Journal of Physics D: Applied Physics

RESEARCH NOTE

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J. Phys. D: Appl. Phys., 1971, Vol. 4. Printed in Great Britain

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Internal friction in indium single crystals

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MS. received 24*th April* 1970, *in revised form* 17*th August* 1970

Abstract. Low frequency, internal friction measurements made as a function of temperature on pure indium in single crystal form have shown three effects which could be resolved into the following three separate mechanisms: (i) A frequency insensitive peak in internal friction observed at about 170 K is believed to be associated with the motion of twin boundaries; (ii) A relaxation measured at about 273 K is interpreted as originating from the motion of dislocations interacting with vacancies; (iii) A background relaxation between 320 K and the melting point is attributed to dislocation climb.

1. Introduction

This work cites three internal friction effects recently observed in 99·9999% indium single crystals during an investigation of anelastic effects in In–Tl alloys (de Morton 1969). Low frequency measurements (about 1 Hz) were made on wire specimens in an evacuated inverted torsional pendulum which eliminated tensile creep at temperatures near the melting point (429 K). Further experimental details are given by de Morton (1969). Initial measurements, made at a surface shear strain of 4×10^{-6}, showed two frequency dependent effects:

(i) a peak at about 273 K (subsequently referred to as the 273 K peak), and
(ii) an exponential rise in the background damping above 320 K, as shown in figure 1.

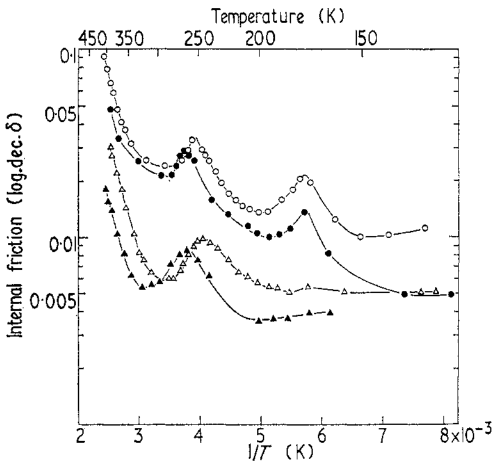

Figure 1. Internal friction of pure indium as a function of strain amplitude and temperature. ○, ●: strain amplitude = 3×10^{-5}, frequency = 0·5, 1·4 Hz respectively. △, ▲: strain amplitude = 4×10^{-6}, frequency = 0·3, 1·7 Hz respectively.

On increasing the strain amplitude to 3×10^{-5} an additional peak was observed at about 170 K (subsequently referred to as the 170 K peak) (figure 1) which was insensitive to frequency; the overall background damping also increased without affecting the character of the two effects already mentioned. These three effects will now be discussed separately.

2. Peak at 170 K

The fact that the peak at 170 K was frequency insensitive (ie not thermally activated) and that it was observed in a monocrystal of high purity eliminates Bordoni, Zener and grain boundary type relaxations as possible explanations for this effect. Indium solidifies as a face centred tetragonal structure (c/a 1·075) which in single crystal form is lightly twinned. The presence of twins suggests a possible explanation for this peak, since it is known that the migration of twin boundaries can give rise to internal friction (Zener 1948). Furthermore, it has been demonstrated (Remant *et al.* 1964) by electron microscopy on thin films of indium that, below 188 K, fine twins (0·1–0·2 μm in width) are nucleated by the movement of dislocations in the films. The dislocations themselves are generated by the combined effects of the electron irradiation and stresses associated with the contaminated layer which formed on the specimen during examination in the microscope. As this layer increases in thickness the twins grow, due to the associated increase in stress, whilst heating the film above 188 K causes the twins to shrink and disappear.

These observations, and the increase in internal friction measured at the higher strain amplitude, prompt the suggestion that dislocations are generated in the internal friction specimen at low temperatures by the stress of measurement, and similarly, these dislocations nucleate fine twins. The internal peak at 170 K could then be accounted for by the reversible stress-assisted motion of twin boundaries producing a reorientation of tetragonal regions in the volume swept by the twin interfaces. At low temperatures (less than 150 K), the stress would be insufficient to propagate twin boundaries since the stress to maintain a given velocity of twin propagation increases with decreasing temperature (Cooper and Washburn 1967) and damping would therefore decrease. At higher temperatures (greater than 200 K), the reduction of micro-twins on heating above 188 K, observed by Remant *et al.*, would also diminish damping. On this basis, a shear stress σ for the propagation of twins in indium can be estimated, since $\sigma = \epsilon G$, where G is the shear modulus ($0{\cdot}65 \times 10^6$ g mm^{-2}; $0{\cdot}64 \times 10^{10}$ N m^{-2}) and ϵ the shear strain (3×10^{-5}). This gives a value of 20 g mm^{-2} ($1{\cdot}96 \times 10^5$ N m^{-2}) for twin propagation in this indium single crystal at 170 K; the shear stress for twin propagation in zinc single crystals at 300 K is 150–450 g mm^{-2} ($1{\cdot}47 \times 10^6$–$4{\cdot}4 \times 10^6$ N m^{-2}) (Cooper and Washburn 1967).

3. Relaxation at 273 K

The activation energy Q_R and frequency factor τ_0 for this relaxation were estimated from the peak shift with frequency, and are shown in table 1. The activation energy and

Table 1. Values of activation energy Q_R, frequency factor τ_0, and relaxation strength ΔG at different strain amplitudes, for the relaxation at 273 K

Strain amplitude	Q_R (eV)	τ_0 (s)	ΔG
4×10^{-6}	$0{\cdot}49 \pm 0{\cdot}09$	5×10^{-11}	0·032
3×10^{-5}	$0{\cdot}52 \pm 0{\cdot}09$	3×10^{-11}	0·095

frequency factor values were essentially independent of strain amplitude, whereas the relaxation strength ΔG increased by a factor of three when the strain amplitude was increased from 4×10^{-6} to 3×10^{-5}. Measurements of the half peak width of this relaxation peak were also within a few per cent of the theoretical half peak width, indicating a relatively simple activated process. Comparison of Q_R with the activation energy for self-diffusion Q_D, which is 0·78 eV (Eckert and Drickamer 1952), showed no correlation. In seeking an explanation for this relaxation, the marked strain–amplitude dependence of internal friction and the unusually high value of τ_0 (5×10^{-11} s) is worth noting. Similar high τ_0

results, for anelastic relaxations on deformed copper and gold, led Okuda and Hasiguti (1963) to consider a relaxation mechanism involving the association of dislocations with vacancies and vacancy complexes. Turner and De Batist (1964) subsequently showed that Okuda and Hasiguti's data presented on a Wert and Marx (1953) type plot gave a straight line with a characteristic slope. The present data also fit closely on this line.

The present results, however, are equilibrium relaxations observed on well annealed single crystals at 0·65 T_m on both heating and cooling. It is suggested that existing dislocations are activated during the measurement and, at an optimum temperature, interact with thermally created defects to produce a viscous drag on the dislocations. Assuming that the most probable lattice defects available at these temperatures to interact with dislocations are vacancies, the activation energy measured should be that for vacancy migration, Q_M. A value of 0·49 eV for Q_M in indium is, however, somewhat higher than expected; for example, in fcc metals Q_M/Q_D is 0·36–0·48 (Cottrell 1964) whereas Q_R/Q_D for the current work is 0·6. However, the value of Q_R measured could conceivably be the sum of Q_M and the binding energy of a vacancy to a dislocation E_B, by analogy with the cold work peaks observed in bcc metals (Shoeck and Mondino 1963). In this case, taking an average value of Q_M/Q_D as 0·42 gives E_B equal to 0·16 eV.

4. Background relaxation

Above 320 K (0·75 T_m), internal friction increases rapidly up to the melting point and, as shown in figure 1, the logarithmic plot is linear with $1/T$ (K). The frequency dependence of this background damping, which is observed in both single and polycrystalline specimens, has suggested a relaxation process of the following relationship (Niblett and Wilks 1960):

$$Q^{-1} = \delta/\pi = A \exp(-Q_B/kT). \tag{1}$$

Estimates of the activation energy Q_B from the slopes of the curves shown in figure 1 give values of 0·24–0·28 eV. These values are about 0·3 Q_D. Schoeck *et al.* (1964) pointed out that the magnitude of background internal friction is too large to be explained solely by a mechanism involving point defects, and suggested that it is probably associated with a thermally activated dislocation relaxation. These workers also questioned the validity of equation (1), and derived the following expression for the background relaxation:

$$Q^{-1} = K\{\omega \exp(Q_B^*/kT)\}^{-n} \tag{2}$$

where k and n are constants, Q_B^* is the true activation energy for the viscous process and ω is the oscillating frequency. The 'apparent' activation energy Q_B is related to Q_B^* by

$$Q_B = nQ_B^*. \tag{3}$$

Since n is characteristically less than unity, Q_B^* is always greater than Q_B; for Al it is approximately equal to Q_D. The frequency dependence of Q^{-1}, indicated in equation (2) and shown in figure 1, has been used to obtain values of n by plotting ln Q^{-1} against ln ω. The value of n obtained was 0·4 which, through equation (3), gives 0·65 eV for Q_B^*. Since this is reasonably close to Q_D, the rate determining process for this background relaxation is probably dislocation climb. If this is so, the technique would be of value in determining the creep behaviour in materials near their melting points.

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