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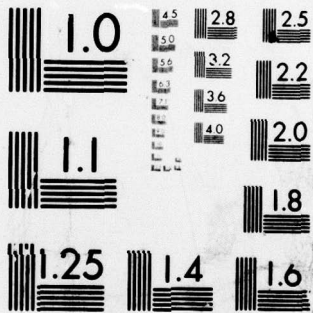UNCLASSIFIED   TR-3   NL

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I. Introduction

The true stress-strain plastic flow and transformation behavior of high-strength TRIP steels were recently measured during both uniform and localized flow as a function of temperature [1]. It was determined that the basic mode of transformation on cooling in these alloys is isothermal, and that the plastic flow at low temperatures is controlled by stress-assisted martensitic transformation. For these conditions, it is of interest to compare the observed transformation and flow behavior with that predicted from our knowledge of the kinetics of isothermal martensitic transformations.

II. Temperature Dependence of Transformation and Flow Behavior

The observed temperature dependence of the yield stress and the stress at which 1% martensite was detected [1] are shown in Fig. 1, indicating that transformation controls flow below the M_s^σ temperature (maximum temperature at which transformation is induced by elastic stress). When plastic flow is controlled by stress-assisted isothermal transformation, yielding will occur at the stress for which the rate of transformation plasticity matches the imposed strain rate. Since the amount of transformation plasticity was found to be linearly related to the amount of transformation in this temperature region [1], this corresponds to a fixed rate of isothermal martensitic transformation.

The kinetics of isothermal martensitic transformations are known to be nucleation controlled with an activation energy that is linearly dependent on the transformation free-energy change, ΔG. The transformation rate can then be expressed as [2]:

$$\dot{f} = n_s V \nu \exp\left(- \frac{A+B\Delta G}{RT}\right) \qquad (1)$$

where f is the volume fraction martensite, n_s the density of nucleation sites, V the instantaneous mean martensitic plate volume, ν the lattice vibration frequency, A and B are constants. The critical ΔG for a fixed $\dot{f}$ is then given by:

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$$\Delta G_{crit}(\dot{f}) = -\frac{1}{B}\left(A + RT\ln\frac{\dot{f}}{n_s \bar{V}\nu}\right) \tag{2}$$

indicating a linear temperature dependence.

The transformation chemical free-energy change (ΔG_{ch}) vs. temperature for the alloy of Fig. 1, estimated from available thermodynamic parameters [3], is plotted in Fig. 2. Calculating the additional thermodynamic assist of the applied tensile stress [4], the total ΔG at which transformation is observed for the low-temperature points in Fig. 1 is plotted to produce the ΔG_{crit} curve of Fig. 2. The linear relation of eq. 2 is verified, and the parameters defined by the straight-line fit then predict the solid curve shown in Fig. 1 representing the stress σ_t required for a fixed rate of stress-assisted isothermal transformation. That transformation is observed below this stress for temperatures above M_s^σ has been attributed [1] to the contribution of <u>strain-induced</u> nucleation defined as the production of new nucleation sites by plastic deformation.

The temperature sensitivity of the transformation kinetics in TRIP steels has been found to be a serious problem since heating effects encountered at moderate strain rates can cause a severe reduction in the uniform ductility associated with the deformation-induced transformation [5]. The temperature dependence of the stress-assisted transformation kinetics is zero at the minimum in the σ_t curve of Fig. 1 at
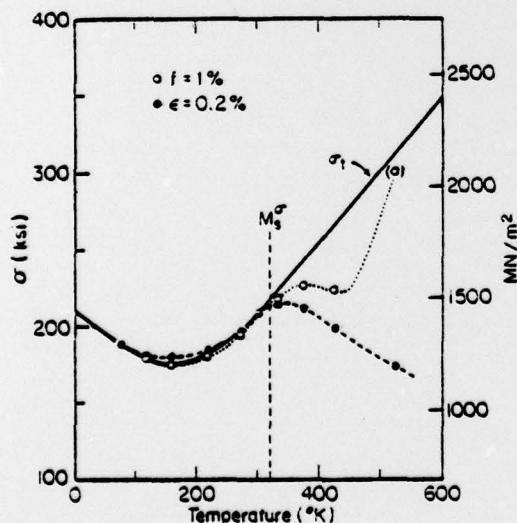

Fig. 1. Temperature dependence of stresses for plastic strain (ϵ) of 0.2% and for fraction martensite (f) of 1% for Fe-9Cr-8Ni-4Mo-2Si-0.8Mn-0.27C TRIP steel (78% RA at 450°C) [1].
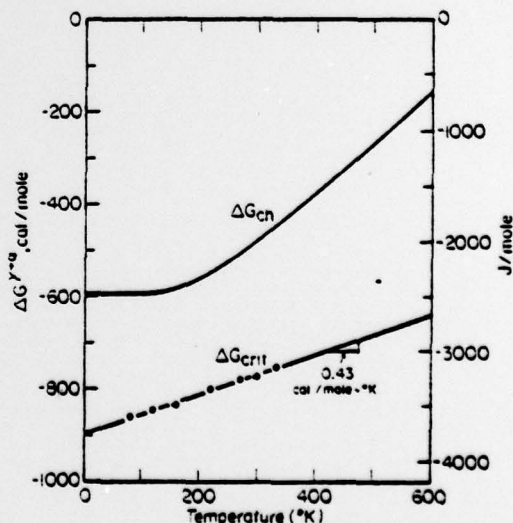
Fig. 2. Temperature dependence of transformation chemical free-energy change (ΔG_{ch}) and critical free-energy change (ΔG_{crit}) at which transformation is observed for alloy of Fig. 1.

~150°K. The temperature sensitivity of the contribution of stress-assisted nucleation to the overall transformation kinetics at ambient temperatures could thus be decreased by moving this minimum closer to 300°K. From Fig. 2 it can be seen that the σ_t minimum occurs when the slope of the $\Delta G_{ch}(T)$ curve (controlled by the transformation entropy change, ΔS_{ch}) matches the slope of the $\Delta G_{crit}(T)$ curve, corresponding to a value of 0.43 cal/mole°K (1.80 J/mole°K) for this alloy. Thermodynamic calculations suggest that substitution of manganese for nickel in these steels could bring the magnitude of the room-temperature transformation entropy change close to this value. In addition, isothermal transformation kinetic data [6] suggest that manganese may also alter the B parameter in eqs. 1 and 2 so as to favorably affect the slope of the $\Delta G_{crit}(T)$ curve.

The temperature dependence of the yield stress for a high-manganese TRIP steel [7] is shown in Fig. 3. Although the as-warm-rolled material does not show a pronounced stress minimum, tempering at 570°C to lower the austenite stability by removing some carbon from solution indicates a minimum above 200°K. From the thermodynamic effect of carbon [3], this is expected to represent a lower limit to the actual temperature of the σ_t minimum when the carbon is in solution. It thus appears that manganese can alter the kinetics of the stress-assisted transformation in the desired way. It has also been predicted that manganese will favorably influence the temperature sensitivity of the strain-induced transformation kinetics [8]. Unfortunately, too high a manganese content is found to reduce ductility as measured by reduction in area [7].

III. Isothermal Stress-Strain Curves

When plastic flow is controlled by stress-assisted transformation, our knowledge of the kinetics of isothermal martensitic transformations can also be used to predict the shape of the $\sigma-\epsilon$ curve. The course of isothermal transformation with time has been successfully modeled using eq. (1) with the number of nucleation sites per volume of sample, n_s, described by the relation [2,6]:

$$n_s = (n_i + pf - N_v)(1 - f) \qquad \cdot (3)$$

where n_i is the initial density of sites, p is an "autocatalytic factor" accounting for new sites produced during transformation, and N_v is the number of martensitic plates per unit volume (accounting for sites which have already operated). The 1-f factor takes into account potential sites which have been "swept up" by the transformation. Substituting eq. (3) in eq. (1) and incorporating an experimentally determined linear dependence of the overall average plate volume, $\overline{V}$, on N_v, a sigmoidal curve of volume fraction martensite, f, vs. time is predicted. The initial transformation rate is controlled by n_i while the autocatalytic p factor can lead to upward curvature. Eventually the decreasing average plate volume can cause the rate of consumption of nuclei to exceed the rate of production, and a saturation level of transformation is reached. This model is found to accurately fit experimentally determined isothermal transformation curves at small amounts of transformation, but the approach to saturation of the calculated curves is somewhat too abrupt relative to the experimental observations [2, 6].
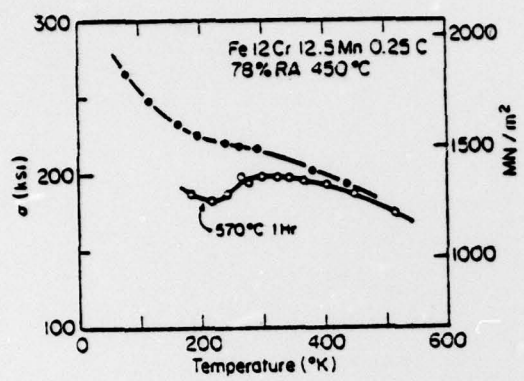
Fig. 3. Temperature dependence of yield stress of high-Mn TRIP steel in warm-rolled condition and after tempering [7].
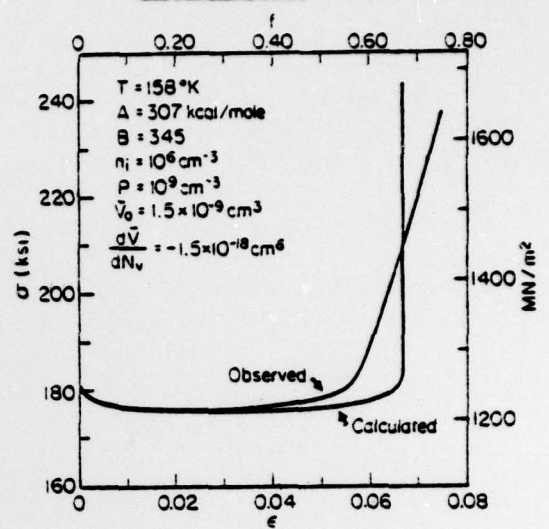
Fig. 4. Comparison of calculated and observed [1] true σ-ϵ curves for TRIP steel of Fig. 1.

Taking into account the thermodynamic effect of applied stress [4], this kinetic model can then express the stress required for a fixed rate of stress-assisted isothermal transformation as a function of the extent of transformation:

$$\sigma_t(f) = -(B\frac{d\Delta G}{d\sigma})^{-1} [A+B\Delta G_{ch} + RT\ln\frac{\dot{f}}{(n_i+pf-N_v)(1-f)V\nu}] . \quad (4)$$

Given the experimentally observed linear relation between transformation plasticity and the extent of transformation [1], we can substitute the simple relation f=kϵ, with k a constant, and obtain a complete constitutive relation predicting the σ-ϵ behavior when plastic flow is controlled by stress-assisted isothermal martensitic transformation. As the last term in eq. (4) suggests, the yield stress will be controlled by n_i, but on further straining (transformation) the p factor can cause the denominator to increase with an attendant drop in stress. As a saturation level of transformation is approached, the denominator approaches zero and the stress must rise rapidly.

Equation (4) was fitted to the σ-ϵ curve measured at 158°K for the TRIP steel of Fig. 1 [1] using the A and B parameters determined from Fig. 2 and typical values [2,6] of the other kinetic parameters. Calculated and observed curves are compared in Fig. 4. The model accurately accounts for the initial stress drop that gives rise to Lüders band formation in these steels, but as expected, the approach to a saturation level, with the attendant high strain hardening, is too abrupt. We believe that this discrepancy is due to the assumption of a singly-activated process, and better agreement could be obtained for both the
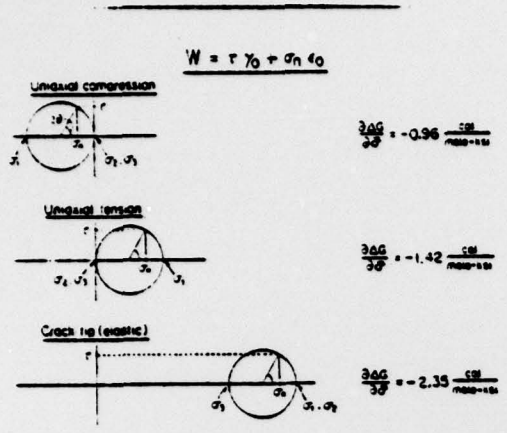
Fig. 5. Calculated work of resolved shear stress, τ, and normal stress, σ_n, assisting transformation shape change with shear and normal strains, γ_o and ϵ_o.
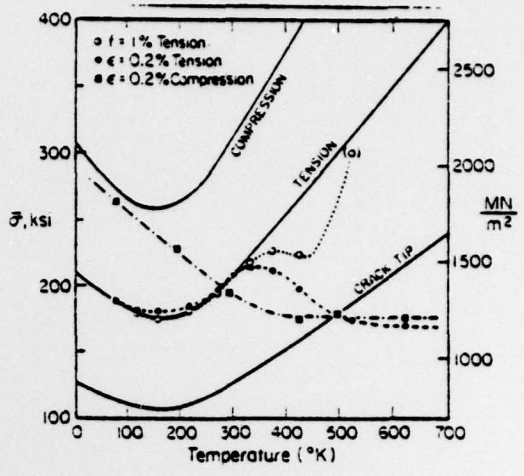
Fig. 6. Calculated transformation stresses and measured yield stresses for TRIP steel of Fig. 1 under different stress states.

conventional f vs. t data and the σ-ϵ curves by adopting a distribution of activation energies as indicated by the isothermal experiments of Magee [9].

IV. Influence of Stress State

Patel and Cohen [4] showed that, due to the transformation volume change, the work done by an applied stress assisting the shape change of a martensitic transformation (calculated for the most favorable plate orientation) will depend on the stress state. Figure 5 shows the relevant Mohr's circle diagrams and calculated values of $\partial\Delta G/\partial\bar{\sigma}$ for various stress states (using equivalent stress, $\bar{\sigma}$) based on the Patel and Cohen procedure. Using these values in eq. 4 we can predict the temperature dependence of the transformation stress, σ_t, for different stress states as indicated by the solid curves in Fig. 6. At temperatures where transformation controls flow, a large strength-differential (S-D) effect, comparing flow stress in tension and in compression, is expected. The measured yield stress in compression for the same alloy represented in Fig. 1 is also shown in Fig. 6 and compared with the tensile yield values.

Below 300°K the expected large S-D effect is observed. The curves suggest that flow is controlled by transformation in tension, but probably by slip in compression. Above 500°K where slip controls flow in both tension and compression, a normal S-D effect of a few percent is observed. However, in the region between 300°K and 500°K where slip is apparently still controlling flow, an anomalous negative S-D effect is observed in which the tensile yield strength is significantly higher

than the compressive yield. In thermodynamic terms, the comparison of flow stress for two stress states is equivalent to comparing alloys of differing thermodynamic stability, and this effect is therefore analogous to a similar observation by Breedis and Robertson [10] in measuring the critical resolved shear stress for slip in Fe-Cr-Ni crystals of different Ni contents. It has been suggested that such "pre-transformation strengthening" effects might arise from an influence of lattice metastability on dislocation mobility as the conditions for spontaneous dislocation dissociations related to martensitic nucleation are approached [11].

The calculated $\sigma_c(T)$ curve for the stress state of an elastic crack tip is also shown in Fig. 6. It appears that transformation will control flow over a wider temperature range and cause a significant reduction in effective flow stress. Such an effect may be desirable from the standpoint of strength/toughness combinations, allowing a high uniaxial tensile strength to be achieved while retaining the crack tip plasticity of a much softer material.

V. Closure

The martensitic transformation represents a unique deformation mechanism in that its kinetics can be studied independently of applied stress and the information obtained then used to predict a variety of aspects of mechanical behavior. The potential exists for the achievement and control of rather unusual mechanical properties.

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