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MEDIUM TEMPERATURE INTERNAL FRICTION IN HIGH PURITY f.c.c. AND h.c.p. METALS

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Abstract.- We study in high purity polycrystalline aluminium and magnesium a strong internal friction peak $P_1$ situated around 0.5 $T_M$ ($T_M =$ melting point in K). This peak presents a maximum as a function of the oscillating stress amplitude. The results are in agreement with the model of dislocation glide controlled by jog climbing and diffusion of vacancies along dislocations.

1. Introduction.- Internal friction peaks attributed to grain-boundary relaxation have been much studied [1]. Recently, Woirgard [2] has shown that the medium and high temperature internal friction spectra in f.c.c. metals is composed of peaks $P_d$ situated at about 0.4 $T_M$ and peaks $P_V$ situated near 0.6 $T_M$. For aluminium, $P_d$ peaks are absent and $P_V$ peaks are composed of three peaks $P_1$, $P_2$, $P_3$. For polycrystals, $P_2$ peak prevails and can be identified to the $K_\theta$ peak [3].

Recently, we have observed in high purity aluminium [4] a strong internal friction peak $P_1$. We have attributed this peak to the dislocation glide controlled by jog climbing and diffusion of vacancies along dislocations. Later, Friedel [5] has attributed this peak to the motion by cross slip of the nodes of polygonized walls. In order to verify either one hypothesis or the other, we propose to present some experiments on high purity metals (Al, Mg) and particularly to study the non-linear effects on the $P_1$ peak.

2. Experimental results.

Aluminium : Measurements were made on high purity (6 N) polycrystalline aluminium with a torsion pendulum described elsewhere [4]. All the samples were annealed 1 h at 500 K under primary vacuum after rolling (the average grain diameter is 1 mm). Figure 1 shows the effect of strain amplitude on the internal friction spectra for a sample cold worked in torsion to 1.2 % and annealed at 580 K (the measurement frequency is about 1 Hz). We observed firstly that the height of the $P_1$ peak situated around 450 K increases with the oscillating strain amplitude $\varepsilon_m$ then the internal friction background monotonically increasing with temperature becomes important. The peak temperature decreases when the vibration amplitude increases.

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The effect of a bias torsion stress of \(2 \times 10^{-5}\) \(\mu\) (\(\mu\) is the shear modulus) on the \(P_1\) peak is shown by figure 2. We can note that the peak height is not very sensitive to a bias stress. Similar results have been obtained for samples without coldwork, deformed by creep or fatigued. The \(P_1\) peak increases by annealing up to 500 K and then is very stable: its height does not vary by annealing up to about 750 K. The activation parameters measured by changing the frequency \((0.03 - 0.6\) Hz) for a vibration amplitude of \(1.2 \times 10^{-5}\), are: activation energy \(E = 1.05 \pm 0.1\) eV.

frequency factor \(f_o = 4 \times 10^{11} s^{-1}\).

For an usual purity (5 N) aluminium, we observe the normal Ke peak (or \(P_2\) peak) situated around 550 K, the \(P_1\) peak being absent.

**Magnesium**: Measurement were made on high purity (6 N) polycrystalline samples, annealed 2 h at 570 K under argon after wire-drawing (average grain size about 1 mm).

Figure 3 shows the effect of the vibration amplitude on the internal friction in the temperature range 300-500 K, for a sample cold-worked in torsion of 2\% at 10 K and annealed at 540 K (frequency is about 1 Hz). As in aluminium, we observe a \(P_1\) peak situated around 400 K; its height increases also with the vibration amplitude then the internal friction background becomes prevailing. The activation energy measured by frequency change is about 1 eV for a strain amplitude \(\epsilon_m = 10^{-5}\).

For a 5 N magnesium, the \(P_1\) peak does no more appear but the classical grain boundary peak is present.

3. **Discussion**.- We have observed a strong internal friction peak \(P_1\) in aluminium and magnesium. This peak is very stable during annealing.

Woirgard [2] has attributed the \(P_1\) and \(P_3\) peaks of aluminium to the dislocation climb by bulk diffusion of vacancies. This hypothesis seems correct for the \(P_2\) and \(P_3\) peaks of which the activation energy is about the energy for the bulk diffusion \(H_V\). But for the \(P_1\) peak, this hypothesis cannot be retained, the activation energy (about 1 eV) being too low for the process proposed by Woirgard. So, we think, that for the \(P_1\) peak [4], the dislocation glide is controlled by jog climbing with diffusion of vacancies along dislocations. This model was criticized by Friedel [5] who suggests that the peak strength must present a saturation as a function of the oscillating stress amplitude due to the limited diffusion of vacancies during a measurement cycle.

Now, we shall discuss in more detail the model of the dislocation glide controlled by jog climbing, illustrated by the figure 4. The dislocation line pinned by jogs jumps from position 1 to position 3 under the effect of an applied stress \(\sigma\) and during this motion, there is an emission or an absorption of a vacancy \(V\). For low values of stress
\( \sigma (\sigma a b L \ll H d) \), the internal friction corresponding to this process is given by |
\[
\delta = \pi \Delta \omega \tau / (1 + \omega^2 \tau^2) \quad \text{with} \quad \Delta = \frac{A L^2}{10} \quad \text{and} \quad \tau = \frac{k T L^2}{10 v_D \mu b^5} \exp \left( \frac{H d}{k T} \right) (1)
\]

Thus, the activation energy of the peak is equal to the activation energy for self-diffusion along dislocation \( H d \).

For high values of stress \( (\sigma a b L > k T) \), we can obtain the internal friction only by numerical calculation \(|6| \) but we can predict that the peak will shift towards low temperatures when \( \sigma \) increases, the activation energy of the peak varying about as \( E = H d - \sigma v \).

This phenomenological model does not take into account the limited motion of vacancies during the oscillating motion of the jog and furthermore, neglects the effect of the dislocation tension line. Indeed, let's consider the dislocation energy before the jump \( W(1) \) and after the jump on a distance \( a \) (with an emission or absorption of a vacancy) \( W(3) \). The difference between these two energies without stress is due to the increase of the dislocation length (see curve a on fig. 4 B) : \( W(3) - W(1) = \mu b^4 / L \). For observing an important relaxation, the two positions must have the same energy. Therefore, without stress, the internal friction peak will be very small. But when the stress increases, the energy \( W(3) \) is decreasing by the stress work. Thus, the two positions 1 and 3 become energetically equivalent when \( \sigma_c a b L = \mu b^4 / L \) or \( \sigma_c / \mu = (b / L)^2 \) (2) and the relaxation strength must increase.

Once as the stress satisfies this condition, the jog motion can be limited by the number of vacancies coming towards the jog. The number of jog jumps cannot exceed about \( 1_d / b \), where \( 1_d \) is the distance travelled by jog during half a cycle \( (1_d = v_D b / 2 f \exp (- H d / k T)) \) with \( f = \text{frequency} = 1 \text{ Hz} \) (3). But the possible maximum displacement of jog is about \( 1_M = L^2 \sigma / \mu b \) and when \( 1_M > 1_d \) (4), the jog motion will be limited and the dislocation behaviour will be elastic and thus the internal friction decreases.

Our experimental results on aluminium and magnesium agree very well with this model as shown by internal friction curves obtained by subtracting the background \( \delta_f = \delta_c \exp (- C t / T) \) (fig. 5 and 6). The critical stress \( \sigma_c \) corresponding to the condition of the peak appearance is about \( 10^{-6} \mu \) : this value is very reasonable and corresponds to a dislocation length of \( 10^3 b \) (the internal stresses are negligible). Then, the peak presents a maximum for a periodic stress \( \sigma_d \). This stress is about \( 3 \times 10^{-6} \mu \) for aluminium and \( 6 \times 10^{-5} \mu \) for magnesium. It seems not affected by a bias stress (fig. 2) : this results is consistent with our model, only the periodic stress limiting the jog motion.

From relation (4), we obtain \( 1_d = 3 b \) for Al and \( = 30 b \) for Mg.
with $L = 1000 \, \text{b}$. From relation (3), we can deduce $H_d$: $H_d$ is about 1 eV for the two materials. This value is near the activation energy of the peak. Furthermore, we can obtain from the peak shift versus the strain amplitude, an order of the activation volume $v = a b L$: we find $v = 1000 \, \text{b}^3$ for aluminium and less for magnesium. From the activation energy of the peak, we can obtain $H_d = E + \sigma v = 1,1 \, \text{eV}$. This value of $H_d$ is in good agreement with the activation energy observed for self-diffusion along dislocations by Murakami and Yoshida [7] and calculated by Mimkes [8].

Thus, the experimental results obtained for Al and Mg are consistent with the model of dislocation glide controlled by jog climbing proposed for the $P_1$ peak. The model proposed by Friedel [5] does not give an account of the experimental results, in particular the decrease of the peak height observed for high stresses. Further, it is difficult to attribute the $P_1$ peak to the dislocation motion by cross-slip in magnesium, the activation energy for cross-slip being higher for hexagonal metals [9].

4. References

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Fig. 1: Internal friction spectra for different vibrating stress amplitudes in 6N aluminium.

Fig. 2: $P_1$ peak in aluminium for different oscillating frequencies and with or without static stress $\sigma_3$.

Fig. 3: Internal friction in 6N magnesium for different strain amplitudes.
Fig. 4: A) Configurations of a jogged dislocation: 1 - before, 3 - after formation and migration of a vacancy. B) Energy diagram: a) torsion line effect, b) formation and migration energy of vacancies, c) applied stress work.

Fig. 5: P1 peaks in aluminium for different strain amplitudes (see fig. 1) without background and relaxation strength or maximum temperature versus strain amplitude.

Fig. 6: P1 peaks in magnesium for different strain amplitudes (see fig. 3) without background and relaxation strength versus strain amplitude.