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P. Guyot

April, 1966
ON THE MECHANISMS OF DEFORMATION IN DISPERSION HARDENED ALLOYS

P. Guyot*

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April, 1966

ABSTRACT

The mechanisms of plastic deformation occurring in alloys with a f.c.c. matrix and in which the incoherent second phase dispersed particles are generally not deformed with the matrix, are described. At low temperatures these mechanisms are well understood. The cutting of the mobile dislocations through the forest agrees with the experimental results; the hardening is due to the steep increase of dislocation density with the strain. The second phase plays an indirect part (pinning or multiplication of dislocations). At a high enough stress (or strain) cross slipping can be induced, determining a cell structure.

At high temperatures various processes can be effective: cross slipping and particle limited climb determine the motion of dislocations in the matrix; for temperatures near the melting point of the matrix, these alloys generally present very high activation energies steeply varying with the stress and the thermal activation of the junction reactions between attractive dislocations explains quite well the experimental measurements.

* The author was Postdoctoral Research Metallurgist at the Euratom, C.C.R. Ispra (Italy).
I. INTRODUCTION

The strengthening of materials by a dispersed second phase has been the subject of considerable investigation in recent years. The results of experimental investigation and theoretical approaches show that the hardening of these alloys depends principally upon whether the discrete second phase particles are sheared by the dislocations moving in the matrix and this is particularly well reflected in the tensile curves [1].

The degree of coherency of the matrix-precipitate interface is one controlling condition for the shearing, such, for instance, as is obtained in materials containing Guinier-Preston zones.

The incoherent precipitates, on the contrary, are usually not sheared by the dislocations moving in the matrix (except possibly at high deformations and for small size of precipitates). When a glide dislocation encounters such precipitates intersecting its slip plane, it may expand between them, leaving residual loops around each precipitate (Orowan's mechanism [2]), or alternatively by-pass them by some more complicated cross slip process [3]; for temperatures at which the diffusion occurs, the climb of dislocations may also control the deformation [4, 5, 7].

Oxide dispersions are generally incoherent, and so we can include them in the general group of alloys with incoherent precipitates. Properties of the oxide dispersions are (1) their stability due to the high free energy of formation of the oxide, which assures the stability of the alloy's dislocation structure at high temperatures; and (2) the large value of the shear modulus of oxide, so that only very small size
particles can be sheared (an approximate computation for \( \text{Al}/\text{Al}_2\text{O}_3 \) alloy shows that the shearing of dislocation free particles can occur if their radius is smaller than about 15Å).

While the initial flow stress of this group of alloys has been considerably studied and related to the structure parameters characterizing the dispersed phase [2, 7, 8], the deformation kinetics have never been extensively considered, at least so far as the deformation equation is concerned. We also find in the literature numerous results concerning the substructures observed by electron microscopy [3, 9, 10, 11]; but these observations have not been conducted in parallel with the determination of the parameters characterizing the mechanisms of plastic deformation.

It is the purpose of this paper to give a short review of the results obtained by the author on the polycrystalline \( \text{Al}_2\text{O}_3/\text{Al} \) system, from the point of view both of deformation kinetics and of electron microscopy. The results are coupled with those obtained by Mitchell, Mitra and Dorn [12] on the \( \text{Al}/\text{Al}_2\text{Cu} \) system. The good agreement between these alloys shows that the flow mechanisms involved seem to be general to alloys with incoherent precipitates in a f.c.c. matrix of high stacking fault energy. An electron microscopy study of TD-Nickel by Heimendahl and Thomas [13] confirms the conclusion.

At low temperatures the dislocations in the matrix move by glide (intersection mechanism, cross slip), while at high temperatures self diffusion (climb of dislocations) controls the deformation, together with some more complicated processes of which we briefly give theoretical approaches.
II. MATERIALS

The Al$_2$O$_3$/Al alloys were obtained by sintering commercial aluminum powders and consequently will be designated SAP (sintered aluminum powder). The weight fractions in alumina were next to 4, 7 and 10 percent.

The Al$_2$Cu/Al alloys were obtained by precipitation in an Al solid solution matrix.

The TD-Ni alloys were obtained by precipitation in an Al solid solution matrix.

The TD-Ni alloy was a system of approximately 2 vol pct ThO$_2$ dispersed in a matrix of commercially pure nickel, and was also fabricated by powder-metallurgical techniques.

The composition and the structural parameters, as determined by electron microscopy, which characterize the alloys here described are given in Table 1. While the Al$_2$Cu and ThO$_2$ particles are roughly spherical in shape, the alumina particles in SAP are platelets with linear size between 200 and 2000Å, and a thickness of about 200Å; the values of $r$ given in Table 1 for the mean planar particle radius, are in this case the radius of the circle of the same area as that of the mean intersection area of platelets by a random plane, i.e. $r = (2Le/\pi)^{1/2}$, where $L$ is half of the mean linear size and $e$ the thickness [5].

III. DEFORMATION MECHANISMS

It is convenient to study the deformation processes by the systematic determination of the activation energy and activation volume which usually characterize a thermally activated process. The mechanisms as
a function of the temperature of deformation are further described. But it will be shown that for a given temperature the level of the applied stress (or the strain) can induce radically different mechanisms.

3.1 Deformation mechanisms at low temperatures (ABC in Fig. 1)

3.1.1 Forest process

In this temperature range, and under small stresses for creep (or at low strain for tension at constant strain rate), the mobile dislocations cannot leave the glide plane; their glide, in pure metals, can be hindered both by long range internal stresses and by short range obstacles provided by the "trees", dislocations piercing their glide plane. The deformation is then controlled by the thermally activated jog formation during the cutting through those trees and by the movement in long range stress field.

If we now consider the tensile curves for both Al/Al$_2$O$_3$ and Al/CuAl$_2$ systems, we note, especially for small strains, a strain hardening rate which is much higher than that for polycrystalline aluminum [5, 6, 12] largely depending on the temperature.

The variations of the flow stress for a given strain rate, with temperature (for three strains increasing from 1 to 3), are shown in Fig. 1.

The curves of type (1) or (2) are similar to those of pure metals whose hardening is controlled by the forest process as mentioned before. This process has been extensively considered by different authors [14, 15]: the AB part on Fig. 1 corresponds to the thermally activated jog formation, whereas in the BC part, the forest opposes the development
of loops only by long range internal stresses $\tau_i$. In f.c.c. structures, the long range stresses are principally due to the attractive trees giving junction reactions with the mobile dislocations [16].

An estimate of the energy of formation of a pair of jogs, $2U_{fc}$, can be deduced from the measurement of $T^*$, according to Friedel's theory [15] as described in reference [5]. The results so obtained are gathered in Table 2, for Al/Al$_2$O$_3$, Al/Al$_2$Cu alloys, and for two purities of Al. Applying the forest theory to the thermal decrease of $\tau$ at low temperatures, we thus obtain for both alloys and aluminum a common value of $2U_{fc}$ namely $2U_{fc} = 0.37 \pm 0.05$ e.v. A better estimate of $2U_{fc}$ is given by a direct measurement of the activation energy beyond $T_c$ [15]; the value so obtained, for SAP 11.7 is $= 0.35$ e.v. [5, 6], in good agreement with the former mean value, and of a reasonable order of magnitude for the energy of formation of a pair of jogs in aluminum. Thus, the activation energy is compatible with the forest process controlling the plastic deformation in these alloys as in pure metals.

The measurement of the activation volume from strain rate change tensile tests, allows the determination of the force-distance diagram, showing the elastic interaction between a repulsive tree (not giving a junction reaction) and a mobile dislocation. This has been done by Dorn et al. for Al/Al$_2$Cu and pure aluminum [12] and by the author for SAP and impure aluminum [5, 6]. The results are shown in Fig. 2. The fact that these F-x curves differ but little for the SAP, Al-Cu, Al,

$^*$ $T_\text{c}$ is next to 173°C for SAP, for a strain rate of $2.7 \times 10^{-4}$ s$^{-1}$ and at small strains ($<2.10^{-3}$).
gives an added evidence for the presence of the intersection mechanism for all materials.

3.1.2 Cross slip and self diffusion

For large strains (curve (3) in Fig. 1), the shape of τ/μ - T curve can no longer be explained solely with the mechanism of section 3.1.1. The absence of a plateau in the τ/μ - T curve seems to be due to a strongly thermally activated mechanism. In fact, measuring the activation energy, ΔH, at constant strain rate as a function of temperature, we establish that the change in shape of the τ/μ - T curves is reflected by the ΔH - T curves, Fig. 3 for SAP 11.7 [5, 6]. At small strains, ΔH flattens out at 2Ufc for T > Tc, suggesting a forest process; but for large strains, ΔH increases beyond Tc and reaches a plateau, at about 1.6 e.v. for high enough temperatures, clearly corresponding solely to the operation of a diffusion process. Below, the variation of ΔH with T probably corresponds to ill defined cross slip processes complicated by the effect of internal stresses. These phenomena should determine the relaxation of the internal stresses leading to the thermal decrease of τ. The observation by electron microscopy of plastically deformed specimens shows an homogeneous distribution of dislocations when the deformation is low, while after high deformations the dislocations are tangled and have a cell structure [11, 13, 17] as shown in Figs. 4, 5 and 6. Cells in SAP were only observed for high deformations and beyond the plateau τ(T).

As we previously saw the cross slip of screw dislocations should lead to a decrease in τ with increasing temperature and for high strains,
we can reasonably conclude that our observations agree with the idea that the cell formation is related to:

(a) the cross slip of screw dislocations
(b) the appearance of stage III of the stress strain curve
(c) the tensile stress decrease with temperature after the plateau.

Thomas et al. [28] suggested that the tangles and cells result from the multiplication of dislocations by operation of Frank-Read sources left by cross-slip of dislocations adjacent to particles. This could explain for instance why the cells observed in TD-Ni appear to have corners located at the largest particles [28]. The Thomas mechanism of formation of cells is slightly different from the above—which is in fact the normal cell formation by cross-slip in pure metals—because requiring the further operation of sources resulting from cross-slip. Until more evidence is obtained, it can not be considered as firmly established. As will be shown in the next paragraph, the role of particles as multiplication centers is evident but not well defined.

3.1.3 Densities of Dislocations

The measurement of the athermal stress $\tau_1$ (part BC on Fig. 1) directly leads to an estimate of the density of dislocations with $\rho = \left( \frac{\mu b}{\beta^*_{\mathbf{1}}} \right)^2$ after Saada's relation [16], where $\beta = \frac{1}{4}$ in f.c.c. structures. On the other hand, when the forest process is controlling the deformation, the measurement of the activation volume $v^*$ gives an other possibility of determination of $\rho$, namely $\rho = (1.5 b^2/v^*)$ [5].
Obviously, a direct determination of \( \rho \) is feasible by transmission electron microscopy. The variations of \( \rho \) with the strain, as determined from the activation volume and electron microscopy, are given in Table 3, for SAP 4.5 deformed at room temperature.

The sources of error have been extensively considered in [5] and [11]. From Table 3, for strains smaller than about 2\%, we see that the agreement between both methods is quite good. For high strains, when cross slip becomes operative and introduces a cell structure, the density as determined from the activation volumes, is on the whole close to the density as determined by electron microscopy in the cell walls (the density in the walls is between 2 and 4 times higher than the mean density). Thus, it seems that the activation volume is characteristic of a forest process for which the mobile dislocations have to cut through the cell walls. This result is in good agreement with the assumption used by Ashby [17] to derive an expression for the work hardening.

The absolute values of the dislocation densities are high, such as are only achieved in pure metals after severe cold working without subsequent annealing. The high initial value of this density in SAP, prior to tensile strain is probably due to the fabrication process (sintering and extrusion).

From the preceding study we can conclude that the hardening of these alloys has the same origin as in pure f.c.c. metals. Their larger strain hardening is due to a more rapid increase of the dislocation density with plastic strain. This has been shown by Dorn et al. for Al/Al\(_2\)Cu,
by the author for SAP [5, 6], and Heimendahl and Thomas for TD-Ni [12].

For Al/Al$_2$Cu, $(dT/dE)_T$ decreases with increasing particle spacing $2R_S$, for a given volume fraction $f$; for SAP the work hardening rate increases with increasing $f$ or decreasing $2R_S$. Although no quantitative relation has been proposed, these results show that the particles are thus playing an indirect role of multiplying the dislocation density. Nevertheless the exact mode of multiplying is not well understood. Although the direct generation of loops from particles by the volume indentation effect seems improbable, the particles can be active by creating prismatic loops when dislocations by-pass them by cross slip (Ashby [17]). The particles can also act as barriers to moving dislocations nucleated at other sources, such as Frank-Read sources, grain boundaries and twin boundaries in TD-Ni. This explains why in both cases the cell size seems to be governed by the particle spacing.

An interesting feature appears when studying the grain size influence. As shown by Heimendahl and Thomas for TD-Ni [13], and the author for SAP [5], the grain size has no influence on the flow stress at 0.1 or 0.2 pct tensile stress as well as on the ultimate tensile stress. Transmission electron microscopies taken by Ruedl [18] on tensile strained recrystallized SAP show that for the same plastic strain, the dislocation density is about the same for as-extruded and recrystallized specimens. Thus in recrystallized specimens, the dislocation density, initially low, increases very rapidly with deformation, reaching after a few tenths pct of strain the same dislocation density, and therefore the same flow stress, as
in the as-extruded material.*

Furthermore, the Fisher-Hart-Pry theory [19] of strain hardening, based on the result of planar dislocation loop formation about the dispersed particles is invalid, since as we saw, the role of these is to multiply the dislocations and enhance the cross slip processes leading to the cell structures.

3.2 Deformation mechanisms at high temperatures (above C** in Fig. 1)

The athermal stress $\tau_1$ in Fig. 1 (BC part) is relieved in SAP at about 100°C for a strain of 0.2 pct. This thermal decrease of the stress is due in pure metals to cross slip and climb mechanisms. At least in aluminum, the activation energies for both mechanisms are quite well known (1.2 e.v. for cross slip, 1.5 e.v. for climb [20], close to the self-diffusion energy).

The activation energies at constant strain rates $\dot{\varepsilon}$ and the activation volume, as determined from tensile tests for different grades of SAP and impure aluminum (99.5%) are presented as a function of temperature in Figs. 7 and 8 [4, 5]. These plots show that in SAP and polycrystalline Al the activation energy and activation volume have a similar temperature

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*This point reinforces the validity of my model of the forest process; the flow stress at strains $\geq 0.1 - 0.2$ pct is effectively determined by the dislocation density, and the objection of Ansell in the present meeting is questionable. The situation for the initial yield stress (or flow stress at very small strain) could perhaps be different in recrystallized material because of the initially low density of dislocations; the particles could then control the yield stress. But this is not analyzed here.

**$\tau(C)$ is next to 373°K for SAP for a strain rate of $2.7 \times 10^{-4}$ s$^{-1}$ and a strain of $2.10^{-3}$. 
dependence as long as \( T \) does not exceed about 350°C. Dorn et al. [24] also found for creep at high temperatures of Al/Al\(_2\)Cu an activation energy close to that of self diffusion in pure aluminum. Beyond 350°C, \( \Delta H \) and \( v^* \) for SAP differ notably from the values in aluminum and rapidly increase with temperature increase.

Consequently, between 100 and 200°C, we can conclude that the movement of dislocations is controlled by cross slip, whereas between 200 and 300°C, the climb of dislocations by diffusion of jogs is the determining process. A more detailed analysis of these processes, is described by the author in [5].

The high activation energies measured above 350°C show that the high temperature deformation mechanisms in SAP are radically different from that in aluminum, since in the latter the climb of dislocations is the controlling process up to the melting point, with a constant activation energy closed to the self diffusion energy [20]. The variation of the activation energy with the temperature is apparent, because the tensile tests were conducted under low stress at high temperature and vice versa. So the activation energy steeply increases with decreasing stress, as reported in [4, 5]. Similar activation energies for creep have been observed by other investigators in SAP [22, 23]. In fact, high values of activation energy, reaching at high enough temperatures ten times or more the self diffusion energy seem to be a general feature for dispersion hardened alloys: indium-glass composite [24], TD-Ni [25], Ni-Al\(_2\)O\(_3\) alloys [26].

Ansell and Weertman [23] developed a semi-quantitative theory
based on the generation of dislocations from grain boundaries as the controlling mechanism; the activation energy of such a process is obviously strongly stress dependent. Nevertheless they have not made theoretical estimations of the activation energy and volume in this case, and so no comparison can be made with experiment. Furthermore such process would give a grain size dependence, in contradiction with experiment [5, 13].

The author presented [4, 5] and earlier model rationalizing the preceding results, consisting of the thermally activated by-passing by glide of the particles intersecting the slip plane of a mobile dislocation. Nevertheless, even though the activation energy of such a process agrees well enough with the experimental results, the calculated activation volumes are between 10 and 100 times greater than the experimental ones.

We will propose here a second model, in theory generally applicable to alloys with f.c.c. matrix, and satisfying better the experimental results for SAP; it consists of the thermal activation of the dissolution of the junction reactions between attractive dislocations, as shown in Fig. 9a and b. Under an applied stress $\tau$ in the glide plane of the mobile dislocation CD, the (a) configuration takes an equilibrium position shown in (b). The thermal fluctuations modify the equilibrium configurations and can lead to the break up of M'N' and thus to the by-passing of AB by CD. Our calculations were conducted for the simple case where CD is parallel to MN. The details are given in [5] and [27]. The activation energy $\Delta H$ so obtained obviously depends on the inclination $\phi$ of the tree AB to the glide plane of CD. Figure 10 shows the variations of $\Delta H$ with $\tau$ for different values of $\phi$ and two values of $AB = CD/2 = 2\ell$. (These
extreme values of $l$ encompass the values of $l$ as determined by electron microscopy on specimens of SAP 4 strained at high temperature.)

We expect that this model will apply only if at high temperatures the dislocation density is large enough for the attractive trees to be the principal barriers to the mobile dislocations. The recovery by climb, strongly limited by the particles, and which ultimately explains the stability of these alloys at high temperatures, is sufficiently low to be able to observe, on SAP specimens strained at 500°C, large densities of dislocations and junction reactions [11], as shown in Fig. 10. For the large $\phi$, the agreement with experiment is satisfactory both for $\Delta H$ and for $v^*$ ($v^*$ is here given by the slope of $\Delta H - \tau$); the agreement is, however, not at all good for small $\phi$ [5].

Nevertheless the discrepancy is reduced if, instead of only considering a single controlling mechanism we consider two mechanisms in parallel, climb and dissolution of junctions [5].

IV. SUMMARY

We described in this paper the mechanisms of plastic deformation occurring in some alloys hardened by a dispersion of incoherent second phase. The mechanisms can be analyzed with respect to the deformation temperatures.

Below room temperature (at small strains in tensile deformation) the variations of the flow stress of SAP and Al/Al$_2$Cu with temperature and strain rate are well explained by cutting through a forest. The greater strength of the alloys, as compared with a pure matrix, would
correspond to a steeper increase of dislocation density with the strain, in which the particles are playing an indirect role of multiplication. At high enough strains, cross slip leading to a cell structure and diffusion can be induced in SAP.

Above room temperature (at small strains) various processes can be effective; cross slip and climb determine the moving of dislocations in the matrix of SAP and Al₂Cu/Al alloys, explaining the thermal decrease of the stress for a given strain, as in pure metals. For temperatures nearer to the melting point of the matrix, SAP, TD-Ni and some other alloys present very high activation energies. The thermal activation of dislocations pinned in their glide plane by junction reactions with attractive trees seems to control, perhaps in parallel with climb, the plastic deformation in SAP.

ACKNOWLEDGMENTS

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REFERENCES

3. Ashby, M. F., Electron Microscopy and Strength of Crystals, Inter-
    ASM (June, 1963).
14. Seeger, A., Dislocations and Mechanical Properties of Crystals,
18. Ruedi, E., Private communication. To be published.
Table 1. Structure Parameters of the Alloys

<table>
<thead>
<tr>
<th>Alloy type</th>
<th>Dispersed phase</th>
<th>f(%)</th>
<th>r(μ)</th>
<th>2R_S(μ)</th>
<th>grain size</th>
</tr>
</thead>
<tbody>
<tr>
<td>SAP 4*</td>
<td>Al_2O_3 γ</td>
<td>2.75</td>
<td>0.0253</td>
<td>0.208</td>
<td></td>
</tr>
<tr>
<td>SAP 7</td>
<td>&quot;</td>
<td>4.85</td>
<td>0.0293</td>
<td>0.182</td>
<td>2 - 5</td>
</tr>
<tr>
<td>SAP 10</td>
<td>&quot;</td>
<td>7.00</td>
<td>0.0314</td>
<td>0.162</td>
<td></td>
</tr>
<tr>
<td>4M</td>
<td>Al_2Cu</td>
<td>4.5</td>
<td>0.298</td>
<td>2.8</td>
<td>430</td>
</tr>
<tr>
<td>4C</td>
<td>&quot;</td>
<td>4.5</td>
<td>1.08</td>
<td>11.4</td>
<td>410</td>
</tr>
<tr>
<td>5F</td>
<td>&quot;</td>
<td>5.9</td>
<td>0.158</td>
<td>1.3</td>
<td>430</td>
</tr>
<tr>
<td>TD-Ni</td>
<td>ThO_2</td>
<td>≈2</td>
<td>0.0196</td>
<td>0.28</td>
<td>25 &lt;&lt; 125</td>
</tr>
</tbody>
</table>

f = volume fraction of dispersed phase.

r = mean planar particle radius.

2R_S = mean planar distance

*The number is the weight fraction of alumina.
Table 2. Energies of formation of a pair of jogs.

<table>
<thead>
<tr>
<th>Alloy</th>
<th>SAP 4.5</th>
<th>SAP 11.7</th>
<th>4M</th>
<th>4C</th>
<th>5F</th>
<th>Al 99.5%</th>
<th>Al 99.9%</th>
</tr>
</thead>
<tbody>
<tr>
<td>$2U_{fc}(\text{e.V})$</td>
<td>0.43</td>
<td>0.42</td>
<td>0.35</td>
<td>0.41</td>
<td>0.34</td>
<td>0.33</td>
<td>0.37</td>
</tr>
</tbody>
</table>

Table 3. Dislocation densities as a function of tensile strain at room temperature. SAP 4.5

<table>
<thead>
<tr>
<th>Strain $\varepsilon$%</th>
<th>mean density, cm$^{-2}$ (electron microscopy)</th>
<th>mean density, cm$^{-2}$ (activation volume)</th>
<th>distribution</th>
</tr>
</thead>
<tbody>
<tr>
<td>0</td>
<td>$(5 \pm 2) \times 10^9$</td>
<td>homogeneous</td>
<td></td>
</tr>
<tr>
<td>0.05</td>
<td>$(7 \pm 2) \times 10^9$</td>
<td>&quot;</td>
<td></td>
</tr>
<tr>
<td>0.2</td>
<td>$(1.05 \pm 0.3) \times 10^{10}$</td>
<td>$(1.9 \pm 1.2) \times 10^{10}$</td>
<td>&quot;</td>
</tr>
<tr>
<td>1</td>
<td>$(1.2 \pm 0.3) \times 10^{10}$</td>
<td>$(2.3 \pm 1.1) \times 10^{10}$</td>
<td>&quot;</td>
</tr>
<tr>
<td>2</td>
<td>$(3.6 \pm 1.0) \times 10^{10}$</td>
<td>$(4.9 \pm 1.1) \times 10^{10}$</td>
<td>cell structure</td>
</tr>
<tr>
<td>3</td>
<td>$(1.6 \pm 0.3) \times 10^{10}$</td>
<td>$(4.9 \pm 0.9) \times 10^{10}$</td>
<td>&quot;</td>
</tr>
<tr>
<td>5</td>
<td>$(6.2 \pm 1.3) \times 10^{10}$</td>
<td>$(6.9 \pm 1.1) \times 10^{10}$</td>
<td>&quot;</td>
</tr>
<tr>
<td>6</td>
<td>$(2 \pm 0.3) \times 10^{10}$</td>
<td>$(6.9 \pm 1.1) \times 10^{10}$</td>
<td>&quot;</td>
</tr>
</tbody>
</table>
FIGURE CAPTIONS

Figure 1. Variations of the flow stress, for different strains, as a function of temperature (the flow stress is corrected for the variations of the shear modulus μ with temperature).

Figure 2. Force distance diagram.

Figure 3. Variations of the activation energy function of temperature, at constant strain rate (\( \dot{\varepsilon} = 2.7 \times 10^{-4} \text{ s}^{-1} \)) and for different strains. SAP 11.7.

Figure 4. SAP 4.5, 0.2 pct tensile plastic strain at 196°C. Homogeneous distribution of dislocations. After [11].

Figure 5. SAP 4.5, 6 pct tensile plastic strain at 20°C. Well developed cell structure. After [11].

Figure 6. TD-Ni, 10 pct tensile plastic strain at 20°C. Early stages of cell formation. After M. von Heimendahl and G. Thomas [13].

Figure 7. Activation energy at constant strain rate as a function of temperature for SAP 3.6, 8.3, 11.7 and Al-99.5%.

\[ H = -T \dot{\varepsilon} \left( \frac{3}{3T} \right) \dot{\varepsilon}. \quad \dot{\varepsilon} = 2.7 \times 10^{-4} \text{ s}^{-1}. \quad \varepsilon = 2.1 \times 10^{-3}. \]

Figure 8. Activation volume as a function of temperature for SAP 3.6, 7.0, 11.7 and Al-99.5%. \( v^* = kT \left( \frac{\partial \ln \mu}{\partial T} \right). \quad \dot{\varepsilon} = 3.5 \times 10^{-4} \text{ s}^{-1}. \quad \varepsilon = 2.1 \times 10^{-3}. \)

Figure 9. Junction reaction between AB (tree) and CD (mobile) attractive dislocations.

(a) without applied stress

(b) with applied stress

Figure 10. Activation energy as a function of stress for the dissolution
of a junction reaction in aluminum.

Figure 11. SAP 4.5 0.2 pct tensile strain at 500°C. Junction reactions at A and B. After [11].
Fig. 2

- POLYCRYSTAL Al 99.5% [6]
- SINGLE CRYSTAL Al [12]
- Cu/Al DISPERSION ALLOYS [12]
- SAP 11.7 [6]
- SAP 4.5 [6]
Fig. 3
Fig. 8
Fig. 9
Fig. 10
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