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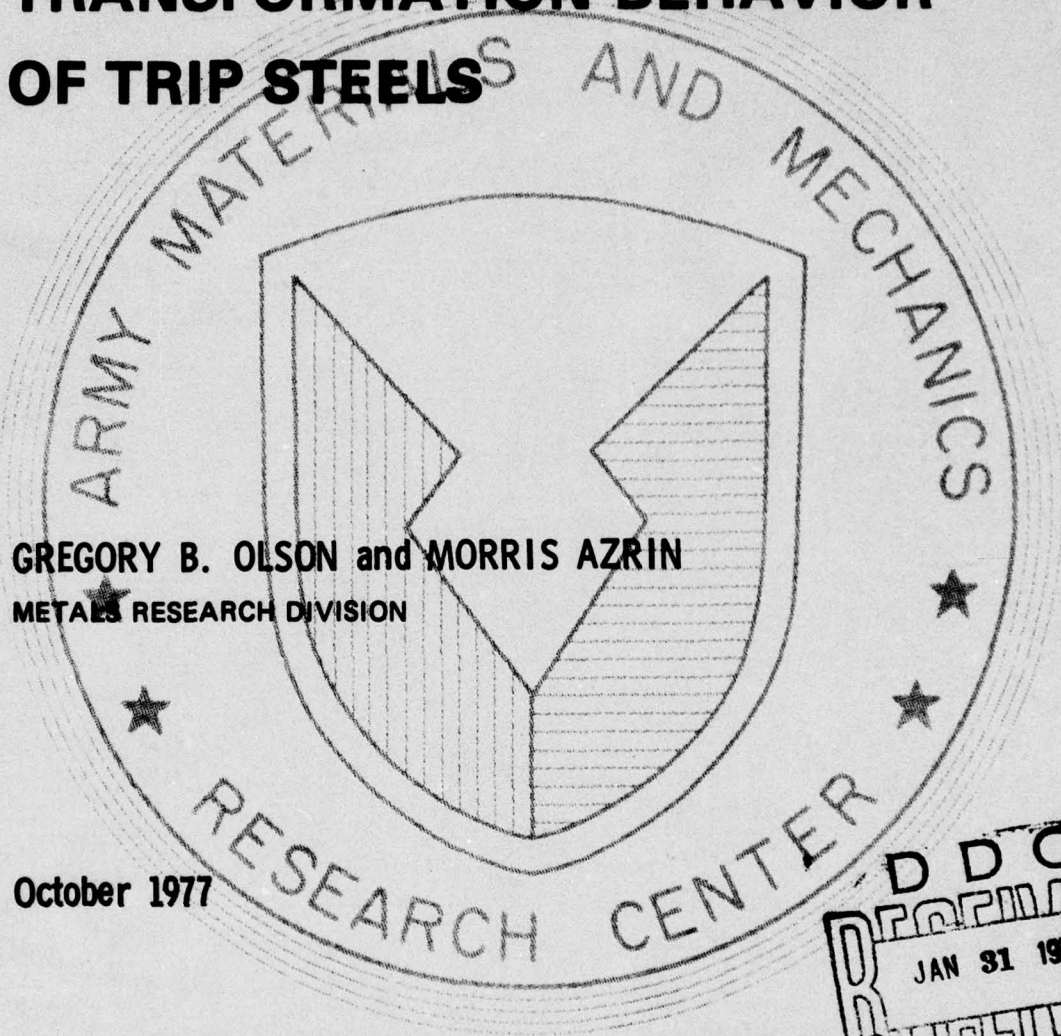
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METALS RESEARCH DIVISION

October 1977

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ABSTRACT

True stress (σ), true strain (ϵ), and volume fraction martensite (f) were measured during both uniform and localized flow as a function of temperature on TRIP steels in both the solution-treated and warm-rolled conditions. The transformation curves (f versus ϵ) of materials in both conditions have a sigmoidal shape at temperatures above M_S^σ (maximum temperature at which transformation is induced by elastic stress) but approach initially linear behavior at temperatures below M_S^σ where the flow is controlled by transformation plasticity. The martensite which forms spontaneously on cooling or by stress-assisted transformation below M_S^σ exhibits a plate morphology. Additional martensite units produced by strain-induced nucleation at shear band intersections become important above M_S^σ . Comparison of σ - ϵ and f - ϵ curves indicate that a "rule of mixtures" relation based on the "static" strengthening effect of the transformation product describes the plastic flow behavior reasonably well above M_S^σ , but there is also a dynamic "transformation softening" contribution which becomes dominant below M_S^σ due to the operation of transformation plasticity as a deformation mechanism. Temperature sensitivity of the transformation kinetics and associated flow behavior is greatest above M_S^σ . Less temperature-sensitive TRIP steels could be obtained by designing alloys to operate with optimum mechanical properties below M_S^σ .

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INTRODUCTION

Recent true-stress, true-strain measurements¹ of the tensile flow properties of high-strength TRIP steels² designed to incorporate the initial region of localized flow associated with Lüders band formation have called attention to a large apparent strain-rate sensitivity of the uniform ductility. This apparent strain-rate effect was interpreted as originating from the temperature dependence of the kinetics of the deformation-induced martensitic transformation which controls the work-hardening behavior of these metastable austenitic steels. Efforts to understand the transformation kinetics and its relation to flow properties in high-strength TRIP steels have been hampered by a lack of experimental information from the region of initial inhomogeneous flow that represents such an important part of the deformation behavior of these materials. The present study was undertaken to extend true-stress, true-strain, and volume fraction martensite measurements to the region of Lüders band formation in order to elucidate the transformation kinetics and its relation to flow properties as a function of temperature, and to compare these findings with the behavior of lower strength metastable austenites which deform homogeneously and for which theoretical concepts are fairly well developed.

MATERIALS AND EXPERIMENTAL PROCEDURE

The composition (in weight percent) of the TRIP steel studied is 0.27C, 0.78Mn, 2.13Si, 8.06Ni, 9.11Cr, and 4.05Mo, similar to that of the A-3 alloy studied by Zackay et al.² A 90-lb (40-kg) heat was air melted from high purity materials and two ingots were poured in tapered iron molds under an argon atmosphere. The ingots were hot pressed to 1-1/2-inch (3.8-cm) square bars at 1150 C and homogenized in vacuum at 1260 C for 12 hours. The bars were hot rolled to 5/8 inch (1.6 cm) at 1150 C then solution treated at 1200 C for 1 hour and water quenched. After machining to 1/2-inch (1.27-cm) thickness to remove surface scale, the material was warm rolled to a 78% reduction in thickness at 450 C. After grinding to a 0.080-inch (0.203-cm) thickness, flat tension specimens were cut with a gage section 0.200 inch (0.503 cm) wide and 1.000 inch (2.54 cm) long.

To study this alloy in the solution-treated condition, a few tension specimens were vacuum encapsulated in quartz tubes and solution treated at 1200 C for 1 hour followed by water quenching. Neither the solution-treated nor the warm-rolled material showed any evidence of martensitic transformation on cooling down to 4.2 K. However, the solution-treated material contained some martensite after isothermal holding at 78 K for 8 hours. For purposes of calibration of the magnetic technique, a slightly less stable alloy composition was required so that various amounts of martensite could be produced by cooling. Accordingly, a portion of the alloy was uniformly decarburized to 0.19 carbon and solution-treated at 1200 C. This material gave rapid isothermal transformation at temperatures below -90 C.

The apparatus, instrumentation, calibration, and data-analysis techniques used for the continuous measurement of true stress (σ), true strain (ϵ), and volume

1. AZRIN, M., OLSON, G. B., and GAGNE, R. A. *Inhomogeneous Deformation and Strain-Rate Effects in High-Strength TRIP Steels*. Mat. Sci. and Eng., v. 23, no. 1, April 1976, p. 33-41; also Army Materials and Mechanics Research Center, AMMRC TR 73-12, March 1973.
2. ZACKAY, V. F., PARKER, E. R., FAHR, D., and BUSCH, R. *The Enhancement of Ductility in High-Strength Steels*. Trans. ASM, v. 60, 1967, p. 252-259.

fraction martensite (f) during both uniform and nonuniform flow are described in detail elsewhere.³ Tension tests were performed on a 30,000-lb-capacity (13,600-kg) Tinius Olsen testing machine at an imposed strain rate of 0.02 min^{-1} . Uniform strain was monitored by crosshead displacement, and local strain during the initial nonuniform flow was determined from travelling calipers which continuously traced the specimen width profile. The specimen gage section was surrounded by a wire coil which acted as a reactive element of an LC resonant circuit forming the frequency-determining part of a transistorized audio oscillator. The oscillator frequency responded to changes in the average a-c permeability of the specimen, and thus served to monitor the total martensite content. Extensive calibration accounted for the influence of specimen shape, position of the paramagnetic specimen grips, and the strain dependence of permeability. Each specimen was individually calibrated after testing by determining the martensite content in the uniformly deformed portion of the gage section via density measurements and optical metallography. During nonuniform flow, the complete specimen strain profile and measured total martensite content were used to obtain the *average* martensite volume fraction versus *average* strain in the deforming portion (Lüders band) of the specimen.

To test the accuracy of the resulting transformation curves (f versus ϵ), the transformation curve of the warm-rolled material at room temperature was independently determined by measuring the density of the central portion of forming Lüders bands cut from separate specimens strained to various early stages of Lüders band development. The points thus obtained are compared with the continuous transformation curve determined from a single specimen by the magnetic technique in Figure 1; the agreement is quite reasonable. The magnetic technique was then employed to evaluate the σ versus ϵ and f versus ϵ curves at different temperatures by immersing the specimen, coil, and grip assembly in a suitable medium. Temperatures were controlled within $\pm 2 \text{ C}$.

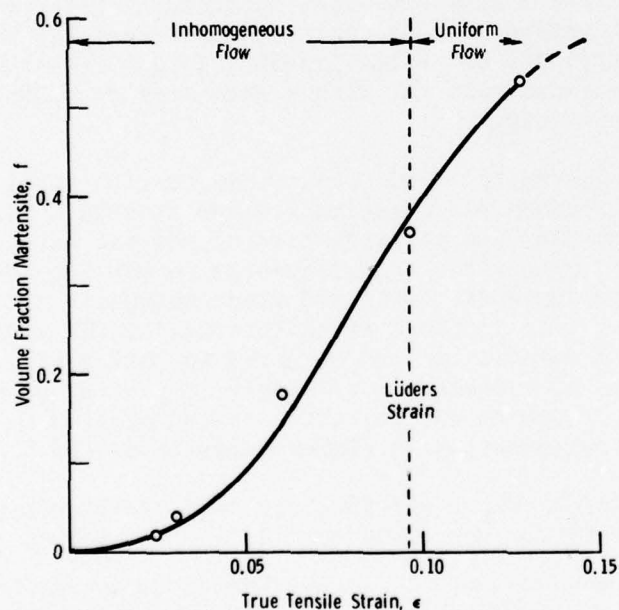


Figure 1. Transformation curve (f versus ϵ) obtained at room temperature for warm-rolled (78% RA, 450 C) TRIP steel. Solid curve is obtained from continuous magnetic measurements during a single test. Open circles were obtained from density measurements from separate interrupted tests.

3. OLSON, G. B., AZRIN, M., and BAND, H. *AC-Permeability Measurement of Strain-Induced Martensite During Tensile Deformation of TRIP Steels*. Army Materials and Mechanics Research Center, AMMRC TR 77-10, March 1977.

RESULTS AND DISCUSSION

Transformation and Stress-Strain Curves

The true stress-strain and associated volume fraction martensite versus true-strain curves measured at various temperatures are shown in Figures 2, 3, and 4. Figures 2 and 3 present the curves for the 0.27C material in the warm-rolled and solution-treated conditions, and the curves in Figure 4 apply to the solution-treated 0.19C material. The data for the solution-treated 0.19C material are expected to be somewhat less accurate due to uncertainties arising from deformation outside the specimen gage section. The higher carbon specimens fractured at lower strains before such effects became significant.

Only the warm-rolled material exhibited Lüders band deformation behavior. The heavy points in Figure 2 represent values obtained at the full Lüders strain. Thus the portion of each curve to the right of this point corresponds to measurements

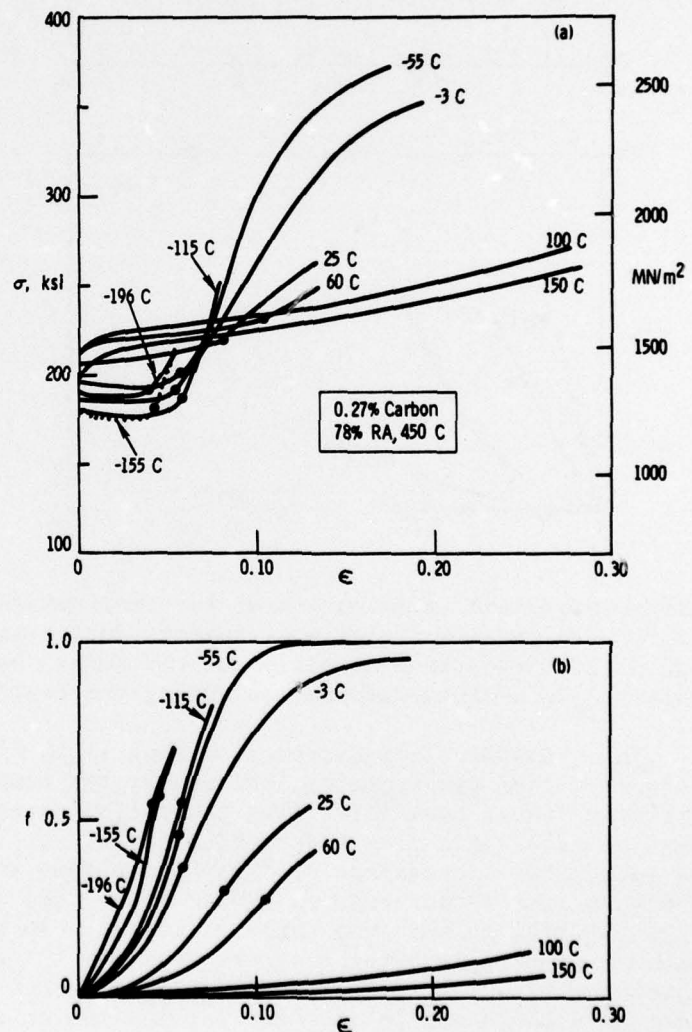


Figure 2. Tensile properties and transformation curves for 0.27% carbon TRIP steel in warm-rolled condition (78% RA, 450 C): (a) σ - ϵ curves and (b) f - ϵ curves. Heavy points represent values at Lüders strain.

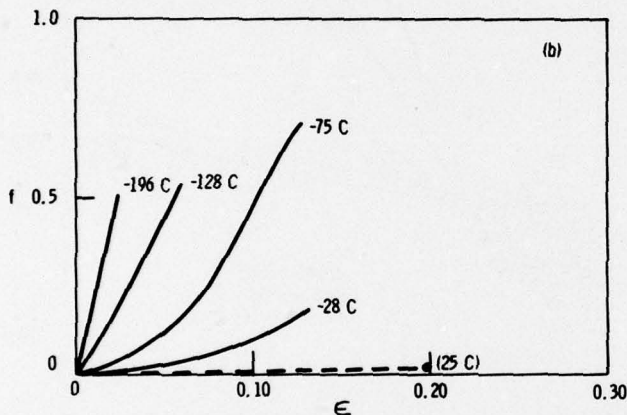
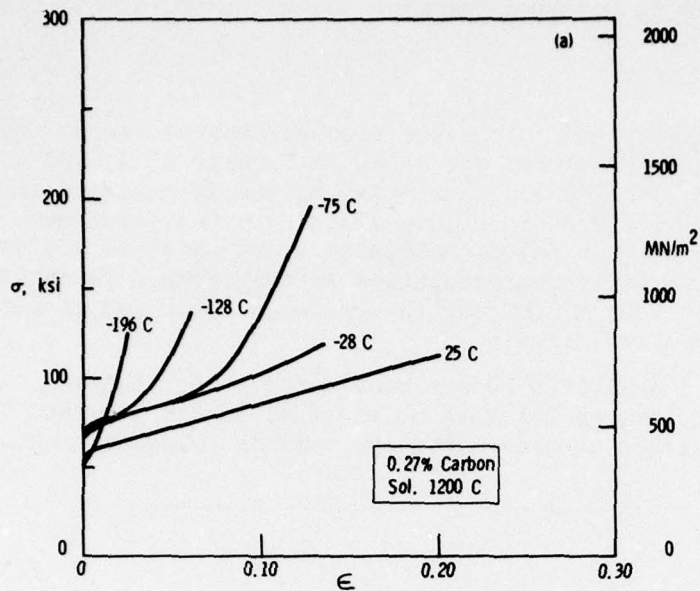


Figure 3. Tensile properties and transformation curves for 0.27% carbon TRIP steel in solution-treated (1200 C) condition: (a) σ - ϵ curves and (b) f - ϵ curves.

during homogeneous deformation of the specimen gage section, while the portion to the left is based on local measurements during initial formation of the Lüders band (i.e., prior to propagation of the Lüders band). The warm-rolled material exhibited no uniform deformation during the tests at 100 and 150 C.

The transformation curves (f versus ϵ) in Figures 2, 3, and 4 have a sigmoidal shape at higher temperatures, whereas at low temperatures the curves approach an initially linear behavior. This transition correlates with a basic change in the relation of plastic flow and transformation which is best understood by examining the temperature dependence of the yield stress in Figures 5 and 6. The solid curves in Figure 5 show the measured 0.2% offset yield stress of the fully austenitic 0.27C material in both the warm-rolled and solution-treated conditions, while the dashed curves represent the stress at which 1% martensite is detected. Similar curves for the 0.19C alloy are given in Figure 6. (Properties of the fully austenitic state could not be measured for the latter alloy below -90 C due to the onset

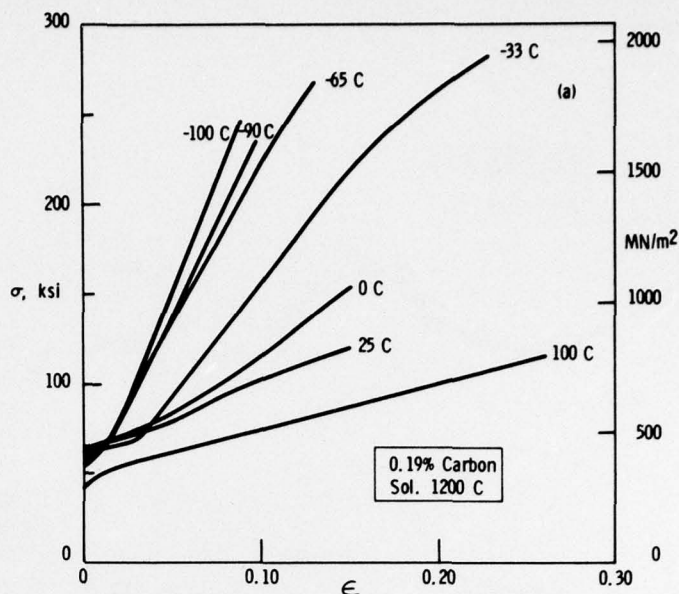
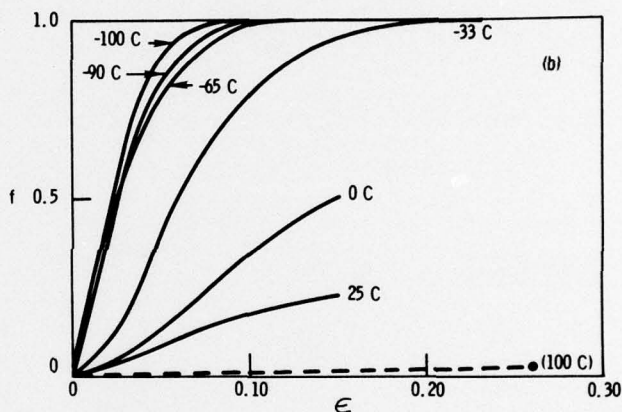


Figure 4. Tensile properties and transformation curves for 0.19% carbon TRIP steel in solution-treated (1200 C) condition: (a) σ - ϵ curves and (b) f - ϵ curves.



of spontaneous isothermal transformation before loading.) A reversal of the temperature dependence of the yield stress occurs at (or very near) a crossover of the dashed and solid curves at a temperature labelled as M_S^σ , following the suggestion of Richman and Bolling;⁴ M_S^σ is the highest temperature at which martensitic transformation is induced by elastic stress. Below this temperature, the transformation is believed to be initiated predominantly by nucleation at the same pre-existing sites responsible for the spontaneous transformation on cooling, but assisted by elastic stress; this mechanism is defined as *stress-assisted* nucleation.⁵ In this temperature regime, the measured yield stress falls significantly below that expected for a stable austenite. That the 1% martensite curves fall just below the 0.2% yield stress curves strongly suggests that plastic flow is here caused by the transformation. When the strains or "transformation plasticity" associated with martensitic transformation under stress constitute the dominant deformation

4. RICHMAN, R. H., and BOLLING, G. F. *Stress, Deformation, and Martensitic Transformation*. Met. Trans., v. 2, September 1971, p. 2451-2462.

5. OLSON, G. B., and COHEN, M. *Mechanism for the Strain-Induced Nucleation of Martensitic Transformations*. J. of Less-Common Metals, v. 28, 1972, p. 107-118.

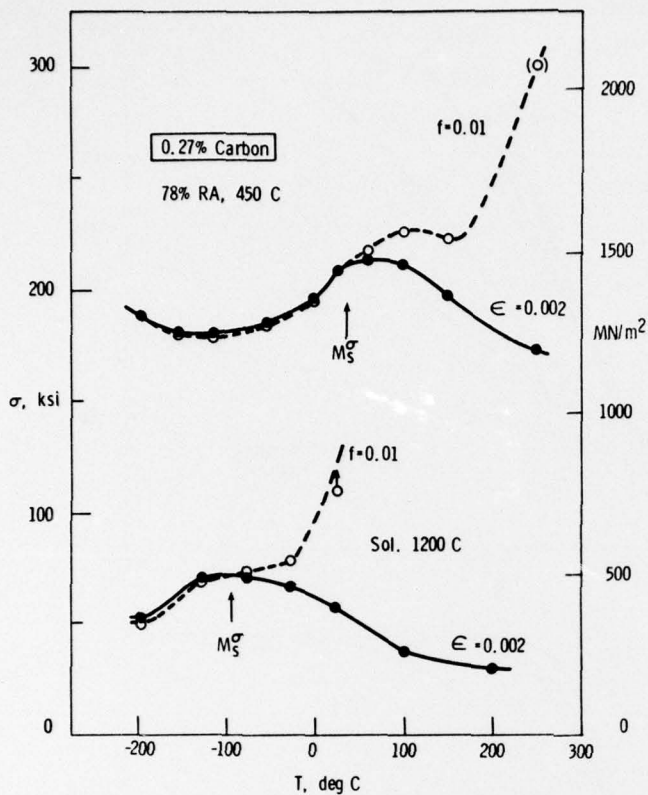


Figure 5. Temperature dependence of yield stress and stress for martensite transformation in 0.27% carbon TRIP steel. Upper curves are obtained from warm-rolled material (78% RA, 450 C). Lower curves represent solution-treated (1200 C) material. Solid curves and dark circles indicate 0.2% offset yield stress. Dashed curves and open circles represent stress at which 1% martensite is detected. (Open point at 250 C is estimated from martensite content in necked region of specimen examined metallographically after test.)

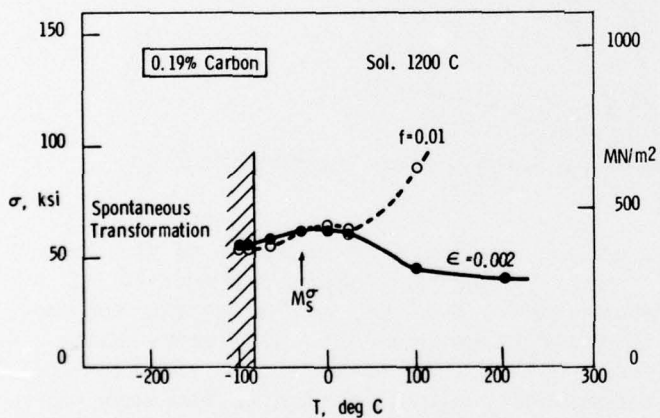


Figure 6. Temperature dependence of yield stress and stress for martensite transformation of 0.19% carbon TRIP steel in solution-treated (1200 C) condition. Solid curves and dark circles indicate 0.2% offset yield stress. Dashed curves and open circles represent stress at which 1% martensite is detected. Austenite properties could not be reliably measured below -90 C due to spontaneous transformation before loading.

mode, a linear relation between plastic strain and the amount of transformation can be expected. This is consistent with the trend in the shape of the observed f - ϵ curves toward an initially linear behavior at low temperatures.

At temperatures above M_s^σ , significant plastic flow precedes the transformation, and an additional contribution to transformation can arise from the production of new nucleation sites by plastic deformation, defined as *strain-induced nucleation*.⁵ An analysis of the transformation kinetics based on a shear-band

intersection mechanism of strain-induced nucleation predicts a sigmoidal shape for the f - ϵ curve.⁶ This is consistent with the observed transformation curve shape at higher temperatures.

In alloys exhibiting athermal martensitic transformations, the stress required for stress-assisted nucleation below M_S^0 usually shows a linear temperature dependence, approaching zero at the M_S temperature. However, a minimum in this stress below M_S^0 for the warm-rolled material (Figure 5) results from the isothermal character of the spontaneous transformation in this alloy.⁷ The curves of Figures 5 and 6 suggest that, if the stress for stress-assisted nucleation below M_S^0 is extrapolated to temperatures above M_S^0 , the actual stress at which martensite is first observed (above M_S^0) is lower than would be expected from stress-assisted nucleation alone. Indeed the curve in Figure 6 suggests a region above M_S^0 where the stress at which martensite is observed continues to run somewhat parallel to the yield stress. Such effects are indicative of the additional contribution to transformation from strain-induced nucleation.

The comparison of the warm-rolled and solution-treated materials in Figure 5 clearly demonstrates the influence of thermomechanical treatment on the kinetics of subsequent martensitic transformation. In an absolute sense, the alloy is made more stable by the thermomechanical treatment. A higher stress is required for stress-assisted transformation after warm rolling than before. This is consistent with the observation that spontaneous isothermal transformation takes place in the solution-treated material after prolonged holding at 78 K (-196 C), whereas no spontaneous transformation could be made to occur after warm rolling. In a *relative* sense, however, the thermomechanical treatment has made the material *less* stable with respect to transformation during deformation. Figure 5 indicates that warm rolling has exerted a larger influence on the stress for slip compared to that for stress-assisted transformation, resulting in an increase in the M_S^0 temperature. Such a decrease in relative stability has been attributed⁸ to the thermodynamic effect of carbide precipitation (removal of some carbon from solution) during warm rolling. However, a basic difference in the effect of strengthening mechanisms on slip dislocations versus transformation dislocations is another likely contribution. This decreased relative stability exerts a large influence on the room-temperature behavior. Comparison of the room-temperature transformation curves in Figures 2 and 3 shows that although a negligible amount of martensite is produced by deformation of material in the solution-treated condition, a substantial amount is produced by deformation after warm rolling. The data of Figures 2 and 3 further suggest that the decreased relative stability increases the M_d temperature for uniaxial tension (maximum temperature at which detectable martensite is formed by uniaxial tensile deformation) from approximately 25 C to greater than 150 C. The observed differences in absolute stability and relative stability with respect to slip are thus consistent with the observation reported for athermally transforming TRIP steels⁹ that thermomechanical treatment can lower the M_S temperature while simultaneously raising M_d .

6. OLSON, G. B., and COHEN, M. *Kinetics of Strain-Induced Martensitic Nucleation*. Met. Trans., v. 6A, 1975, p. 791-795.
7. OLSON, G. B., and COHEN, M. *Stress-Assisted Isothermal Martensitic Nucleation in TRIP Steels*. Presented at the 1975 Materials Science Symposium, ASM, and TMS-AIME, Cincinnati, Ohio, November 13, 1975.
8. GERBERICH, W. W., THOMAS, G., PARKER, E. R., and ZACKAY, V. F. *Metastable Austenites: Decomposition and Strength*. University of California, Berkeley, UCRL-20308, August 1970.
9. ZACKAY, V. F. *Some Anticipated Developments in Physical Metallurgical Research*. University of California, Berkeley, UCRL-18676, January 1969.

Martensite Morphology

As reported for lower strength Fe-Ni-based TRIP steels,¹⁰ the different conditions above and below the M_S^0 temperature are accompanied by changes in the morphology of martensite produced during deformation, and this might to some extent be related to differences in the dominant mechanisms of martensitic nucleation. Figure 7a shows the microstructure of the solution-treated 0.27C alloy after isothermal transformation at -196 C, and Figures 7b to f show the microstructures after deformation at increasing temperatures. The martensite formed by deformation at low temperatures has predominantly the same lenticular plate morphology as the product of the spontaneous transformation without external stress. This is consistent with the view that the transformation mechanism is identical, but thermodynamically assisted by the applied elastic stress. As the temperature of deformation is increased, these plates become less regularly shaped, probably due to some interference with growth from the increased amount of simultaneous slip in the austenite. A second feature in the microstructure which finally becomes dominant well above M_S^0 (Figures 7e and f) arises from small units found predominantly at the intersection

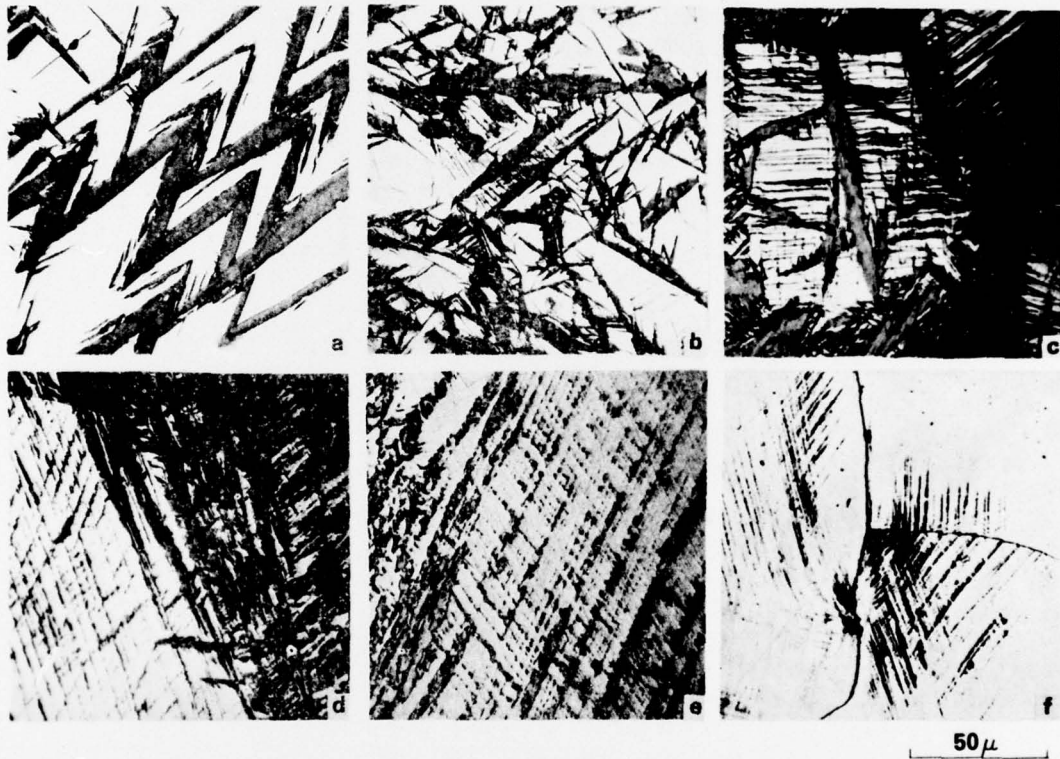


Figure 7. Morphology of martensites formed in 0.27% carbon TRIP steel solution treated at 1200 C. (a) Martensite formed by isothermal holding at -196 C for 8 hours without applied stress, (b to f) martensite formed by deformation at various temperatures: (b) -196 C, (c) -128 C, (d) -75 C, (e) -28 C, and (f) 25 C.

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10. MAXWELL, P. C., GOLDBERG, A., and SHYNE, J. C. *Stress-Assisted and Strain-Induced Martensites in Fe-Ni-C Alloys*. *Met. Trans.*, v. 5, 1974, p. 1305-1318.

of shear bands in the austenite. These are consistent with the shear-band intersection mechanism of *strain-induced* nucleation observed in austenites of low intrinsic stacking-fault energy.¹¹

Figures 8a to f show the microstructure of the warm-rolled material as the test temperature is increased. Again, deformation at the lowest temperatures produces fairly coarse plate martensite which becomes finer and less regular as the deformation temperature is raised. This is accompanied by a larger contribution from the fine shear band intersection morphology (Figures 8c to e). At the highest temperatures, the transformation product is rather finely dispersed and the morphology poorly defined. This is in part due to the fact that the martensite forms extremely gradually with plastic strain such that by the time an appreciable quantity of martensite has been generated, the martensite itself is highly deformed. The microstructures in Figures 8e and f are taken from the highly deformed necked region of the tension specimens.

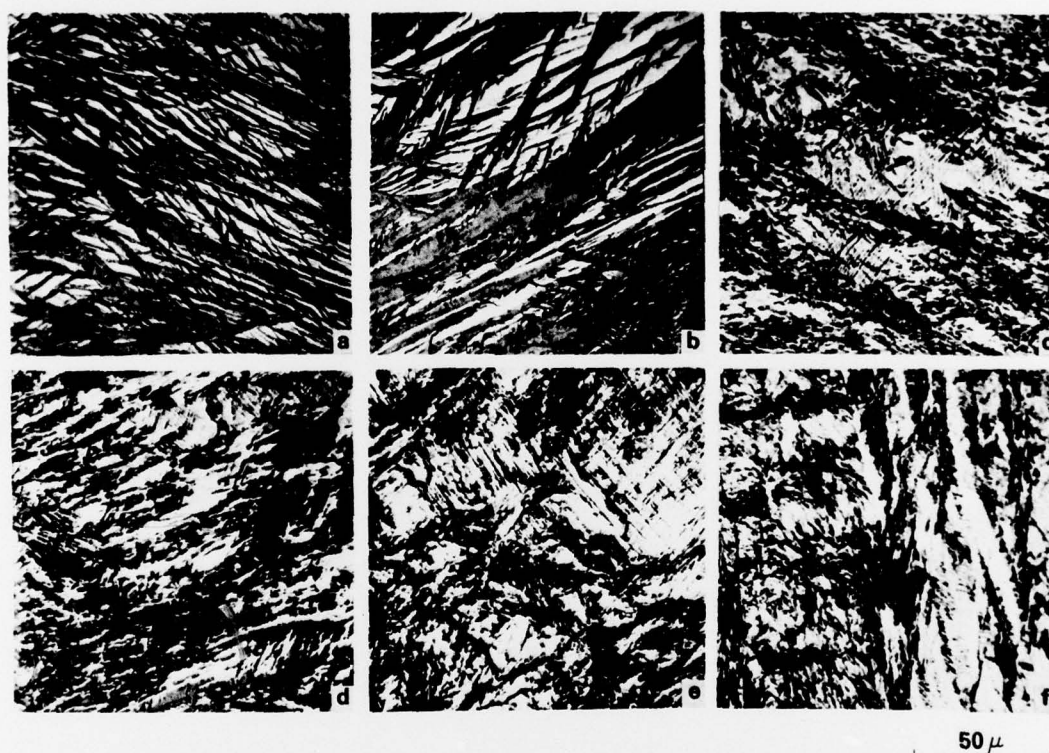


Figure 8. Morphology of martensites formed in 0.27% carbon TRIP steel in warm-rolled (78% RA, 450 C) condition. Martensite formed by deformation at: (a) -196 C, (b) -115 C, (c) 25 C, (d) 60 C, (e) 100 C, and (f) 150 C.

19-066-133/AMC-77

11. LECROISEY, F., and PINEAU, A. *Martensitic Transformations Induced by Plastic Deformation in the Fe-Ni-Cr-C System*. *Met. Trans.*, v. 3, 1972, p. 387-396.

Although it is tempting to try to correlate martensite morphology with the mechanism of nucleation (i.e., stress-assisted versus strain-induced), the only clear case in which this is possible occurs when the product is found at shear band intersections; such intersections clearly represent new nucleation sites induced by plastic strain. The situation is not as clear for the plate-shaped product produced by elastic stress at lower temperatures, for even in the spontaneous transformation without stress the majority of the plates are attributed to autocatalytic nucleation. A major contribution to autocatalysis is believed to be strain-induced nucleation resulting from plastic accommodation in the austenite surrounding a forming plate.¹² Hence, many of the coarse plates formed below M_S^σ may actually arise from strain-induced nucleation, though possibly by a mechanism different from the shear-band intersection mechanism which is so important above M_S^σ . Below M_S^σ , then, only the very first plates to form, initiating plastic flow and thus determining the yield stress, can be unambiguously attributed to stress-assisted nucleation from pre-existing nucleation sites.

Comparison of Figures 7b and 8a reveals an interesting difference in the arrangement of martensite plates formed under conditions where plastic flow is controlled by the transformation. Transformation plasticity, or the macroscopic strain accompanying martensitic transformation under stress, has generally been attributed to a biasing of the plastic accommodation processes around a forming martensite plate, with only a small contribution from orientation of the plates themselves.¹³ This is based on studies of the transformation from lower strength austenites in which no obvious biasing of the distribution of plate variants under stress is observed. This also appears to be the case for the solution-treated material shown in Figure 7b in which the distribution of plates appears as random as the distribution formed without stress in Figure 7a. The plates formed under stress in the high-strength TRIP steel of Figure 8a, however, clearly show a predominance of nearly parallel plates. This stronger biasing of the plate variants might be attributed to the higher stress levels (~200 ksi or 1400 MPa) required to induce transformation in the thermomechanically processed TRIP steels. Here, rather than a small perturbation on the large chemical driving force for transformation, the orientation-dependent thermodynamic assist of applied stress in this case may account for a significantly larger fraction of the total driving force.*

Stronger biasing of plate variants in the high-strength TRIP steel suggests an additional contribution to transformation plasticity which should increase the amount of strain per unit of transformation under conditions where transformation is the controlling deformation process. This is supported by the initial slopes of the f versus ϵ curves in Figures 2, 3, and 4 at the lowest temperatures. The curves for both the 0.19C and 0.27C solution-treated materials reach the same slope corresponding to a true tensile strain of 0.05% per 1% martensite. The curves for the high-strength warm-rolled material, however, approach a slope which corresponds to a larger strain of 0.08 to 0.10% per 1% martensite.

*An additional contribution to the biasing of the plate variants might arise from anisotropy of the dislocation substructure produced by thermomechanical treatment.

12. KNOROVSKY, G. A. *Autocatalysis of Martensitic Transformations*. Sc. D. Thesis, Massachusetts Institute of Technology, Department of Materials Science and Engineering, Cambridge, MA, February 1977.

13. SATTler, H. P., and WASSERMAN, G. *Transformation Plasticity During the Martensitic Transformation of Iron with 30% Ni*. *J. Less Common Metals*, v. 28, 1972, p. 119.

Mechanical Behavior

A "rule of mixtures" relation has been proposed to relate the flow stress of TRIP steels to the deformation-induced martensite content:⁸

$$\sigma = \sigma_A + f\Delta\sigma. \quad (1)$$

Here σ_A is the austenite flow stress and $\Delta\sigma$ is the strength difference between austenite and martensite. A quantitative test of Equation 1 with the data obtained in this study would require independent measurement of the σ - ϵ behavior of stable austenite and martensite. However, the σ - ϵ and f - ϵ curves of Figures 2, 3, and 4 allow at least a qualitative comparison leading to the behavior depicted schematically in Figure 9.

Austenite flow properties measured at higher temperatures and martensite flow properties measured at high strains at lower temperatures in earlier work¹ suggest that the σ - ϵ curves of stable austenite and martensite in warm-worked TRIP steels can be reasonably represented by linear work hardening with comparable rates for both phases. Accordingly, the dotted lines labelled σ_A and σ_M in Figure 9 represent the flow properties of stable austenite and martensite. The sigmoidal transformation curve observed at temperatures above M_S^σ and its associated σ - ϵ curve are denoted by the solid curves in Figure 9a. The dashed σ - ϵ curve is the prediction of Equation 1. As the figure suggests, the curves measured above M_S^σ are in qualitative agreement with the rule of mixtures in that the σ - ϵ curves possess the same basic sigmoidal shape as the f - ϵ curves. The σ - ϵ curve measured at 60 C for the warm-rolled material is consistent with the earlier observation¹ that Lüders-band formation can arise from a low initial work-hardening rate which need not be accompanied by a drop in true stress. A low initial work-hardening rate is predicted by Equation 1 for a material exhibiting a sigmoidal transformation curve. However, as the dashed and solid curves of Figure 9a suggest, the data obtained here indicate an additional contribution to the low initial work-hardening rate which causes the measured σ - ϵ curve to fall below the prediction of Equation 1. This effect can be

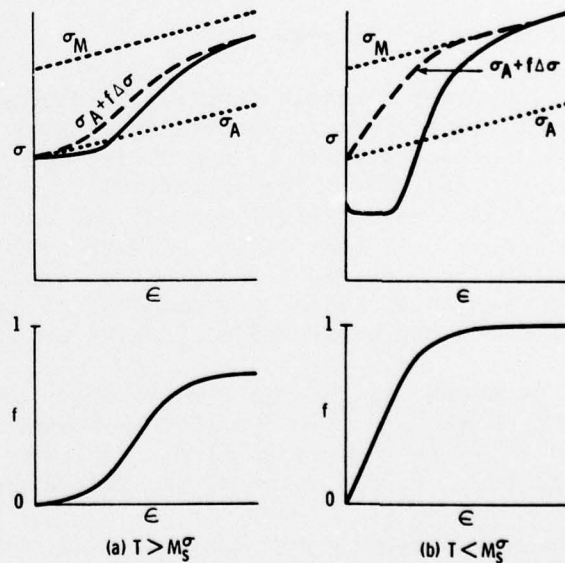


Figure 9. Schematic relation of tensile flow properties and transformation behavior for high-strength TRIP steel. Dotted lines represent σ - ϵ behavior of stable austenite (σ_A) and martensite (σ_M). Solid curves show observed shape of σ - ϵ and f - ϵ curves of metastable austenite. Dashed curve shows predicted σ - ϵ curve from "rule of mixtures" (Eq. 1), based on shape of associated f - ϵ curve. (a) Behavior of material at temperatures above M_S^σ and (b) behavior below M_S^σ .

seen by comparison of the σ - ϵ curves at 60 C and 100 C in Figure 2. The 100 C curve can be taken to represent the properties of stable austenite since only a small amount of martensite is formed at this temperature. Although the 60 C and 100 C curves begin at the same yield stress, the 60 C curve falls below the 100 C curve when transformation begins. Equation 1 requires that the flow stress of the two-phase mixture must always be above that of the stable austenite. Clearly, in addition to the basic transformation hardening effect represented by Equation 1, there is a softening phenomenon associated with martensitic transformation during deformation.

Although a small effect above M_S^σ , "transformation softening" is a dominant factor at temperatures below M_S^σ where transformation plasticity becomes the controlling deformation mechanism. This is particularly so for the higher strength warm-rolled material as depicted schematically in Figure 9b. Comparison of the solid and dashed σ - ϵ curves shows that a true stress drop is observed where Equation 1 would predict a maximum rate of work hardening. This discrepancy is due to the basic assumption of Equation 1 that the transforming material, at a given strain, can be treated as a "static" two-phase composite material. This does not take into account the operation of the transformation itself as a deformation mechanism. When transformation occurs at stresses below σ_A , the flow stress is the stress required to maintain an imposed rate of transformation plasticity. The basic shape of the σ - ϵ curve under these conditions can be accounted for with existing models for the kinetics of isothermal martensitic nucleation. The stress drop on yielding is an expected result of autocatalytic nucleation which increases the density of nucleation sites, thus lowering the nucleation frequency required for a given rate of transformation.* These effects are not strong enough to produce a macroscopic true stress drop in the lower strength austenites of Figures 3 and 4, but a softening contribution upon yielding is still evident at the lowest temperatures. Although the f - ϵ curves become linear, the σ - ϵ curves consistently show upward curvature. Hence, in relating mechanical behavior to the kinetics of martensitic transformation during deformation, we must consider not only the "static" effect of two-phase hardening, but also the "dynamic" softening contribution of transformation plasticity.

Temperature Sensitivity

The plastic strain required to produce a given amount of martensite as a function of temperature is plotted in Figures 10 and 11 for the 0.27C and 0.19C material. It is apparent that at M_S^σ and above, the transformation kinetics are extremely temperature sensitive. The current TRIP steels unfortunately give the best combination of strength and ductility near M_S^σ and are therefore designed with M_S^σ near ambient temperature. In this region of high temperature sensitivity, only a small amount of adiabatic heating is required to drastically alter the shape of the stress-strain curve with an attendant serious loss of uniform ductility. This is responsible for the severe strain-rate effects which can be observed in high-strength TRIP steels.¹

As shown best by the high-strength material in Figure 10a, at temperatures below M_S^σ where flow is controlled by transformation plasticity, the temperature sensitivity of the transformation kinetics and associated flow properties is greatly diminished. If the shape of the stress-strain curve under these conditions could

*OLSON, G. B., and COHEN, M., M.I.T., unpublished research.

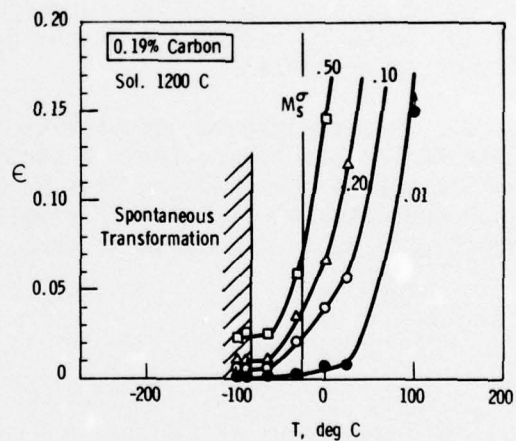
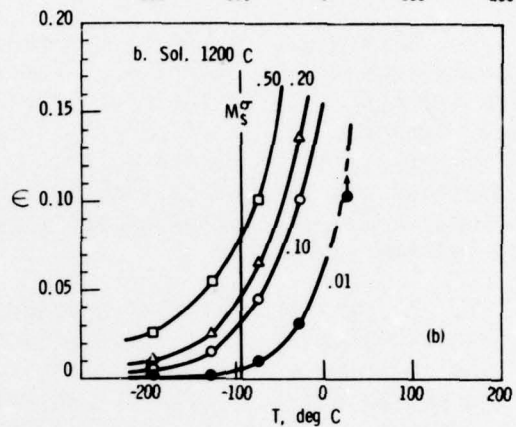
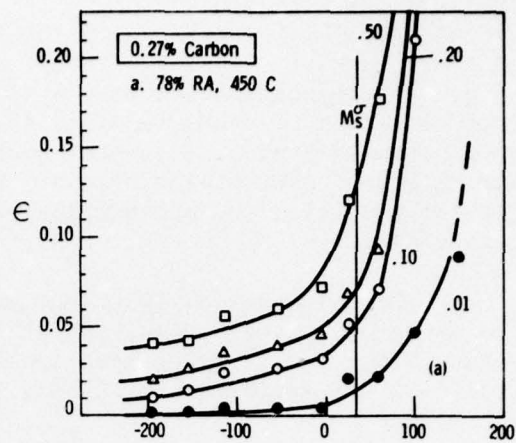


Figure 10. Temperature dependence of plastic strain required to give fixed volume fractions of martensite in 0.27% carbon TRIP steel: (a) warm-rolled material (78% RA, 450 C) and (b) solution-treated material (1200 C).

Figure 11. Temperature dependence of plastic strain required to give fixed volume fractions of martensite in 0.19% carbon TRIP steel in solution-treated (1200 C) condition.

be controlled to give the desired optimum combination of strength and ductility, TRIP steels designed to operate in this temperature regime should be far more suitable for dynamic applications where high strain rates and adiabatic heating effects are encountered. It is hoped that progress in understanding the mechanism and kinetics of isothermal martensitic nucleation and its interaction with applied stress will allow the development of this type of TRIP steel, and the full potential of this unusual class of alloys can then be realized.

CONCLUSIONS

a. The transformation curves (f versus ϵ) measured with respect to true plastic strain for TRIP steels in both the solution-treated and warm-rolled conditions have a sigmoidal shape at temperatures above M_S^σ consistent with the behavior of other metastable austenitic steels. Below M_S^σ , where stress-assisted transformation becomes the controlling deformation mechanism, the curves approach an initially linear behavior.

b. Thermomechanical treatment makes the TRIP steel more stable in an absolute sense as shown by the suppression of the spontaneous transformation on cooling. At the same time, the alloy is made *less* stable relative to slip, as attested by the increase in the M_S^σ temperature and a larger amount of martensite for a given plastic true strain.

c. Martensite formed by spontaneous transformation on cooling and by stress-assisted transformation below M_S^σ has a plate morphology. Unlike the behavior of lower strength steels, the distribution of plates in the high-strength thermomechanically treated TRIP steel shows a very strong biasing effect of external stress. The martensite units formed at higher plastic strains are less regularly shaped. In addition, an increasing contribution from fine units formed at shear band intersections is found at deformation temperatures above M_S^σ where strain-induced nucleation becomes dominant.

d. A "rule of mixtures" approach which considers only the "static" strengthening contribution of the transformation product provides a reasonable approximation for the relation of the deformation-induced transformation kinetics (f versus ϵ) to the mechanical flow behavior (σ versus ϵ) at temperatures above M_S^σ , although with less accuracy at the lowest strains. The discrepancy at low strains can be attributed to a dynamic "transformation softening" contribution which becomes dominant below M_S^σ and is due to the operation of transformation plasticity as a deformation mechanism.

e. The temperature sensitivity of the transformation kinetics with respect to plastic strain is greatest at temperatures above M_S^σ . If the transformation kinetics and its relation to flow properties could be properly controlled, the temperature sensitivity of TRIP steels could be minimized by designing alloys to operate with optimum mechanical properties below the M_S^σ temperature.

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True stress (σ), true strain (ϵ), and volume fraction martensite (f) were measured during both uniform and localized flow as a function of temperature on TRIP steels in both the solution-treated and warm-rolled conditions. The transformation curves (f versus ϵ) of materials in both conditions have a sigmoidal shape at temperatures above M_s (maximum temperature at which transformation is induced by elastic stress) but approach initially linear behavior at temperatures below M_s where the flow is controlled by transformation plasticity. The martensite which forms spontaneously on cooling or by stress-assisted transformation below M_s exhibits a plate morphology. Additional martensite units produced by strain-induced nucleation at shear band intersections become important above M_s . Comparison of σ - ϵ and f - ϵ curves indicate that a "rule of mixtures" relation based on the "static" strengthening effect of the transformation product describes the plastic flow behavior reasonably well above M_s , but there is also a dynamic "transformation softening" contribution which becomes dominant below M_s due to the operation of transformation plasticity as a deformation mechanism. Temperature sensitivity of the transformation kinetics and associated flow behavior is greatest above M_s . Less temperature-sensitive TRIP steels could be obtained by designing alloys to operate with optimum mechanical properties below M_s .

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