

Communications

On the Deformation Behavior of Dual-Phase Steels

LUIS F. RAMOS, DAVID K. MATLOCK, AND GEORGE KRAUSS

Low carbon steels processed to replace pearlite by martensite exhibit a decrease in yield strength, an absence of discontinuous yielding, and an increase in ultimate tensile strength when compared to the same steels prepared with a ferrite-pearlite microstructure.¹⁻⁴ Generally, the desirable mixture of ferrite and other phases is produced by intercritical annealing, *i.e.* heating between A_{c1} and A_{c3} , and steels so treated have been identified as dual-phase steels. Despite the name dual-phase, it is now recognized that a combination of three phases, ferrite, martensite, and retained austenite,^{5,6} may provide the best structure for applications where the good formability and strength of dual-phase steels are of considerable interest.

A complete understanding of the deformation behavior of the complex dual-phase structures has not yet been developed, but a number of important observations have been made. The low initial flow stress and continuous yielding behavior in the dual-phase steels is attributed to the austenite to martensite transformation within the constraints of the ferrite matrix,^{1,3} and micrographs of high dislocation densities adjacent to martensite areas have been presented.³ The tensile strength and martensite volume fraction (MVF) in dual-phase steels have been related by the "rule of mixtures"^{1,4,7} and both total elongation² and the strain hardening parameter n^4 have also been found to depend upon the amount of martensite. Several investigators, as noted above, have found retained austenite by X-ray diffraction and transmission electron microscopy and emphasize the possible importance of retained austenite in controlling the overall deformation behavior of dual-phase steels.

This note presents data that shows that the stress-strain curves of dual-phase Fe-C-Mn-Si steels cannot be characterized by a single value of n , the strain hardening exponent, and suggests that the deformation of dual-phase steels may occur in several well-defined stages. It is also shown that the effects of MVF on deformation behavior of dual-phase steels cannot be isolated from those of carbon content of the martensite.

The composition of the steels used in this study are given in Table I. Table II lists the intercritical heat treatments and the microstructural parameters of MVF and pct C in the martensite. The grain size of all samples was 10 μm . The steels were selected to

Table I. Composition (Wt. Pct) of the Steels, Balance Fe

Code	C	Mn	Si	S	V	Ti	P	Al	N
HT-1	0.052	0.90	<0.01	0.020	<0.002	0.005	0.013	0.006	0.0038
HT-6	0.110	0.78	0.28	0.015	<0.002	0.003	0.012	0.074	0.0049

Table II. Summary of Heat Treatments*

Sample	Steel	Heat Treat. Temp., °C	Mart. Volume Fraction, Pct	Pct C in Martensite
A	HT-1	745	6	0.64
B	HT-1	780	10	0.42
C	HT-1	800	15	0.30
D	HT-1	820	25	0.24
E	HT-1	840	35	0.15
F	HT-6	745	15	0.64
G	HT-6	780	25	0.42
H	HT-6	800	35	0.30

*All samples were heat treated for 6 min and iced brine quenched. Constant grain size = 0.010 mm.

permit variation of either MVF or the pct C in the martensite while the other parameter was held constant. The microstructural parameters listed in Table II were predicted from the phase diagram and the volume fraction of second phase was confirmed with metallography. Tensile specimens with a reduced gage length of 31.25 mm and a cross-section of 6.25×2 mm were machined prior to heat treatment. All tensile tests were performed at room temperature on an Instron at a strain rate of $6.7 \times 10^{-4} \text{ s}^{-1}$.

The observed stress-strain data, as shown in Fig. 1, change with both MVF and pct C in martensite. Figures 2 and 3 show that the standard tensile properties are a function of both MVF and pct C in martensite. Also from Fig. 1, it is apparent in specimens with a constant MVF, that the initial strain hardening characteristics depend on the pct C in martensite, *i.e.* the martensite strength and/or morphology.

In general, the deformation behavior of a multi-phase microstructure is characterized by the deformation of each phase plus the interaction between the phases. In dual-phase steels, the deformation behavior

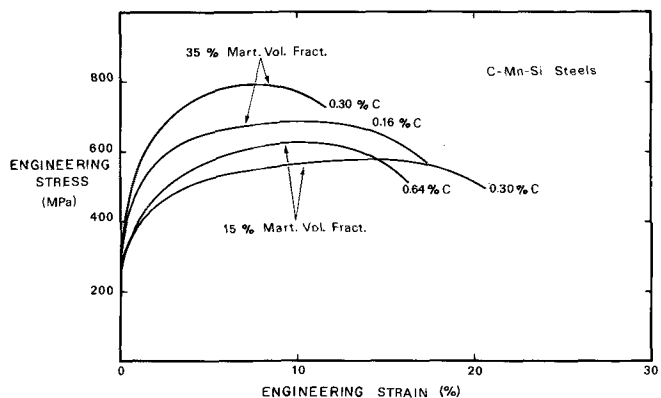


Fig. 1—Engineering stress-strain curves of Fe-C-Mn-Si steels with dual-phase structure. Martensite carbon contents and martensite volume fractions as shown.

LUIS F. RAMOS, DAVID K. MATLOCK and GEORGE KRAUSS are Graduate Student, Associate Professor, and AMAX Professor of Physical Metallurgy, respectively, Department of Metallurgical Engineering, Colorado School of Mines, Golden, CO 80401.

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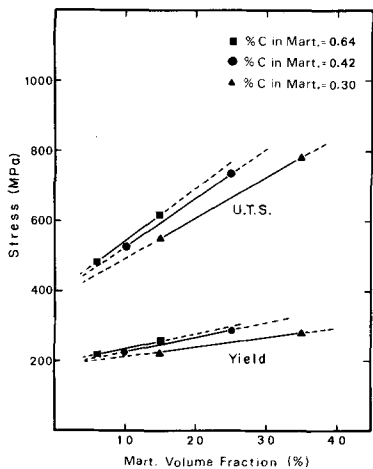


Fig. 2—Yield and tensile strengths as function of martensite volume fraction for Fe-C-Mn-Si steels with different martensite carbon contents.

is expected to depend on the deformation of ferrite, of martensite, of retained austenite, if present, and on the compatibility between the phases. Furthermore, the observed behavior should depend on the relative strengths and volume fractions of each phase. For example, for a constant MVF, as the pct C in martensite is increased, it is anticipated that the strain distribution between the ferrite and martensite will change, thus altering both initial yielding and strain hardening at low strains. Note that the pct C in the ferrite depends directly on the pct C in the martensites. Although variations in pct C in ferrite have been shown to affect microyielding in Fe-C alloys,⁸ the data in Fig. 2 which shows an increase in yield strength with increase in pct C in martensite (which corresponds to a decrease of pct C in ferrite) implies that the effects of the pct C in martensite and MVF overshadow variations in ferrite strength due to carbon. The balance of this note shows, via analysis of stress-strain curves, a method which will potentially allow one to better analyze the strengthening mechanisms in dual-phase steels.

The mathematical expressions which best

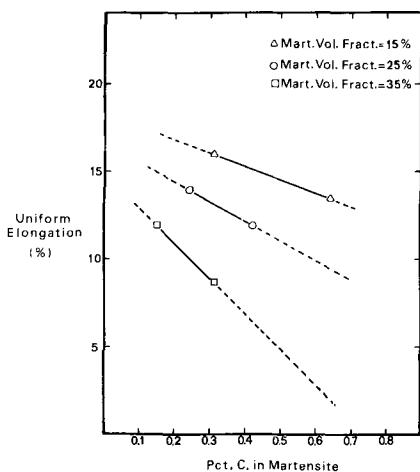


Fig. 3—Uniform elongation as function of pct C in martensite for Fe-C-Mn-Si steels with various martensite volume fractions.

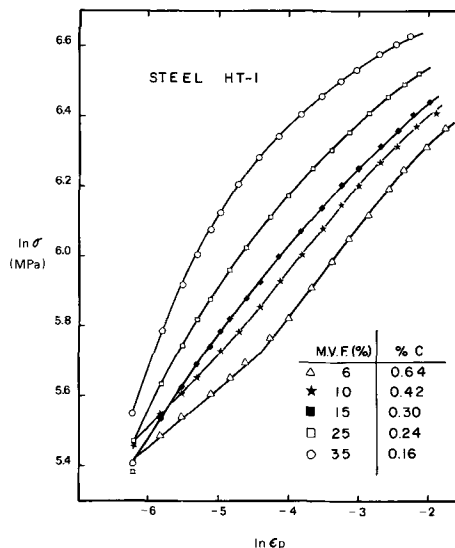


Fig. 4—Analysis of stress-strain curves of HT-1 according to Eq. [1].

describe⁸⁻¹⁰ the strain hardening behavior of iron and steel in uniaxial deformation are:

$$\sigma = K\epsilon^n \text{ (Hollomon}^{12}) \quad [1]$$

$$\sigma = \sigma_0 + K'\epsilon^{n'} \text{ (Ludwick}^{13}) \quad [2]$$

where σ and ϵ are true stress and true plastic strain, and the other parameters are constants. Equation [1] has been used to describe the stress-strain behavior of dual-phase steels,^{2,4} in which it is assumed that a single value of n described the strain hardening behavior. For steel HT-1, Fig. 4 shows a plot of $\ln \sigma$ vs $\ln \epsilon$ for dual-phase steels with different MVF. These data are clearly not linear, and show the effect of MVF and that a single value of n in Eq. [1] does not adequately describe the deformation of dual-phase steels.

Monteiro *et al*¹¹ have shown that a convenient method for delineating stages in strain hardening is by the application of the Jaoutt¹⁴-Crussard¹⁵ analysis

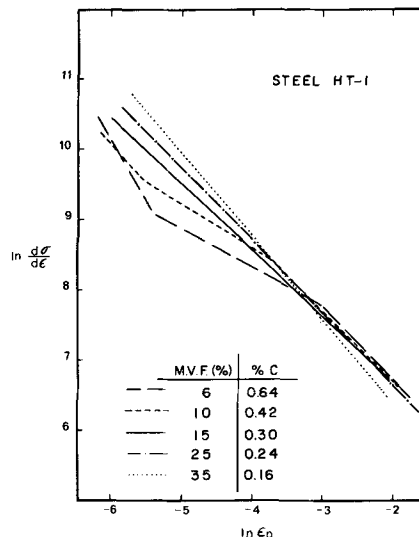


Fig. 5—Stress-strain curves of steel HT-1 according to Jaoutt-Crussard analysis.

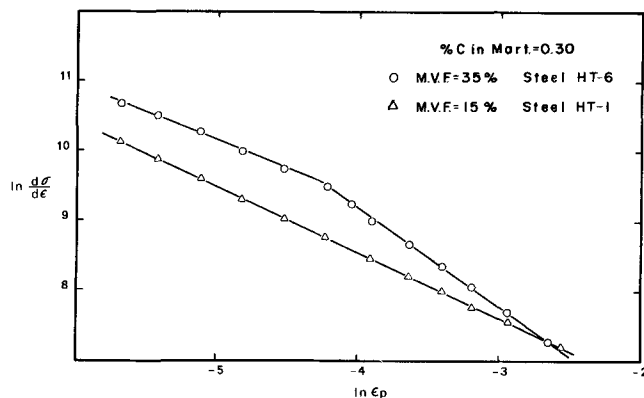


Fig. 6—Jaout-Crussard analysis of stress-strain curves. Martensite volume fractions vary and martensite carbon content is constant as shown.

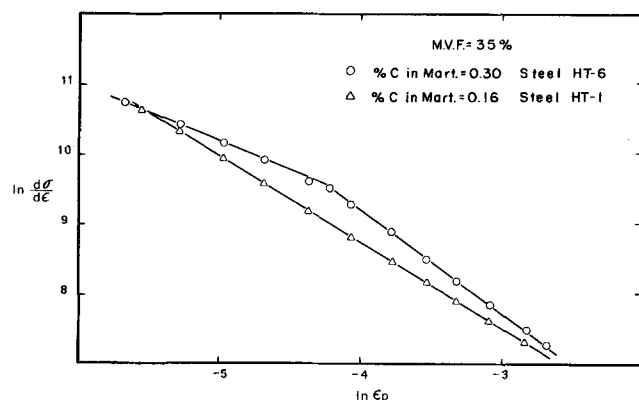


Fig. 7—Jaout-Crussard analysis of stress-strain curves. Martensite carbon content varies but martensite volume fraction is constant as shown.

to Eq. [2]. In the Jaout-Crussard analysis, $\log(d\sigma/d\epsilon)$ is plotted vs $\log \epsilon$. Figure 5 shows the data presented in Fig. 4 as a Jaout-Crussard plot. Note in this figure the effect of MVF and pct C in martensite on the strain hardening behavior at low strains and the similarity in behavior at high strains. Figure 6 shows similar results for a constant pct C in martensite but with two MVF, and Fig. 7 shows results for a constant MVF but different levels of pct C in martensite.

Figure 5 suggests that at higher strains the rate of strain hardening is similar for the various starting materials. Thus, differences in strain hardening behavior and uniform elongation arise from the effects of MVF and pct C in martensite on the strain hardening behavior at low strains as shown in Figs. 5 to 7.

In summary, the results described in this note show the independent effect of MVF and pct C in martensite on the deformation behavior of dual-phase steels and clearly indicate that the interactive effects primarily occur at low strains. It remains to be seen if other microstructural parameters such as retained austenite or various types of carbide particle distributions exert a similar effect on the early stages of deformation as amplified by the Jaout-Crussard analysis of the stress-strain curves.

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1. S. Hayami and T. Furukawa: *Microalloying 75 Proceedings*, p. 311, Washington, D.C., Oct. 1975.
2. M. S. Rashid: General Motors Research Publication, CMR-2326, Warren, MI, Feb. 1977.
3. G. Tither and M. Lafite: *J. Metals*, 1975, vol. 27, pp. 15-23.
4. R. G. Davies: *Met. Trans. A*, 1978, vol. 9A, pp. 41-52.
5. A. R. Marder: Presented at TMS-AIME Fall Meeting, Chicago, IL, 1977.
6. P. J. Vander Arend and J. M. Rigsbee: Presented at TMS-AIME Fall Meeting, Chicago, IL, 1977.
7. J. Y. Koo and G. Thomas: *Met. Trans. A*, 1977, vol. 8A, pp. 525-28.
8. R. G. Davies and R. C. Ku: *Trans. TMS-AIME*, 1966, vol. 236, pp. 1691-96.
9. W. B. Morrison: *Trans. ASM*, 1966, vol. 59, pp. 824-46.
10. H. J. Kleemola and M. A. Nieminen: *Met. Trans.*, 1974, vol. 5, pp. 1863-66.
11. S. N. Monteiro and R. E. Reed-Hill: *Met. Trans.*, 1971, vol. 2, pp. 2947-49.
12. J. H. Hollomon: *Trans. AIME*, 1945, vol. 162, pp. 268-90.
13. P. Ludwik: *Element der Technologischen Mechanik*, p. 32, Julius Springer, Berlin, 1909.
14. B. Jaout: *J. Mech. Phys. Solids*, 1957, vol. 5, pp. 95-114.
15. C. Crussard: *Rev. Metall. (Paris)*, 1953, vol. 10, pp. 697-710.

The Effect of Antimony on Sintering of Silver in the Initial Stage

P. Y. Y. MAAK AND K. T. AUST

A previous study by Takahata *et al*¹ gave some indication that the addition of a solute (1.16 wt pct Sb) to spherical silver powder increased the shrinkage rate at 800 and 850°C in the early stage of sintering (up to about 0.5 pct shrinkage). However, the difference in shrinkage between the pure Ag and Ag-Sb powders was quite small in this initial stage. It was possible, therefore, that this result may have been partly associated with experimental error in measuring shrinkage (about ± 0.0018 maximum) and in determining the true origin of the isothermal shrinkage curves. It was evident that further studies at lower temperatures would be required before unequivocal statements could be made regarding this early stage of sintering.

In the present study, the effect of Sb (1.16 wt pct) on the sintering kinetics of Ag in the early stage was investigated using lower sintering temperatures (700 to 800°C) and shorter sintering times (up to 3 to 4 h) in argon atmosphere. The material characteristics, preparation of specimens for the sintering studies, shrinkage measurements and metallographic study of the sintered specimens are described in Ref. 1.

Isothermal shrinkage curves of $\Delta L/L_0$ vs sintering time (t) are shown in Fig. 1 for specimens sintered at 700, 740, and 780°C and in Fig. 2 for the data at 720, 760 and 800°C. These curves indicate that the Ag-Sb alloy shrinks at a faster rate than the pure Ag in the first 10 to 20 min of sintering at the same temperature. However, with continued sintering at the higher temperatures the rate of shrinkage of the alloy pow-

P. Y. Y. MAAK and K. T. AUST are Graduate Student and Professor, respectively, Department of Metallurgy and Materials Science, University of Toronto, Toronto, Canada M5S 1A4. The present work is based on a B.A.Sc. thesis of P. Y. Y. Maak, University of Toronto, 1977.

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