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Internal friction due to twinning dislocations in Cu–Ge solid solution alloys

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Abstract

Anelastic behaviour of Cu–Ge solid solution alloys has been studied by low frequency internal friction measurements. Plastic deformation gives rise to characteristic internal friction in the temperature range between 80 K and 350 K, which is attributed to the motion of twinning dislocations. By using single-crystal specimens of a Cu–8%Ge alloy, the dependence of the magnitude of the internal friction on the crystal orientation and the degree of deformation has been examined. The results suggest that the internal friction is due to local oscillatory motion of individual twinning dislocations. The internal friction is suppressed completely by electron irradiation, which can be explained in terms of pinning of the dislocations by point defects.

1. Introduction

Copper- and silver-based f.c.c. solid solution alloys deform by twinning at ambient temperatures, after the work hardening by slip deformation [1]. Narita *et al.* found that Cu–8at.%Ge crystals exhibit a large pseudo-elastic hysteresis loop in the stress–strain curve when subjected to a loading–unloading cycle in the twinning stage, and attributed this phenomenon to reversible movements of the boundaries of microtwins [2]. The present authors studied the internal friction of Cu–Ge, Cu–Si and Cu–Al alloys to see whether such mobile boundaries exhibit anelastic effects [3]. It was found that deformed specimens display characteristic internal friction, which increases linearly with temperature in the range from 80 K to 280 K and falls rapidly above 300 K. Since the magnitudes of the internal friction and the pseudoelastic hysteresis exhibit similar dependence on the solute concentration and the degree of deformation, the internal friction is ascribed to the motion of twin boundaries or twinning dislocations.

In the present work, we have studied the anelastic effect in more detail for Cu–8at.%Ge alloy. First, we report the variations in the internal friction with the crystal orientation and the degree of deformation, which have been investigated by using single-crystal specimens. Second, we discuss the mechanism of the decay of the internal friction above room temperature on the basis of the results of isothermal annealing and electron irradiation experiments.

2. Experimental procedure

Cu–8at.%Ge alloy was prepared by the same procedure as in the previous work [3]. Polycrystalline wires of 1 mm in diameter and 60 mm in length were used for internal friction measurements. Single crystals of the same size were grown by the Bridgman method in a graphite mould with a growth rate of 30 mm h^{-1}. Specimens were deformed either in tension or in torsion at room temperature. Internal friction was measured in an inverted torsion pendulum apparatus with vibrational frequencies of about 1 Hz, in the temperature range from 80 K to 400 K at a heating rate of 1 K min^{-1}.

Electron irradiation was performed by using the linear accelerator at the Research Institute for Advanced Science and Technology, University of Osaka Prefecture. Polycrystalline specimens were irradiated with 200 MeV electrons up to a dose of 1.2×10^{12} m^{-2}, with the temperature being kept below 100 K. The concentration of vacancies introduced by the irradiation is estimated to be of the order of 10^{-6}. After the irradiation the specimens were stored in liquid nitrogen until internal friction measurements were made.

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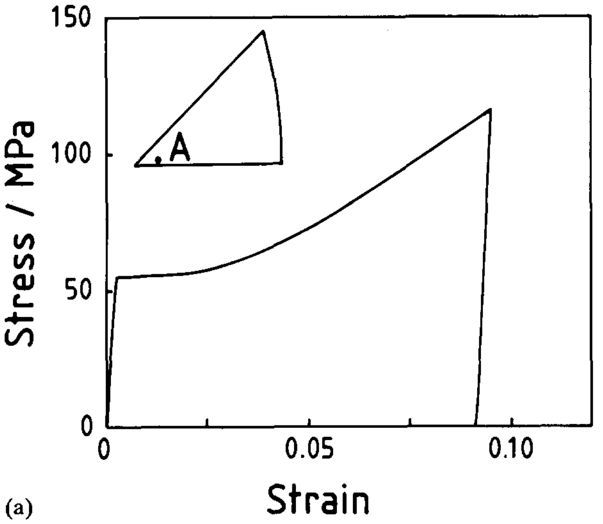
3. Results and discussion

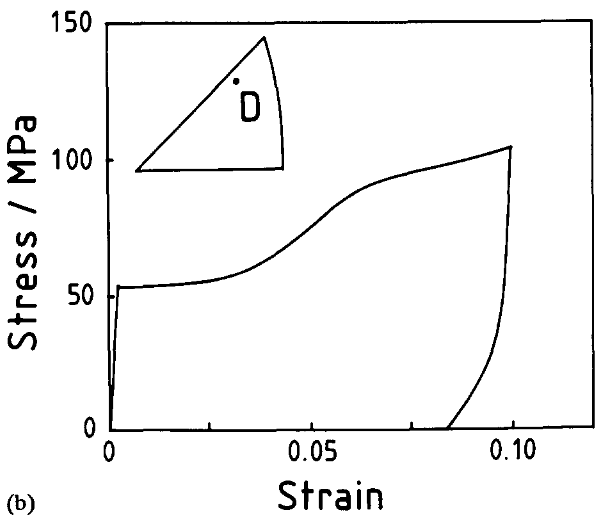
3.1. Pseudoelasticity and anelasticity of single crystals

The stress–strain curves of single crystals A and D under tensile deformation are shown in Fig. 1. The crystal axis is close to [100] for A, and to [111] for D, as shown in the stereographic triangle. It should be noted that the unloading part is almost straight for A but is curved for D. The magnitude of the strain recovered by the extra contraction (the pseudoelastic strain) is of the order of 0.01 in the latter.

Figure 2 shows the internal friction and the vibrational frequency of five single crystals, A to E, deformed by 8% in tension. The square of the vibrational frequency, which is proportional to the shear modulus, is normalized to the value at 100 K. The orientation varies from [100] to [111] in the order from A to E. The internal friction *vs.* temperature curves of crystals C and D are nearly the same as that of E, and are omitted for clarity. Both the internal friction and the modulus anomaly (the upward variation associated with the internal friction maximum) of crystal A are similar to those observed for polycrystals of the same alloy [3]. One can see that the magnitude of the internal friction is decreased as the crystal axis approaches the [111] direction.



(a)



(b)

Fig. 1. The resolved shear stress–resolved shear strain curves of Cu–8at.%Ge single crystals under tensile testing at room temperature.

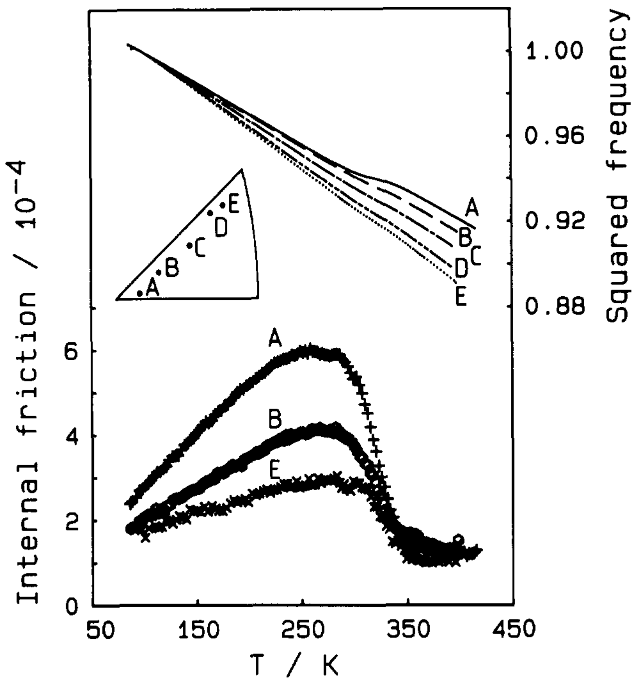


Fig. 2. The internal friction (lower) and the square of the vibrational frequency (normalized at 100 K, upper) of single crystals deformed by 8% in tension.

As exemplified in Fig. 1 and reported by Narita *et al.* [2], the pseudoelastic effect is more pronounced for tensile axes closer to the [111]–[110] line in the stereographic triangle. The orientation dependence of the internal friction shown in Fig. 2 is opposite to this trend. It should be noted that deformation was made in tension and internal friction was measured in torsion. Since the orientations of the specimens are different, the effective shear stress acting on the relevant slip or twinning system is also different. Nevertheless, a simple calculation shows that the variation in the average stress amplitudes on the primary twinning system with the orientation is less than 10% between [100] and [111]. Moreover, this internal friction is known to be independent of the strain amplitude in the range of amplitudes employed in the present measurements, namely below 10^{-5} [4]. Therefore, the dependence of the magnitude of the internal friction on the crystal orientation is a real effect.

Next, internal friction was measured as a function of the amount of deformation on a crystal whose axis lies midway between [111] and [110]. Tension tests and internal friction measurements were repeated by increasing the cumulative plastic strain to 2%, 4%, 8% and 11%. The stress–strain curve is similar to that of Fig. 1(b). The pseudoelastic strain, defined as the width of the hysteresis loop in the stress–strain curve, and the height of the internal friction maximum at each stage of deformation are plotted against the amount of deformation in Fig. 3. One can see that the internal friction rises faster than the pseudoelastic hysteresis.

In the previous investigation on polycrystalline specimens, it was found that the variation in the magnitude of the pseudoelastic hysteresis and that of the internal friction with the solute concentration and the degree of deformation were similar to each other. In contrast, the magnitudes of the two effects depend differently on the degree of deformation for the single crystals of the Cu–8%Ge alloy (Fig. 3). Furthermore, Fig. 2 indicates that the amount of twins or twin boundaries contributing to the pseudoelastic strain is not related directly to the magnitude of the internal friction. From these results obtained for the single crystals, we conclude that the mechanism of the internal friction is not identical to that of the pseudoelastic hysteresis. While the pseudoelastic effect is considered to arise from collective, long-range motion of twinning dislocations, the internal friction is probably due to the local oscillatory motion of individual partial dislocations.

3.2. On the decay of the internal friction above room temperature

As reported in the previous paper [3], the internal friction maximum is not a relaxation peak but is an apparent peak due to the "annealing effect" above room temperature; the internal friction is very low and flat between 80 K and 400 K for a second run measurement following a first run immediately after deformation. The fall in the internal friction in the first run is accompanied by a modulus hardening and self-twisting of the wire specimen [3]. These features suggest that the internal friction is diminished because recovery of deformation substructures begins to occur at room temperature. However, it was also found that the defects responsible for the internal friction are not annihilated but are only immobilized temporarily; the internal friction reappears by giving a light deformation (see Fig. 14 of ref. 3). In the present work we have made some further experiments to obtain a better understanding of this phenomenon.

Figure 4 shows the decay of internal friction and the evolution of the modulus (squared frequency) during isothermal annealing at 290 K, 300 K and 310 K. Each specimen was deformed by 10% in torsion at room temperature, and internal friction was measured at each temperature. In deformed pure metal specimens, a similar recovery of internal friction and modulus is observed, which is sometimes called the Köster effect [5]. Granato *et al.* [6] proposed a theory of this phenomenon which assumes that the time variation results from pinning of dislocations by point defects produced by deformation. Although the detailed mechanism of the internal friction is not known, the magnitude of the anelastic strain due to dislocations is strongly dependent on the loop length of a dislocation line. Here, we shall tentatively apply the vibrating string model [5] in the analysis of the decay, although the applicability of such a model is somewhat questionable for a low frequency of about 1 Hz.

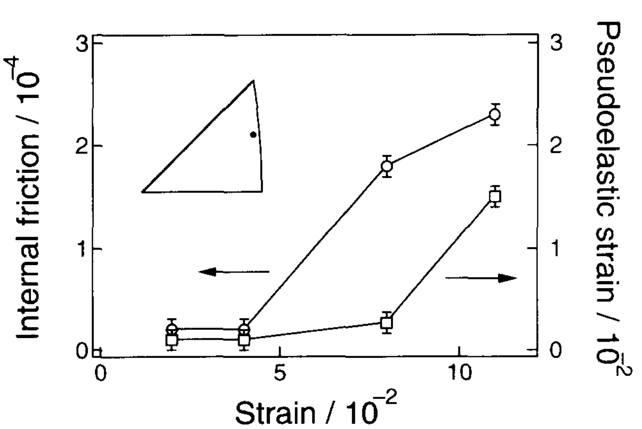


Fig. 3. The variation with the amount of plastic deformation in the pseudoelastic strain under tensile deformation (□) and the magnitude of the internal friction (○) of a single-crystal specimen.

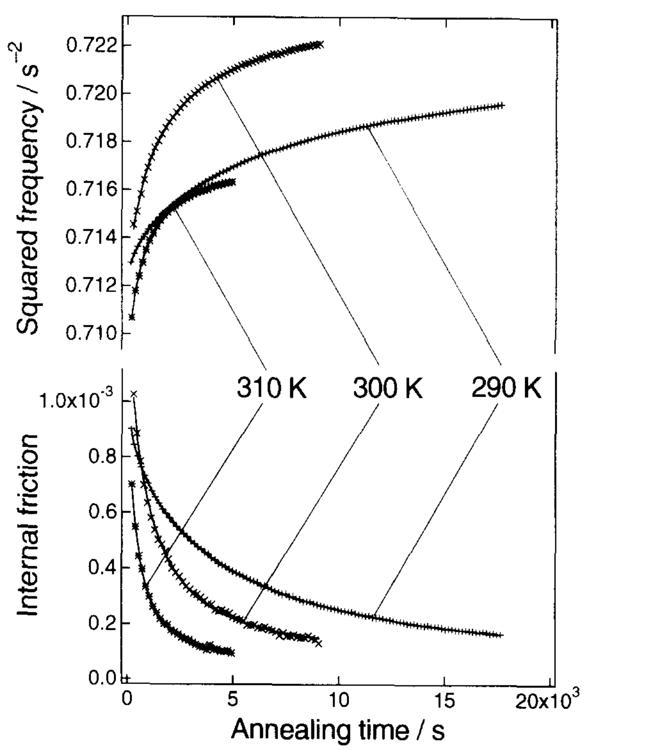


Fig. 4. The decay of the internal friction and the evolution of the modulus (squared frequency) during isothermal annealing. The specimens are polycrystals deformed by 10% in torsion.

According to the theory of Granato *et al.*, the dependence on time of the internal friction Q^{-1} and that of the relative change $\Delta M/M$ in modulus are given by

$$Q^{-1}=a(1+\beta t^{2/3})^{-4} \quad (1)$$

$$\Delta M/M=b(1+\beta t^{2/3})^{-2} \quad (2)$$

where a, b and β are coefficients depending on the dislocation density, the concentration of point defects etc. The rate of recovery is determined by β, which involves the diffusion coefficient of the point defects, thereby possessing a temperature dependence of the Arrhenius type. The values of β for the three temperatures are obtained by fitting eqn. (1) to each experimental decay curve of the internal friction. The variation in β with temperature is found to obey the Arrhenius law, and the activation energy for the diffusion of the point defects responsible for the pinning is calculated to be 0.8 ± 0.1 eV from the temperature dependence of β. This value is reasonably close to the vacancy migration energy in pure copper, 0.71 eV [7]. The activation energy can be obtained independently from the modulus evolution by using eqn. (2); this gives the same value as that obtained from the decay of the internal friction.

The effect of electron irradiation on the internal friction is shown in Fig. 5. Before irradiation the specimen had been deformed by 10% in torsion at room temperature and cooled rapidly to 80 K, and internal friction was measured up to 280 K (curve 1). The internal friction is suppressed by the irradiation almost completely (curve 2). This result is naturally explained by pinning of dislocations by point defects created by the irradiation. After this measurement, the specimen was cooled again and deformed *in situ* (strain, about 1%). Curve 3 shows the internal friction after the light deformation; the reappearance demonstrates that the dislocations have been unpinned to contribute to internal friction again. A small peak is found at about 380 K in curves 2 and 3, but was not studied in detail.

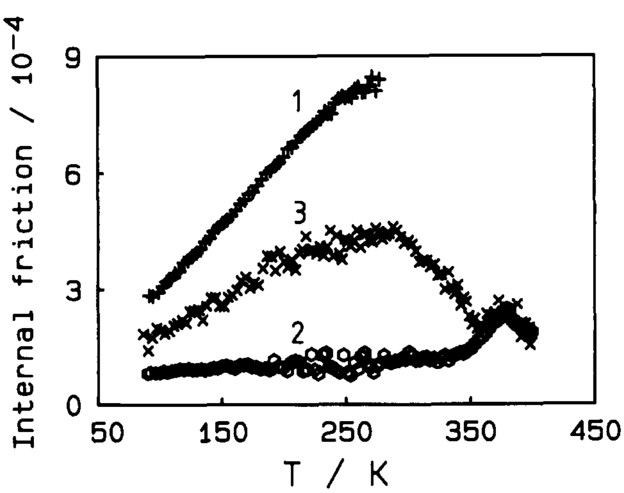


Fig. 5. Effect of electron irradiation on the internal friction of a polycrystalline specimen: curve 1, immediately after deformation by 10% in torsion at room temperature; curve 2, after irradiation; curve 3, after additional deformation (strain, about 1%).

We discussed in the previous paper that rearrangement of dislocations into more stable configurations was also a possible origin of the decay of the internal friction; the spontaneous twisting accompanying the decay suggested that such a process was taking place. Although this still remains a possibility, the results described in this section and particularly the value of the activation energy seem to support the interpretation that the decay is caused by pinning of dislocations by the migration of point defects created during plastic deformation.

4. Conclusions

The dependence of the internal friction on the crystal orientation and the degree of plastic deformation has been studied by using single crystals of a Cu–8at.%Ge alloy. The results suggest that the mechanism of the internal friction is local motion of individual dislocations. The internal friction is suppressed completely by electron irradiation through the dislocation pinning by point defects. The activation energy of the recovery (the decay of the internal friction above room temperature) has been found to be 0.8 ± 0.1 eV, which is close to the vacancy migration energy in copper. The recovery effect is also attributable to the pinning of dislocations by point defects.

Acknowledgment

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