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The Bauschinger Effect and Residual Microstresses in Alpha Brass

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The Bauschinger effect in alpha brass, specifically the lowering of the yield strength in the direction opposite to the preceding plastic deformation, was studied by an examination of both tensile and compressive elastic limits measured at the quarter-cycle stages throughout a complete cycle of uniaxial plastic strain of one percent amplitude. X-ray measurements of axial residual directed microstresses indicated that the latter could be correlated with the decrease of the elastic limit measured in the direction opposite to the preceding deformation only at the first quarter-cycle of deformation. After 3 quarter-cycles the axial residual stress was always tensile regardless of the direction of previous deformation; whereas the elastic limits continue to show strong directionality.

1. Introduction

Although it is nearly 80 years since Bauschinger [1][1] first reported the effect which may be described very broadly as the anisotropic modification of the plastic properties of a metal specimen that has undergone a preceding plastic deformation, a simple quantitative explanation for the phenomenon has not yet been established. Perhaps the most common aspect of this effect is the raising of the yield strength of a plastically deformed metallic specimen in the direction of previous deformation and the lowering of the yield strength in the opposite or reverse direction. The basis of an explanation often advanced for this phenomenon is the difference in degree of susceptibility to slip in differently oriented crystal grains or regions of differing hardness within the specimen [2]. This inhomogeniety of slip is thought to give rise to a system of residual internal stresses upon release of the deforming load; and, upon reapplication of load, these residual stresses presumably modify the external stress required for yielding.

Residual stresses in metals can be calculated from measurements obtained by means of X-ray diffraction [3]. If the stresses are random in sign and magnitude in the diffracting material, the X-ray diffraction lines are observed to be broadened; if the stresses are nonrandom, the peak of the line will be shifted. This latter effect, the change in Bragg angle of diffraction after plastic deformation of a specimen, implies a nonzero net stress in the diffracting material, which is usually observed to be of such a sign as to oppose the prior deformation. S. L. Smith and W. A. Wood [4] observed such residual stresses by means of X-rays and believed them to be related to the Bauschinger effect.

Recently a detailed quantitative study was made of this type of residual "oriented microstress" and its relation to the Bauschinger effect by B. M. Rovinskiy and V. M. Sinayskiy [5]. These experimenters worked with three groups of specimens of 40X steel (Russian designation), each group having had a different thermal or strain-rate history. The specimens, after having been annealed, were compressed uniaxially by different amounts, ranging from 0.30 percent to 1.46 percent. Although the authors do not give details of their X-ray procedures, they state that back-reflection X-ray patterns of the (310) spacing (using cobalt radiation) after these prestrains enabled them to calculate the residual microstress, $\underline{\sigma_R}$, in each case. The direction of $\underline{\sigma_R}$ is not specified, but it is presumed to be axial. The conventional 0.2 percent yield strengths in compression and in tension were then determined, and the difference, $\underline{\Delta\sigma}$, between them was obtained for each degree of prestrain. Both the yield strength difference and the residual stress were found to go through a maximum at a deformation near 1 percent. The value of prestrain that resulted in these simultaneous maximums near 1 percent depended on the thermal and strain-rate history of the specimen. The ratio of $\underline{\sigma_R}$ to $\underline{\Delta\sigma}$ was found to differ from one group of specimens to another, but to vary by only a few percent within a group. These authors observed that this ratio, in most cases somewhat greater than unity, was comparable to that observed for the ratio of the X-ray stress, measured on a specimen under load, to the applied mechanical stress. They concluded that, if one keeps in mind the coefficient of correspondence between stresses determined by X-ray and mechanical methods, one can predict the magnitude of the reduction of the yield strength during reversed loading of a specimen from the size of the residual oriented microstress observed with X-rays. Thus the Bauschinger effect after uniaxial plastic deformation is, according to this picture, the result of the establishment of a residual oriented microstress that can be summed with the macrostress created by the external load. Or, in equation form,

$$\sigma_A=\sigma_0+\alpha\sigma_R,$$

where $\underline{\sigma_A}$ is the yield strength of the annealed mate-

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

rial, σ_0 is the yield strength observed in the reverse direction after a plastic deformation of the specimen, α is constant nearly equal to unity, and σ_R is the oriented axial residual microstress measured by X-ray diffraction. Or, from another viewpoint, the reduction of the yield strength is

$$\Delta\sigma = \sigma_A - \sigma_0 = \alpha\sigma_R.$$

Since the residual stresses in a specimen are balanced internally by stresses of the opposite sign, there is a further corollary implicit in this model, that is, that the region or crystals that support the additive stress observed by means of X-rays is that same region in which plastic flow first begins, thus determining the yield strength.

This article on the Bauschinger effect by Rovinskiy and Sinayskiy was of sufficient interest that it was decided to investigate the Bauschinger effect and the state of residual stress at appropriate stages in specimens that had undergone a complete cycle of plastic deformation, unlike those of Rovinskiy and Sinayskiy, which had been plastically deformed in one direction only.

2. Experimental Material, Procedure, and Results

When a metal specimen containing a residual internal stress system is sectioned, it is assumed that at the exposed surface the stress component normal to the surface vanishes and the components of stress parallel to the surface are not affected. A few years ago an X-ray study of residual stresses on various sections of plastically deformed iron by C. J. Newton and H. C. Vacher [7] yielded results consistent with this long-held hypothesis. The procedure in that study employed an X-ray beam normally incident on the sections of the specimen with the diffraction pattern recorded on photographic film. The observed lattice spacing was compared with that of the annealed material. A similar procedure was used in this study of sections of alpha brass specimens, except that the diffracted X-rays were detected by means of a diffractometer (counter) technique instead of by film.

The alpha (Cu=70, Zn=30, Pb<0.07 percent) brass specimens which served both for elastic limit measurement and for X-ray residual stress measurements, were machined from 1 by ⅜ in. bars, to be 7½ in. long with a gage length of 1¼ in. and a reduced section measuring 0.625 by 0.373 in. X-ray examination indicated no appreciable preferred orientation in the material. The plastic extension was performed on a hydraulic testing machine using Templin grips, the plastic compression was made possible by careful machining of the ends of the specimens and by the use of massive steel guides around the specimen undergoing compression. During the course of interrupted straining, resistance wire strain gages were used to follow the amount of strain. The routine of the mechanical testing and strain measurements, along with a discussion of the difficulties that sometimes arise when one attempts to measure very small permanent sets, and hence elastic limits, with resistance gages, has been previously reported [8]. The elastic limit was taken to be the stress required to give the specimen a permanent set of 2×10^{-5}, as indicated by gages not previously subjected to large plastic strains.

The values of the limit for various conditions of the brass specimens are listed in table 1. It is difficult to specify precisely the reliability of these values, most of which are the average of several observations. The precision of the testing machine is high, but variations from one specimen to another may be large. A reasonable estimate of average uncertainty might be about $\pm1\times10^3$ psi.

TABLE 1. *Elastic limits and oriented residual stresses*

Plastic deformation state of specimen			Elastic limits in 10^3 psi		Axial residual stress in 10^3 psi
Description	Net plastic prestrain (%)	Cum. plastic prestrain (%)	In tension	In compression	
Annealed	0	0	14	13	(Reference zero).
Positive cycle (initial deformation extension)					
E	+1	1	15	3	−9
E–C	0	2	4	16	+2
E–C–C–	−1	3	5	20	+5
E–C–C–E	0	4	20	3	+7
Negative cycle (initial deformation compression)					
C	−1	1	4	16	+8
C–E	0	2	17	3	−3
C–E–E–	+1	3	----------	----------	+4
C–E–E–C	0	4	----------	----------	+8

E=Extended plastically 1%.
C=Compressed plastically 1%.
Average uncertainty approximately $\pm1\times10^3$ psi.
Different specimens were used for the X-ray residual stress measurements (last column) than were used for the elastic limit determinations.

One can see in table 1 that essentially symmetrical results of elastic limits with respect to plastic extension or compression were obtained when the directions of limit determination and prestrain are taken into account. The effect of plastic deformation was to raise slightly the elastic limit measured in the same direction as the deformation and to lower by a large amount the limit measured in the opposite direction; this is in accord with the Bauschinger effect. When a second plastic deformation, reversed in direction, was made, the elastic limits were affected by roughly the same amounts relative to the values pertaining to the original material in the annealed state, but reversed in direction. This demonstrated that it is the direction of the immediately preceding deformation that is of major importance with regard to the elastic limits. The third step in the positive cycle of plastic deformation, at a minus 1 percent net strain, showed a small increase in both tensile and compressive elastic limits. And finally, the last step of the cycle, an extension of 1 percent to reverse the net strain to zero, resulted in a strong reversal of the elastic limits, as expected.

Elastic limit determinations were made during only the first two steps in the negative cycle of deformation, since the results were obviously symmetrical with those of the positive cycle.

Longitudinal and crosssections were cut from the reduced sections of new test specimens at quarter-cycle stages of one complete cycle each of positive and negative 1 percent plastic deformation. The sections were mounted without heat in acrylic cement and carefully mechanically polished, electro-polished, and etched until all surface disturbance was removed. The X-ray diffraction measurements were made on a commercial horizontal diffractometer with cobalt $K\alpha_1$ characteristic radiation, using a proportional counter as a detector. Each specimen was continuously rotated about an axis normal to its surface in order to bring more grains into diffracting position. The position of the peak of the (400) line, near the 2θ diffraction angle of 153°, was determined by the analytical method of three-point parabola fitting of D. P. Koistinen and R. E. Marburger [9], as described in the SAE Information Report TR–182 [10]. Corrections were made in the intensity at each point for the Lorentz and polarization factors, and all values of the calculated Bragg angle were corrected for specimen temperature variation. The precision of peak determination was approximatly $\pm 0.02°$ in 2θ.

The value of the crystal lattice spacing observed normal to the surface of a given section compared to that of the annealed material enabled one to calculate the sum of the principal stresses parallel to the surface of the section. Because of the uniaxial manner of deformation, it was assumed that one principal stress σ_A, was axial in the specimen and two equal principal stresses, each σ_T, were at right angles to the axis. Hence, measurements on the cross section yielded a value, $2\sigma_T$, twice the transverse stress, and those on the longitudinal section yielded $(\sigma_A+\sigma_T)$, the sum of the axial stress plus the transverse stress. Therefore, the axial residual stress was calculated for each pre-strain condition by subtracting one-half the sum of the stresses found in the cross section from the sum of the stresses found on the longitudinal section. That is, $\sigma_A=(\sigma_A+\sigma_T)-\frac{1}{2}(2\sigma_T)$. The resulting values of axial residual stress are listed in table 1. Because of the involved nature of the calculation and the many possible sources of error, only an estimate of average uncertainty, $\pm 1\times 10^3$ psi, is expressed as a measure of precision of the results.

The behavior here again shows a kind of symmetry between the cycle of plastic deformation begun with an axial extension and the cycle begun with axial compression. After the first 1 percent plastic deformation, a residual axial stress is observed that is opposite in sign from the deformation; that is to say, after extension, a residual compressive stress is observed, and, after compression, a residual tensile stress is observed. That is the common behavior that has often been reported. After the second quarter step in the deformation cycles, the observed stress, though smaller in magnitude, has reversed sign so that it continues to be directed opposite to the deformation. After the third and fourth quarter steps, however, the residual stress is tensile in all cases, regardless of direction of deformation. This behavior is shown graphically in figure 1.

These residual stresses, resulting in a change in the X-ray diffraction peak position, are oriented microstresses, indicating a nonzero average component of stress present in the coherently diffracting material of many grains of the specimen in the X-ray beam. Another type of microstress, in which the residual stresses may be random in magnitude and sign as one goes from grain to grain, results in a broadening of the diffraction line. Two other factors in the specimen may increase line breadth: random microstress that varies within each grain (microstress of the "third kind," as it is sometimes called) and very small particle size. Previous study of individual diffraction spots on stationary film patterns made with a monochromatic misorientation goniometer [6], using some of the same specimens, however, revealed that neither of these factors were significant under the conditions of this investigation.

Line breadth measurements were made on the (400) diffraction line of the brass using cobalt $K\alpha_1$ radiation, employing the diffractometer with a 3° beam slit, a 0.1° receiving slit, Soller slits, scanning rate of 0.2° per min, rate meter time constant of 8 sec, recording chart speed of 0.4 in. per min, and specimen spinner in use. Since the $\alpha_1-\alpha_2$ doublet was not fully resolved in these patterns, one-half the breadth of the low-angle side of the α_1 line was measured at half-peak intensity and multiplied by two. Measurements on the cross and longitudinal sections of a specimen were very nearly equal and were averaged together. The observed breadth was corrected for instrumental broadening by the method

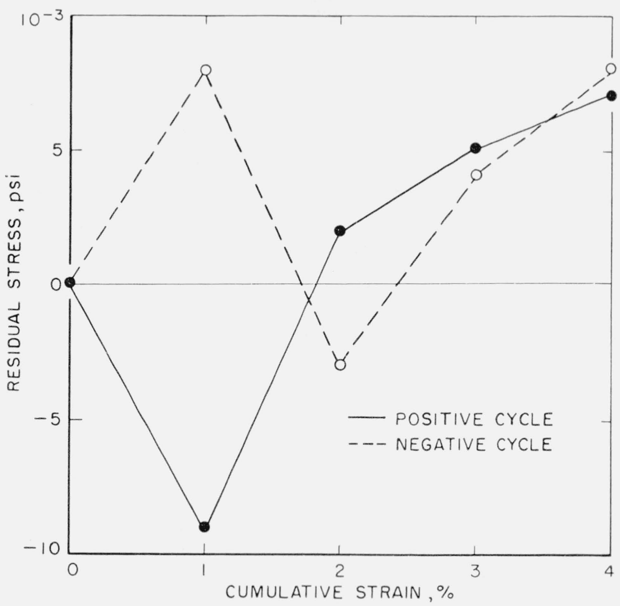

FIGURE 1. *Axial oriented residual stress versus cumulative plastic strain through one complete cycle of deformation.*

described by H. P. Klug and L. E. Alexander [11] for high-angle diffractometer lines. The maximum random microstress, without regard to sign, is given, according to B. D. Cullity [12], by

$$\sigma_{Mx}=\frac{E}{4\tan\theta}\,(\beta)$$

where σ_{Mx} is the maximum random tensile or compressive microstress, E is Young's modulus of the material, θ is the Bragg angle of diffraction, and β is the corrected line breadth. The values of β and σ_{Mx} are to be found in table 2. The precision in β is estimated to be about $\pm 0.05°$, which is reflected as an uncertainty of $\pm 0.7\times 10^3$ psi in σ_{Mx}.

It is apparent that the maximum random microstress is of the same order of magnitude as the residual directed microstresses. One may also see that there is no significant difference in the random stress level in the annealed specimen and those of the first two quarter steps of both the deformation cycles, after which the stresses rise markedly.

3. Discussion

The essential conclusion of Rovinskiy and Sinayskiy was that the Bauschinger effect can be explained by summing the residual directed microstress, determined by X-ray diffraction, with the externally applied macrostress to account for the lowered elastic limit measured in the direction opposite to a preceding plastic deformation (which we shall call the "reverse elastic limit"). The results of this study clearly demonstrate that this conclusion cannot with validity be extended to those cases where the immediately preceding deformation has itself been preceded by a plastic deformation in the opposite direction. This inequality of the residual stress and the decrease of the reverse elastic limit, $\Delta\sigma$, beyond the first quarter-cycle is presented graphically in figure 2, where both quantities are plotted against the cumulative strain. Throughout at least the first four quarter-cycle steps of a cycle of deformation, the Bauschinger effect, as can be seen in table 1, depends to a very high degree upon only the immediately preceding deformation. The effect of earlier deformations seems to be merely to increase slightly the difference between the forward and reverse elastic limits. An entirely different situation exists with regard to the X-ray determined residual axial stresses. As can be seen in table 1 and figure 2, only the first quarter-cycle shows an axial residual stress that approximates the Bauschinger decrease in reverse elastic limit. The residual stress at the second quarter-cycle, $+2\times 10^3$ psi, agrees in sign but not in magnitude with the decrease, $+10\times 10^3$ psi, in the limit. Clearly something other than the immediately preceding deformation is beginning to have an important effect. After the third and fourth quarter-cycles of deformation, the direction effect of the immediately preceding deformation is increasingly masked by this cumulative factor. At this point, specimens in the positive cycle and the negative cycle are indistinguishable on the basis of the axial residual stresses observed, whereas the Bauschinger effect, that is, the directionality in elastic limit, is undiminished.

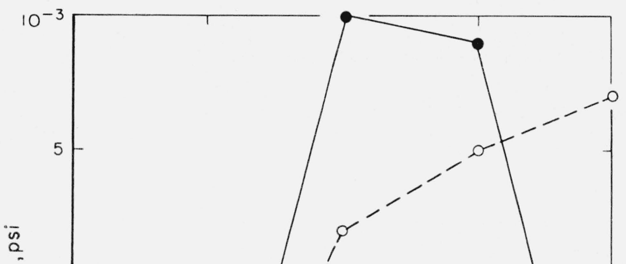

FIGURE 2. *Decrease in elastic limit and axial oriented residual stress versus cumulative plastic strain through one complete cycle of deformation.*

σ_A=Elastic limit of annealed material.
σ_o=Elastic limit observed in reverse direction after deformation.
σ_R=Axial oriented residual microstress measured by X-ray diffraction.

The validity of the various methods of calculating residual stresses from X-ray measurements continues to be open to some question. A new approach has recently been used by D. M. Vasil'ev [13,14] to compute the internal stress system from measurements on sections. He assumes as usual that the stress components lying in the plane of the cut are not affected by the cut. He does not make the usual assumption, however, that the component normal to the surface of the cut is relaxed to zero; but rather that, in the equations relating strain to stresses in this region reached by X-rays, it appears multiplied by a factor k, to be evaluated, which lies between zero and one. He computes k from data obtained from additional X-ray measurements. His resulting values of k vary typically from 0.2 in the case of aluminum (99.99 percent pure) examined with Cr Kα radiation to 0.7 in the case of Fe_3C examined with Cr Kα radiation. A surprising result of his measurements and calculations is that all principal stresses, both axial and transverse, after either uniaxial extension or compression, are always compressive. This is obviously in conflict with the usual result that the residual axial stress is opposite to the direction of deformation, as was used by Rovinskiy and Sinayskiy for the basis of their explanation of the Bauschinger effect. It is also in conflict with our result that, after cyclic axial deformation, the residual axial stress becomes tensile regardless of the direction of the last axial deformation.

No attempt was made in this study to evaluate k by the somewhat complex procedure of multiple inclined incidence X-ray exposures of Vasil'ev. However, calculations were made of the axial stresses using a k of 0.4, Vasil'ev's value for copper with cobalt radiation. These calculations yielded results containing both tensile and compressive stresses, at least qualitatively similar to those obtained by the accepted procedure and listed in table 1. These newer values were not added to the table, however, because they did not seem to add appreciably to the results already presented, and, moreover, there does not appear to be adequate evidence that the value of 0.4 used for k was more nearly correct than the usual value zero.

While the directed residual stresses seem gradually to lose any relationship with the elastic limits as the plastic strain cycle passes the three-quarter mark, the values of the maximum random microstress calculated from diffraction line breadth measurements and listed in table 2 show an abrupt change after the third quarter step in the deformation cycle. Since no such marked change in the Bauschinger effect is observed at this point, it seems clear that the latter, at least in the cyclic case, is the result of some factor or factors other than either of these types of microstresses revealed by X-ray diffraction.

TABLE 2. *Results of line breadth measurements*

Plastic condition of specimen	Corrected line breadth $\beta(2\theta^\circ)$	Max. random microstress σMx (10^3 psi)
Annealed	0.35	5.2
Positive cycle of deformation		
E	0.30	4.5
E–C	.35	5.2
E–C–C	.50	7.5
E–C–C–E	.55	8.2
Negative cycle of deformation		
C	0.35	5.2
C–E	.35	5.2
C–E–E	.50	7.5
C–E–E–C	.50	7.5

E=extended 1%. C=compressed 1%.
Average uncertainty in β approximately $\pm 0.05^\circ$.
Average uncertainty in σMx approximately $\pm 0.7 \times 10^3$ psi.

It may be that the Rovinskiy and Sinayskiy model of the part played by the oriented microstress is correct for the first deformation of an annealed material; but that, as the material is subjected to further strain of a cyclic nature, another type of residual stress arises that is always tensile in character and increasing, perhaps asympotically, in magnitude, while that associated with the Bauschinger effect is outweighed or masked, after the third quarter stage, altogether. The rise of this conjectural new type of residual stress is also indicated at this point in the cycle by the increase in line broadening. The Bauschinger effect, in the meantime, is not affected at all; the elastic limits are as strongly directional at the end of a cycle of deformation as at the beginning. One must conclude, therefore, that an explanation of the Bauschinger effect that is valid in all cases must be based on some model other than the additive action of those types of residual microstress that are revealed by X-ray diffraction effects of peak shift or broadening.

The plastic properties of metals in general are being explained with increasing success on the basis of the dislocation theory of behavior as exemplified in the work of A. H. Cottrell [15], N. F. Mott [16], and many others. Some application of the theory has been made to the problem of the mechanism of the Bauschinger effect, as for example in the work of R. L. Wooley [17] and that of S. N. Buckley and K. M. Entwistle [18]. A critical review of these theories is beyond the scope of this paper; it appears, however, that there does not yet seem to be a theory sufficiently complete to enable one to predict quantitatively the behavior of the forward and reverse elastic limits of a cyclically deformed polycrystalline specimen such as used in this study. No doubt the complexity of the physical situation in experiments such as this with polycrystalline brass is such that it would be difficult, if not impossible, to apply dislocation theory quantitatively and with rigor. It has been suggested that the Bauschinger effect is a result of the nonsymmetry of the stress potential associated with a dislocation arrangement after flow, as, for example, arising from a pile-up against obstacles on the slip plane or from the interaction of other dislocations on parallel slip planes. If this is true, the elastic stresses in the crystal lattice would probably be quite localized, on the scale of perhaps 10^2 or 10^3 interatomic distances; whereas the stresses associated with X-ray measurements are averaged over much larger dimensions. Even if the tool of X-ray microstress measurement were sufficiently sensitive to reveal the nature of a localized dislocation configuration after an initial deformation of the order of 1 percent, it should probably not be surprising that this information would become masked by more gross structural effects and would cease to be a major contributor to the X-ray microstress after the cumulative effects of reversed plastic strain past the middle of a strain cycle became important in the specimen.

4. Summary and Conclusions

The principal experimental facts reported in this study may be summarized as follows:

(a) The elastic limits indicated a strong Bauschinger effect in alpha brass throughout a complete cycle of axial plastic deformation of 1 percent amplitude.

(b) There was no marked change in the effect as the cycle progresses; only a slight hardening was apparent.

(c) The axial oriented residual microstress was not additive to the applied stress to account for the Bauschinger effect in the latter part of the cycle; in fact, after the third-quarter stage, the microstress appears always to become positive in sign (tensile).

(d) The random maximum microstress (X-ray line broadening) remains near the value found in the annealed material until it exhibits a marked change at the three-quarter cycle condition of the specimen.

These facts point to the following conclusions:

(a) The effect of the cyclic deformation on the elastic limits seems to require two mechanisms (at least): one to account for the forward limit hardening; a second to account for the large reduction of the reverse limit. Neither mechanism would appear to be strongly affected by the cumulative strain of one complete cycle of 1 percent amplitude of plastic deformation.

(b) Since both the oriented and the maximum random microstresses measured by X-ray techniques show a marked change in character after three-quarters of a cycle of deformation while the Bauschinger effect does not, they are not directly related to the Bauschinger effect, at least under these conditions of reversed plastic deformation.

(c) The cyclic uniaxial working of this material leads to a residual stress in the axial direction. Determined by X-ray diffraction, this is a positive (tensile) stress.

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