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# Fatigue Stability of Residual Stress in Shot Peened Alloys

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Fatigue specimens of Ni-20 pct Co and Ni-60 pct Co alloys were shot peened to an Almen intensity of 0.008N2 and cycled at two levels of alternating stress. The surface residual stress was monitored during cycling by the two exposure X-ray method. Microhardness surveys were made across cross sections of specimens representing each test condition. It was found that the surface residual stress decayed partially during the plastic cycling of the lower strength-higher stacking fault energy (SFE) 20 pct Co alloy and that little such decay occurred in either the plastic or elastic cycling of the higher strength-lower SFE 60 pct Co alloy. There was virtually no change of microhardness during cycling in either alloy, hence the residual stress relaxation in the 20 pct Co alloy is interpreted as a dynamic recovery process.

SHOT peening generally improves fatigue resistance because of residual compressive stresses created by peening in the surfaces of fatigue specimens. Subsequent plastic alternating strains, however, will reduce this residual stress. Earlier studies[1,2] have shown that the hardness of a material, as measured by static compressive or tensile flow stress, can decrease during fatigue cycling. More recent studies[3,4] have shown that broadened X-ray lines from a cold worked material sharpen with subsequent fatigue cycling of the material. Compressive residual stresses induced by shot peening decrease both in surface and subsurface layers as a function of fatigue stress and surface depth.[5,6]

The current study sought to ascertain what effects SFE and alternating stress have on surface residual stress induced by shot peening in two Ni-Co alloys. Thus, the aim of our study was not fatigue *softening* per se, as referred to above (decrease of hardness), but the stability or instability of residual elastic stress. However, since residual elastic stress produced by shot peening exists only as a consequence of plastic (local) deformation, factors which influence softening or in general any dislocation movement, will be relevant to present considerations.[7,8]

The alloys studied were 20 pct Co-80 pct Ni and 60 pct Co-40 pct Ni. The SFE of the 20 pct Co alloy is 183 ergs per sq cm (0.183 N/m) and the SFE of the 60 pct Co alloy is 39 ergs per sq cm (0.039 N/m).[9] The X-ray two angle peak shift method of measuring residual surface stress provided a nondestructive method of monitoring this parameter as a function of the number of fatigue cycles.

## I. EXPERIMENTAL PROCEDURES

### 1.1 Specimen Preparation

Table I. Composition of Alloys (Wt Pct)

| Heat | Ni | Co | C | Al | Mg |
|---|---|---|---|---|---|
| 59809 | Bal. | 19.5 | 0.004 | 0.08 | 0.011 |
| 59810 | 41.4 | Bal. | 0.016 | 0.08 | 0.011 |

Table II. Shot Peening Parameters for Almen Intensity 0.008N2

| Parameter | |
|---|---|
| Shot Diameter | 0.0032 to 0.0046 cm |
| Air Pressure | 25 psi (172,700 N/m$^2$) |
| Coverage | 200 pct |
| Time per Piece | two min |
| Nozzle Size | 0.95 cm |
| Angle of Peening | 90 deg |
| Peening Distance | 10.16 cm |

The compositions of the nickel-cobalt alloys used are presented in Table I. These alloys were prepared by the International Nickel Company. To facilitate discussion, heat number 59809 will be referred to as the 20 pct Co alloy and heat number 59810 will be referred to as the 60 pct Co alloy.

Cantilever fatigue specimens were prepared from 0.203 cm thick sheet material according to Sonntag Scientific Corporation Drawing Number 90881-S. After shaping the specimens, all surfaces were finished with 2/0 emery polishing paper and annealed at 600°C for 1 h in an argon atmosphere. Electropolishing removed any oxide scale formed during the stress-relief anneal. Subsequently, both alloys were glass-bead shot peened to an Almen intensity of 0.008N2.

Parameters used to attain the 0.08N2 Almen intensity are listed in Table II. Using the recrystallization techniques of Valentine,[10] the residual stress layer was measured and found to be 0.0064 cm deep for both alloys.

### 1.2 X-Ray Diffraction Method

Residual stress measurements were made using the two-exposure X-ray diffractometer technique.[11] In the current work chromium $K_\alpha$ radiation diffracted from the $\{200\}$ crystallographic planes provided the best resolution and peak heights in the high $2\theta$ region between 133 and 134 deg $2\theta$.

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Theoretically the stress constant, $K$, is a function only of Young's modulus, the Bragg angle, Poisson's ratio, and the angle of rotation, $\psi$, during measurement. However, this relation does not allow for anisotropy in polycrystalline materials, hence, it is generally preferable to obtain $K$ experimentally by elastically straining an annealed material by known increments and measuring the corresponding change in Bragg angle. Thus, stress always will be proportional to $\Delta 2\theta$ for a given crystallographic reflection and $\psi$ rotation. Application of only elastic stress is important since uniaxially applied plastic stress produces anomalous residual stress when measured with X-rays.[12,13]

### 1.3 Testing Procedures

The X-ray stress measurements were performed on a General Electric XRD-5 diffractometer equipped with residual stress specimen stage and a proportional counter mounted on a radial track support assembly. Vanadium filtered Chromium $K_\alpha$ radiation (50 kV, 25 mA) was used. All X-ray residual stress measurements were made with fatigue specimens at normal incidence ($\psi$ = 0°C), and with the specimen rotated 40 deg from normal incidence ($\psi$ = 40°C). Prior to each experiment, the X-ray tube and radial track were aligned. Measurements were made on an annealed sheet of 60 pct Co alloy to determine standard deviation of the data.

The stress constant, $K$, was obtained by elastically straining an annealed 1 × 3 in. strip of each alloy at known strains and measuring the corresponding ($2\theta_{\psi=0} - 2\theta_{\psi=40}$) difference. The elastic strains, measured with a Magnaflux Series TA-06-250BB strain gage, were applied with a four-point bending jig which was capable of centering the strips on the X-ray goniometer axis (Fig. 1). The indicated strain was converted to stress using Hooke's Law.

Before fatigue cycling on a Sonntag SF-2-U constant-force fatigue machine, the residual stress in each specimen was measured. Additional stress measurements were performed at 100, 500, 1000, 5000, 10,000, 50,000, 200,000, 400,000, and 2,000,000 cycles as well as after failure of each specimen. To minimize any effects of fluctuations in X-ray power level, measurements on each specimen between 0 and 400,000 cycles were obtained during a single day. Maximum fiber stresses of ±35,000 and ±45,000 psi ($\pm 241 \times 10^6$ and $\pm 310 \times 10^6$ N/m$^2$) were calculated from elasticity and imposed during fatigue cycling, and the rate of cycling was 1800 cycles per minute.

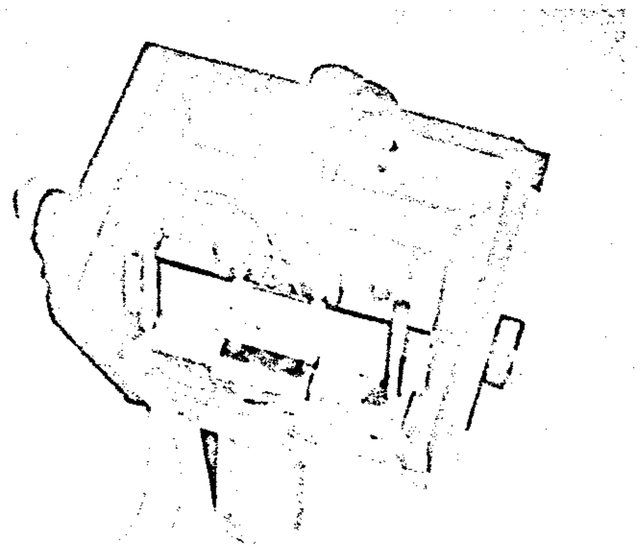

Fig. 1—Four point bending jig capable to centering and straining calibration strips on the axis of the X-ray goniometer. Reduced to 50 pct of original size.

Determination of the $2\theta$ diffraction peak positions were performed using the 3-point parabola method of Koistinen and Marburger.[14] A Fortran program which received uncorrected inverse intensity data, performed appropriate Lorentz-Polarization factor and specimen absorption corrections, determined the $2\theta$ diffraction peak positions at $\psi = 0$ and $\psi = 40$ deg, and calculated surface residual stresses, was used on a Univac 1108 computer. The inverse intensity data consisted of averages of ten 100,000 unit counts.

## II. RESULTS

### 2.1 Stress Factor Determination

The stress factors, $K_{20}$ and $K_{60}$, obtained from the least square slopes of the plot in Fig. 2 are 177,840 psi per deg $2\theta$ ($1211.8 \times 10^6$ N/m$^2$ per deg $2\theta$) and 320,600 psi per deg $2\theta$ ($2210.4 \times 10^6$ N/m$^2$ per deg $2\theta$) for the 20 pct Co and 60 pct Co alloys, respectively.

### 2.2 Standard Deviation of X-ray Residual Stress Measurements

X-ray residual stress measurements on the annealed 60 pct Co sheet produced a standard deviation

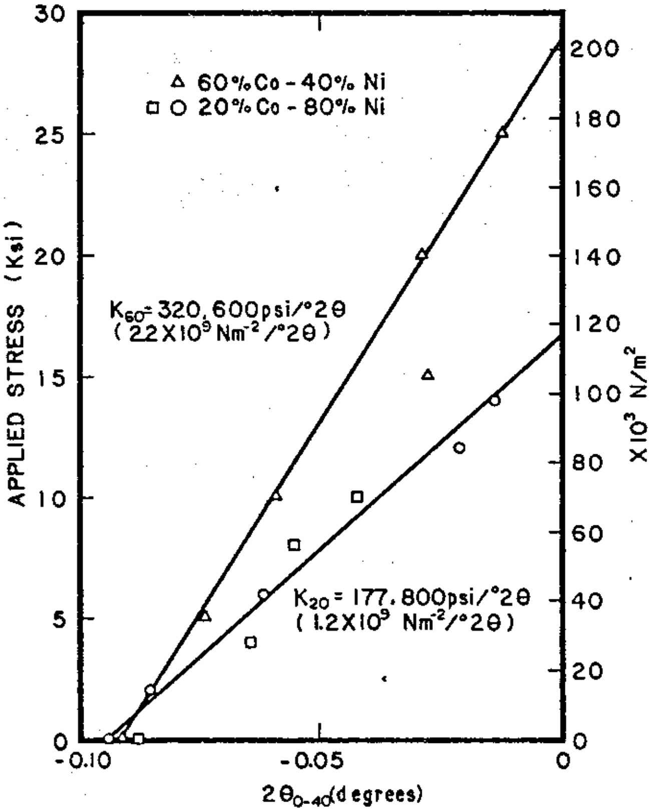

Fig. 2—Plot of applied stress vs angular position of {220} diffraction peak during calibration of alloys.

Table III. Residual Stress* Data

| Fatigue Specimen No. | Alternating Stress (ksi) (Calculated) | Initial Surface Residual Stress (psi) | Average Surface Residual Stress at Saturation (psi) | Change in Surface Residual Stress (psi) | Pct Change | Failure (Cycles) |
|---|---|---|---|---|---|---|
| 20 pct Co–80 pct Ni | | | | | | |
| F-2 | 45 | −62,000 | −24,000 | −38,000 | −61.4 | 76,000 |
| F-5 | 45 | −60,000 | −38,000 | −22,000 | −36.8 | 99,000 |
| F-6 | 35 | −21,000 | −+5,000 | −26,000 | −124.0 | 2,191,000 |
| F-7 | 45 | −37,000 | −−5,000 | −32,000 | −86.1 | 76,000 |
| F-8 | 35 | −40,000 | −21,000 | −19,000 | −47.5 | No Failure |
| F-9 | 45 | −52,000 | −34,000 | −18,000 | −34.6 | 354,000 |
| F-10 | 45 | −48,000 | −32,000 | −16,000 | −33.2 | 129,000 |
| 60 pct Co–40 pct Ni | | | | | | |
| F-2 | 35 | −121,000 | −123,000 | +2,000 | +1.7 | No Failure |
| F-4 | 45 | −119,000 | −104,000 | −15,000 | −12.6 | No Failure |
| F-5 | 45 | −129,000 | −119,000 | −10,000 | −7.5 | No Failure |

*1 psi = $6.89 \times 10^3$ N/m$^2$.
1 ksi = 1000 psi.

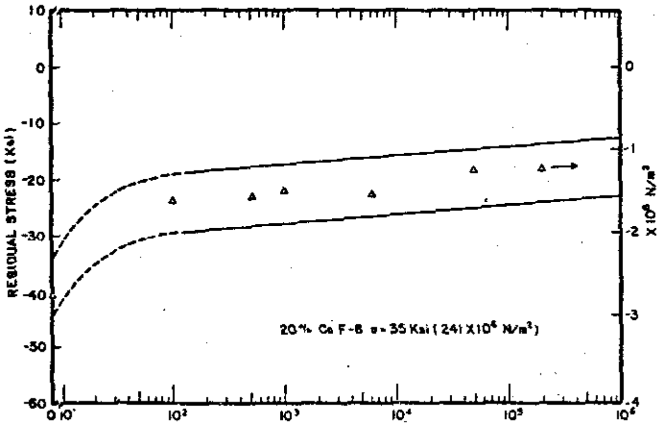

Fig. 3—Residual stress as a function of cycles at ±35,000 psi (±241 × $10^6$ N per sq m) for the 20 pct Co alloy.

of $\Delta 2\theta_{(0-40)} = \pm 0.03$ deg $2\theta$ for the series of stress measurements. This deviation corresponds to stress deviations of ±5300 psi ($\pm 36.5 \times 10^6$ N/m$^2$) and ±9600 psi ($\pm 66.2 \times 10^6$ N/m$^2$) for the 20 pct Co and 60 pct Co alloys, respectively.

### 2.3 Residual Stress and Hardness

The residual stress data resulting from cycling the shot-peened fatigue specimens are given in Table III. The method of plotting is illustrated by Fig. 3. The "Change in Surface Residual Stress" values in Table III are the difference between the average plateau residual stress values and the initial residual stress values. The negative and positive signs in Columns 3 and 4 of Table III indicate compressive and tensile residual stresses, respectively.

Fig. 3 depicts the surface residual stress data for the 20 pct Co fatigue specimen F-8 cycled at an alternating stress of 35,000 psi ($241 \times 10^6$ N/m$^2$). The bands represent the standard deviation of ±5300 psi ($\pm 36.5 \times 10^6$ N/m$^2$) while the arrow indicates a fatigue life of greater than $10^6$ cycles. The uncertainty of surface residual stress behavior between 0 and 100 cycles is represented by the dashed portion of the standard deviation line. Stress relaxation is evident in that the magnitude of the surface residual stress decreases as a function of cycling. For this particular specimen, the surface residual stress decreased 19,000 psi ($131 \times 10^7$ N/m$^2$) with the greatest percentage of this decrease occurring during the first 100 cycles. The slope of the six points in the saturation or plateau region indicate stress relaxation is continuing with cycling although not at as rapid a rate as in the first 100 cycles.

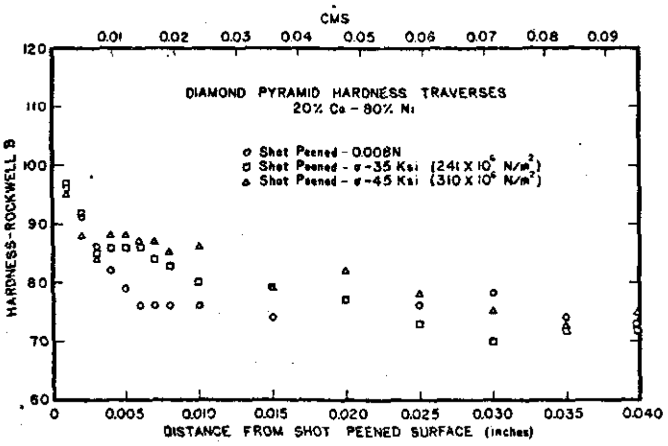

Fig. 4—Diamond pyramid hardness vs depth below peened surface for the 20 pct Co alloy before and after fatigue.

The data for the 20 pct Co specimens cycled at the higher stress level of 45 ksi ($310 \times 10^6$ N/m$^2$) also show the occurrence of fatigue caused stress relaxation.

Residual stress data for the 60 pct Co fatigue specimens cycled at alternating stresses of 35 ksi ($241 \times 10^6$ N/m$^2$) and 45 ksi ($310 \times 10^6$ N/m$^2$) showed little stress relaxation during fatigue compared with the 20 pct Co alloy. It is of interest that none of the three 60 pct Co specimens failed in cycling through $2 \times 10^6$ cycles. The F-5 specimen was cycled through $11.7 \times 10^6$ cycles at 45,000 psi ($310 \times 10^6$ N/m$^2$) without failure.

Diamond pyramid hardness (DPH) traverses (0.100 kg load) were made perpendicular to shot-peened surfaces on cross sections of fatigue specimens in both cycled and uncycled regions. Plots of traverses are presented in Figs. 4 and 5. In Fig. 4 the uncycled condition for the 20 pct Co alloy shows a hardness decrease with increasing depth. It is apparent that shot peening affected the material surface to a depth of 0.006 in. (0.015 cm) which is deeper than first determined by metallography. The traverses for cycled specimens are identical to a depth of 0.003 in. (0.008 cm), indicating that a change in hardness at the surface did not occur due to fatigue cycling. Of interest is the region between depths of 0.003 in. (0.008 cm) and 0.010 in. (0.025 cm) where the hardness of the cycled condition is higher than for the uncycled condition. From the 0.010 in. (0.025 cm) depth to the approximate center of each specimen (0.035 in., 0.089 cm), the traverses approach a common hardness.

Hardness traverses for the 60 pct Co fatigue speciments are presented in Fig. 5. Again the traverse for the shot-peened uncycled condition shows an affected layer 0.006 in. (0.015 cm) in depth. Traverses of the cycled and uncycled specimens are similar. The absence of deviation between the three traverses indi-

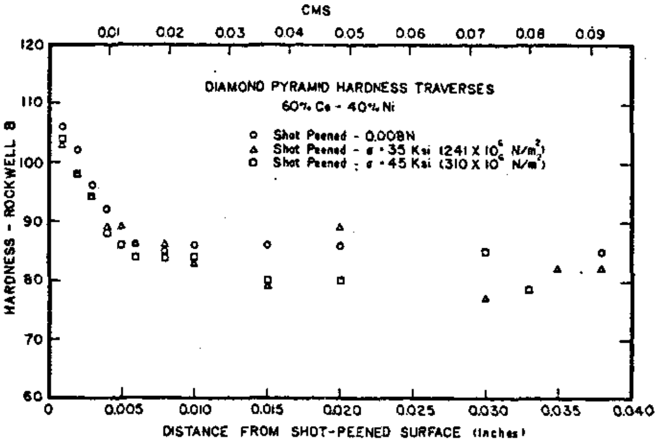

Fig. 5—Diamond pyramid hardness vs depth below peened surface for the 60 pct Co alloy before and after fatigue.

cates fatigue cycling did not affect the specimen hardness in the 60 pct Co alloy.

In the 20 pct Co alloy, fatigue stress relaxation occurred without sensible heating during cycling. Furthermore, identical hardness traverses were obtained from the surface to a depth of 0.003 in. (0.008 cm) on cycled and uncycled specimens. The reduction in surface residual stresses *without* a simultaneous reduction in hardness indicates that stress relief, similar to that which occurs during recovery annealing, occurred. Recovery during fatigue cycling has been reported by others in that X-ray lines,[3,4,15] broadened by cold working, became sharper with cycling. An interesting feature of the current results is that the recovery occurs only in the 20 pct Co alloy. This point will be discussed in terms of SFE and yield stress levels.

## III. DISCUSSION

It is important to point out that the yield stresses of the two alloys (without shot peening) are 18 and 43 ksi[18] for the 20 pct and 60 pct Co alloys, respectively ($124 \times 10^6$ N/m$^2$ and $296 \times 10^6$ N/m$^2$). Thus for the 20 pct Co alloy, both applied alternating stresses were above the (unpeened) yield point. For the 60 pct Co alloy one alternating stress (35 ksi) was below the yield stress and the other alternating stress (45 ksi) was above the yield stress. Shot peening should have had little effect on the yield stress of either alloy, because as is seen in Figs. 4 and 5, the effects of the peening were local to the surfaces. Another problem, however, is introduced by the now known fact[18] that the alloys (unpeened) fatigue harden so as to give Rockwell B hardness increases of 33 pct and 29.4 pct for the 20 pct and 60 pct Co alloys, respectively, when cycled to failure at an alternating stress of 45 ksi. Thus it is not surprising that the alternating plastic strains at both applied stress levels caused "shake down" or reduction in residual stress. In this regard, however, it is to be noted that for the 60 pct Co alloy little change of residual stress occurred even at the higher alternating stress of 45 ksi.

In addition to the fact that the two alloys have different yield points, we would like to suggest that another parameter of importance here is the stacking fault energy.

The 20 pct and 60 pct Co alloys have SFE values[9] of 183 and 39 ergs per sq cm, respectively (0.183 N/m and 0.039 N/m). This difference is in a direction such as to make dislocation motion, whether by planar slip, cross-slip, or even climb, much easier in the 20 pct Co alloy than in the 60 pct Co alloy. The influence of SFE on work hardening in the Ni-Co system has been studied by several groups[16,17,18] and the outcome has been that the rate of work hardening increases with increasing percentages of Co or decreasing SFE. This energy therefore has a strong influence on the work hardening which accompanies each of shot peening and fatigue. It is qualitatively understandable then that both the yield stress before peening and the residual stress after peening are higher in the 60 pct Co (lower SFE) alloy. The relaxation of long range residual stress also should be facilitated by easier dislocation motion; *i.e.*, in the 20 pct Co alloy.

In Cu which has a SFE of about 40 ergs per sq cm[19] (0.040 N/m), softening occurs at strain amplitudes producing failure in less than $10^4$ cycles.[20] At similar strain amplitudes, Cu-7.5 pct Al, a material in which the SFE is about 1.7 ergs per sq cm[20] (0.0017 N/m), experiences incomplete softening. Copper, interestingly, is a wavy slip mode material while Cu-7.5 pct Al is a planar slip mode material. Considering the current alloys, the 20 pct Co alloy shows far more metallographic evidence[21] of cross slip than does the 60 pct Co alloy although both alloys are more nearly planar slip than wavy slip type materials.

Neither alloy showed an appreciable change in microhardness during cycling in the outermost 0.01 cm. According to Manson and Hirschberg[22] any material with a ratio of U.T.S./Y.S. > 1.4 will cyclically harden rather than soften. In the unpeened condition both alloys have ratios of U.T.S./T. S. far in excess in 1.4,[18] and more to the point, are known to fatigue harden.[18] If then, due to shot peening, the outer regions of a specimen attain the stress level of the U.T.S., then further fatigue hardening in that location cannot occur. It is quite possible, however, that some subsurface fatigue hardening can occur, especially in the softer 20 pct Co alloy. This would account for the subsurface hardening noted in Fig. 4 between 0.008 and 0.025 cm.

Since we find relaxation of residual stress in the absence of any hardness decrease in the 20 pct Co alloy, it is appropriate to describe the process as dynamic recovery in the same sense as the recovery produced in stress relief anneals prior to recrystallization. It is to be noted in this connection that there were no significant changes in microstructure which would preclude a recovery interpretation.

## IV. CONCLUSIONS

1. Residual stress caused by shot peening was stable in a low SFE alloy of Ni-60 pct Co during cycling at an alternating stress slightly above the yield stress of the unpeened alloy.
2. Residual stress caused by shot peening decayed in a high SFE Ni-20 pct Co alloy, during cycling at alternating stresses above the yield point of the unpeened alloy. This is a dynamic recovery process since no significant hardness or microstructural changes occur.
3. The probable causes of the decay of residual stress

in the high SFE alloy are the greater ease of dislocation motion and the fact that cyclic stresses far exceeded the yield stress in this alloy.

4. The most interesting direction for future work on this problem would be to study the effect of alternating stress on the residual stress in the 20 pct Co alloy at a level well below the yield stress of that alloy.

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## REFERENCES

1. H. N. Polakowski: *Proceedings of the ASTM,* 1952, vol. 52, pp. 1986-97.
2. H. N. Polakowski and A. Palchoudhuri: *Proceedings of the ASTM,* 1954, vol. 54, pp. 701-12.
3. W. A. Wood and R. L. Segall: *Proceedings of the International Conference on Fatigue of Metals,* Institute of Mechanical Engineers, New York, 1956, pp. 225-28.
4. R. C. Baggerly and R. M. N. Pelloux: in *Advances in X-Ray Analysis,* pp. 236-48, J. B. Newkirk, G. R. Mallett, and C. S. Barrett, eds., Plenum Press, New York, 1969.
5. S. Taira and Y. Murakami: *Proceedings of the Third Japan Congress on Testing Materials,* Japan Society for Testing Materials, Kyoto, Japan, 1960, pp. 52-55.
6. A. L. Esquivel and K. R. Evans: *Exp. Mech.,* 1968, vol. 8, pp. 496-503.
7. R. K. Ham and T. Broom: *Phil. Mag.,* 1962, vol. 7, pp. 95-103.
8. M. Klesnil and P. Lukas: *J. Iron Steel Inst.,* 1967, vol. 205, pp. 746-49.
9. B. E. P. Beeston, I. L. Dillamore, and R. E. Smallman: *Met. Sci. J.,* 1968, vol. 2, p. 12.
10. K. B. Valentine: *Trans. ASM,* 1948, vol. 40, pp. 420-34.
11. A. L. Christenson, D. P. Korstein, R. E. Marburger, M. Semchyshen, and W. P. Evans: *SAE Technical Report 182,* SAE Inc., New York, 1960.
12. B. D. Cullity: *J. Appl. Phys.,* 1964, vol. 35, pp. 1951-57.
13. R. E. Ricklefs and W. O. Evans: in *Advances in X-Ray Analysis,* pp. 273-83, J. B. Newkirk and G. R. Mallett, eds., Plenum Press, New York, 1967.
14. D. P. Koistinen and R. E. Margurger: *Trans. ASM,* 1959, vol. 51, pp. 537-55.
15. S. Taira and K. Honda: *Proceedings of the Fifth Japan Congress on Testing Materials,* p. 8, Japan Society for Testing Materials, Kyoto, Japan, 1962.
16. V. A. Pavilov, N. U. Noskova, and R. I. Kuznestsov: *Phys. Metals Metallogr.,* 1967, vol. 24, pp. 171-89.
17. C. K. L. Davies, P. W. Davies, and B. Wilshire: *Phil. Mag.,* 1965, vol. 12, pp. 827-39.
18. C. M. Wan: Ph.D. Thesis, University of Utah, 1973.
19. D. A. Avery and W. A. Backofen: *Fracture of Solids,* p. 339, D. P. Drucker and J. J. Gilman, eds., Interscience Publishers, New York, 1962.
20. C. E. Feltner and C. Laird: *Acta Met.,* 1967, vol. 15, pp. 1621-33.
21. B. Souther and J. G. Byrne: unpublished results, 1965, B. Southern, c/o Stevens Institute of Technology, Hoboken, New Jersey; J. G. Byrne, Division of Materials Science and Engineering, University of Utah, Salt Lake City, Utah.
22. S. S. Manson and M. Hirschberg: *Fatigue—An Interdisciplinary Approach,* p. 133, Syracuse University Press, Syracuse, New York, 1964.