Statistical Aspects of Martensitic Nucleation

G. B. Olson, K. Tsuzaki*, and Morris Cohen

Department of Materials Science and Engineering
MIT, Cambridge, MA 02139

ABSTRACT

Analysis of the Cech-Tarnbull small-particle martensitic transformation experiments in terms of heterogeneous nucleation theory defines an exponential nucleation-site potency distribution. Sensitive acoustic-emission detection of martensitic nucleation events shows that the same form of distribution describes the behavior of bulk polycrystals, and the influence of heat treatment on the distribution amplitude can be identified in this way. On the assumption of a random distribution of pre-existing nucleation-site orientations, the effect of applied stress on the effective potency distribution has been calculated, thus accounting for an observed nonlinear stress dependence of the transformation kinetics. The model also predicts a transformation "yield locus" for multiaxial stress.

INTRODUCTION

While materials science has much in common with condensed matter science, an important difference is that it must include a science of "dirt" if we are to have any hope of understanding the phenomena governing the behavior of the

* Present address: Kyoto University, Sakyo-Ku, Kyoto 606, Japan.

technologically important subset of condensed matter that we call materials. An excellent example of the pioneering role of Professor David Turnbull in establishing this branch of science is the classic Cech-Turnbull small-particle experiment of 1956, which quantified the heterogeneous nature of martensitic nucleation and established principles of broad relevance to heterogeneous phenomena controlled by sparse defects. We here assess the pivotal role of this experiment in light of recent developments, with particular attention to the emerging technology of transformation plasticity.

THE NUCLEATION-SITE POTENCY DISTRIBUTION

1. Small Particle Behavior

The results of the Cech-Turnbull experiment on an Fe-30Ni alloy are summarized in Fig. 1, showing the fraction $\chi$ of particles which contain detectable martensite on cooling to various temperatures as a function of particle size. If a number density $N_v$ of nucleation sites is randomly distributed throughout the volume, the probability $p$ that a particle of volume $V_p$ contains at least one such site is (Cohen and Olson, 1976):

$$p = 1 - \exp (-N_v V_p).$$

(1)

If sites occur only on particle surfaces, a similar function of surface area obtains. Since the transformation of bulk materials demands a volume distribution of sites, the two possible alternatives are that nucleation occurs according to a volume distribution or a combination of volume and surface distributions. Of these possibilities, the simple volume distribution best describes the behavior observed in Fig. 1. Identifying $\chi$ with $p$, the best fit of Eq. (1) is represented by the solid curves in Fig. 1 and these then define the value of $N_v$ for sites of sufficient potency to nucleate the transformation on cooling to a given temperature $T$. From knowledge of the fcc-bcc trans-
Fig. 1. Fraction $\chi$ of Fe-30Ni particles showing detectable martensitic transformation on cooling to various temperatures ($^\circ$C) vs. particle diameter (Cech and Turnbull, 1956). The curves represent best fits of Eq. (1).
Fig. 2. Cumulative number density $U_v = N_v$ of operational nucleation sites as a function of molar chemical free energy change $\Delta G^{ch}$ and corresponding cooling temperature, $T$ (Cohen and Olson, 1976). The curve is the best fit of Eqs. (2) and (3).
formation thermodynamics of this alloy, the potency of nucleation sites can be equivalently expressed in terms of the critical volume chemical free energy change $\Delta g_{\text{ch}}$ for their operation, and the cumulative potency distribution thus defined by the Cech-Turnbull data of Fig. 1 is plotted in Fig. 2 (Cohen and Olson, 1976).

Based on a simple model of barrierless heterogeneous classical martensitic nucleation via elastic interaction with superdislocation-like linear defects similar to those commonly observed in strain-induced nucleation experiments, the potency of a nucleation site operating at a given volume free energy change $\Delta g$ can be expressed in terms of a defect size parameter of the form (Olson and Cohen, 1976, 1981):

$$
n = \frac{-2\gamma/d}{\Delta g + g_{\text{ef}} + w_f}
$$

where $\gamma$ is the nucleus specific interfacial energy, $d$ is the close-packed interplanar spacing, $g_{\text{ef}}$ is an elastic coherency strain energy, and $w_f$ is the frictional work of interfacial motion. With estimated values of $\gamma=0.15$ J/m$^2$ and $g_{\text{ef}}+w_f = 6.1 \times 10^7$ J/m$^3$, the thermodynamic potency distribution of Fig. 2 is consistent with a cumulative structural defect potency distribution of the form:

$$
N_v(n) = N_v^0 \exp (-\alpha n)
$$

with $\alpha$ a constant shape factor and $N_v^0$ the total number density of defects of all potencies. The solid curve in Fig. 2 is the best fit of Eqs. (2) and (3) giving $\alpha = 0.84$ and $N_v^0 = 2 \times 10^{17}$ m$^{-3}$. The form of Eq. (3) was theoretically derived (Cohen and Olson, 1976) for a simple defect model appropriate to the nucleation mechanism represented by Eq. (2), and is typical of experimentally observed distribution functions for sparsely distributed nonequilibrium structural defects as encountered in fracture and fatigue. The Cech-Turnbull experiment thus defines a structural "dirt" function which may be of comparable fundamental significance to heterogeneous phenomena in
materials as the Boltzmann function is to thermally activated phenomena. The distribution of Fig. 2 implies that the number of defects initiating transformation at the $M_s$ temperature of underformed bulk materials can be quite low; the search for the scarce highly potent defects described by Eq. (3) seems to have more in common with high-energy particle physics than condensed matter physics. Observations of nucleation at more plentiful low potency defects in metallic alloys generally provide qualitative support for the classical heterogeneous nucleation mechanism of Eq. (2). The most quantitative confirmation of theory has come from recent observations of nucleation at well-characterized low potency defects in small particles of ZrO$_2$ ceramics (Chen and Chiao, 1985a). The statistics of transformation behavior in the ZrO$_2$ particles is found to conform to the same form of distribution function as the Cech-Turnbull experiment, extending the available data to smaller particle sizes and confirming the generality of Eq. (3) (Chen et al., 1985).

2. Bulk Polycrystal Behavior

Measurements have now been extended to the extreme high potency tail of the defect distribution using highly sensitive acoustic emission measurements to detect initial nucleation events in macroscopic polycrystalline specimens of an Fe-32.3Ni-0.004C alloy. The acoustic-emission transformation start temperature $M_{s\ AE}$ was measured using a Dunegan S9204 175kHz resonance transducer with a Panametrics 5660B preamplifier, a 100-300 khz band pass filter, and a Dunegan 301 totalizer operated at a total system amplification of 50-90dB. The conventional transformation burst temperature $M_b$ was also monitored by simultaneous electrical resistivity measurements. The grain size of specimens of $5 \times 10^{-8}$ m$^3$ volume was varied between 75 and 1000$\mu$m by austenitizing at 850-1350°C followed by a one-hour stress-relief anneal at 500°C. Figure 3 shows $M_{s\ AE}$ (open symbols) and $M_b$ (closed symbols) measured at a cooling rate of 0.5°C/min, plotted as a function of grain diameter $D$ and corresponding linear intercept $l$. Cir-
Fig. 3. Grain-size dependence of acoustic-emission $M_s^{AE}$ (open symbols) and electrical-resistivity $M_b$ (closed symbols). Circles represent experiments conducted at 70 dB gain while triangles denote 90 dB gain. SC indicates single crystal value from similar alloy.

Circles depict experiments run at an acoustic emission system gain of 70 dB, while triangles represent a 90 dB gain. The $M_s^{AE}$ is insensitive to gain at larger grain sizes, but the highest gain is required to detect the earliest events in the finer grained material. Optical metallography revealed that the first event in the latter case corresponded to a small cluster of plates presumably autocatalytically connected to a single initial nucleation event. Serial sectioning at 50 $\mu$m intervals showed that in two instances a single martensitic plate had been detected.
Assuming the same defect-potency distribution form \((\alpha = 0.84)\) as derived from the Cech-Turnbull data, the solid \(M_s^{AE}\) curve in Fig. 3 corresponds to the prediction of Eqs. (1)-(3) with a grain-size dependent \(N_v^0\) of the form:

\[ N_v^0 = k_1 + k_2 \overline{D}^{-m} \]  

with \(m = 2.6, k_1 = 1.73 \times 10^{12} \text{ m}^{-3}, \) and \(k_2 = 1.4 \times 10^6 \) (\(\overline{D}\) in meters). The value of \(k_1\) is estimated from the single-crystal \(M_s\) of a similar alloy (Machlin and Cohen, 1951) denoted by SC in Fig. 3. Over the 75-1000 \(\mu\)m grain-size range examined, \(N_v^0\) then varies by three orders of magnitude from \(9.0 \times 10^{13}\) to \(7.4 \times 10^{16} \text{ m}^{-3},\) extrapolating to the value derived from the Cech-Turnbull small-particle data at \(\overline{D} \approx 50\mu\)m. Although observed initial nucleation events were invariably related to grain boundaries, the \(\overline{D}\) dependence of \(N_v^0\) is stronger than the \(\overline{D}^{-1}\) dependence of the boundary area per volume, suggesting that more defective boundaries are associated with the finer grain sizes produced by lower temperature austenitizing.

The \(M_b\) temperature shows the form of \(\overline{D}\) dependence commonly observed in previous studies and reflects a grain-refinement inhibition of the spread of autocatalytic nucleation across boundaries, analogous to Hall-Petch strengthening. The solid \(M_b\) curve in Fig. 3 represents a relation of the form:

\[ M_b = k_3 + k_4 \overline{D}^{-1/2} \]  

with \(k_3 = -27.9^\circ\text{C}\) and \(k_4 = -5.67^\circ\text{C} \sqrt{\text{mm}}.\) This reflects the \(\overline{D}\) dependence of the effective potency of autocatalytic nucleation. Above a critical \(\overline{D}\) of \(\sim 500\mu\)m, the macroscopic transformation bursting behavior is triggered by a single nucleation event controlled by \(M_s^{AE}.\) Bursting at lower temperatures than the extrapolated \(M_b(\overline{D})\) relation leads to a catastrophic overdriven autocatalysis in which 40-50% of the sample transforms in a single burst.
Bars in Fig. 3 depict the standard deviation of the experimental scatter of $M_s^{AE}$ and $M_b$ relative to Eqs. (4) and (5). The larger scatter of the $M_s^{AE}$ measurements associated with the first nucleation event is in good quantitative agreement with the corresponding confidence limits predicted by Eqs. (1) to (3). In addition, the statistical nature of the initial nucleation behavior governed by the defect-potency distribution predicts a specimen-size dependence of $M_s^{AE}$. This was confirmed in specimens with 950 $\mu$m grain size where the initial event triggers bursting. Over a range of specimen volumes of $1.7 - 13.3 \times 10^{-8}$ m$^3$, the $M_s$ temperature exhibited a systematic $20^\circ$C variation, consistent with prediction. The $M_b$ temperature of the finer $220 \mu$m grain size material was size independent over the same specimen volume range, indicating a deterministic nature of the autocatalytic spreading, in contrast to the statistical nature of initial nucleation.

**EFFECT OF APPLIED STRESS**

1. The Mechanical Driving-Force Distribution

   When martensitic transformation occurs under an applied stress as originally treated by Patel and Cohen (1953), the total volume free-energy change or "driving force" for transformation, $\Delta g$, is the sum of a chemical $\Delta g^{ch}$ and a mechanical $\Delta g^\sigma$ contribution. The $\Delta g^\sigma$ term is orientation dependent and, for a uniaxial stress, $\sigma$, can be expressed by:

   $$\Delta g^\sigma = \frac{\sigma}{2}\{\gamma_0 \sin 2\theta \cos \alpha + \epsilon_0(1 + \cos 2\theta)\} \quad (6)$$

   where $\gamma_0$ and $\epsilon_0$ are the transformation shear and normal strains, $\theta$ is the angle between the stress axis and the habit normal, and $\alpha$ is the angle between the transformation shear direction and the maximum shear stress direction resolved on the habit. The value of $\Delta g^\sigma$ lies in the range from
Fig. 4. Mechanical driving force distribution for random orientation distribution of nucleation sites under uniaxial applied stress.

\[
\Delta g_{\text{min}}^\theta = -\frac{1}{2} \left( -|\sigma| \sqrt{\gamma_0^2 + \varepsilon_0^2} + \sigma \varepsilon_0 \right)
\]

to

\[
\Delta g_{\text{max}}^\theta = -\frac{1}{2} \left( |\sigma| \sqrt{\gamma^2 + \varepsilon^2} + \sigma \varepsilon_0 \right).
\]

For a distribution of nucleation-site orientations, the effective site-potency distribution expressed in terms of the applied stress for site operation will be broadened relative to the structural potency distribution of Eq. (3). In the case of a random orientation distribution, this broadening can be predicted from a mechanical driving-force distribution computed from the fraction of \( \theta \) and \( \alpha \) values in Eq. (6) which yield a particular value of \( \Delta g^\theta \). Computed distributions for uniaxial tension and compression are shown in Fig. 4. The normalized total number of sites is plotted as a function of a normalized mechanical
driving force $\xi = \Delta g^\sigma / \Delta g^\sigma_{\text{max}}$. Due to the $\varepsilon_0$ dilational term, the distribution has two peaks at $\xi$ positions determined by the strain ratio $R = \varepsilon_0 / \gamma_0$. The distributions plotted in Fig. 4 are for $R = 0.2$, appropriate to the transformation in steels. The positions of the distribution centers indicate that nucleation sites with positive $\xi$ will be more plentiful than those with negative $\xi$ in tension and vice versa for compression.

Combining the driving-force distribution of Fig. 4 with the thermodynamic defect-potency distribution defined by Eqs. (2) and (3), the cumulative operational nucleation-site potency distribution for cooling under constant stress can be expressed as:

$$
N_v(\Delta g^\text{ch}_{\sigma}, \sigma) = \int_{\Delta g^\sigma = \Delta g^\sigma_{\text{min}}}^{\Delta g^\sigma_{\text{max}}} \frac{dN_v^0(\Delta g^\sigma)}{d\Delta g^\sigma} 
\times \exp \left( \frac{2a\gamma/d}{\Delta g^\text{ch} + \Delta g^\sigma + g^e + w_f} \right) d\Delta g^\sigma . \tag{7}
$$

For $\gamma_0 = 0.2$, Fig. 5 shows the predicted applied tensile stress for various critical values of $N_v$ for transformation, $N_v^c$, as a function of the increment in $\Delta g^\text{ch}$ relative to its value at the stress-free $M_s$ temperature, $M_s^0$. The $\Delta g^\text{ch}$ increment can be directly related to cooling temperature as in Fig. 2. Also represented are a normalized chemical driving force

$$
\phi = -\frac{\Delta g^\text{ch} + g^e + w_f}{2\gamma/d}
$$

and a normalized stress

$$
\Sigma = -\frac{\Delta g^\sigma_{\text{max}}}{2\gamma/d} .
$$

The predictions are compared with experimental results of Fisher (1974) for Fe20Ni 0.34C and Patel and Cohen (1953) for Fe20Ni 0.52C obtained for the stress dependence of the
Fig. 5. Stress-driving-force plot corresponding to experimental measurements of the stress-dependence of the $M_s$ temperature. Solid curves depict statistical model for $N_v^{c}/N_v^{o}$ from $10^{-4}$ to $10^{-13}$. Dashed curve represents $\Delta g^a = \Delta g^a_{\text{max}}$ model.

$M_s$ temperature. The Fisher data are in reasonable agreement with the nonlinear predictions of the statistical model with $N_v^{c}/N_v^{o} \approx 10^{-7}$. The latter value is typical of the $N_v^{c}$ corresponding to $M_b$ in the finer-grained material in Fig. 3. At high stress, however, the curves of Fig. 5 approach a limiting slope corresponding to $\Delta g^a = \Delta g^a_{\text{max}}$ represented by the dashed line. The latter corresponds to the prediction for a fully biased orientation distribution in which all nucleation sites are of the optimum orientation.

The data of Patel and Cohen appear to be in better agreement with the $\Delta g^a_{\text{max}}$ prediction. This can be rationalized if
the stronger bursting tendency of their higher-carbon alloy provides the supercritical behavior found in the coarse-grained material of Fig. 3 for which a single initial event triggers a burst. The prediction for this case corresponds to $N_v^c/N_v^o = 10^{-13}$ in reasonable agreement with the observed behavior, apart from an offset in $\Delta g^{ch}$ equivalent to a 4°C temperature interval. This is smaller than the 95% confidence limit of 8°C defined by the observed $M_s$ scatter in Fig. 3. This limit is denoted by the bracket in the lower right of Fig. 5. Within such experimental scatter, then, the difference between these two alloys can be rationalized in terms of the dependence of $N_v^c/N_v^0$ on the severity of autocatalysis.

While the behavior described thus far applies to cooling under constant stress, further calculations based on the statistical model predict enhanced transformation under conditions of isothermal loading to the same level of $\Delta g^{ch}$ and $\sigma$, in good agreement with such experimental comparisons reported by Bolling and Richman (1970). Cooling to temperature before monotonic loading allows the operation of some sites which would otherwise have an unfavorable interaction with applied stress. Extending the predictions of Fig. 5 into compressive stress, the statistical model also has the property of predicting a smooth curve in contrast to the sharp corner associated with a $\Delta g^s = \Delta g_{\text{max}}^s$ model.

2. Applications to Transformation Plasticity

Martensitic transformation under stress can provide an effective deformation mechanism for which constitutive relations can be derived from transformation kinetic theory. The mechanical behavior associated with retained austenite in martensitic steels and the toughening influence of metastable dispersed phases in certain ceramic systems such as partially stabilized zirconia have called attention to the special transformation plasticity behavior of dispersed particles in a solid matrix. When the dispersed metastable particles are sufficiently isolated that autocatalytic nucleation interactions can be
neglected, the constitutive relations for stress-assisted transformation plasticity can be predicted directly from the nucleation site potency distribution.

Following the considerations applied to the data of Fig. 1, the fraction of a large number of particles to transform to martensite can be described by the probability \( p \) of Eq. (1). When deformation is controlled by stress-assisted transformation, the transformed fraction \( f \) and the resulting plastic strain \( \varepsilon \) can be linearly related:

\[
f = \frac{\varepsilon}{\varepsilon_1}
\]

where \( \varepsilon_1 \) is the total strain at \( f = 1 \). Combining Eqs. (1)-(3) with the considerations previously described for isothermal monotonic loading, the isothermal \( \sigma-\varepsilon \) curve for transformation plasticity from dispersed particles of size \( V_p \) can be described by:

\[
f = \frac{\varepsilon}{\varepsilon_1} = 1 - \exp \left\{ -V_p \int_{\Delta g_{\text{min}}}^{\Delta g_{\max}} \frac{\text{d}N_\nu(\Delta g^\sigma)}{\text{d}\Delta g^\sigma} \times \exp \left( \frac{2\alpha\gamma/d}{\Delta g^\text{ch} + \Delta g^\sigma + g^\text{el} + w_f} \right) \text{d}\Delta g^\sigma \right. \\
+ \int_{\Delta g_{\text{max}}}^{\Delta g_{\text{max}}} \frac{\text{d}N_\nu(\Delta g^\sigma)}{\text{d}\Delta g^\sigma} \times \exp \left( \frac{2\alpha\gamma/d}{\Delta g^\text{ch} + \Delta g^\sigma + g^\text{el} + w_f} \right) \text{d}\Delta g^\sigma \left. \right\}.
\]

In the case of the simple \( \Delta g^\sigma = \Delta g_{\text{max}}^\sigma \) model the relation becomes
Fig. 6. Predicted $\sigma$-$\epsilon$ curves for transformation plasticity from non-interacting dispersed particles in uniaxial tension and compression. Particle volumes correspond to particle diameters of 10 to 50 $\mu$m, using $N_v^o$ from potency distribution of Fig. 2.

\[
f = \frac{\epsilon}{\epsilon_1} = 1 - \exp \left[ -V_p N_v^o \right.
\]
\[
\times \exp \left( \frac{2\alpha / d}{\Delta g_{\text{ch}} + \Delta g_{\text{max}} + g^e + w_f} \right) \]. \quad (10)
\]
Calculated normalized stress-strain curves ($\Sigma$-f) at $\phi = 0.07$ for $N_v^0 V_p = 10^2$, $10^3$, $10^4$ are depicted in Fig. 6. Unnormalized stress is shown along the right vertical axis. The stress-strain curve based on the $\Delta g_{\text{max}}^\sigma$ prediction for $N_v^0 V_p = 10^4$ is shown for comparison. For large particles where nucleation sites are plentiful, some particles transform to martensite during cooling before deformation ($N_v^0 V_p = 10^4$) and the transformation flow stress is low. For small particles, a high strain hardening arises from early exhaustion of the most potent sites, and further strain requires operation of less potent sites at higher stresses. The compression case shows a slightly higher strain hardening rate than in tension arising from the wider mechanical driving force distribution in compression. The stress level is higher than that of the $\Delta g_{\text{max}}^\sigma$ prediction by about a factor of two.

As the stress-strain curve of the $\Delta g_{\text{max}}^\sigma$ prediction can be obtained analytically, it is useful to express the stress-strain curve of the random orientation prediction by that of the $\Delta g_{\text{max}}^\sigma$ prediction as:

$$\Sigma_{\alpha}(f) = \beta \Sigma_m(f)$$  \hspace{1cm} (11)

where $\Sigma_{\alpha}(f)$ and $\Sigma_m(f)$ are normalized stresses of the random orientation prediction and the $\Delta g_{\text{max}}^\sigma$ prediction, respectively. The coefficient $\beta$ tends to be slightly smaller with increase in $f$ and with decrease in $N_v^0 V_p$, but is almost constant. The value of $\beta$ in uniaxial tension ($\beta_t$) is around 2.2 and the ratio $\beta_c/\beta_t$ is around 1.07, where $\beta_c$ corresponds to compression.

Recent compression experiments performed on dispersed ZrO$_2$ ceramics under superimposed hydrostatic pressure provide $\sigma$-$\epsilon$ curves very similar in shape to those predicted in Fig. 6, but show significant quantitative departures which are rationalized in terms of autocatalytic interactions among particles in the high volume-fraction dispersions investigated (Chen and Reyes Morel, 1986).
Fig. 7. Biaxial-stress transformation yield-locus shapes at three levels of normalized chemical driving force, $\phi_{ch}$. Sharp-cornered locus represents $\Delta g^{a}_{\max}$ model. $N_{v}^{c}/N_{v}^{a} = 5 \times 10^{-6}$.

The statistical model can also be used to calculate a transformation "yield" locus for multiaxial stress. Generalizing Eq. (6) and the $\Delta g^{a}$ distribution of Fig. 4 to other stress states and applying the type of analysis represented by Eq. (9), a computed yield locus for biaxial stress is represented in Fig. 7 for three values of normalized chemical driving force,

$$\phi_{ch} = \frac{\Delta g^{ch} + g^{cl} + w_{f}}{2\gamma/d}.$$
\( N^C_v / N^O_v = 5 \times 10^{-6} \) was deduced from the yielding behavior of austenitic TRIP steels. Stresses are normalized to the transformation stress in uniaxial tension, \( \sigma_o \). The smooth "egg-shaped" yield locus is compared with the sharp-cornered shape given by the \( \Delta g^o = \Delta g^\text{max} \) model. At low \( \phi_{ch} \), corresponding to higher transformation stresses, the locus moves toward the \( \Delta g^\text{max} \) limit, consistent with the high-stress limiting slopes depicted in Fig. 5. Further calculations for triaxial stress states indicate that the yield condition in the low-stress limit (high \( \phi_{ch} \)) can be reasonably approximated by a linear combination of tensor invariants, \( \sigma_h \) and \( \bar{\sigma} \). For tensile stress states ranging from plane stress to the plane-strain crack-tip case, this can be represented by:

\[
\bar{\sigma} = \sigma_s - k \sigma_h
\]  

(12)

with \( \sigma_s \) the yield stress in pure shear and \( k = 0.794 \) for the transformation in steels. Predictions of transformation behavior at a crack tip can thus be obtained from observations in uniaxial tension. Such predictions are in good agreement with recent experimental observations of crack-tip transformation behavior and associated toughening in high-strength TRIP steels (Léal, 1984).

**CLOSURE**

The martensitic nucleation-site potency distribution provides a rational basis for many statistical aspects of the kinetics of heterogeneous martensitic transformations, including specimen-size effects and the magnitude of experimental scatter. When combined with the mechanical driving-force distribution associated with a random distribution of nucleation site orientations relative to an applied stress, the statistical model also accounts for observed nonlinear stress effects, describes potency-distribution-based strain hardening in dispersed-phase transformation plasticity, and allows prediction of the transformation behavior under multiaxial stress.
A research program is now underway in which the predicted austenite-stabilizing effect of particle-size refinement is exploited to control the stability of dispersed austenite for optimum transformation plasticity behavior during plane-strain tensile deformation in order to provide enhanced stretch formability in high-strength microalloyed sheet steels, and enhanced fracture toughness in ultrahigh-strength martensitic alloy steels. These developments are a direct extension of the defect-potency distribution defined by the classic Cech-Turnbull small-particle experiments of 1956.

ACKNOWLEDGEMENTS

Research at MIT on the mechanism and kinetics of martensitic transformations is sponsored by NSF under Grant DMR-7915196. Research on transformation plasticity is sponsored by DOE under Grant DE-F02-84ER45154.

REFERENCES