Effect of Rolling in the Intercritical Region on the Tensile Properties of Dual-Phase Steel

E. Ahmad and R. Priestner

(Submitted 26 January 1998; in revised form 19 May 1998)

A steel containing 0.088 wt% C, 1.2 wt% Mn, and 0.78 wt% Cr was rolled at intercritical temperature (790 °C) and quenched to produce dual-phase microstructure. Rolling caused anisotropic increase in tensile strength and little change in ductility. The results suggest that rolling increased strength by a combination of strengthening of the ferrite and an increase in the stress transferred to the martensite. Up to 20% rolling reduction strengthened the ferrite by work hardening, larger reductions then reduced the strength of ferrite, anisotropically, due to increased recovery. Subgrains in ferrite were observed after rolling in the intercritical region which can contribute to the ultimate strength of the rolled material.

Keywords dual phase steel, intercritical region, strain, stress, tensile properties

1. Introduction

Dual-phase steels offer a combination of properties in tension (Ref 1-3), which make them unique among high-strength, low-alloy (HSLA) steels. Their main characteristics are continuous yielding, low elastic limit and 0.2% offset yield strength, high work hardening rate in early stages of deformation, and usually high uniform and total elongation. The low elastic limit and absence of discontinuous yielding in dual-phase steels has been explained in terms of residual stresses and high dislocation density in the ferrite matrix surrounding the martensite particles (Ref 4). Lattice shear and volume expansion accompanies the austenite to martensite transformation (Ref 5, 6) during cooling from the ferrite plus austenite phase field and causes distortion in the product (martensite) and the matrix (ferrite). This distortion generates residual stresses (Ref 7, 8) and a high density of mobile dislocation in the ferrite adjacent to martensite particles and causes plastic flow at low plastic strains (Ref 1). Because the plastic flow begins at many sites throughout the specimens, discontinuous yielding is suppressed (Ref 9).

There appears to be general agreement that the strength of dual-phase steels is linearly proportional to the percentage of martensite in the structure (Ref 10-12) and also to the carbon content of the martensite (Ref 13). For the same material, the flow stress at 1% strain is linearly dependent upon the martensite volume fraction. At a given carbon level, the higher the intercritical annealing temperature, the larger the flow stress (Ref 10). Other studies (Ref 14-15) have shown that strength of dual-phase steel is dependent only on the volume fraction of martensite and not on the carbon content of martensite. Similar effects of volume fractions of martensite and its carbon content on tensile properties were studied in connection with controlled rolling in the intercritical annealing region (Ref 16). Rolling in the intercritical regions can produce fibers of martensite in the ferrite matrix resulting in an anisotropy in the tensile properties. The aim of the present study was to investigate the effect of rolling on the longitudinal and transverse tensile properties of the warm rolled, dual-phase steel.

E. Ahmad, PINSTECH, Nilore, Islamabad; and **R. Priestner,** Manchester Materials Science Center, Manchester, UK.

2. Experimental Work

The composition of the steel is described in Table 1. The material was provided by British Steel Corporation in the form of hot rolled, 12 mm slabs. Metallographic examination of as-received microstructure showed it consisted mainly of ferrite and pearlite.

2.1 Thermomechanical Processing for Mechanical Testing

Rolling reduction of 0, 20, 30, and 50% were planned after annealing at 790 °C. This temperature was selected from another set of experiments in which a plot was drawn between volume fraction of austenite and the intercritical heat treatment (ICHT) temperature. The volume fraction of austenite was 50% at 790 °C. The final dimensions of the specimens after rolling were such that tensile specimens in both longitudinal and trans-

Table 1 Chemical composition of the steel used

Element	Composition, wt%
C	0.088
Si	0.26
Mn	1.2
P	0.01
S	0.009
Cr	0.78
Mo	0.04
Ni	0.15
Cu	0.20
Sn	0.017
Fe	bal

verse directions could be machined from each of the slabs. It was required that the final thickness be 5 mm (for consistency of rolling) after each rolling reduction. Therefore, the specimens were machined initially to different lengths and thicknesses. Very little increase in width was obtained after rolling. A hole of 2 mm diam was drilled through the width of each specimen, to contain a thermocouple with its junction in the center of the specimen. The hole was packed with alumina wool. Outside the specimen, the thermocouple wires were insulated with ceramic beads and sheaths. The specimen was mounted on a steel rod to facilitate handling during heat treatment and rolling.

To record the history of temperature changes inside the specimen during rolling and quenching in different media, a flatbed millivolt recorder was used. The chart recorder output of millivolt versus time was used to estimate the cooling rate. The speed of the chart recorder was changed according to the cooling rate.

The rolling speed of 17 mm/min was used for all specimens. All rolling reductions were performed in a single pass except the 50% reduction that was completed in two passes. After the required rolling reduction, the specimens were quenched in iced brine solution at –5 °C so that all the austenite could transform to martensite.

2.2 Tensile Testing

Tensile testing was conducted on an Instron machine (Instron Corporation, Canton, MA) having 100 kN load capacity. All tests were performed at the crosshead speed of 0.5 mm/min in normal atmosphere. An extensometer with a 25 mm gauge length was used, and plots of load versus strain were obtained in each case.

2.3 Metallography

Long transverse sections were used from the rolled and not-rolled specimens. All the specimens were etched in the 2% nital solution that revealed ferrite-carbide constituents as black areas, leaving martensite unetched and differentiated from ferrite with a bold outline. A Swift point counter (James Swift, Basingstoke, England) was used to determine the fractions of various phases. A magnification of 1000× was used for counting 1000 to 1500 points for each specimen. A scanning electron microscope was used to examine the substructure formed in the rolled specimens.

3. Results and Discussion

The tensile data of the specimens are listed in Table 2. The specimens are coded according to the quenching medium used, the planned reduction, and the direction of tensile testing with respect to rolling reduction. For example, in BR50L, "BR" represents brine, "50" refers to planned percent rolling reduction, and "L" represents the longitudinal testing direction. The carbon contents of martensite were estimated from the approximate carbon content of the ferrite (0.01%), the carbon content of steel (0.08%), and the volume fraction of the martensite.

The true stress-true strain curves of 0, 19, 29 and 48% rolling reductions, to the onset of necking, are shown (Fig. 1a, b, c, and d). The comparison of behaviors in longitudinal and transverse directions shows that the true stress at the instability was approximately the same at 0% reduction, but increased more in the rolling than in the transverse direction as the rolling reduction increased. At 0 and 19% rolling reductions, ductility was greater in the transverse direction, but at the higher reductions it was greater in the longitudinal direction. These trends are illustrated (Fig. 2a, b).

The martensite particles became more fibrous with increasing rolling strain. The fibrous structure of specimens BR50L and BR50T and the equiaxed structure of specimens BR0L and BR0T are shown (Fig. 3a, b, respectively). In the longitudinal direction the martensite fibers are lengthened and are reduced in thickness by rolling. In the transverse direction they are not lengthened but thinned in the direction normal to the rolling plane. Thus the aspect ratio of the fibrous martensite particles will be greater in the longitudinal than in the transverse test pieces.

Table 2 Tensile data of various rolling reductions and quenched in iced brine solution

Specimen code	Reduction obtained, %	Martensite, %	Carbon content of martensite, %	Testing direction	Maximum true stress, MPa	True uniform strain, %	Total elongation, %	UTS, MPa	Uniform elongation, %	0.2% Proof stress, MPa
BR0L	0	42.2	0.19	Longitudinal	848.9	8.64	17.2	778.6	9.02	345
BR0T	0	42.2	0.19	Transverse	851.4	11.18	…	761.3	11.83	386
BR20L	19.2	50.6	0.16	Longitudinal	978.5	5.61	14.0	925.1	5.8	546
BR20T	19.2	50.6	0.16	Transverse	961.0	8.15	13.0	885.8	8.5	507
BR30L	28.7	50.3	0.17	Longitudinal	1022.6	10.66	16.4	919.26	11.25	540
BR30T	28.7	50.3	0.17	Transverse	901.8	6.64	11.6	843.91	6.86	430
BR50L	48.0	45.5	0.18	Longitudinal	1020.5	9.70	18.0	926.1	10.2	384
BR50T	48.0	45.5	0.18	Transverse	938.5	7.6	11.6	869.8	7.9	461
0.12% steel (a)	…	44.3	0.28	…	…	…	…	788	14.0	…
0.16% steel(a)	…	50.3	0.32	…	…	…	…	920	10.2	…
ICHT at 780 °C(b)	0	55.0	…	…	…	…	12.0	924.09	6.39	…
ICHT at 780 °C(b)	50	48.9	…	…	…	…	8.9	1014.76	8.9	…

(a) Ref 13. (b) Ref 16

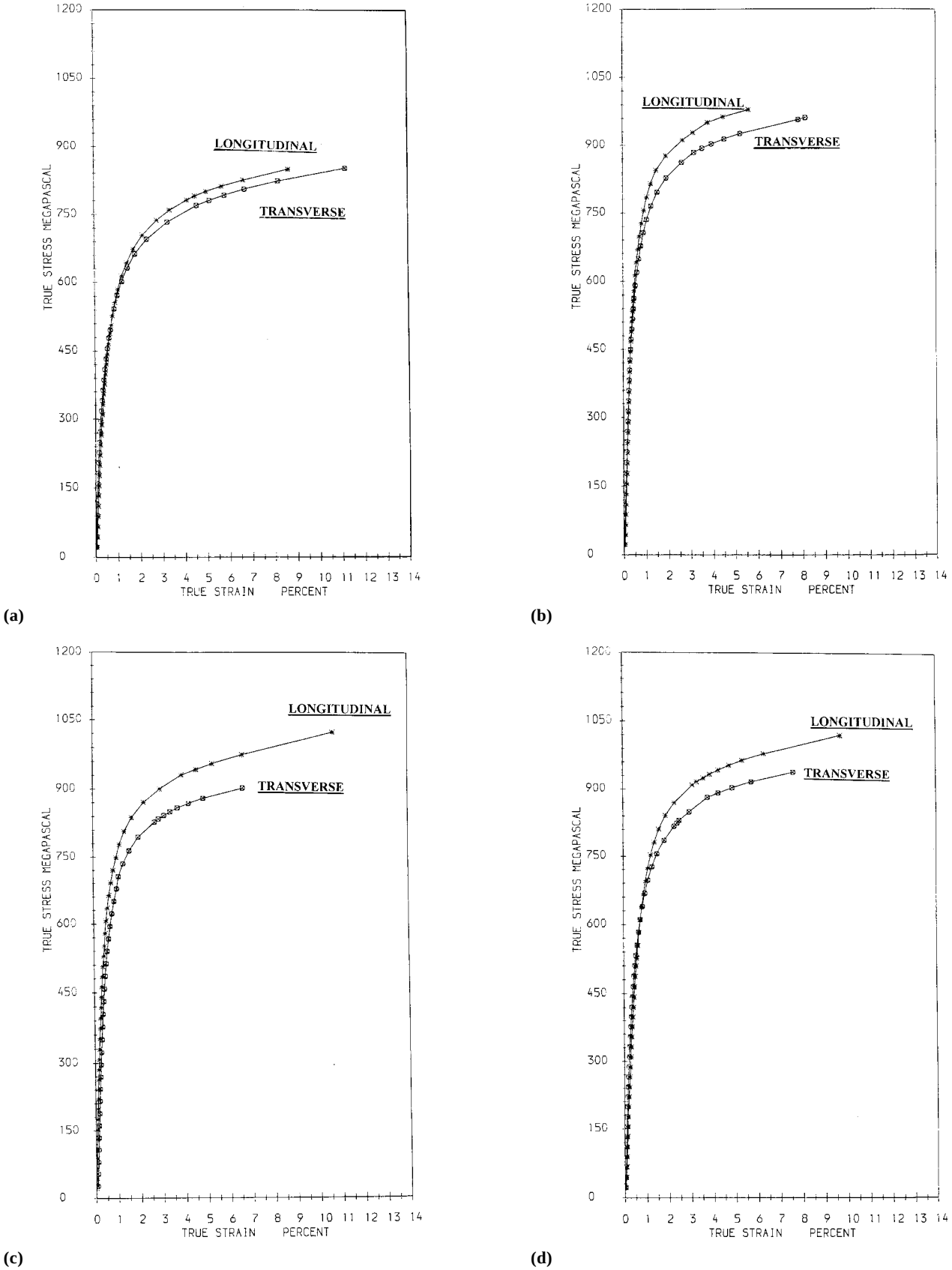


Fig. 1 True stress-true strain curves after intercritical annealing at 790 °C followed by (a) 0%, (b) 19%, (c) 29%, and (d) 48% rolling reductions. All the specimens were quenched in iced brine solution.

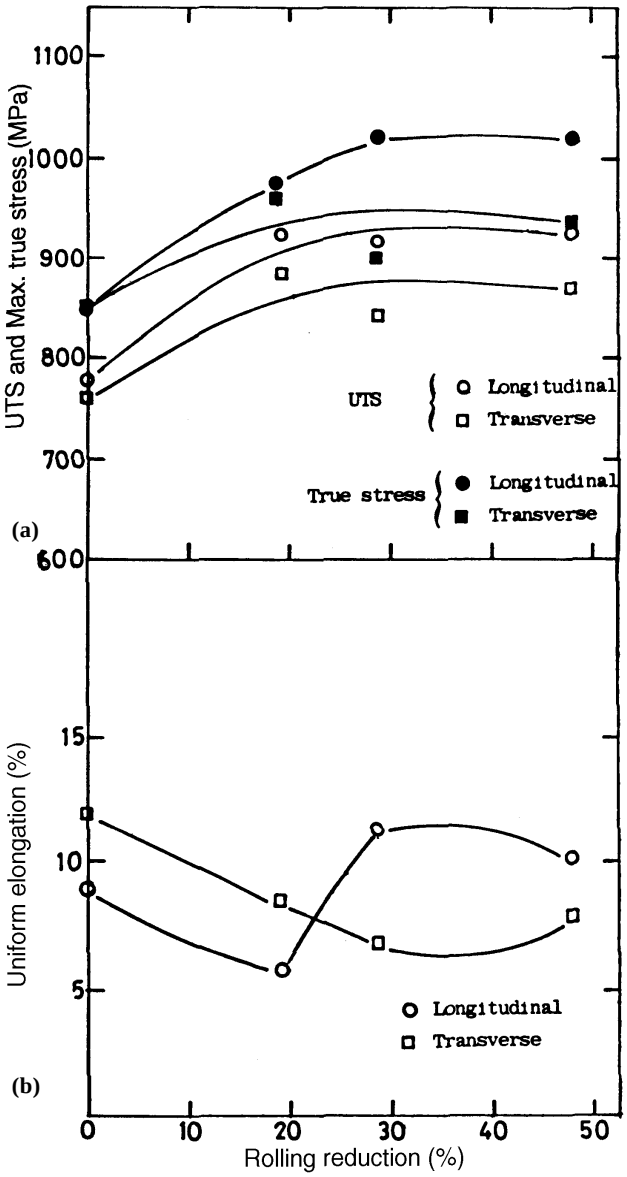
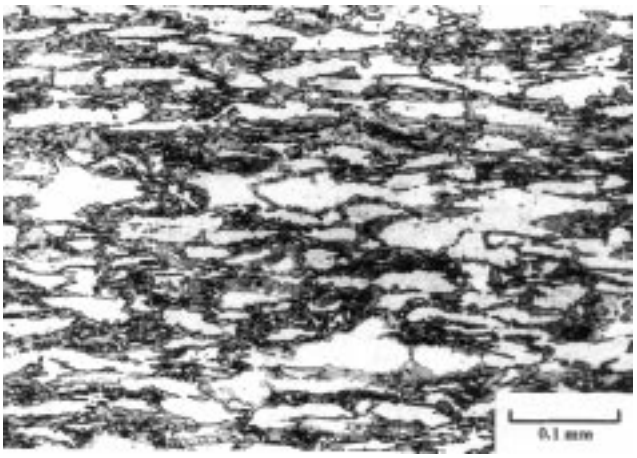
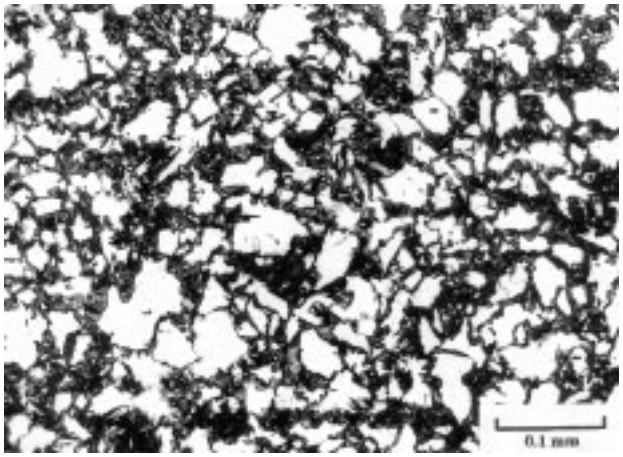


Fig. 2 Effect of rolling reduction on (a) UTS and maximum true stress and (b) uniform elongation, after quenching in iced brine solution

The comparison of true strain values for longitudinal and transverse specimens, up to the onset of necking, also shows that for rolling reductions of 0 and 19%, ductility was greater in the transverse direction. For reductions of 29 and 48%, ductility was greater in the rolling direction. The formation of substructure in the ferrite as a result of 48% reduction is shown in Fig. 4. Tanaka (Ref 17) and Sarwar and Priestner (Ref 16) also observed the formation of substructure in the ferrite after deformation in the intercritical temperature range and subsequent cooling to room temperature. Tanaka showed that even heavy deformation in the fully austenitic region produced only grain refinement of the resultant ferritic structure after cooling, whereas deformation of a mixture of ferrite and austenite produced a substructure in the ferrite which contributed to the strength of the material.



(a)



(b)

Fig. 3 (a) Fibrous martensite in specimens BR50L and BR50T, 250×. (b) An equiaxed structure in specimens BR0L and BR0T, 250×. (Art has been reduced to 65% of its original size for printing.)

These observations suggest that improvement of strength with rolling reduction was caused by two principal changes in the microstructure: (a) The formation of fibrous martensite increased the area of contact with the matrix (ferrite) compared to the equiaxed structure of unrolled martensite. Therefore, stress transfer from the matrix to the fiber might be more efficient; and (b) Deformation produced subgrains in the ferrite, which might have increased the strength of the matrix.

A comparison of ultimate tensile strength (UTS) and uniform elongation with some results reported by Speich and Miller (Ref 13) and those by Sarwar and Priestner (Ref 16) are made in Table 1. The tensile data of Speich and Miller with different carbon contents of the steels showed a dependence of UTS and uniform elongation on carbon content in the martensite. In the present work, unrolled specimens of similar martensite content but lower carbon concentration in the steel and the martensite, had similar strength, with lower ductility, to that of the 0.12% carbon steel tested by Speich and Miller. Rolling at the intercritical temperature increased both strength and ductility in the rolling direction, whereas the strength increase ob-

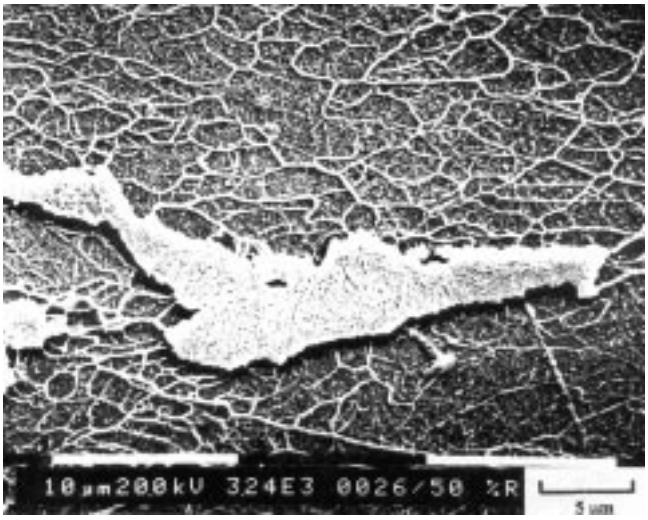


Fig. 4 The substructure formed in the ferrite after 48% reduction and brine quench specimen

served by Speich and Miller as a result of increasing carbon content was accomplished by a decrease in ductility. Sarwar and Priestner used 0.166 wt% carbon steel for warm rolling in the intercritical region. They concluded that warm rolling increased the strength of the steel. In the present studies the increase in strength after warm rolling of 48% is more pronounced than that observed by Sarwar and Priestner with no effect on uniform elongation. Although the UTS in both rolled and not rolled conditions was higher than that observed in the present studies, this may have been due to high carbon content of the martensite as steel with high carbon percentage was used. However, the uniform elongation is markedly higher in the present case.

The 0.2% proof stress values are listed in Table 1. Specimen BR50L had very low proof stress but the highest UTS. Although specimen BR20L had a high proof stress, rolling generally did not increase the proof stress as much as it increased UTS. Thus rolling improved the overall work hardening rate between the proof stress and UTS.

4. Conclusions

- Anisotropy in the tensile properties was developed after warm rolling. The strength of the steel is greater in the longitudinal direction than in the transverse direction at all rolling reductions.
- The formation of the fibrous martensite and subgrains in the ferrite contributed to the increase in strength both in the longitudinal and transverse directions.
- Warm rolling in the intercritical region increased the work hardening rate by better stress transfer from the matrix (ferrite) to the fiber (martensite) during plastic deformation.

References

1. G.T. Eldis, *Structure and Properties of Dual-Phase Steels,* R.A. Kot and J.W. Morris, Ed., AIME, 1979, p 202-220
2. N.J. Kim, *Scr. Metall.*, Vol 18, 1984, p 817-820
3. N.K. Balliger and T. Gladman, *Met. Sci.*, 1981, p 95-108
4. J.M. Rigsbee and P.J. VanderArend, *Formable HSLA and Dual-Phase Steels*, A.T. Davenport, Ed., AIME, 1979, p 56-86
5. C.L. Magee and R.G. Davies, *Acta Metall.*, Vol 20, 1972, p 1031
6. J.M. Moyer and G.S. Ansell, *Metall. Trans. A*, Vol 6, 1975, p 1785
7. G. Tither and M. Laviye, *J. Met.*, Vol 27, 1975, p 15
8. W.C. Leslie and R.J. Sober, *Trans. ASM*, Vol 60, 1967, p 459
9. J.D. Baird, *Iron and Steel*, Part 1, Vol 63, 1963, p 186-191, 326-334, 368-374
10. R.G. Davies, *Metall. Trans. A*, Vol 9, 1978, p 671-679
11. R.D. Lawson, D.K. Matlock, and G. Krauss, *Fundamentals of Dual-Phase Steels*, R.A. Kott and B.L. Bramfitt, Ed., AIME, 1981, p 347-381
12. J.Y. Koo, M.J. Young, and G. Thomas, *Metall. Trans. A*, Vol 9, 1980, p 852-854
13. G.R. Speich and R.L. Miller, *Structure and Properties of Dual-Phase Steels*, R.A. Kott and J.W. Moriss, Ed., AIME, 1979, p 145-182
14. R.G. Davies, *Metall. Trans. A*, Vol 9, 1978, p 41-52
15. J.Y. Koo and G. Thomas, *Mater. Sci. Eng.*, Vol 24, 1976, p 187-198
16. M. Sarwar and R. Priestner, *J. Mater. Sci.*, Vol 31, 1996, p 2091-2095
17. T. Tanaka, *Int. Met. Rev.*, Vol 4, 1985, p 185