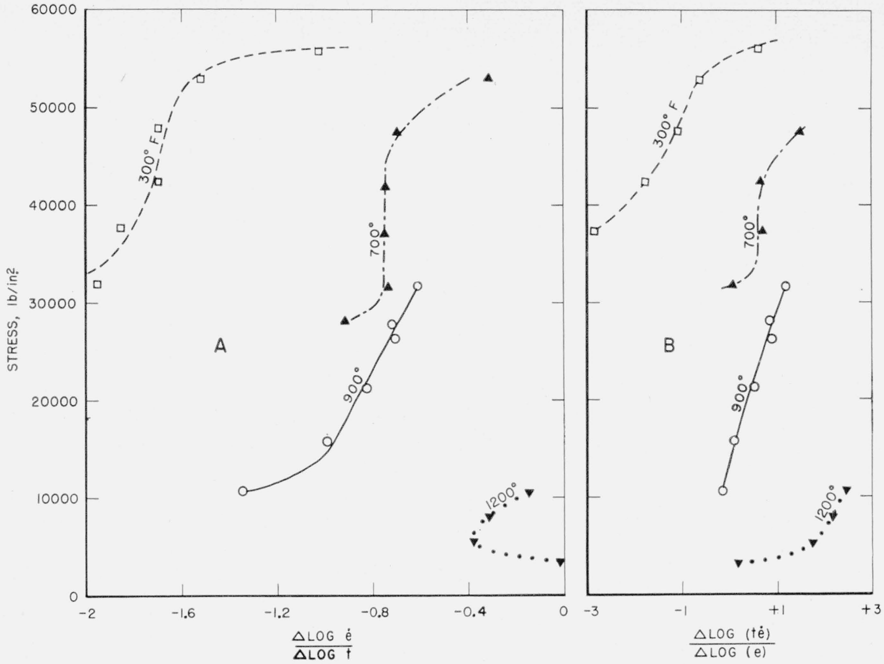


FIGURE 7. *Relations of stress to the slopes of the log creep rate-log time curves and of the log product of time and creep rate-log strain curves during the first stage of creep of 70-percent-*Ni*–30-percent-*Cu *specimens at different temperatures.*

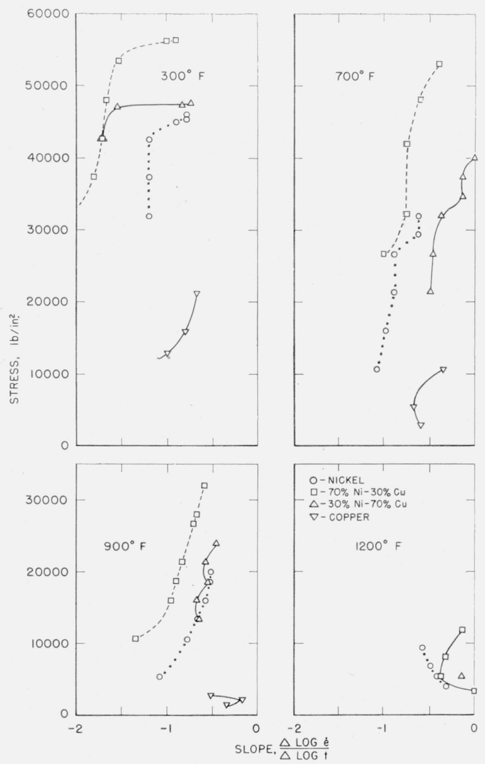


FIGURE 8. *Relation of stress to the slopes of the log creep rate-log time curves during first stage of creep at different temperatures.*

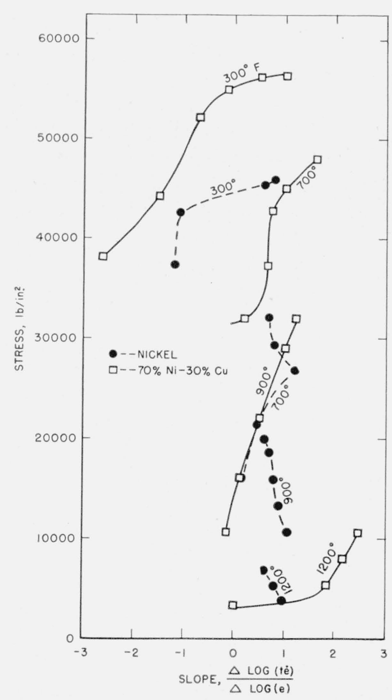


FIGURE 9. *Relation of stress to the slopes of the log product of the time and creep rate-log strain curves during the first stage of creep.*

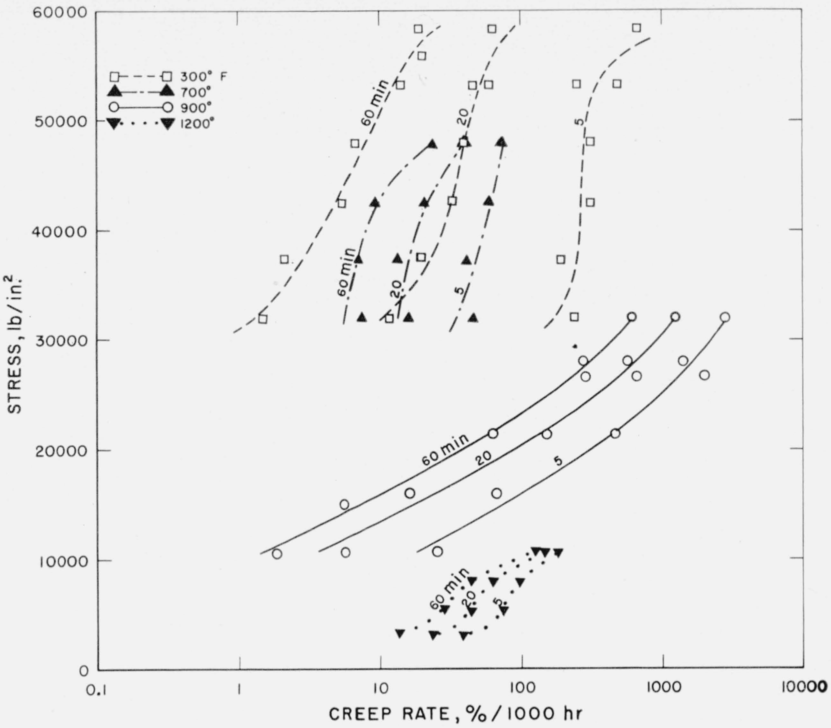


FIGURE 10. *Relation of stress to first stage creep rate of 70-percent-*Ni*–30-percent-*Cu *alloy at different temperatures and times.*

less than that of the 30-percent-Ni–70-percent-Cu alloy. Thus, the strengthening effects due to alloying that was observed in the first stage are still evident during the second stage of creep. At 1,200° F, however, the nickel specimens appeared stronger than those of the 30-percent-Ni–70-percent-Cu alloy, indicating that factors affecting the strengths at low temperatures are not necessarily as effective at the higher test temperatures. Similar strengthening effects were obtained previously in short-time tensile tests on high-purity and commercial grades of the 70-percent-Ni–30-percent-Cu alloy [6]. Above the threshold stresses at which creep begins at 300° F (fig. 11), small changes in stress caused a marked change in creep rate. However, the relatively high threshold stresses at 300° may be attributed to a combination of a high degree of strain hardening and aging accompanied by a low rate of recovery. As the temperature is raised, (figs. 11 and 12) recovery predominates and the threshold stresses needed to produce equal second stage rates become less.

The relations between temperature and stress necessary to produce different second-stage creep rates are shown for the 70-percent-Ni–30-percent-Cu alloy in figure 13. As the shape of this family of curves is typical of the behavior of all the metals used in this investigation, the data for the other metals are not shown. However, the short-time tensile strengths of the 70-percent-Ni–30-percent-Cu alloy are included for comparison. In the temperature range 300° to 700° F, the rate of change in stress with temperature increases with increasing creep rates; whereas, the reverse is true in the 900° to 1,200° F range. However, within the 700° to 900° F range, these slopes appear to be independent of creep rate. Apparently, both high- and low-temperature mechanisms are active within this latter range of temperatures.

The relation between stress and nickel content to produce various second-stage creep rates at different temperatures is shown in figures 14 and 15. In general, the slopes of the stress-nickel content curves to produce second-stage creep rates of 0.5 percent, 1 percent, 10 percent, or 100 percent per 1,000 hr are a maximum as the nickel content is increased from zero to 30 percent. It is observed that the variation of stress with nickel content at any creep rate is generally greater at 700 ° F than at other test temperatures. Several phenomena are considered to be responsible for the latter observation. Of the temperatures used, simultaneous straining and recrystallization of copper have been reported at 700° F [7]; this temperature is within the vicinity of the Curie point of nickel [3]; serrated stress-strain curves are obtained for specimens of the alloys in short-time tensile tests [7].

The relation between stress and nickel content to produce different creep rates at different temperatures is shown in figure 15. Although the 70-percent-nickel alloy specimens showed the best creep resistance at all temperatures and stresses, the parallelism of some of the line segments indicated that the strengthening effects of nickel were practically independent of the second-stage creep rate.

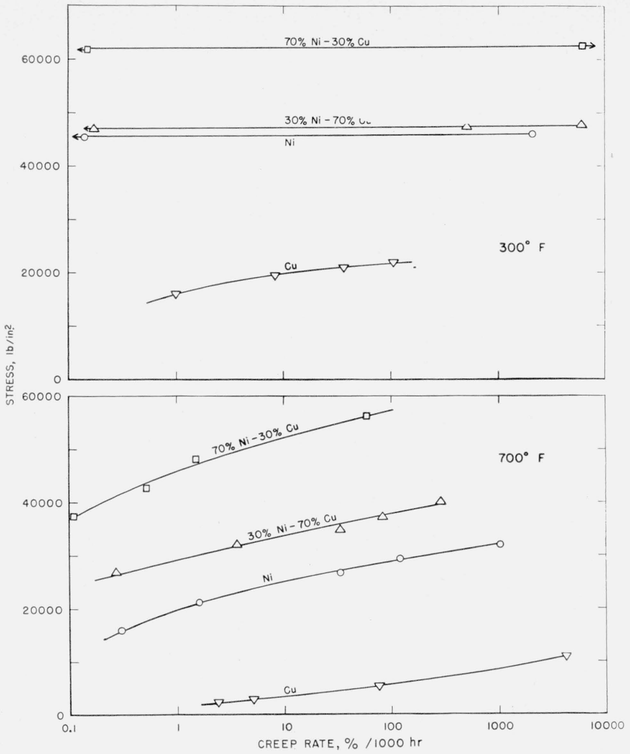


FIGURE 11. *Influence of stress on the average creep rate during the second stage at 300° and 700° F.*

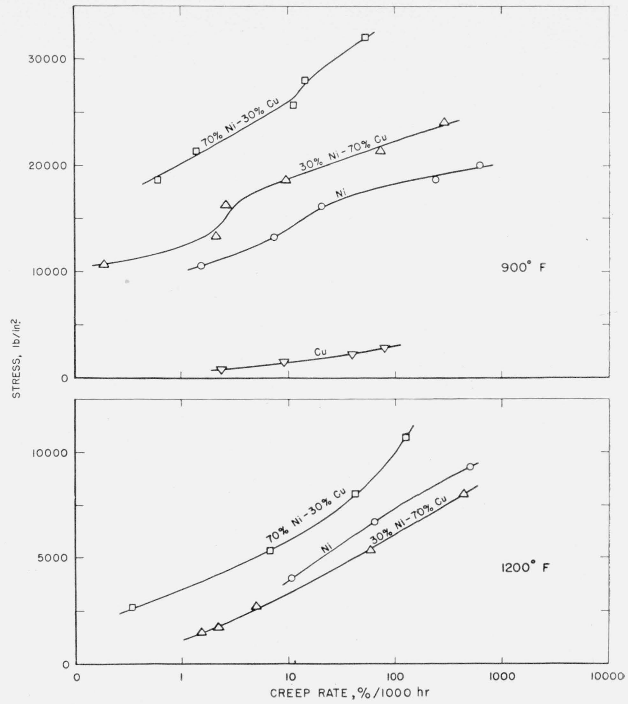


FIGURE 12. *Influence of stress on the average creep rate during the second stage at 900° and 1,200° F.*

Larson and Miller [19] have proposed time-temperature relationships for describing creep behavior and rupture strength. For any given stress

$$T_R(20+\log t)=K$$

or

$$T_R(20-\log r)=C$$

where T_R=temperature in degrees Rankine, t=fracture time in hours, r=second-stage creep rate in percent per 1,000 hr and C and K are constants. Also if the logarithm of the stress is plotted against C or K, all values for specimens of each material should fall on a single curve in the absence of major structural changes. Conformance to these predictions is obtained for the materials used in this investigation as shown in figure 16. The discontinuous curves for copper are attributed, in part, to recrystallization and extensive grain growth during creep. Moreover, the relative positions of the nickel and the 30-percent nickel alloy indicate the factors contributing to the strengthening of the nickel due to alloying are most effective within the parameter values of 16 to 26×10^3.

e. Creep Rate-Ductility Relations

One of the most important structure-sensitive properties of metals is ductility. Previous investigations have shown that no consistent trend existed between ductility values and the rate of straining at different temperatures. Therefore, predictions of ductility values, based on fundamental concepts, does not appear possible at this time. However, it is generally conceded that the ductility of metals depends on the initial concentration of lattice defects and the formation and motion of new defects throughout the crystals. Apparently, as observed in short-time tensile tests on materials of the nickel-copper system [7], substantial changes in the ductility-temperature relations are made by alloying. Changes of this type have been analyzed by Cottrell [14] who considered them to be related to the creation of vacancies and interstitial defects during cold-working.

The relation between strain values, obtained one hour after application of load and the logarithm of the second-stage creep rates for specimens tested at 700°, 900°, and 1,200° F is shown in figure 17. Sufficient data were not available for analyzing the behavior of the specimens at 300° F. The initial strain values, corresponding to equal second-stage creep rates at 700° and at 900° F, increase as the copper content of the specimens is decreased from 100 to 30 percent. The strain values at 700° F for the nickel are approximately the same as those of the 30-percent-Ni–70-percent-Cu alloy; whereas, at 900° F the corresponding strain values for the nickel are greater than those of the 30-percent-nickel alloy. Moreover at 1,200° F, the strain values of the nickel are higher than those for either of the alloys. These observations indicate that the flow characteristics of these metals are strongly dependent on the copper content.

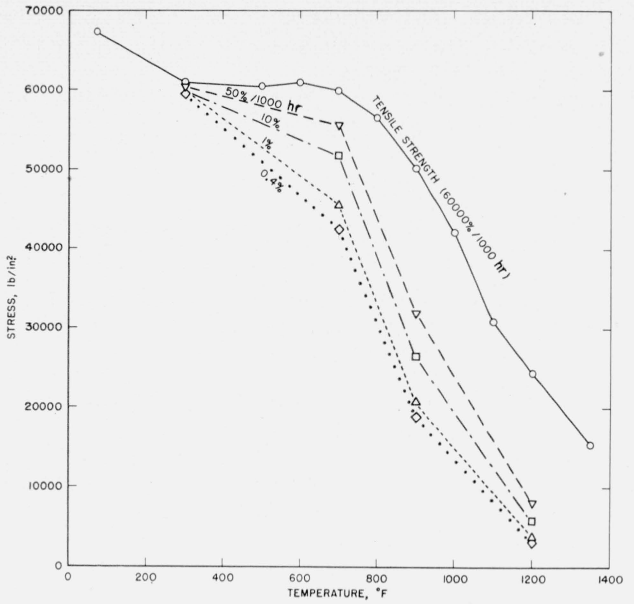


FIGURE 13. *Variation of stress with temperature required to produce various second-stage rates of 70-percent-*Ni*–30-percent-*Cu *alloy.*

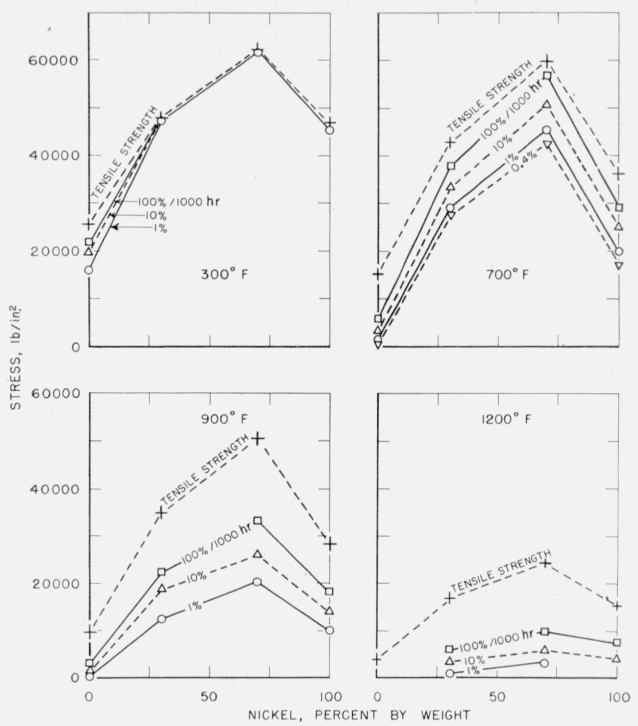


FIGURE 15. *Variation of stress with nickel content to produce different second-stage creep rates at various temperatures.*

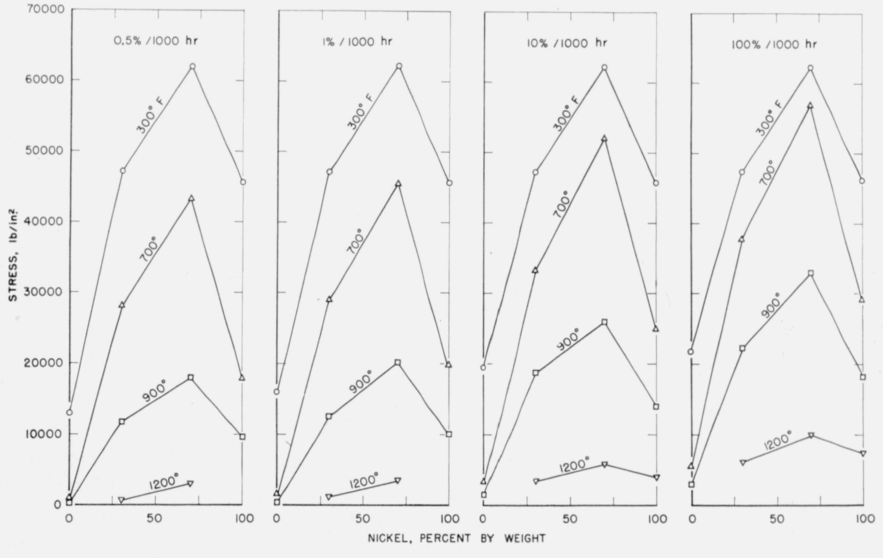


FIGURE 14. *Variation of stress with nickel content to produce equal second-stage creep rates at different temperatures.*

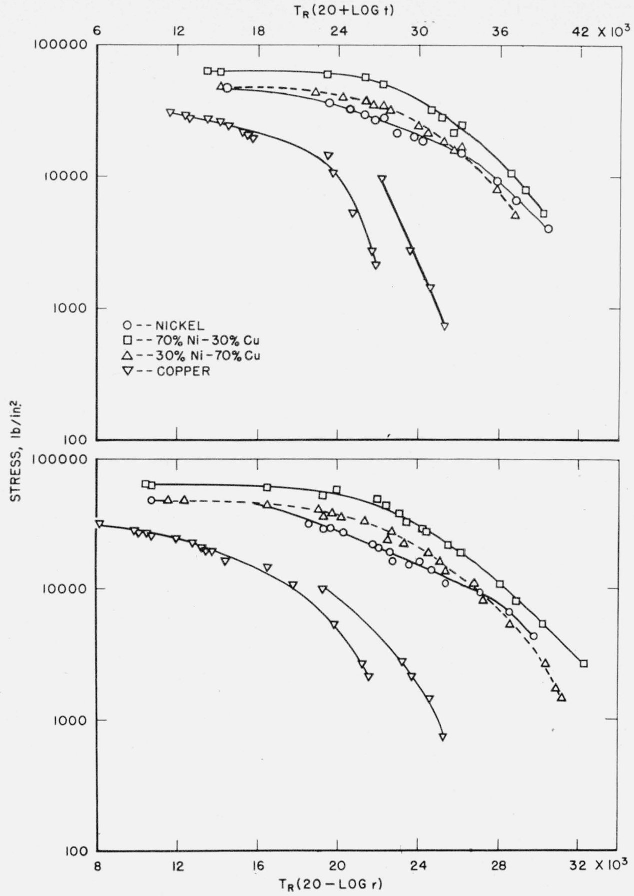


FIGURE 16. *Relation between creep stress and fracture time or creep-rate parameters.*

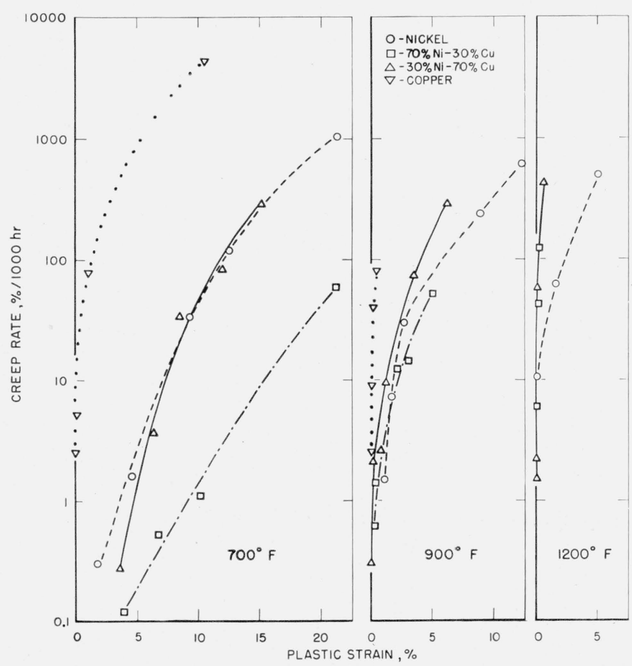


FIGURE 17. *Relation of strain, measured 1 hr after application of load, to the second-stage creep rate at different temperatures.*

The relation between second-stage creep rate and elongation or reduction of area values for specimens tested to complete fracture at 700°, 900°, and 1,200° F is shown in figure 18. It is noteworthy that the ductility of the specimens of the alloys was intermediate between the low values for copper and the high values of the nickel.

f. Third Stage of Creep, Specimen Contour, and Post-Test Hardness

Several processes are discussed in the literature as being the causes of the initiation and the propagation of accelerating creep in the third stage. Among these processes are the following: rising stresses resulting from a decrease in cross-sectional area of the specimen, structural changes, extensive recovery or recrystallization, and nucleation and density of void nuclei. Machlin [20] has recently developed a theory based on the probability of the growth of pre-existent voids by vacancy condensation. Each of the above concepts are applicable to a limited extent to the data for the present investigation. No local contraction (necking) nor voids of a microscopic size were observed for any of the specimens that were examined after stopping the tests prior to the third stage. Therefore, it is concluded that the necking and the growth of cracks, were a consequence of accelerating creep in the third stage. Similar observations were made previously for the copper [1] and for the nickel [5]. With the exception of nickel, previously discussed [5], the degree of necking, in general, decreased with increase in temperature and with decrease in creep rate. This typical behavior is illustrated by the curves shown for the 70-percent-Ni-30-percent-Cu alloy specimens in figure 19A. Moreover, the specimens exhibiting the greatest tendency to contract locally (B–18 and B–13) both showed a tendency toward an increase in hardness with increase in contraction of area values (fig. 19B).

For the specimens tested at 1,200° F (B–3, B–5 and B–8), the opposite effect on hardness was observed. The tendency to soften as the contraction of area values increased was attributed both to the relatively high temperature and decrease in creep rate. This behavior was accompanied by a progressively increasing number of microscopic cracks both near the surface and in the interior of the specimens.

The relations between maximum post-test hardness and second-stage creep rate are shown in figure 20 for specimens of the nickel and the two alloys. The general trends were for the hardness to increase with a decrease in test temperature (creep rate constant) or with increase in creep rate (temperature constant). This same observation was made for the copper; however, the hardness values were too low to be included in the figure. Alloying the copper tended to raise the hardness values markedly; however, with one exception the maximum hardness value for the nickel was higher than the corresponding values of the 30-percent-Ni–70-percent-Cu alloy.

Apparently, some of these and previous data are consistent with the observations on the lattice param-

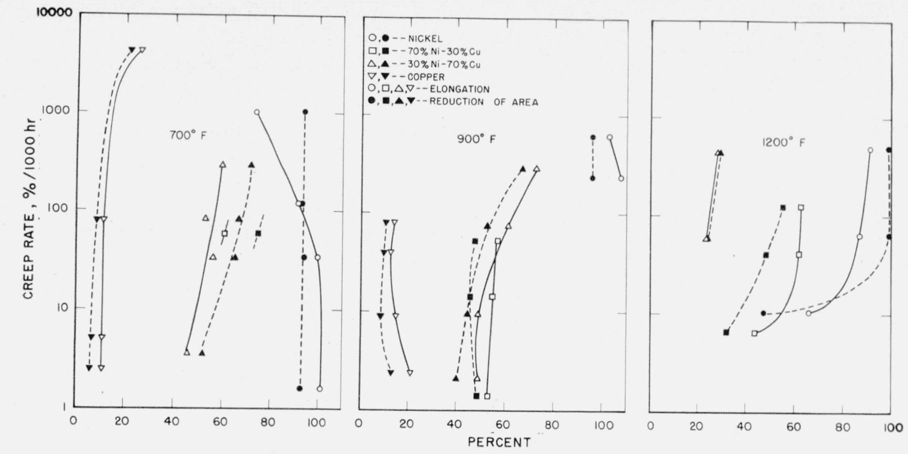


FIGURE 18. *Effect of second-stage creep rate on elongation or reduction of area.*

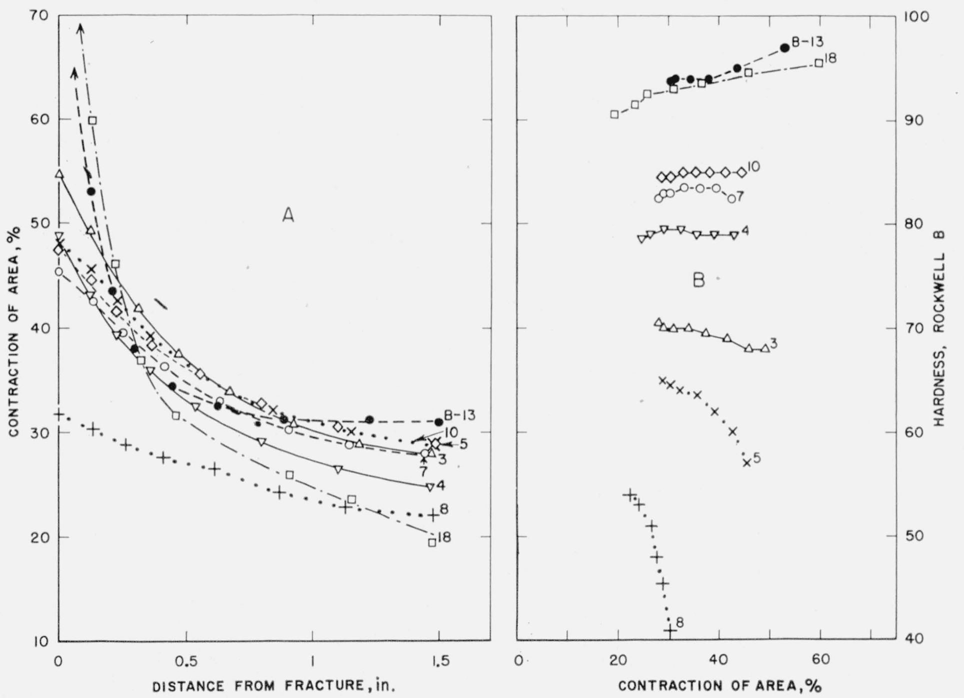


FIGURE 19. *Specimen contours and hardness distributions of specimens of 70-percent-*Ni*–30-percent-*Cu *alloy tested to complete fracture*

Specimen number	Tempera-ture	Average creep rate, second stage
	° *F*	*%/1,000 hr*
B–18	300	220,000
B–13	700	59
B–10	900	51.8
B– 7	900	14.3
B– 4	900	1.4
B– 3	1,200	125
B– 5	1,200	41
B– 8	1,200	6.6

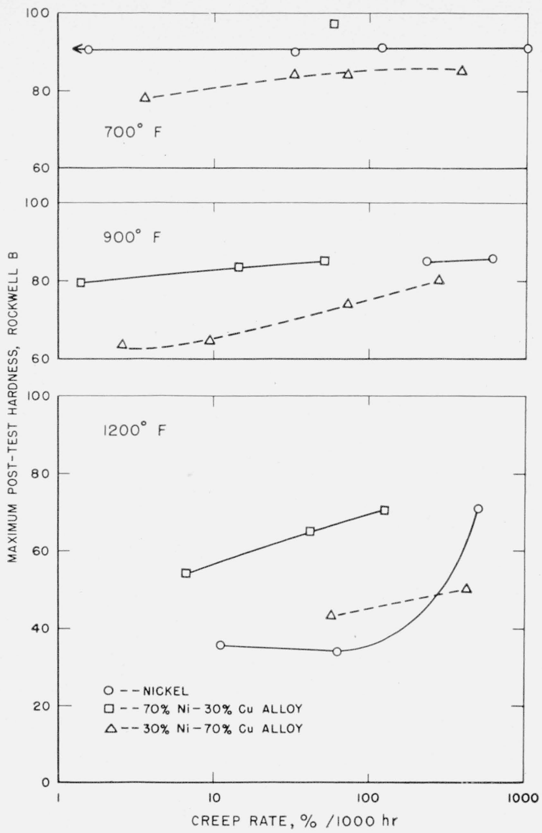


FIGURE 20. *Effect of second-stage creep rate on the maximum post-test hardness values of specimens tested to complete fracture at different temperatures.*

eters of the copper-nickel system recently presented by Coles [21].

g. Activation Energy

Determinations of activation energies from data obtained in creep tests on different metals have been made in recent years. Some investigators have found that the activiation energy associated with creep is approximately equal to that for self-diffusion of the material. Other data depart rather markedly from this observation. Also controversial are the influences of prior-strain history, strain, temperature, grain and subgrain sizes, and the metallurgical changes occurring during creep. In addition, differences in values of activation energy may be attributed to the use of different basic assumptions in the calculations. For example, the activation energy for creep of the nickel used in this investigation was calculated as about 53,500 calories per mole according to the analysis of Manjoine and Mudge [23] and about 65,000 calories per mole by using the dislocation climb mechanism proposed by Weertman and Shahinian [22]. It was, therefore, thought unfruitful to pursue further any analysis of activation energy until a more complete appraisal of the basic factors affecting the calculations can be made.

3.2. Effect of Prior-Strain History on Creep Behavior

a. Strain-Time and Stress-Strain Curves

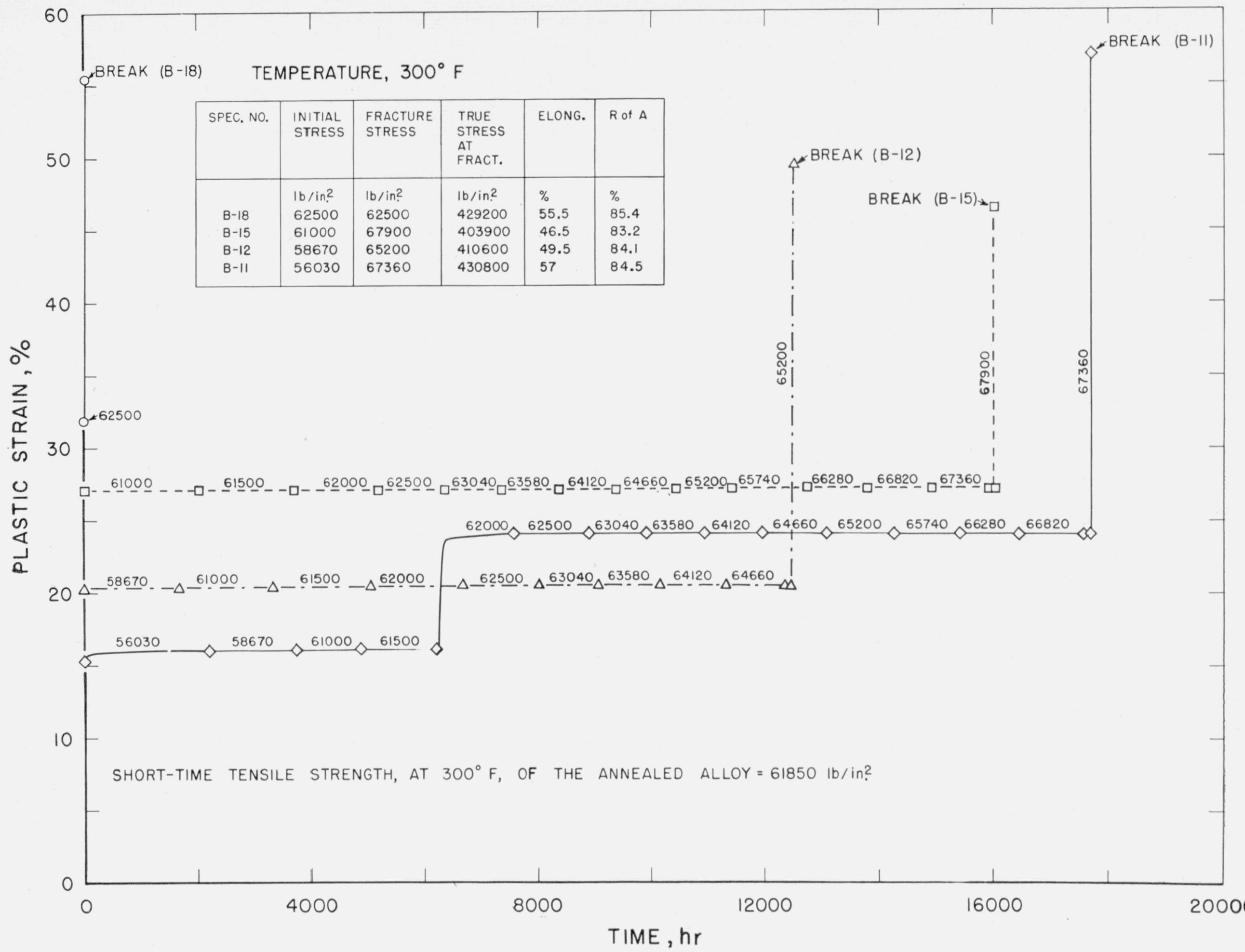
An analysis of experimental data led Ludwik [24] to propose that the strain-hardening characteristics of a material were a function of the instantaneous strain, strain rate, and temperature and were practically independent of the prior strain history. Many fundamental data have since been accumulated to test the validity of this equation of state. It is now the generally accepted view that this relation is generally invalid and that the data tending to confirm this and similar proposals may be considered as only special cases of the general problem of plastic flow.

Wood [25], in summarizing the factors affecting the structural changes in metals during deformation, indicated that the parent grains are fragmented. The size of the substructures so formed increases with increase in test temperature and with decrease in rate of straining. The boundaries of the substructures can act as barriers to the motion of dislocations or as sites where stable barriers form. During a constant-stress creep test the substructure tended toward an equilibrium size characteristic of the stress and test temperature. Systematic changes in subgrain size, accompanied by changes in resistance to deformation, was predicted for specimens as the test temperature or strain rate was altered. A quantitative test of this hypothesis was recently made by Hazlett and Hansen [10] who indicated that the shape of the creep curve could be materially altered by changes in the initial substructure of the base material and further altered by alloying. The prestraining of the nickel in that investigation was done exclusively at room temperature and followed by a recovery treatment at 1,290° F.

The influence of prior straining in creep on the subsequent creep behavior at elevated temperature and tensile properties at room temperature of nickel and copper has been discussed in previous publications [1, 2, 3, 5]. The present paper is concerned with similar effects on the 70-percent-Ni–30-percent-Cu and 30-percent-Ni–70-percent-Cu alloys. The test conditions and results are summarized in tables 5 and 6 and analyzed graphically in figures 21 to 27, inclusively.

The effect of the magnitude of the initial stress and the subsequent rate of loading on the creep behavior of specimens of 70-percent-Ni–30-percent-Cu alloy at 300° F is shown in figure 21. If the only modes of deformation, existing for these specimens, were the ones suggested by Wood [25], then the curves for specimens tested at equal stresses should merge after a time. Apparently, an aging phenomenon, such as was shown for nickel [5] existed for this alloy under these test conditions. Although the reduction of area values are approximately independent of the rate of loading, elongation and the true stress at fracture appear to be dependent on this factor.

The influence of alloying the nickel with copper on the resistance to creep at 300° F of slowly loaded specimens can be deduced from a comparison of the difference in stresses, in excess of the tensile strengths,



SPEC. NO.	INITIAL STRESS	FRACTURE STRESS	TRUE STRESS AT FRACT.	ELONG.	R of A
	lb/in.²	lb/in.²	lb/in.²	%	%
B-18	62500	62500	429200	55.5	85.4
B-15	61000	67900	403900	46.5	83.2
B-12	58670	65200	410600	49.5	84.1
B-11	56030	67360	430800	57	84.5

FIGURE 21. *Effect of rate of loading at 300° F on the strain-time characteristics of 70-percent-*Ni–*30-percent-*Cu *alloy.*

required to produce fracture in 1,000 hr. The 70-percent-Ni–30-percent-Cu alloy required a stress at least 3,350 lb/in.² in excess of its short time tensile strength (fig. 21) whereas the nickel required only 1,600 lb/in.² or less, in excess of its tensile strength to produce complete fracture in 1,000 hours in slowly loaded specimens [5]; the short-time tensile strengths at 300° F of the alloy and nickel were 61,850 and 46,400 lb/in.², respectively. This increase in creep strengths at 300° F of both the alloy and the nickel is primarily attributed to a combination of strain-aging and strain-hardening characteristics. The results show that the strain aging and hardening are much more prominent in the specimen loaded slowly than in the specimens loaded rapidly to the selected creep stresses.

The strain-aging characteristics at 300° F of the alloy are further illustrated, in figure 22, by a comparison of the relative positions of the stress-strain curves associated with the slowly loaded specimens (dashed curves) and the specimens loaded at 5,330 lb/in.² hr (solid curve). In the absence of strain aging, the curves for the four specimens should coincide.

The effect of straining a 70-percent-Ni–30-percent-Cu specimen at a lower stress at 900° F on the subsequent creep behavior at a higher stress and at the same temperature, is shown in figure 23. Although the second stage creep rate was approximately the same for the specimen (B–14) prestrained 2.23 percent at 18,670 lb/in.² as that for the specimen (B–4) not prestrained, the behavior of the two specimens in the first stage was somewhat different. The duration of the first stage of the former specimen was relatively short in comparison with that of the latter. These observations are consistent with Wood's theory describing the tendency of the subgrains to approach an equilibrium size. The fact that this behavior is not typical at 900 °F is shown by the results reproduced in figure 24. The specimens were prestrained different amounts at temperatures of 300°, 700°, 900°, or 1,200° F. The specimens prestrained at 300°, 700°, and 1,200° F showed a tendency toward an increase in second stage creep rate and a decrease in fracture time when compared to the specimen loaded at the standard rate and strained at 900° F. Straining slowly at 900° F (B–14) to a value of 5 percent tended to decrease the second stage creep

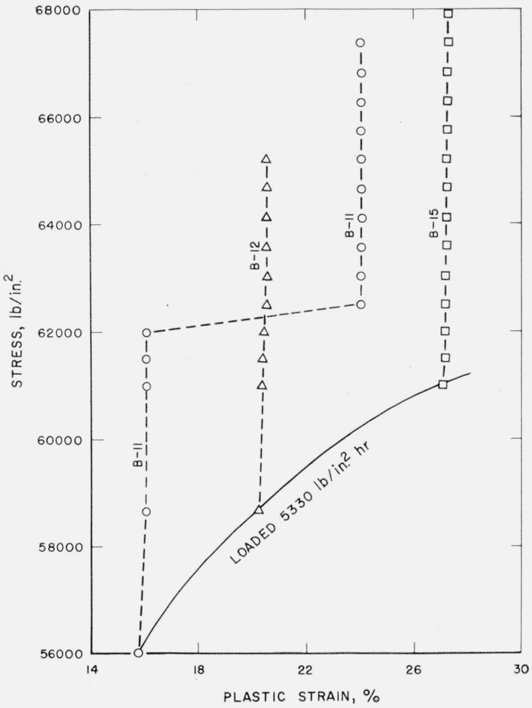


FIGURE 22. *Effect of rate of loading at 300° F on the stress-strain relations of 70-percent-*Ni*–30-percent-*Cu *alloy.*

For rate of loading schedule, see figure 21.

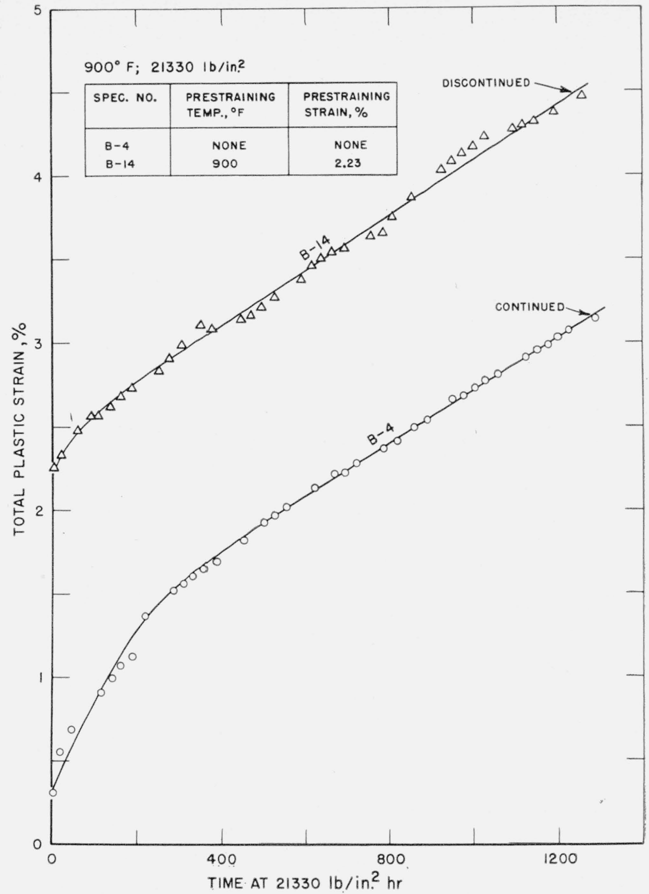


FIGURE 23. *Influence of prestraining at 900° F on the strain-time relations of specimens of 70-percent-*Ni*–30-percent-*Cu *alloy tested with a stress of 21,330 lb/in².*

For rate of loading schedule of specimen B–14, see table 5.

rate and increase the fracture time and elongation. The strain associated with the first stage of creep increased as the prestraining temperature was increased. Moreover, the first stage was practically eliminated for the specimens prestrained relatively large amounts at 300° or 700° F (B–17 and B–9).

The effect of prestraining at 900° F on the creep characteristics at 1,200° F of the 70-percent-Ni–30-percent-Cu alloy is shown in figure 25. Prestraining caused a detectable first stage, a decrease in second-stage creep rate, elongation, and fracture time at 1,200° F.

The influence of prestraining temperature and the amount of prestrain on the creep characteristics of The 30-percent-Ni–70-percent-Cu alloy tested at 900° F with a stress of 18,670 lb/in.² is shown in figure 26. Compared to the reference specimen (C–8) the fracture time was decreased and the second-stage creep rate increased for all the prestrained specimens. The maximum increase in creep rate was observed for the specimen (C–13) prestrained at 1,200° F to 10.22 percent whereas the minimum increase was observed for the specimen (C–21) prestrained at 900° F to a small strain value. Apparently, the substructures developed in the latter specimen had only a slight effect on the second-stage rate, but caused a marked decrease in fracture time and elongation. Generally, for the specimens of 30-percent-Ni–70-percent-Cu an increase in the prestraining temperature (amount of prestrain constant) was accompanied by an increase in second-stage creep rate and a decrease in values of elongation or reduction of area. Increasing the amount of prestrain at 300° or 900° F tended to cause an increase in second-stage creep rate, elongation, and reduction of area for the specimens; whereas the increase in creep rate for the specimens prestrained at 1,200° F was accompanied by a decrease in elongation.

Although prior-strain history appeared to have a pronounced effect on the subsequent mechanical properties at 900° F of the 30-percent-Ni–70-percent-Cu alloy, the properties at 300° F were practically independent of the rate of loading (tables 3 and 6). The latter observation is not in agreement with the results obtained on the 70-percent-Ni–30-percent-Cu alloy and may be due, in part, to the differences in the degree of strain aging in the metals.

b. Specimen Contour and Post-Test Hardness

The effect of prior-strain history on the necking characteristics and the post-test hardness of specimens of the 30-percent-Ni–70-percent-Cu alloy, fractured at 900° F with a stress of 18,670 lb/in.²

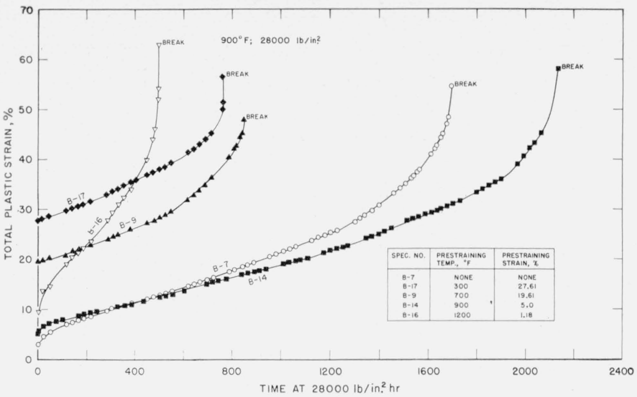


FIGURE 24. *Influence of prior-strain history on the strain-time relations at 900° F of specimens of 70-percent-*Ni*–30-percent-*Cu *alloy tested to complete fracture with a stress of 28,000 lb/in².*

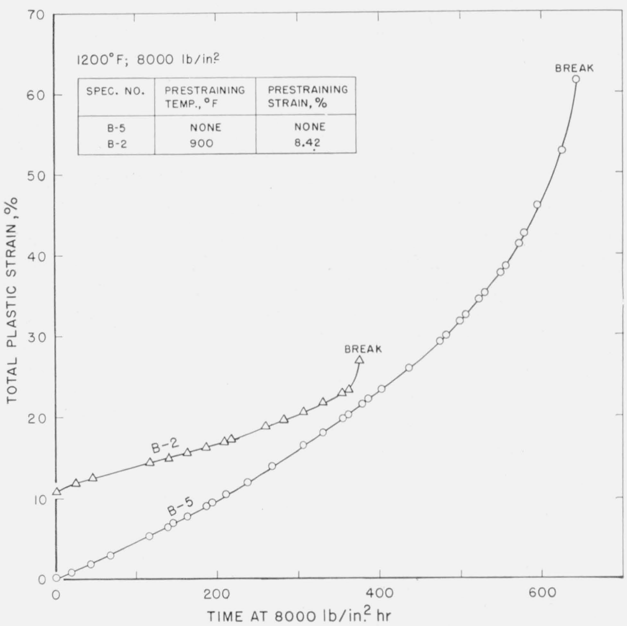


FIGURE 25. *Influence of prior strain history on the strain-time relations at 1,200° F of specimens of 70-percent-*Ni*–30-percent-*Cu *alloy tested to complete fracture with a stress of 8,000 lb/in².*

is shown in figure 27. No necking of the specimens was observed during the first or second stage of creep. Therefore, it was concluded that the specimen contours after fracture (figure 27a) were due, in part, to the different mechanisms existing in the third stage. A further examination of the data indicates that the specimen contour was affected more by prestraining temperature than by the subsequent second-stage creep rate. For example, the greatest tendency to neck was shown by specimen C–24 and the least tendency by specimen C–13. The prestraining temperature of the latter specimen was 1,200° F and the second stage creep rate ex-

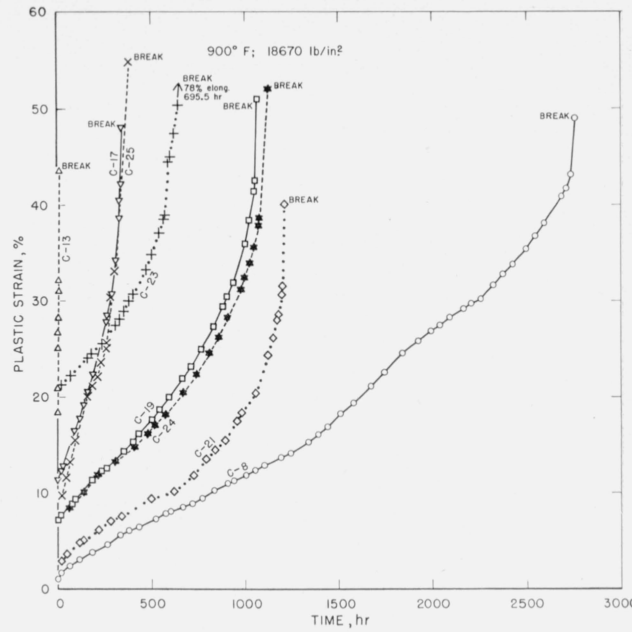


FIGURE 26. *Effect of prior strain history on the strain-time relations at 900° F of specimens of 30-percent-*Ni*–70-percent-*Cu *alloy tested to fracture with a stress of 18,670 lb/in².*

Specimen number	Prestrain temperature	Prestrain	Avg. creep rate, second stage
	° F	*Percent*	*%/1,000 hr*
C–8	None	None	9.5
C–23	300	20.7	19.2
C–24	300	7.74	18.0
C–19	700	7.33	21.5
C–17	900	10.12	57
C–21	900	0.865	11.6
C–13	1,200	10.22	905
C–25	1,200	2.374	60

ceeded that of the former specimen, which was prestrained at 300° F, by a factor of about 50. In general, prestraining at the lower temperatures increased the severity of the neck, the post-test hardness (fig. 27B), and the rate of strain hardening. It is evident that both the strain-hardening and necking characteristics of the specimens, tested under the same conditions, were strongly dependent on the initial structure, temperature, and chemical composition of the metal.

4. Summary

Creep tests were made at 300°, 700°, 900° and 1,200° F on initially-annealed specimens of copper, nickel, and 70-percent-Ni–30-percent-Cu and 30-percent-Ni–70-percent-Cu alloys. The study was extended to include an evaluation of the rate of loading and prior-strain history on the creep characteristics, hardness, and contours of specimens of the two alloys.

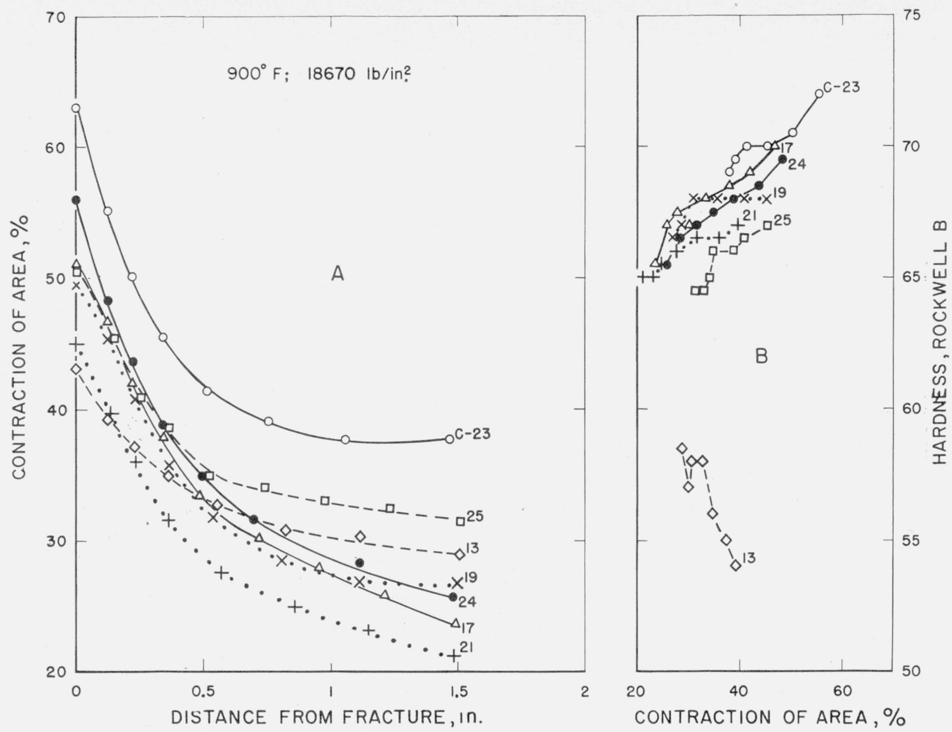


FIGURE 27. *Effect of prior-strain history on specimen contour and hardness distributions of specimens of 30-percent-*Ni–*70-percent-*Cu *alloy tested to fracture at 900° F with a stress of 18,670 lb/in².*

Refer to figure 26 for prior strain history and second-stage creep rates of the specimens.

The initial resistance to flow was increased by alloying the component metals. The maximum resistance was obtained in specimens of the 70-percent-Ni–30-percent-Cu alloy.

Conformance to both the theory of exhaustion and of generation of lattice defects during the first stage of creep was obtained over limited ranges of temperatures and stresses.

The shape of the stress-creep rate curves during the first and second stages was altered markedly as the test temperature was increased.

The stress dependence of second stage creep rate was interpreted on the basis of a hyperbolic sine law based on the chemical-rate theory. Strict conformance was limited by the relative balance between strain hardening and recovery of the specimens at the different temperatures. At the same temperature the resistance to creep in the second stage was highest for the 70-percent-Ni–30-percent-Cu alloy and lowest for the copper; however, the increase in creep resistance above that of the component metal was greater when 30 percent of nickel was added to the copper than when 30 percent of copper was added to the nickel.

The creep characteristics of the specimens conformed closely to a theory based on a Cottrell effect at low temperatures and to a theory based on recovery at high temperatures; for specimens tested at low stresses, the results were not easily explainable by either of these theories.

With the exception of the copper specimens tested at 900° F, the creep rates or fracture times of each metal were expressed in terms of single parameters which varied monotonically with values of the applied creep stress.

Elongation or reduction of area values of the alloys was intermediate between the high values for the nickel and low values for the copper.

Post-test hardness values and the rate of strain hardening of the alloys decreased and the tendency toward uniform contraction of specimens increased with increase in temperature or decrease in second-stage creep rate.

Strain aging induced by slowly loading specimens at 300° F, was more pronounced for the 70-percent-Ni–30-percent-Cu than for the 30-percent-Ni–70-percent-Cu alloy or for that previously obtained for the nickel.

In general, the shape of the creep curves, mechanical properties, and the fracture times were apparently affected by the substructure induced in the specimens by prestraining; however, all phases of Wood's theory of subgrain formation were not generally in agreement with present data.

Prestraining the specimens at the higher temperatures reduced the post-test hardness values, the rate of strain hardening, and the tendency toward a formation of an acute neck.

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