The Physics of Metals and Metallography, Vol. 81, No. 4, 1996, pp. 374–380.

Comparative Analysis of Stress Jumps in Metals and Intermetallic Compounds: I. Description of Two-Stage Straining

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Received November 2, 1995

Abstract—The Cottrell–Stokes two-stage straining of fcc metals and intermetallic compounds whose yield stress exhibits an anomalous temperature dependence is analyzed theoretically. The magnitude of stress jumps brought about by changes in temperature is expressed via characteristics of dislocation transformations to short- and long-lived barriers. The manner in which the thermally activated blocking of dislocation sources affects the magnitude of stress jumps in intermetallic compounds is investigated. Several alternative combinations of conditions are discussed, under which stress jumps similar to those observed in metals occur in intermetallic compounds with changing temperature.

INTRODUCTION

The well known experiments of Cottrell and Stokes [1], in which stress jumps that arise as a result of changes in temperature or strain rate are measured, make it possible to distinguish between the contributions made to the flow stress by various obstacles inhibiting the movement of dislocations. Above all, these are contributions due to long-range and short-range obstacles. Usually such contributions are supposed to be additive; they are responsible for irreversible and reversible changes in flow stress, respectively, with changes in temperature or strain rate [2].

Recently, the field of research on stress jumps has expanded considerably owing to prestraining experiments on intermetallic compounds. A rather motley picture has emerged. For example, in their experiments with TiAl, Stucke *et al.* [3] found that the change in stress due to a decrease in temperature seemed to be similar, to a certain extent, to what is observed in Cottrell–Stokes experiments on fcc metals. On the other hand, in their study of Ni_3Al, Dimiduk and Parthasarathy [4] noted a decrease in stress, and an unusually large one for that matter.

The anomalous temperature dependence of yield stress observed in the intermetallic compounds considered brings us to several questions: How does this temperature anomaly affect stress jumps upon changes in temperature (strain rate)? What are the conditions under which the changes in stress can be almost the same for intermetallic compounds and metals, despite the above anomaly (part I of this study)? Why is it that the stress jumps occurring in intermetallic compounds are so large (part II)?

We attempt to describe stress jumps in metals and intermetallic compounds within a unified scheme. In doing so, we find it efficient to use the concept we have set forth in [5, 6]. It offers a way to consider various types of dislocation transformations both aided and unaided by thermal fluctuations. Our analysis suggests several experiments that offer an insight into the phenomenon we are concerned with (see part II).

1. EQUATIONS OF PLASTIC DEFORMATION

The dislocation transformation diagram shown in Fig. 1 is the simplest of those that schematically depict the reactions prerequisite for stress jumps to occur. In what follows, it will be referred to as the *C* diagram, as is done in [6]. It includes an elementary diagram of the lobe type, which describes $g \rightleftharpoons s$ transformations, and an elementary diagram of the ray type, which describes $g \longrightarrow s'$ transformations. They demonstrate that glissile (g) dislocations can go over into sessile configurations of the s and s' types. These configurations differ in the ratio between their lifetime and observation time t, equal to $\varepsilon/\dot{\varepsilon}$. More specifically,

$$\begin{aligned} t_s &\ll t, \\ t_{s'} &\gg t, \end{aligned} \tag{1.1}$$

that is, s-type barriers are short-lived, and s'-type barriers are long-lived. In other words, g dislocations continuously go over into s dislocations and back, whereas s'-type dislocations do not transform back to glissile dislocations over the observation time. That is, they are indestructible barriers.

For the C diagram, the equations of detailed balance may be written as

$$\begin{aligned} \dot{N}_g &= -N_g\nu_{gs} + N_s\nu_{sg} - N_g\nu_{gs'} + M, \\ \dot{N}_s &= N_g\nu_{gs} - N_s\nu_{sg}, \qquad (1.2) \\ \dot{N}_{s'} &= N_g\nu_{gs'}. \end{aligned}$$

Here N_g is the density of glissile dislocations; N_s and $N_{s'}$ are the densities of sessile dislocations of the s and s' types; M is the strength of dislocation sources; ν_{gs}, $\nu_{gs'}$, and ν_{sg} are the frequencies of $g \longrightarrow s$, $g \longrightarrow s'$, and $s \longrightarrow g$ transitions, respectively, ν_{sg} being inversely proportional to the lifetime t_s introduced previously.

As will be recalled, N_g is connected to the strain rate $\dot{\varepsilon}$ by a relation of the form

$$\dot{\varepsilon} = bvfN_g, \qquad (1.3)$$

where b is the modulus of the Burgers vector, v is the velocity of a glissile dislocation, and f is the Schmid factor. Solutions to simultaneous equations (1.2) are analyzed in [5, 6], where the applicable approximations are also formulated. Here, we will limit ourselves to the expressions essential for the subsequent analysis of stress jumps.

Because $N_g \ll N_s$ [6], the total dislocation density ρ for the C diagram may be written as

$$\rho = N_s + N_{s'}. \qquad (1.4)$$

Whatever the blocking mechanisms involved, a glissile dislocation must overcome the elastic opposition of its environment that consists of s and s' dislocations. Therefore, a necessary condition for plastic flow to take place is

$$\sigma = \kappa\sqrt{\rho}, \quad \kappa = \alpha\mu b/f. \qquad (1.5)$$

Here, μ is the shear modulus and α is a nondimensional parameter, which depends, among other things, on the dislocation orientation. Later, we will go back to the form in which the relation between σ and ρ can be written.

For convenience, we introduce partial stresses σ_s and $\sigma_{s'}$

$$\sigma_s^2 = \kappa^2 N_s, \quad \sigma_{s'}^2 = \kappa^2 N_{s'}, \qquad (1.6)$$

which describe the transformation of the glissile dislocations into the short-lived (lobe-type diagram) and long-lived (ray-type diagram) barriers. Because N_s and $N_{s'}$ are additive, it follows that

$$\sigma^2 = \sigma_s^2 + \sigma_{s'}^2. \qquad (1.7)$$

Using the solutions to (1.2) for N_s and $N_{s'}$ and relations (1.6), we immediately obtain

$$\sigma_s^2 = \frac{1}{\gamma\lambda_s}\dot{\varepsilon}t_s, \quad \sigma_{s'}^2 = \frac{1}{\gamma}\overline{\left(\frac{1}{\lambda_{s'}}\right)}\varepsilon, \qquad (1.8)$$

where

$$\gamma = \frac{bf}{\kappa^2}, \quad \overline{\left(\frac{1}{\lambda_{s'}}\right)} = \frac{1}{\varepsilon}\int_0^{\varepsilon}\frac{1}{\lambda_{s'}}d\varepsilon. \qquad (1.9)$$

Here, the free path lengths λ_s and $\lambda_{s'}$ are inversely proportional to the previously defined frequencies ν_{gs} and $\nu_{gs'}$, respectively.

Thus, for the C diagram we have

$$\sigma^2 = \frac{\dot{\varepsilon}}{\gamma}\left[\frac{t_s}{\lambda_s} + \overline{\left(\frac{1}{\lambda_{s'}}\right)}\frac{\varepsilon}{\dot{\varepsilon}}\right]. \qquad (1.10)$$

As (1.10) implies, the contributions to σ^2 from short- and long-lived barriers, determined by $1/\lambda_s$ and $\overline{1/\lambda_{s'}}$, respectively, have different weight factors. They are t_s and $\varepsilon/\dot{\varepsilon}$, respectively. It is important that $t_s \ll \varepsilon/\dot{\varepsilon}$, according to (1.1). Therefore, if $1/\lambda_s$ and $\overline{1/\lambda_{s'}}$ are comparable in value or even $\overline{1/\lambda_{s'}} < 1/\lambda_s$, then the dominant contribution to σ should come from the $g \longrightarrow s'$ transformation to long-lived barriers. On the contrary, the $g \rightleftharpoons s$ transformation to short-lived barriers can be dominant only if the corresponding free path is considerably shorter than it is for the $g \longrightarrow s'$ transformation. In other words, the ray diagram easily wins over the lobe diagram.

In view of the above statement, we assume that

$$\sigma_s \ll \sigma_{s'}. \qquad (1.11)$$

Then, equation (1.7) takes the form

$$\sigma = \sigma_{s'} + \Delta\sigma, \quad \Delta\sigma = \frac{\sigma_s}{2}\frac{\sigma_s}{\sigma_{s'}}. \qquad (1.12)$$

Using (1.8), we write $\Delta\sigma$ as

$$\Delta\sigma = \frac{1}{2\gamma\lambda_s}\frac{\dot{\varepsilon}t_s}{\sigma_{s'}}. \qquad (1.13)$$

In form, (1.13) is similar to the expression known from [7]

$$\sigma = \sigma_G + \sigma_S. \qquad (1.14)$$

Here, σ_G and σ_S determine the contributions from long-range and short-range obstacles, respectively (see Introduction). The main difference between (1.12) and (1.14) is that in (1.12) the elastic opposition of dislocations is described by the entire expression, whereas in (1.14) this is done solely by the term σ_G. According to (1.7), the contributions σ_s and $\sigma_{s'}$ form a quadratic sum, and it is only in the approximation defined in (1.11) that

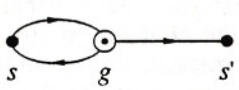

Fig. 1. C diagram of dislocation transformations.

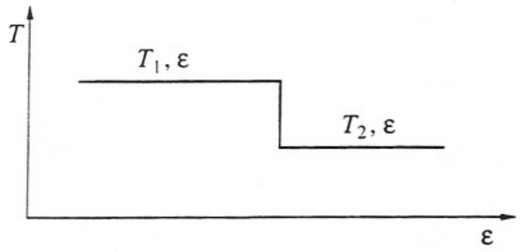

Fig. 2. Schematic representation of two-stage straining.

they are additive, whereas (1.14) does not use such an approximation at all. Moreover, in (1.14) only σ_S is associated with thermally activated processes. In (1.12), both terms are involved because $g \longrightarrow s'$ transformations can be thermally activated, for example, in intermetallic compounds. Lastly, the contributions from long-range and short-range obstacles in (1.14) are separated, whereas $\Delta\sigma$ in (1.12) contains both σ_s and $\sigma_{s'}$.

2. STRESS JUMPS DUE TO CHANGES IN TEMPERATURE OR STRAIN RATE IN METALS

The Cottrell–Stokes experiment is usually carried out as follows. At first, the specimen is deformed at temperature T_1 to a certain predetermined amount of strain ε (a few percent). Then, the temperature is changed and the specimen is deformed at T_2 likewise to ε. We will limit ourselves to the case (Fig. 2) where $T_1 > T_2$ so as to avoid considering the recovery of the dislocation structure. Using the expressions derived above, let us calculate the stress jumps occurring upon a decrease in temperature at stage II of work hardening in fcc metals.

2.1. Description of Stage II of Work Hardening in fcc Metals

Suppose that two types of dislocation transformation take place at stage II. One happens because Lomer–Cottrell barriers arrest dislocations. The other occurs when dislocations intersect the "forest". Assuming that the Lomer–Cottrell barriers are indestructible, we will describe the arrest of dislocations as $g \longrightarrow s'$ transformations, for which the corresponding stress $\sigma_{s'}$ is given by (1.8). In describing the intersections of the forest by dislocations as $g \rightleftharpoons s$ transformations, we will assume that at stage II the characteristic time t_s satisfies relation (1.1) and that the applicable stress σ_s is given by (1.8). Of the two types of transformations we are concerned with only the $s \longrightarrow g$ transformation, which represents the intersection of a forest by dislocations and is a thermally activated process. Its t_s may be written as

$$t_s = \tilde{t}_s \exp\left(\frac{U_{sg}}{kT}\right), \tag{2.1}$$

where U_{sg} is the activation energy for an $s \longrightarrow g$ transformation.

Because a great number of Lomer–Cottrell barriers are formed at stage II, the term $\overline{1/\lambda_{s'}}$ is apparently smaller than $1/\lambda_s$, albeit only slightly. Then, as follows from (1.10), the $g \longrightarrow s'$ transformation is a dominant one, and the stress σ is defined by (1.12). Note that, because (1.11) is satisfied, $\Delta\sigma$ is such that

$$\Delta\sigma \ll \sigma_{s'}. \tag{2.2}$$

Similar to [7], we take it that $\lambda_{s'}$ is inversely proportional to the strain ε as reckoned from the onset of stage II, that is, $\lambda_{s'} = \Lambda_{s'}/\varepsilon$. Then, using (1.9), we immediately obtain

$$\sigma_{s'} = \theta\varepsilon, \quad \theta = \frac{\alpha\mu}{f}\sqrt{\frac{b}{2f\Lambda_{s'}}}, \tag{2.3}$$

where θ is the work-hardening rate.[1]

2.2. Analysis of Stress Jumps

In two-stage straining, the stresses σ_1 and σ_2 at the first and second stage are defined by expression (1.12) taken at the appropriate temperatures, with $\sigma_{s'}$ defined by (2.3) and $\Delta\sigma$, by (1.13). In keeping with (2.2), the increments $\Delta\sigma_1$ and $\Delta\sigma_2$ have almost no effect on the flow stress at a fixed temperature. Otherwise, the strain-hardening factor at stage II would vary with temperature. Upon a change in temperature, however, the stress jump

$$\Delta\sigma^{(T)} = \sigma_2 - \sigma_1 \tag{2.4}$$

is determined solely by the difference of small increments

$$\Delta\sigma^{(T)} = \Delta\sigma_2 - \Delta\sigma_1. \tag{2.5}$$

This is because the contribution $\sigma_{s'}$ from the dominant $g \longrightarrow s'$ transformation, equal to $\theta\varepsilon$, is independent of temperature, unless one takes into consideration the temperature dependence of the shear modulus. However, the stress jump "memorizes" the dominant transformation. Indeed, as can be seen from (2.5) and (1.12), the $\Delta\sigma^{(T)}$ is not the difference between the stresses σ_s at different steps of deformation. It determines the corresponding increment $\Delta\sigma$ only if it is multiplied by the ratio $\sigma_s/2\sigma_{s'}$. Thus, $g \rightleftharpoons s$ transformations, which do not reveal themselves at stage II, that is, remain "hidden", show up when a change in temperature takes place. The difference between the associated increments, $\Delta\sigma_2 - \Delta\sigma_1$, is not infinitesimal when a strong

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[1] To allow for a nonuniform distribution of dislocations, that is, for the fact that a number of dislocations, say n of them, are piled up at a barrier, it will suffice to multiply κ in (1.5) by $\sqrt{n}$. Accordingly, we multiply by $\sqrt{n}$ the right-hand side of expression (2.3) for the work hardening rate, which thus becomes identical (to within a certain numerical coefficient) to the expression known from [7].

change in temperature takes place and governs the magnitude of $\Delta\sigma^{(T)}$.

Using (1.13), we may re-write (2.5) as

$$\Delta\sigma^{(T)} = \Delta\sigma_1\left(\frac{t_s(T_2, \sigma_2)}{t_s(T_1, \sigma_1)} - 1\right). \qquad (2.6)$$

Hence, it immediately follows that $\Delta\sigma^{(T)} > 0$ at $T_2 < T_1$ because, according to (2.1), the lifetime t_s increases with decreasing temperature.

In a similar way, we find the stress increment $\Delta\sigma^{(\dot{\varepsilon})}$ caused by an increase in the strain rate $\dot{\varepsilon}$

$$\Delta\sigma^{(\dot{\varepsilon})} = \Delta\sigma_1\left(\frac{\dot{\varepsilon}_2 t_s(\sigma_2)}{\dot{\varepsilon}_1 t_s(\sigma_1)} - 1\right). \qquad (2.7)$$

Because $\sigma_2 = \sigma_1 + \Delta\sigma^{(T)}$, expression (2.6) is an equation from which one can find $\Delta\sigma^{(T)}$ if one knows the behavior of λ_s and U_{sg} with changing σ. For example, if we assume that λ_s is inversely proportional to ε^2 and the activation energy U_{sg} only slightly varies with σ, we obtain $\Delta\sigma^{(T)}/\sigma \cong \text{const}$, which is the well-known Cottrell–Stokes relation [1, 2]. To demonstrate, it immediately follows from (1.13) and (2.3) that, under the assumptions made, $\Delta\sigma/\sigma_{s'}$ is independent of ε. Therefore, noting that in the case at hand t_s defined by (2.3) is independent of σ and $\sigma_1 \cong \sigma_{s'}(T_1)$, we immediately obtain the Cottrell–Stokes relation from (2.6).

It is likewise possible that the Cottrell–Stokes relation will be satisfied under a different set of conditions. We will not analyze solutions to equation (2.6). Instead, we will only attempt to define the role that dislocation transformations to long-lived and short-lived barriers play in plastic deformation under both fixed and varying external parameters.

3. TWO-STAGE STRAINING OF INTERMETALLIC COMPOUNDS

3.1. Experimental Findings

Experiments similar to Cottrell and Stokes's were made on intermetallic compounds displaying a temperature anomaly of flow stress. In their studies of the straining behavior of Ni_3Al, Davies and Stoloff [8] and Copley and Kear [9] found that at $T_1 > T_2$ the stress decreased rather than increased as in the case above; that is, $\Delta\sigma^{(T)} < 0$.

The increments $\Delta\sigma^{(\dot{\varepsilon})}$ caused by changes in the strain rate look more usual. Above all, as Thornton *et al.* [10] have found in Ni_3Al and Takeuchi and Kuramoto [11] in Ni_3Ga, an increase in the strain rate is accompanied, similar to metals, by a positive change in stress $\Delta\sigma^{(\dot{\varepsilon})}$, but the stress change is small compared to the stress itself. Moreover, $\Delta\sigma^{(\dot{\varepsilon})}$ in intermetallic compounds was found to be significantly smaller than in metals.

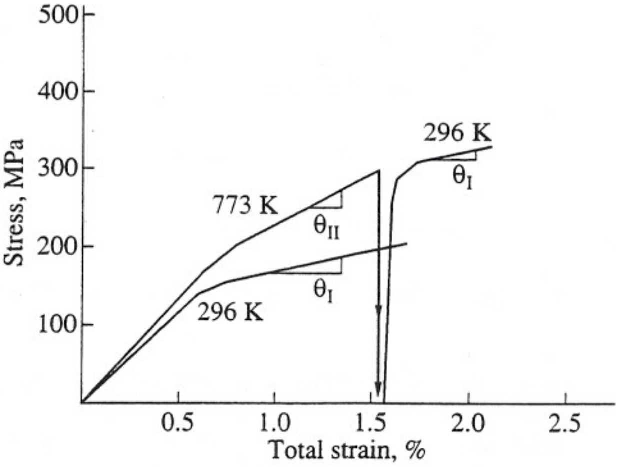

Fig. 3. Stress–strain curves for TiAl single crystals. Lower curve, compression at room temperature; upper curves, two-stage straining [3].

In the analysis that follows we will also use the results reported by Thornton *et al.* [10] for $Ni_3(Al,Cr)$, where the specimen was prestrained (to about 20% at room temperature), and the yield stress was then measured within the anomalous behavior region. They observed that up to about 600°C the specimen retained the high yield stress imparted by prestrain.

Stucke *et al.* [3], and Dimiduk and Parthasarathy [4] experimented with TiAl and Ni_3Al, respectively, using a scheme (see Fig. 2) similar to the one used in Cottrell–Stokes experiments. These intermetallic compounds were found to respond differently to two-stage straining, although both exhibited an anomalous behavior of their yield stress in a broad temperature range. Stress–strain curves for TiAl and Ni_3Al are given in Figs. 3 and 4, respectively. In both cases, T_1, which corresponds to the first straining stage, falls within the interval where $\sigma_y(T)$ shows an anomalous behavior, and the second stage proceeds at room temperature. Also, Figs. 3 and 4 show stress–strain curves taken at room temperature without prestrain.

Referring to Fig. 3, it can be seen that in the case of TiAl changes in temperature leave the stress almost unchanged. According to Stucke *et al.* [3], the change in stress, $\Delta\sigma^{(T)}$, is about 10 MPa. Hence, the observed behavior of TiAl looks more like that of metals in the conventional Cottrell–Stokes experiments, as noted previously.

In the case of Ni_3Al, however, as can be seen from Fig. 4, a sharp decrease in stress is observed on passing to the second stage. If we take as σ_2 the stress at which the hardening factor settles at a constant value, then $\sigma_1/\sigma_2 \approx 3.5$. The strain-hardening factor at the second-stage temperature T_2 likewise responds to prestrain differently in the intermetallic compounds involved. Comparison of the curves in Figs. 3 and 4 shows that prestrain leaves this factor almost unchanged in TiAl, but increases it significantly in Ni_3Al.

3.2. Blocking of Dislocation Sources

Intermetallic compounds owe their anomalous stress–strain behavior with temperature first of all to the fact that thermal activation is responsible not only for the transformations of blocked dislocations into glissile dislocations, which is usual, but also for the dislocation blocking processes itself. In other words, thermal activation causes dislocations to transform to barriers, which may be regarded as indestructible in a certain temperature interval. In our opinion, dislocations are blocked as soon as they multiply, and this occurrence determines the stress required to activate dislocation sources [12].

Among all the possible types, let us choose dislocation sources that have a critical configuration. Such a configuration must be achieved before the subsequent loss of stability can lead to the multiplication of dislocations. Let there be, as an example, an ordinary Frank–Read dislocation source, in which the spacing between the points of pinning is $2l$ and which emits a superdislocation whose Burgers vector is $2b$. Unless it is blocked, the superdislocation will acquire a critical shape (in the simplest case, this will be a semicircle) at a stress σf equal to $\mu b/l$.

Suppose that some thermally activated $g \longrightarrow s'_F$ transformation, with a characteristic free path λ_F, results in dislocation barriers that may be taken as indestructible at a given temperature. Then, if $l > \lambda_F$, a dislocation segment will be too long (compared to λ_F) to achieve a critical shape; instead, it will acquire a blocked configuration. If, on the other hand, $l < \lambda_F$, the segment will have enough time to pass through a critical configuration before it is blocked.

The values of l at which superdislocations can multiply and at which dislocations are blocked as they bow out are separated by a smeared boundary. However, the boundary becomes smeared near λ_F, so it is safe to adopt λ_F as the critical size of a dislocation segment.

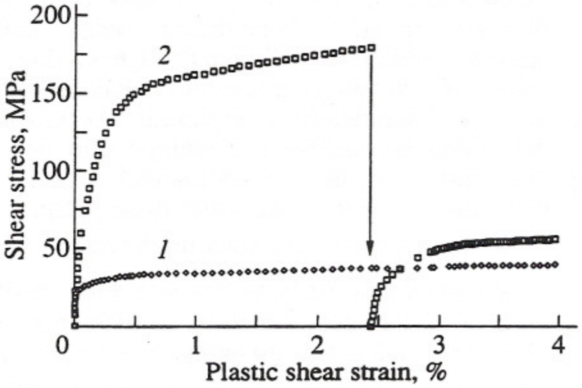

Fig. 4. Resolved shear stress plotted against shear strain in Ni_3Al: (*1*) at 300 K; (*2*) at 875 K (at the first stage) and 300 K (at the second stage) [4].

Then, the stress required to activate Frank–Read sources may be written as

$$\sigma_F(T)f = \frac{\mu b}{\lambda_F(T)}. \qquad (3.1)$$

In writing (3.1), no other dislocation blocking mechanisms are considered except the thermal activation. As (3.1) implies, the dislocation blocking significantly changes the situation. Indeed, in the absence of blocking, an increase in their length makes it increasingly more convenient for dislocation segments to serve as sources. In the presence of blocking, the lengths of these same segments fall in the forbidden interval $l > \lambda_F$, where sources just cannot be switched on. Moreover, if the barrier remains indestructible, no increase in stress can make the source for which $l > \lambda_F$ operative. Thus, whether or not it is possible to activate a source depends on the ratio between the values of the source length and the free path related to the transformation of a dislocation to an indestructible barrier.

For a thermally activated $g \longrightarrow s'_F$ transformation, $\lambda_F(T)$ may be written as

$$\lambda_F(T) = \tilde{\lambda}_F \exp\left(\frac{U^F_{gs'}}{kT}\right), \qquad (3.2)$$

where $U^F_{gs'}$ is the activation energy for blocking a dislocation that belongs to the source. Using (3.1) and (3.2), we immediately obtain an expression that makes allowance for the anomalous temperature behavior of the stress required to activate dislocation sources

$$\sigma_F(T) = \tilde{\sigma}_F \exp\left(-\frac{U^F_{gs'}}{kT}\right). \qquad (3.3)$$

However, the switching-on of dislocation sources alone is not sufficient to initiate plastic flow. Its onset may be visualized as follows. Dislocations do not begin to multiply uniformly throughout the crystal; rather, they do so separately, most favorable regions initially isolated from one another. Later, the regions join together and take up a sizable proportion of the space, similar to what happens in the case of what is known as percolation. By extending the analogy, this instant may be thought of as marking the start of plastic flow. For this to happen, however, it is necessary that the increasing dislocation density and the external stress should be connected by a relation of the form (1.5). In that case, glissile dislocations will be able to overcome the elastic opposition of their environment. To sum up, the condition for plastic flow to take place may be written as

$$\sigma \approx \max\{\sigma_F, \kappa\sqrt{\rho}\}. \qquad (3.4)$$

For brevity, this condition may be termed a "double key"; it permits the plastic flow to start as a macroprocess.

It is not at once that the relation defined in (1.5) takes effect after dislocation sources become operative. A time interval is needed for the growing dislocation density and the operative stress to strike a balance. This interval determines the transition region between elastic and plastic deformation in the stress–strain curve.

In a way, there is an alternative to the above view. Frequently, the operative stress is written as

$$\sigma = \sigma_0 + \kappa\sqrt{\rho}, \tag{3.5}$$

where σ_0 is a certain starting stress, say, σ_F. In what follows, we will compare the possibility of using either the "double key" (3.4) or relation (3.5) in interpreting the experimental findings.

3.3. Two-Stage Straining without a Change of the Slip System

Consider two-stage straining when the temperature T_1 lies within the region where the yield stress exhibits an anomalous behavior. We will use the dislocation transformation diagram shown in Fig. 1. Here, the $g \rightleftharpoons s$ transformations are intersections of a forest, as in Section 2, but the $g \longrightarrow s'$ transformations are the thermally activated transformations, typical of intermetallic compounds, that convert dislocations to barriers indestructible at the temperatures under study. The stress σ_s associated with a forest intersection is significantly lower than the stress $\sigma_{s'}$ associated with the formation of barriers. That is, relation (1.11) is satisfied, and the resultant stress can be found from (1.12) and (1.13).

In finding $\sigma_{s'}$, we assume that the initial condition is $\sigma_0 = \sigma_F$. If we limit ourselves to a small amount of strain, then, in view of (1.8), we will get

$$\sigma_{s'}(T) = \sigma_F(T) + \theta(T)\varepsilon, \qquad \theta(T) = \frac{1}{2\gamma\lambda_{s'}(T)\sigma_F(T)}. \tag{3.6}$$

Let us write $\lambda_{s'}$ as

$$\lambda_{s'}(T) = \tilde{\lambda}_{s'}\exp\left(\frac{U_{gs'}}{kT}\right), \tag{3.7}$$

where $U_{gs'}$ is the activation energy for a $g \longrightarrow s'$ transformation. In contrast to the activation energy $U^F_{gs'}$ defined above, we are concerned here with the blocking of dislocations no longer associated with a source. Therefore, the temperature dependence of the work hardening rate takes the form

$$\theta(T) = \tilde{\theta}\exp\left(-\frac{(U_{gs'} - U^F_{gs'})}{kT}\right). \tag{3.8}$$

As follows from (3.8), for the function $\theta(T)$ to show an anomalous behavior with temperature, it is necessary that $U^F_{gs'} < U_{gs'}$.

Thus, the stress $\sigma_1(T_1)$ achieved toward the end of the first straining stage may be written as

$$\sigma_1(T_1) = \sigma_{s'}(T_1) + \Delta\sigma_1(T_1), \qquad \Delta\sigma_1(T_1) = \frac{1}{2\gamma\lambda_s}\frac{\dot{\varepsilon}t_s(T_1)}{\sigma_{s'}(T_1)}. \tag{3.9}$$

Here, $\sigma_{s'}(T_1)$ and $t_s(T_1)$ are defined by (3.3), (3.6), (3.8), and (2.1), respectively.

As it cools to T_2, which is the temperature of the second straining stage, the specimen retains the microstructure produced at the first stage and having a total dislocation density ρ_1. For this reason, if the slip system operative at the second stage is the same as before, the microstructure will present to the dislocations of this system an opposition $\kappa\sqrt{\rho_1}$, equal to σ_1. Hence, if plastic flow at the second stage begins at a stress $\sigma_2(T_2)$, equation (3.4) will take the form

$$\sigma_2(T_2) \approx \max\{\sigma_F(T_2), \sigma_1(T_1)\}. \tag{3.10}$$

In compliance with (3.3), we have $\sigma_F(T_2) < \sigma_F(T_1)$. Therefore, noting (1.12) and (3.6), we have $\sigma_F(T_2) < \sigma_1(T_1)$. This immediately implies that

$$\sigma_2(T_2) \approx \sigma_1(T_1). \tag{3.11}$$

On the contrary, if equation (3.5) applies, writing it consecutively for the end of the first stage and for the start of the second immediately reveals that a decrease in temperature is accompanied by a decrease in stress by an amount approximately equal to the difference between the stresses required to make a slip system go operative, that is,

$$\sigma_1(T_1) - \sigma_2(T_2) \cong \sigma_F(T_1) - \sigma_F(T_2). \tag{3.12}$$

As the curves of Fig. 3 show, equation (3.11), rather than equation (3.12), holds for TiAl. In a way, this validates the concept of the "double key".

The difference between the stresses σ_1 and σ_2 marking the end of the first and the start of the second stage of straining in an intermetallic compound without a change of the slip system is governed solely by $g \rightleftharpoons s$ transformations. Therefore, we may write $\Delta\sigma^{(T)}$ in a form similar to (2.6). In contrast to metals, however, the stress $\sigma_{s'}$, which, according to (3.9), has a direct bearing on $\Delta\sigma_1(T_1)$, shows an anomalous behavior with temperature in intermetallic compounds. As (2.6) implies, $\Delta\sigma^{(T)}$ becomes positive as the temperature decreases. It, thus, turns out that the anomalous behavior of $\sigma(T)$ does not reverse the sign of the stress jump.

If the strain rate varies within the region where $\sigma(T)$ shows an anomalous behavior with temperature, $\Delta\sigma^{(\dot{\varepsilon})}$ will, as in the case of metals, be defined by equation (2.7) [in which $\Delta\sigma_1$ is given by equation (1.13), which contains the stress $\sigma_{s'}$]. Owing to this, in changing $\dot{\varepsilon}$, the memory of the previous $g \longrightarrow s'$ transformations is retained, although $\Delta\sigma^{(\dot{\varepsilon})}$ is governed, similar to $\Delta\sigma^{(T)}$, by $g \rightleftharpoons s$ transformations, which remain "hidden" as long as the strain rate is held constant. As in metals, $\Delta\sigma^{(\dot{\varepsilon})}$ is positive upon an increase in the strain rate. Equation (1.13) implies that the anomalous behavior of $\sigma_{y}(T)$ will produce a definite effect on the temperature dependence of $\Delta\sigma^{(\dot{\varepsilon})}$.

In intermetallic compounds whose yield stress exhibits an anomalous behavior with temperature, large stress jumps (macrojumps) $\Delta\sigma^{(T)}$ comparable with the stress itself may occur. Stress macrojumps in intermetallic compounds are discussed in part II of this study.

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