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HIGH TEMPERATURE RELAXATION MECHANISMS IN Cu-Al SOLID SOLUTIONS

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ABSTRACT
The high temperature damping of copper-aluminium solid solutions has been investigated in single crystals containing various aluminium contents: 3.1 - 5.4 - 7.3 - 9.3 and 11.4 at % of Al. Measurements have been performed in a wide frequency range (10^{-4} Hz - 10 Hz) between room temperature and 1200 K. Two internal friction peaks have been observed in slightly strained specimens. Both the relaxation strength and the activation energy have been found to be sensitive with the aluminium content. The results can be explained with a mechanism involving dislocation climb and independent gliding of Shockley partials in widely split dislocations.

I - INTRODUCTION
High temperature relaxation effects in metals are commonly assigned to mechanisms involving dislocation climb especially in the case of single crystals where no kind of grain boundary sliding can occur. Therefore the stacking fault energy is one of the most important parameters determining the plastic behavior, through the splitting of dislocations. Cu-Al solid solutions can give valuable results since the composition dependence of $\gamma$ is now well documented.

II - EXPERIMENTAL METHOD
The experimental method has been discussed elsewhere /1/. A variable frequency torsional pendulum was used allowing isothermal internal friction measurements for vibration frequencies ranging between $10^{-4}$ and 10 Hz at maximum strain amplitudes between 5.10^{-6} and 10^{-5}. In that strain amplitude range no clear influence of the applied stress level could be detected.

Single crystals were spark-machined in the form of flat bars (50x5x1 mm) parallel to $<111>$ directions. Specimens were prepared from 5 N copper containing various amounts of aluminium: 3.1 - 5.4 - 7.3 - 9.3 and 11.4 at %.

III - EXPERIMENTAL RESULTS
No clear peaks were found in specimens with low aluminium contents (3.1 and 5.4 at %) in the initial state but a small peak appeared, superimposed onto an exponential low frequency background, after the specimens were submitted to a 1 % flexure strain. Isothermal damping spectra obtained at increasing temperatures on well stabilized states are shown in figure 1.

In specimens with larger aluminium contents (9.3 and 11.4 at %) a peak was detected in the initial state which was significantly increased during subsequent straining (1 % in flexure). The peak was also shifted towards the high frequencies. This
double effect is clearly illustrated in figure 2 where the isothermal damping curves are replotted versus the measurement temperature. In figure 3 it can be observed that annealing at very high temperatures increases the height of peak and does not modify its frequency location.

In figure 4 are plotted the variations of the value of the maximum damping with the annealing temperature in the case of a specimen containing 11.4 at % of Al. The apparently complex behavior can be simply explained when the damping maximum is analyzed in two elementary peaks: a "low" temperature peak $P_1$ and a high temperature one $P_2$. When the specimen is first heated from $A$ to $B$ (figure 4) the amplitude of the $P_1$ peak decreases and remains constant during cooling (from $B$ to $C$). During further heating the height of the $P_1$ peak goes on decreasing (from $B$ to $D$) and again remains stable during cooling (from $D$ to $E$). A higher temperature annealing (from $D$ to $F$) increases the height of $P_2$ which attains a roughly stable value (from $F$ to $C$ cooling).

These results were obtained assuming a low frequency background obeying the relation $1/\omega^n$, /1,2,3,4/.

The activation parameters of $P_1$ and $P_2$ are listed in Table 1.

<table>
<thead>
<tr>
<th>% of Al</th>
<th>$T_m$ (K)</th>
<th>$T_m$ (Hz)</th>
<th>$H_p$ (eV)</th>
<th>$H_p$/H</th>
<th>$C_0$ (s)</th>
</tr>
</thead>
<tbody>
<tr>
<td>3.1</td>
<td>$P_1$ 1348</td>
<td>.67</td>
<td>1.99</td>
<td>.98</td>
<td>$10^{-11}$</td>
</tr>
<tr>
<td></td>
<td>$P_2$</td>
<td>.90</td>
<td>1.64</td>
<td>.84</td>
<td>$10^{-16}$</td>
</tr>
<tr>
<td>5.4</td>
<td>$P_1$ 1341</td>
<td>.75</td>
<td>1.57</td>
<td>.77</td>
<td>$10^{-9}$</td>
</tr>
<tr>
<td></td>
<td>$P_2$</td>
<td>100</td>
<td>1.24</td>
<td>.61</td>
<td>$10^{-6}$</td>
</tr>
<tr>
<td>7.3</td>
<td>$P_1$ 1335</td>
<td>.80</td>
<td>1.50</td>
<td>.73</td>
<td>$10^{-9}$</td>
</tr>
<tr>
<td></td>
<td>$P_2$</td>
<td>100</td>
<td>1.30</td>
<td>.64</td>
<td>$10^{-6}$</td>
</tr>
<tr>
<td>9.3</td>
<td>$P_1$ 1328</td>
<td>.90</td>
<td>1.99</td>
<td>.98</td>
<td>$10^{-10}$</td>
</tr>
<tr>
<td></td>
<td>$P_2$</td>
<td>100</td>
<td>1.60</td>
<td>.78</td>
<td>$10^{-7}$</td>
</tr>
<tr>
<td>11.4</td>
<td>$P_1$ 1322</td>
<td>.75</td>
<td>2.28</td>
<td>1.10</td>
<td>$10^{-11}$</td>
</tr>
<tr>
<td></td>
<td>$P_2$</td>
<td>.85</td>
<td>1.99</td>
<td>.98</td>
<td>$10^{-8}$</td>
</tr>
</tbody>
</table>

From table 1 it appears that $P_2$ is a very high temperature peak since it would be located at the melting temperature $T_m$ for a vibration frequency of 1 Hz; It is also observed that the activation energy of $P_1$ is higher than that of $P_2$ but that, conversely, the limiting relaxation time $\tau_0$ is much higher for $P_2$.

The composition dependence of the activation energies is shown in figure 5. It is observed that for both peaks the energy is minimum for an aluminium content between 5.4 and 7.3. A similar result had been previously reported in the case of polycrystals /5/.

The relaxation strength is shown in figure 6, for the case of $P_2$ and for the same annealing conditions. It increases with the aluminium content, while the peak is shifted towards high frequencies.

**IV - DISCUSSION**

For the thermally activated motion of dislocation segments, the limiting relaxation time is of the form /6/:

$$\tau_0 = \frac{a}{\mu d^3} \frac{kT}{\gamma_0 bh}$$
α being a coefficient of the order of 1/6 for pure climb /6/, ν₀ the Debye frequency, λ the mean length of the moving segments and h the distance covered during one activation event. It leads to λ values ranging between 1 and 10 μ for the P₂ peak and between 0.05 - 0.5 μ for the P₁ peak. Thus P₁ can correspond to the motion of short tangled segments belonging to cells formed during straining and P₂ to free dislocations remaining after destruction of the cells during high temperature annealings.

The observed minimum in the activation energy of the peaks seems to indicate that at least two elementary mechanisms are involved, the slowest one controlling the deformation: for example diffusion enhanced dislocation climb and independent glide of Shockley partials in widely split dislocations.

In alloys with a low aluminium content, corresponding to a high stacking fault energy, deformation is controlled by the glide of partials which strongly interact and in the highly concentrated alloys deformation is controlled by climbing of widely split dislocations.

V - REFERENCES

/6/ FRIDEL, J., Dislocations, Pergamon Press.

Figure 1 - Internal friction measurements after successive annealings on heating in Cu 11.4 % at Al.

Figure 2 - Influence of the pre-strain 1 % in Cu 3.1 % at Al.

(a) and (b) pre-strained specimen
(o(a) and (b) initial state.
Figure 3 - Influence of the annealing temperature in pre-strained Cu 9.3 % at Al. Specimen annealed at: (+) 1178 K and (■) 1070 K.

Figure 4 - Variation of the activation energy with aluminium content in copper.

Figure 5 - Variation of the maximum damping with the annealing temperature in Cu 11.4 % at Al. pre-strained 1 % in flexure.

Figure 6 - Internal friction measured at 1020 K after annealing at 1160 K in pre-strained specimens 1 % in flexure
(△) Cu 3.1 % at Al.
(□) Cu 9.3 % at Al.
(•) Cu 11.4 % at Al.