The Mechanical Properties of Zero-Carbon Ferrite-Plus-Martensite Structures

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A study has been made of an Fe-3Ni-3Mo alloy whose structure can be either ferrite or martensite or any combination of ferrite plus martensite. By being able to measure the mechanical properties of the ferrite and martensite phases separately it is possible, using Milieko's theory for composites of two ductile phases, to calculate the properties expected for ferrite-plus-martensite mixtures. This theory assumes: 1) the stress-strain relationship for all structures is a power law; 2) a perfect interface between the phases; and 3) the fibers are aligned parallel to the stress axis. The experimentally determined tensile strength and ductility of the two-phase structures are in good agreement with the theory even though the martensite is not in the form of aligned fibers. The ductilities obtained at tensile strengths greater than 620 MPA (90 ksi) are no better than those for conventional HSLA steels because of the low ductility of the ferrite and low strength of the martensite.

1. INTRODUCTION

In the last few years a new type of steel, consisting of about 15 to 20 pct martensite in a ferrite matrix, has become of technological interest because at a constant tensile strength it exhibits greater formability than a standard high-strength low-alloy (HSLA) steel.[1-5] For example, when VAN-80 (a Jones & Laughlin Steel Co. HSLA steel) is obtained in the martensite plus ferrite condition, the yield stress decreases from 550 to 380 MPa (80 to 55 ksi) while there are increases in the total elongation from about 18 to 26 pct and in the strain hardening exponent, n, from about 0.12 to 0.22; the tensile strength remains approximately constant, independent of structure, at 700 MPa (100 ksi).[2-4] These steels with a martensite plus ferrite structure are often called dual-phase steels.[1-5]

Mileiko[6] and Garmong and Thompson[7] have developed a theory of the mechanical properties of fiber composites of two ductile phases tested in tension parallel to the fiber axis; good agreement between experiment and theory was found for Ni-W, Cu-W and Ag-stainless steel composites.[6,7] It has been shown[4,5] that the changes in tensile strength and ductility of dual phase steels as a function of the percent martensite are in agreement with this theory even though the martensite is not in a fibrous form. The values of the tensile strength and ductility of the component ferrite and martensite used to compare experiment with theory were obtained by extrapolation; a major problem in measuring the properties of the phases individually is that high-carbon martensites (>0.3 pct C), as found in dual-phase VAN-80, are very notch sensitive and have very limited tensile ductility. In addition there may be movement of alloying elements into and out of the ferrite and thus its composition is not easily determined.

To more directly check the applicability of the composite theory to dual-phase alloys a zero carbon Fe-3Ni-3Mo alloy was chosen for investigation. This alloy can be obtained in either the full martensitic or fully ferritic states[8] or in the dual-phase condition. This makes it possible to measure the properties of the components of the dual-phase steel (zero carbon martensites are quite ductile[9]), and thus to compare theory and experiment.

2. THEORY

Mileiko[6] was the first to develop a theory relating the strength and ductility of the two ductile components to the mechanical properties of the composite; the theory is based upon the application of plastic instability criteria to the composite. There are two assumptions as to materials properties in the theory: 1) the bond between fiber and matrix is ideal, so that the strength of the interface is sufficient to prevent the fiber necking without necking of the composite as a whole, that is, the more stable matrix restrains the less stable fiber. The interface in dual-phase steel is atomic in nature and therefore meets this requirement. 2) the relationship between stress and strain for both the composite and components is a power law of the form

$$\sigma = k\epsilon^n \quad [1]$$

where σ is the true stress, ϵ the true strain and k and n are constants. It has already been demonstrated that dual-phase steels[2-5] and ferrites[10] follow reasonably well such a power law. Thus dual-phase steels fulfill the materials criteria for the applicability of the theory to experimental data, although they do not fulfill the geometric requirement of fibers aligned parallel to the tensile axis.

Following Mileiko, the equations relating σ_c and ϵ_c, the tensile strength and true uniform strain of the dual-phase steel, to the volume fraction, V, of the martensite, given σ_m and σ_f, the tensile strength of martensite and ferrite respectively, and ϵ_m and ϵ_f, the true uniform strain of the martensite and ferrite respectively, are as follows:

For uniform strain

$$V = \frac{1}{1 + \beta \dfrac{\epsilon_c - \epsilon_m}{\epsilon_f - \epsilon_c} \epsilon_c^{\epsilon_m - \epsilon_f}} \quad [2]$$

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where

$$\beta = \frac{\sigma_m}{\sigma_f} \frac{\epsilon_f^{\epsilon_f}}{\epsilon_m^{\epsilon_m}} \frac{\exp \epsilon_m}{\exp \epsilon_f}.$$

For tensile strength

$$\sigma_c = V \lambda' \sigma_m + (1 - V)\lambda'' \sigma_f \qquad [3]$$

where

$$\lambda' = (\epsilon_c/\epsilon_m)^{\epsilon_m} \exp(\epsilon_m - \epsilon_c)$$

and

$$\lambda'' = (\epsilon_c/\epsilon_f)^{\epsilon_f} \exp(\epsilon_f - \epsilon_c).$$

3. EXPERIMENTAL PROCEDURE

The composition of the vacuum cast alloy is, in wt pct, 3.0 Ni, 3.0 Mo, 0.7 Mn, 0.3 Si, 0.05 V and 0.01 C with balance iron; the ingot was broken down by hot rolling at 1100°C to a slab 1 cm (0.40 in.) thick. After cold rolling approximately 50 pct it was found that this alloy could be recrystallized to a fully ferritic structure by annealing at 750°C for 1 h; thus by alternative cold working and annealing it was possible to obtain a 1 mm (0.04 in.) thick strip. Tensile samples with a 50 mm (2 in.) gage length and 12.5 mm (0.5 in.) wide were cut from the cold-rolled strip prior to final heat treatment. All tensile testing was carried out at room temperature in an Instron machine at a cross head rate of about 2×10^{-2} mm/s (0.05 in./min).

To produce the required structures two basic heat treatments were employed: 1) isochronal—all specimens were recrystallized to the fully ferritic condition at 750°C and then individual specimens were annealed for 1 h at temperatures starting at 760°C and increasing to 940°C in 20°C intervals; metallographic, hardness and strength measurements showed that it is possible to partially reaustenitize to produce a two phase, $\alpha + \gamma$, region between 760 and 870°C;2) isothermal–the knee of the isothermal transformation curve was found to be about 650°C (Ref. 8) and so a series of specimens were quenched from 950°C to this temperature; it was found that the α-to-γ reaction was initiated within 2 h at 650°C and was complete in about 10 h. Following the completion of all heat treatments the specimens were quenched into brine.

4. RESULTS

a) Metallography

Figure 1 shows, for the isothermal heat treatment, the full range of structures; the structure as quenched from 950°C (Fig. 1(*a*)) is a typical lathe martensite, while Figs. 1(*b*) and (*c*) show mixed martensite plus ferrite structures and Fig. 1(*d*) is the fully ferritic condition with 10 μm equiaxed grains. A similar series of micrographs was obtained from the isochronally heat treated samples. The spots in Fig. 1(*c*) are an etching artifact within ferrite grains which was used as a guide to distinguish between ferrite and martensite in mixed structures. Point counting was used to estimate the martensite content.

b) Mechanical Properties

Plots of log true-stress against log true-strain for the ferrite, martensite and dual-phase structure, Fig. 2 are straight lines with slope n, confirming that all the structures obey Eq. [1]; thus one requirement of Mileiko's theory is fulfilled. There were no Lüders strains or yield-point complications because all the specimens had smooth stress-strain curves; the small vanadium addition effectively getters the residual carbon in the alloy. Stress-strain data for all the structures was plotted as in Fig. 2 and n values were measured.

A further check on the validity of the power law was made by comparing n with the true uniform strain for each specimen, Fig. 3; n equals true uniform strain if the power law of Eq. [1] is obeyed.[11] Figure 3 shows that there is a good correlation between n and true uniform strain except at low n values, that is, when martensite contents were in excess of approximately 65 pct. In the high martensite alloys the elongation is less than expected; these alloys have low n values and thus any geometric inhomogeneities in the sheet specimens can lead to necking and premature failure. In the calculations to compare theory with experiment, the n values were used in place of the true uniform strain.

The value of n for the fully ferritic condition, 0.24, is considerably less than value of 0.31 found for pure iron;[10] this is probably a consequence of the substitutional alloying additions. However, the n value of 0.06 for the fully martensitic alloys is the same as that found in simple Fe-Mn-C alloys[5] and in highly alloyed Fe-Ni-Cr alloys;[9] thus n for martensite appears to be independent of alloy content.

The flow stress at various strain levels is shown in Figs. 4 and 5 as a function of percent martensite. It can be seen that to a first approximation the strength at all strains is directly proportional to the percent martensite; this agrees with results from dual-phase VAN-80[3,4] and Fe-Mn-C.[5] The 0.2 pct flow stress of the ferrite is about twice that found by extrapolation for VAN-80 ferrite;[4] this increase is no doubt due to solid solution strengthening by Ni and Mo.[12] However, the 0.2 pct flow stress for 100 pct martensite, 700 MPa (100 ksi), is in good agreement with the value obtained by Leslie and Sober[13] for low-alloy carbon-free martensite.

5. COMPARISON WITH THEORY

Utilizing the measured values of tensile strength and n for ferrite and martensite, the corresponding values for martensite-plus-ferrite mixtures were calculated from Eqs. [2] and [3]. The line through the tensile strength data in Fig. 4 is that predicted by the theory. It can be seen that there is very good agreement between theory and experiment for the tensile strength. A plot of n *vs* percent martensite, Fig. 6, shows quite good agreement between theory and experiment, with the experimental points being consistently below the theoretical line.

Mileiko's theory provides a good description of the variation of tensile strength and ductility with percent martensite, even though the martensite is not in the

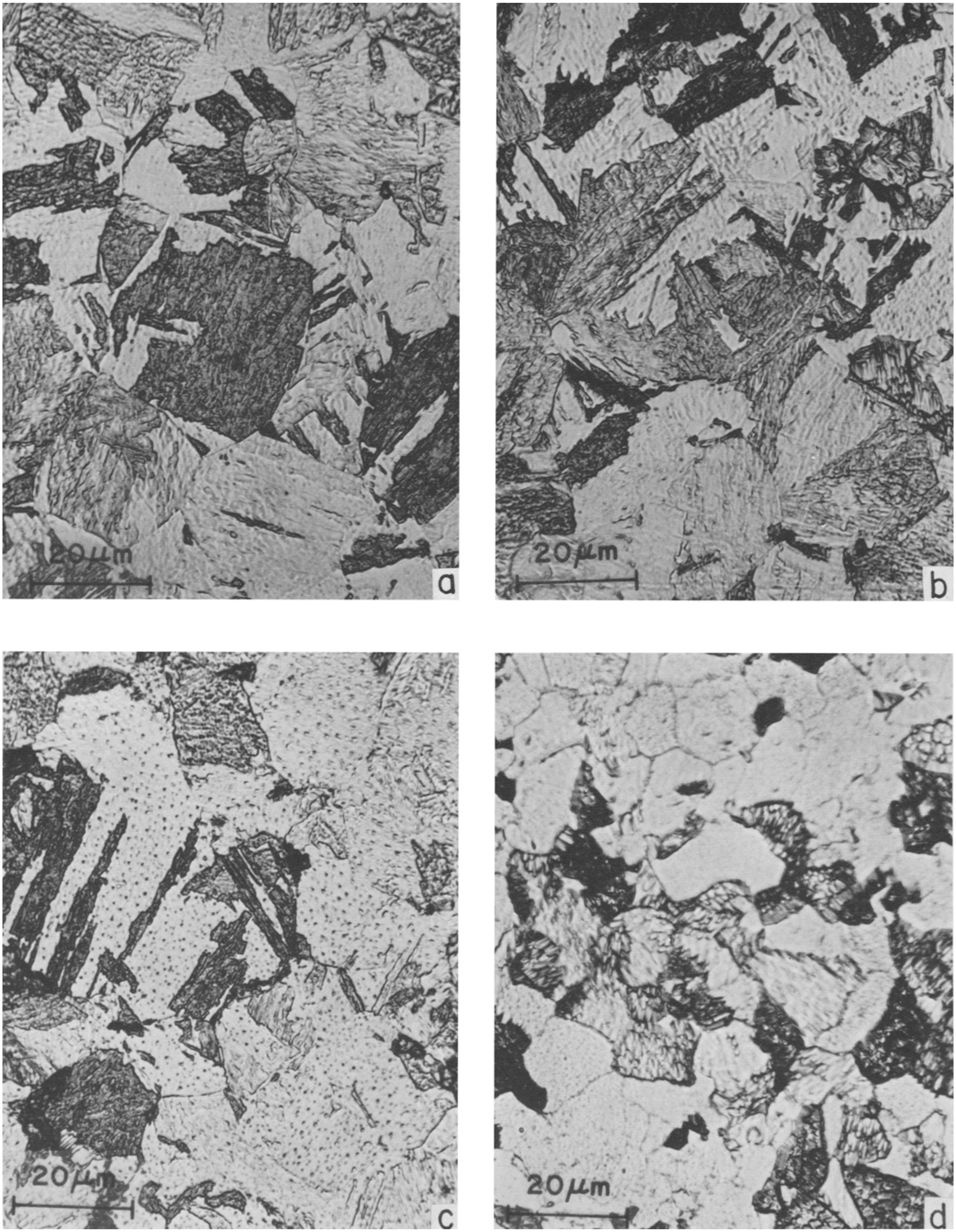

Fig. 1—Fe-3Ni-3Mo alloy isothermally annealed at 650°C: (*a*) 0 h—fully martensite; (*b*) 2 h; (*c*) 4 h—partially transformed; and (*d*) 16 h—fully ferritic.

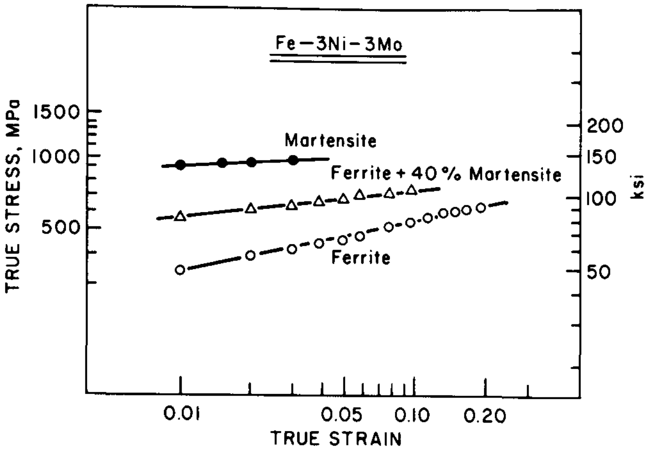

Fig. 2—A plot of log true stress against log true strain for three conditions of Fe-3Ni-3Mo alloy showing that they obey a power law.

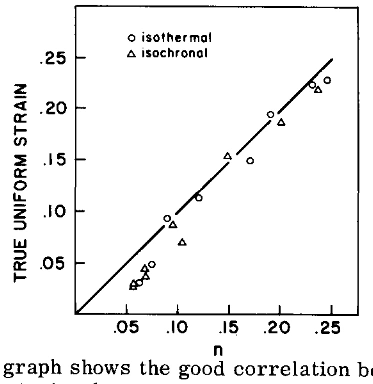

Fig. 3—This graph shows the good correlation between n and true uniform strain; the solid line is for perfect correlation.

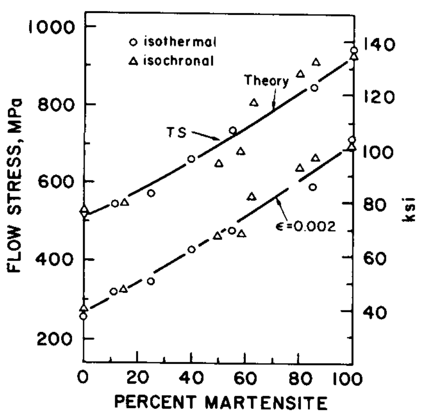

Fig. 4—Tensile strength and flow stress at 0.002 strain as a function of percent martensite; the line through the tensile strength (TS) data is calculated from Mileiko's theory.[6]

form of fibers aligned parallel to the tensile axis. Normally, misaligned or low-aspect-ratio (length/diameter) fibers quickly fracture and thus do not support their share of the load;[14] this is important if the fibers are considerably less ductile than the matrix. The ductility, as measured by reduction in area, of low or zero carbon martensites is essentially the same as for pure ferrite.[9,15] In addition the martensite islands are under hydrostatic pressure because of the volume expansion when austenite transforms to martensite; this could lead to a significant increase in fracture stress and ductility.[16]

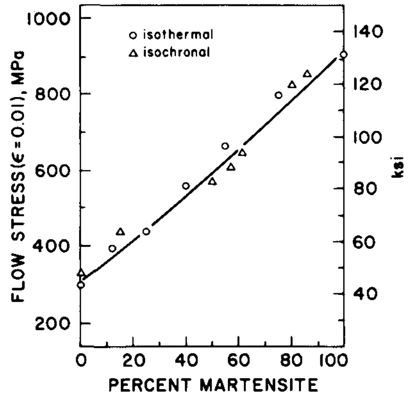

Fig. 5—The flow stress at 0.01 strain as a function of percent martensite.

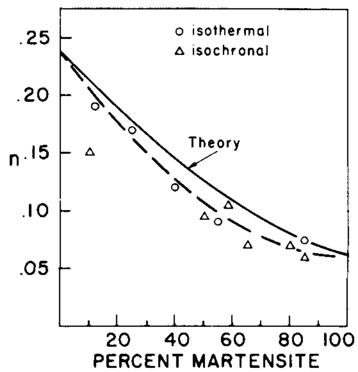

Fig. 6—Stress exponent n as a function of percent martensite compared to the theoretical prediction.

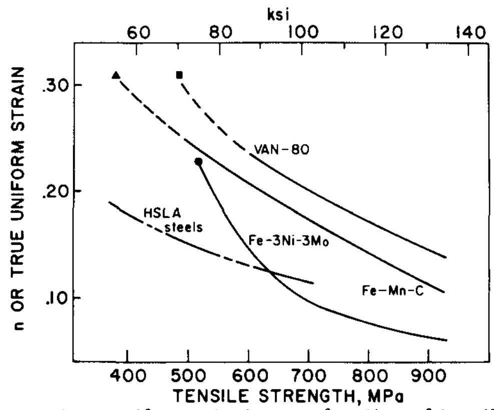

Fig. 7—n or true uniform strain as a function of tensile strength for three dual-phase steels, VAN-80, Fe-Mn-C and Fe-3Ni-3Mo, and conventional HSLA steels.

6. STRENGTH-DUCTILITY RELATIONSHIP

It has been shown previously[1-5] that at all strength levels dual-phase steels have greater ductility than conventional HSLA steels. In Fig. 7 n, or true uniform strain is plotted against tensile strength for three dual-phase steels, VAN-80,[4] Fe-Mn-C,[5] and Fe-3Ni-3Mo, and conventional HSLA steels;[2,4] VAN-80 and the Fe-Mn-C dual-phase steels are clearly superior in ductility at all strength levels. However, at tensile strengths above 620 MPa (90 ksi) conventional HSLA steels have greater elongation than the Fe-3Ni-3Mo dual phase steel. This illustrates the important point that obtaining a steel with the dual-phase structure of ferrite plus martensite does not guarantee better ductility at a given tensile strength.

In earlier papers[4,5] it was suggested that VAN-80 and Fe-Mn-C dual-phase steels have superior ductility at a given tensile strength because they are composed of a special ferrite and high-strength martensite. The ferrite is special in that it has high strength, mainly by virtue of its very fine grain size, and the high ductility (n = 0.31) of low-interstitial iron;[10] normally the higher the strength the lower the ductility. In the Fe-3Ni-3Mo dual phase structures the ferrite is of high strength but with low ductility (n = 0.23), and the martensite has a low tensile strength (940 MPa (136 ksi)). Thus the present results confirm the need for the special ferrite and strong martensite.

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REFERENCES

1. S. Hayami and T. Farukawa: *Proceedings Microalloying Conference*, pp. 78-87, Washington, D.C., 1975.
2. M. S. Rashid: SAE Preprint 760206, February, 1976.
3. J. H. Bucher and E. G. Hamburg: SAE Preprint 770164, February, 1977.
4. R. G. Davies: *Met. Trans. A*, 1978, vol. 9A, pp. 41-52.
5. R. G. Davies: *Met. Trans.*, in press.
6. S. T. Mileiko: *J. Mater. Sci.*, 1969, vol. 4, pp. 974-77.
7. G. Garmong and R. B. Thompson: *Met. Trans.*, 1973, vol. 4, pp. 863-73.
8. A. J. McEvily, R. G. Davies, C. L. Magee, and T. L. Johnston: *Transformation and Hardenability in Steels Symposium*, pp. 179-91, Climax Molybdenum Corporation, Ann Arbor, Mich., 1967.
9. S. Floreen: *Trans. TMS-AIME*, 1966, vol. 236, pp. 1429-40.
10. W. B. Morrison: *Trans. ASM*, 1966, vol. 59, pp. 824-46.
11. *See for example*: J. D. Lubahn and R. P. Folgar: *Plasticity & Creep of Metals*, p. 113, J. Wiley & Sons, Inc., New York, N.Y. 1961.
12. W. P. Rees, B. E. Hopkins, and H. R. Tipler: *J. Iron Steel Inst.*, 1954, vol. 177, pp. 93-110.
13. W. C. Leslie and R. J. Sober: *Trans. ASM*, 1967, vol. 60, pp. 459-84.
14. A. Kelly and W. R. Tyson: *Proc. 2nd Berkeley Conf. on High Strength Materials*, V. F. Zackay, ed., pp. 578-99, J. Wiley & Sons, Inc., New York, N.Y., 1964.
15. W. C. Leslie and R. J. Sober: *Trans. ASM*, 1967, vol. 60, pp. 99-111.
16. P. W. Bridgman: *Studies in Large Plastic Flow & Fracture*, Harvard Press, Cambridge, Mass., 1964.