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KINETICS OF MARTENSITIC INTERFACE MOTION

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Abstract. - The mobility of β-γi martensitic interfaces in thermoelastic β-CuAlNi alloys is investigated using stress-assisted single interface transformations over a temperature range of 180 to 410K and imposed interfacial velocities of 10^{-6} to 10^{-2} m/s. The behavior is consistent with thermally-activated interfacial motion, although an anomalous temperature dependence is observed below 210K which may be due to elastic softening. Measured activation energies of 0.3 to 0.4 eV (4.5 x 10^{-20} to 6.5 x 10^{-20} J) and activation volumes of 10^{3} to 10^{4} atomic volumes suggest that motion of the interface above 210K is controlled by interaction with discrete particles or clusters.

Introduction. - Few measurements of the mobility of martensitic interfaces under controlled conditions have been reported. Dislocation models of interfacial structure suggest that the interfaces should possess mobilities similar to those of slip dislocations (1,2). Thus, motion at low velocities is expected to be thermally-activated (3). This concept has formed the basis of a model for isothermal nucleation kinetics (1). Due to the well-defined geometrical aspects of deformation by martensitic interface motion (4), stress-assisted single interface motion in thermoelastic alloys provides an ideal opportunity to measure interface mobility in the thermally-activated regime. Such single-interface mobility measurements have so far been performed only for the β-γi martensitic transformation in a CuAlNi alloy deformed in uniaxial tension at one temperature (5). Subsequent experiments revealed that γi martensite in this alloy system exhibits a greatly extended temperature range of stability in compressive deformation as compared to tension (6).

The present study was undertaken to extend the single-interface mobility measurements to a wide range of temperatures using compressive deformation of a series of CuAlNi alloys, and to interpret the behavior in terms of the theory of thermally-activated deformation (3) in order to identify possible mechanisms for the thermally-activated motion.

Materials and Experimental Procedures. - CuAlNi alloy single crystals in the form of 5 mm diameter rods were grown by the Bridgman method. Two crystals, alloys A and B, were heat treated to give different stabilities with respect to the β-γi martensitic transformation, Table 1. Parallel-ended compression specimens 20 mm in length were cut from the crystals prior to final heat treatment. The compression axes were oriented a few degrees from [001]β. Constant velocity isothermal compression tests were conducted in appropriate isothermal baths using a 20,000 lb (89,000 N) Instron mechanical testing machine, covering a range of test temperatures and imposed velocities. A theoretical heat transfer analysis and direct measurements by thermo-couples mounted on specimens both revealed a maximum temperature rise at the interface due to the transformation latent heat), of 75K at the highest velocity employed (7.1 x 10^{-3} ms^{-1}). At velocities of 1.4 x 10^{-3} ms^{-1} and below, the temperature rise was less than 25K and only these data were employed for thermal activation analysis. Recorded applied load-crosshead displacement curves were converted to mechanical driving force - interface displacement curves using the known transformation
Table 1.

<table>
<thead>
<tr>
<th>Alloy Composition (wt. pct.)</th>
<th>Heat Treatment</th>
<th>M_σ</th>
<th>M_f</th>
<th>A_s</th>
<th>A_f</th>
</tr>
</thead>
<tbody>
<tr>
<td>A Cu14.3Al14.0Ni</td>
<td>1073K(4 min)-273K</td>
<td>218K</td>
<td>206K</td>
<td>249K</td>
<td>253K</td>
</tr>
<tr>
<td>B Cu15.1Al13.5Ni</td>
<td>1249K(10 min)-273K</td>
<td>157K</td>
<td>142K</td>
<td>166K</td>
<td>183K</td>
</tr>
</tbody>
</table>

Invariant-plane shape strain (7), resolving the applied loads in the shape-strain plane and direction and computing corresponding axial displacements for the most favorably oriented transformation variant (4). The computed correlation of crosshead and interface positions was verified by direct observation of the interface. A typical driving force-displacement plot as observed for T>A_f is shown schematically in Figure 1. Both the resolved shear stress, τ, and total driving force per unit volume, Δg, are indicated. At T_N (or Δ0N), the martensite nucleates at a specimen corner. The load then drops and the interface moves forward at a nearly constant level of τ and Δg. Reversing the direction of crosshead displacement, the interface moves in the reverse direction at a lower level. The average "friction stress" τ_f (and the corresponding Δg) is one-half of the resulting hysteresis for forward and reverse interface motion. If τ_f is assumed equal in magnitude in both directions, the hysteresis midpoint defines T_0 where Δg=0 (5,8). At higher temperatures, some Σ martensite formed at the ends of the specimen before the η| interface passed through. This small end effect did not affect the motion of the η| interface as it approached the center of the specimen.

Results and Discussion.- The measured τ_f and corresponding Δg values for alloys A and B are summarized in Figures 2 and 3, covering a total temperature range of ~180 to 410K and an interfacial displacement curve for isothermal, constant-velocity η| martensitic single-interface transformation. T>A_f.

The value of τ_f in alloy A appears to approach a constant minimum level at the lowest velocities and highest temperatures. This suggests an "athermal" contribu-
tion to $\tau_\infty$ of $\tau_\infty = 22$MN/m$^2$ (with a corresponding athermal $\Delta g$ level of $\Delta g_\infty = -2.2$MJ/m$^3$). The levels of $\tau_\infty$ and $\Delta g_\infty$ in alloy B are not as well defined but they are clearly lower, with an estimated value of $\tau_\infty = 15$MN/m$^2$, $\Delta g_\infty = -1.5$MJ/m$^3$. The athermal contributions are usually attributed to interface-obstacle interactions which are too long range to be surmounted thermally. Possible athermal obstacles are relatively large precipitates and inclusions which could produce a particle pinning force similar to that proposed by Zener (13) for grain boundary migration.

The thermal component of the friction stress, $\tau_f - \tau_\infty$, can be evaluated assuming an interface velocity law of the form

$$v = v_0 \exp \left( -\frac{Q}{kT} \right)$$

(1)

where the activation energy $Q$ is a function of $\tau$ or $\Delta g$, and $v_0$ is a temperature insensitive constant. At a given level of $\tau$ or $\Delta g$, $Q$ can be evaluated from the data plotted in Figures 2 and 3 using the partial derivative

$$Q = -\frac{\partial \ln v}{\partial (1/kT)}.$$ 

(2)

Such an analysis gives activation energies in the range of $Q = 0.3$ to $0.4$ eV ($4.5 \times 10^{-20}$ to $6.5 \times 10^{-20}$ J), and identifies the pre-exponential factor as $v_0 = 3.76 \times 10^4$ m/s.

The exact form of the function $Q(\tau)$, or $Q(\Delta g)$, which offers some insight into the nature of the rate-controlling barrier, is best evaluated from the rate-sensitivity of $\tau_f$ or $\Delta g$. A useful quantity in the analysis of thermally-activated deformation is the derivative

$$\frac{\partial Q}{\partial T} = -\frac{\partial \ln v}{\partial T}.$$ 

(3)

Expressed in units of volume, this derivative is often termed the "stress activation volume". For comparison with model-derived quantities, $V_T$ for the thermally-activated motion of a line dislocation of Burgers vector $b$ is usually represented as $V_T = bA^*$ where $A^*$, the "activation area", is the increment of area swept by the dislocation during the critical event (3). Hence, the critical event for the motion of a line dislocation is regarded as two-dimensional in character.
Discrete dislocation models for the structure of coherent interfaces (including the local structure of the twinned semicoherent interfaces dealt with here) describe the interface in terms of transformation or coherency partial dislocations which propagate the local transformation lattice deformation while maintaining continuity of the crystal lattice \((2,14,15)\). When the coherency dislocations in a particular array are relatively widely spaced, it is expected that the interaction of the individual coherency dislocations with a given obstacle involves an activation area \(A^*\) similar to that for the case of slip dislocations. If the coherency dislocations are closely spaced, the interface will behave as a quasi-continuous distribution of dislocations or a "surface dislocation". Then, when interacting with an obstacle, the critical event is more three-dimensional in character \((2)\). Accordingly, it is useful to replace \(V^*\) by an "activation volume" \(V^*\) defined as the derivative of activation energy with respect to volume driving force.

\[
V^* = \frac{\partial Q}{\partial \delta g} \bigg| _T .
\]

Fig. 4. Thermal component of friction stress (driving force) vs. activation volume for alloys A and B. \(T>210K\).

If we regard the local strain, \(\gamma\), accompanying the motion of a line dislocation as \(\gamma = b/d\) with \(d\) the interplanar spacing, then the local volume driving force is \(\delta g = -\gamma Y = -\gamma b/d\) and the activation volume \(V^*\) associated with an activation area \(A^*\) is \(V^* = dA^*\). Hence, \(V^*\) can conveniently be used to describe the behavior of both line and surface dislocations allowing direct comparison of the interaction of each with a given obstacle. An added benefit of the use of a volume driving force formulation for thermally-activated motion is that \(\delta g\) can be used to describe response to chemical forces as well as mechanical forces. For chemical forces such as those operative in the spontaneous martensitic transformation on cooling, a tensor quantity corresponding to \(\gamma\) does not exist.

From equations (1) and (4), we obtain

\[
V^* = -kT \left( \frac{\partial \ln \gamma}{\partial \delta g} \right)_T .
\]
Using this relation to evaluate \( V^* \) from the experimental data, the variation of \( \Delta g \) with \( V^* \) (and \( \tau_f \) with \( V^*_f \)) for alloys A and B above 210K, corrected for the athermal contributions, is shown in Figure 4. Activation volumes are expressed in terms of the atomic volume \( \Delta V \). Data for the two alloys over the temperature range investigated are in reasonable agreement, suggestive of a single rate controlling mechanism. Plots of this form describe the force-distance characteristics of the controlling obstacle (3), in this case indicating a relatively long range "tail" for which the force decreases approximately as the reciprocal of the distance \( (\%l/r) \). Details of the short range (low \( V^* \)) behavior could not be discerned due to the intervention of the anomalous softening effect at low temperatures.

The overall profile shape and large activation volumes of \( 10^3 \) and \( 10^4 \) are similar to those for dislocation-dislocation interactions, but another likely controlling mechanism would be the interaction with discrete particles such as fine precipitates or solute clusters, possibly in a distribution of sizes. Further helpful information on the nature of the short-range interactions may be obtained using a modulus-normalized analysis of the \( \tau-V^*_f \), \( \Delta g-V^* \) data based on the measured elastic constants.

Little information is currently available for direct comparison of interface and slip dislocation mobilities. Measurements of (110)\( _z \) slip kinetics in a similar \( \beta \)-CuZn alloy over the range of 180 to 300K (16) reveal an athermal friction stress component of 18 MN/m\(^2\) which compares well with the interfacial \( \tau_f \) values of 15 to 22 MN/m\(^2\) observed here. The measurements of slip kinetics also indicate a relatively large activation volume \( (>10^4 \text{ near } 300K) \) comparable to the values reported in Figure 4. A lower activation volume of \( (>10^2 \text{ near } 300K) \) for interfacial motion can be inferred from analysis of isothermal nucleation kinetics in austenitic iron alloys, in agreement with slip kinetics for those alloys (1), but these are measured at lower temperatures than the range examined here. Slip kinetic measurements on stable \( \beta \)-CuAlNi alloys are planned to allow a more direct comparison of interfacial and slip-dislocation mobilities.

Conclusions. - Over a temperature range of 180 to 410K and a velocity range of \( 10^{-6} \) to \( 10^{-3} \text{ m/s} \), the mobility of the \( \beta-\gamma \) interface of CuAlNi is consistent with thermally-activated motion, although an anomalous temperature dependence is observed below 210K which correlates with increased elastic softening. Activation energies of 0.3 to 0.4 eV \( (4.5 \times 10^{-20} \text{ to } 6.5 \times 10^{-20} J) \) were determined, with activation volumes of \( 10^3 \) to \( 10^4 \). The behavior is suggestive of rate control by interaction with discrete particles or solute clusters, but further low temperature mobility information is needed to define the nature of the short-range interaction.

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References
13. SMITH, C.S., Trans. AIME 175 (1948) 47.