Figure 2 Equilibrium diagram of the Cu-Se system.

are not so rapid that the heating and cooling of samples during measurements could be considered as maintaining complete equilibrium. Therefore this part of the equilibrium diagram is established with less precision. In any case this other phase is not Cu_3Se_2 but, more likely, CuSe. This is in agreement with Heyding's observation [1] that in samples which have once been heated above 135°C, Cu_3Se_2 no longer appears.

The homogeneity range of the cubic or high-temperature phase of $Cu_{2-x}Se$ at room-temperature (26°C) determined by means of measurements of electrical conductivity ($Cu_{1.79}Se$-$Cu_{1.73}Se$) is not in agreement with the range determined from X-ray measurements ($Cu_{1.82}Se$ − $Cu_{1.75}Se$) [1]. In view of the drastic and easily measurable changes in conductivity at temperature T_α we must conclude that electric conductivity supplies more reliable information.

References

1. R. D. HEYDING, *Canad. J. Chem.* **44** (1966) 1233.
2. Z. OGORELEC and B. ČELUSTKA, *J. Phys. Chem. Solids* **30** (1969) 149.
3. Z. OGORELEC, R. ROČAK, and J. IVIĆ, *Phys. Stat. Sol.* (a) **6** (1971) K29.
4. B. ČELUSTKA and Z. OGORELEC, *Croat. Chem. Acta* **41** (1969) 73.

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On the amplitude-independent internal friction in crystalline solids

Recent work on the background internal friction in alkali halides [1], high-purity copper [2] and alpha-brasses [3], has established the inapplicability of the Granato-Lücke model to the results,

over a wide frequency spectrum. To explain the observed features, the depth and coherency of the internal microstress-fields have to be taken into account [4]. In this note we present new measurements of the background internal friction in high-purity brass polycrystals containing from 0·1 to 5 at% of zinc. These results

bridge the gap between corresponding ones on copper and on concentrated alpha-brasses [3]; they are briefly discussed in the light of the model outlined in [4].

The experimental procedure was similar to that reported by Spears [3]. Polycrystals of 45μm grain size containing 0·1, 1·0 and 5 at% of zinc were studied, and measurements were correlated with results previously obtained with similar alloys containing up to 30% of zinc [3]. Details relating to the impurity content, and to experimental procedure, were essentially as outlined before [3]. Rod-shaped specimens oscillated longitudinally *in vacuo* at 12 kHz, at a strain amplitude of about 10^{-7}; the internal friction was amplitude independent. No characteristic peaks, e.g. of the Bordoni or Hasiguti types were observed in the temperature range used, 100 to 280 K. The rods were pre-strained various amounts at room temperature immediately before the measurements were made. Results, together with some relating to specimens of concentrated alloys, are shown in Figs. 1 and 2.

A strong dependence of Q^{-1} on prestrain and concentration is evident; it should be noted that the Q^{-1} ordinate in Fig. 1 is scaled up by a factor of 10 for the dilute alloy (a) compared with the concentrated one (b).

The maxima of the curves correspond in all cases approximately to the end of the linear, "stage II" work-hardening region, which, as in polycrystalline copper for example [5], is well defined in annealed brasses. The correlation of Q_{max}, corresponding to peaks in isotherms of the type shown in Fig. 1, with zinc content

Figure 2 Dependence of Q_{max} on zinc content (a) and on temperature (b). Data referring to concentrations of zinc in excess of 5 at.% are taken from reference [3].

(Fig. 2a) and temperature (Fig. 2b) are linear; the deviation indicated by the point corresponding to 25 at.% of zinc, in Fig. 2a, is probably a consequence of enhanced short-range order, possibly near dislocations. Similar anomalies at concentrations corresponding to Cu_3Zn have also been observed in the low-temperature stress relaxation of α-brasses [6].

In the "vibrating kinks" model [4] the length L_r and the activation energy for the displacement of a kinked segment, H_r, are related, such that $H_1/L_1 = H_2/L_2 = \ldots = H_{max}/L_{max}$, where the upper limit, H_{max}, depends upon the strain, and H_1 relates to a single kink. Also, it is assumed that the product of the number of L-segments per unit volume, $N(L)$, and the length, i.e. $N(L).L$ is constant, independent of L within the range of available L-values. With this distribution one then obtains [4]

$$Q^{-1} = \frac{\Omega N_{max} L_{max} G b^2}{2\lambda\sigma_\infty} (kT/H_{max}) \ln(\omega_D/\omega), \quad (1)$$

where Ω is an orientation factor, λ the mean spacing between adjacent geometric kinks on dislocations in the absence of an applied stress, G the shear modulus, σ_∞ a constant equal to about $10^{-6}G$, ω_D a "Debye" frequency, and ω the vibrational frequency of the specimen.

Except for the non-zero intercepts on the abscissa, i.e. about 20 to 50 K in Fig. 2b, the linearity in T, required by Equation 1 is observed. The discrepancy is readily explained if it is considered that Equation 1 is valid only for $L \geqslant L_1$, a "refinement" which was not included in the derivation of the equation.
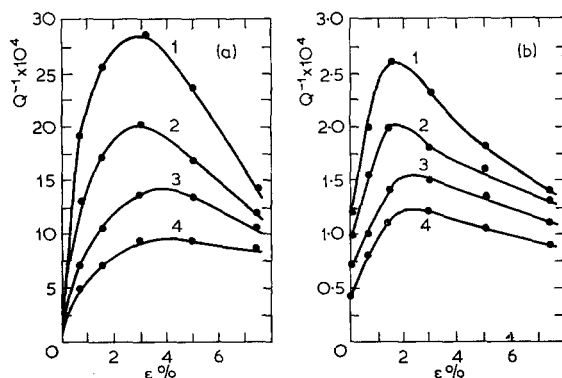

Figure 1 The effect of tensile prestrain on Q^{-1} of polycrystals containing (a) 0.5 at.% Zn and (b) 10 at.% Zn, at 280 K(1), 230 K(2), 180 K(3) and 130 K (4). Results for the concentrated alloy are from reference [3].
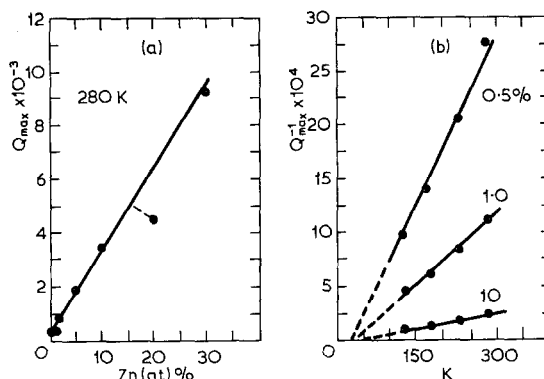
In attempting to interpret the effect of alloying on Q^{-1} it was considered that, at least as a rough approximation, one may have H_{max} $(c) = H_{max}(0)$. $[1 + \alpha c]$, where c is expressed as atomic fraction, and α is a constant. If the first fraction in Equation 1 were not significantly concentration dependent at strains corresponding to Q_{max}, which is however an unsubstantiated hypothesis at present, then the relation shown in Fig. 2a would be expected to hold. However, this point requires further investigation.

The energy barrier H_1 should relate to single kinks. If the extrapolation of the lines to about 30 K is accepted, simple considerations show that values of H_1 are of the order of a few meV, i.e. they are of reasonable magnitude to be interpreted in terms of lattice friction.

A point of interest in relation to Fig. 2a is the continuity of the line; there appears to be no qualitative difference between the behaviour of dilute and concentrated alloys. These observations may have implications for the theory of solid-solution hardening [7].

Den Buurman and Weiner [2] find that in their 99·999% copper, Q^{-1} was higher at 78 K than at room temperature. This does not invalidate Equation 1, for it is well known that under the experimental conditions used by them the "foot" of the range of Bordoni peaks is observable at 78K [8], and this results in a lifting of the background level of internal friction.

Although some features, e.g. the dissociation of dislocations, in the fcc lattice, have not been allowed for explicitly, the present results point clearly to the pronounced effects of the micro-stress-fields on the internal friction of crystalline materials.

It is probable that the broadening of the Bordoni and Hasiguti peaks results from processes of the type here considered. A more detailed, stochastic, treatment of the entire problem is desirable; this might facilitate dispensing with some of the assumptions which had to be introduced in the derivation of Equation 1.

References

1. W. H. ROBINSON, J. Mater. Sci. 7 (1972) 115.
2. R. DEN BUURMAN and D. WEINER, Scripta Met. 5 (1971) 753.
3. C. J. SPEARS, J. de Physique, 32 (1971) C2-183.
4. P. FELTHAM, Scripta Met. 5 (1971) 395.
5. P. FELTHAM and J. D. MEAKIN, Phil. Mag. 2 (1957) 1237.
6. P. FELTHAM, J. Inst. Metals 89 (1960) 211.
7. G. LEHMANN, Brit. J. Appl. Phys. 2 (1969) 126.
8. P. FELTHAM and C. BURDETT, "Realstruktur und Eigenschaften von Reinstmetallen" (Akademie Verlag, Berlin, 1967) p. 615.

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Book Review

Mechanical Properties of Solid Polymers

I. M. Ward

Wiley-Interscience, 1971, pp XV + 375. Price £7

Professor Ward's new book will be welcomed by the increasing number of University teachers giving courses in the physics and mechanical properties of polymers. Previous books in the same field have generally lacked the rigour and didactic quality required for teaching, but these features are well provided by the present volume.

The book is an extended version of lectures given in the Bristol M.Sc. course in the Physics of Materials. The treatment is not abstruse, however, and the book is quite suitable as a text for third, and even second, year honours undergraduates with appropriate backgrounds.

Apart from its usefulness as a textbook, it should also serve as a valuable introduction for research workers and industrial scientists who lack background in the mechanics of polymers but need to consider mechanical properties in the course of their work.

The approach adopted by Professor Ward is largely a continuum one. That is to say that polymeric solids are treated as continuous media