RESISTANCE TO DISLOCATION MOTION IN COPPER-BASED SOLID SOLUTIONS WITH A STATISTICALLY DISORDERED ATOMIC DISTRIBUTION

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The stress σ_F due to friction forces in copper-based solid solutions was determined. Under the conditions of the procedure used to measure σ_F, on the basis of the half-wave hysteresis with polycrystalline samples, the value of d: $\sigma_F = \sigma_{F_0} + K_F d^{-(1/2)}$, where σ_{F_0} is the resistance to dislocation motion in an alloy having an infinite grain size, and K_F is a constant. It is shown that σ_{F_0} is governed by the interaction of moving dislocations with impurity atoms in the case of a statistically disordered atomic distribution. A study was made of the effects of various factors on σ_F and of the nature of the changes in σ_{F_0} caused by alloying.

1. The resistance to dislocation motion (τ_S) is known to be higher in alloys than in pure metals. The contribution of various factors to the increase in τ_S during the alloying is governed by the nature of the alloying elements, the alloy composition, and the alloy state.

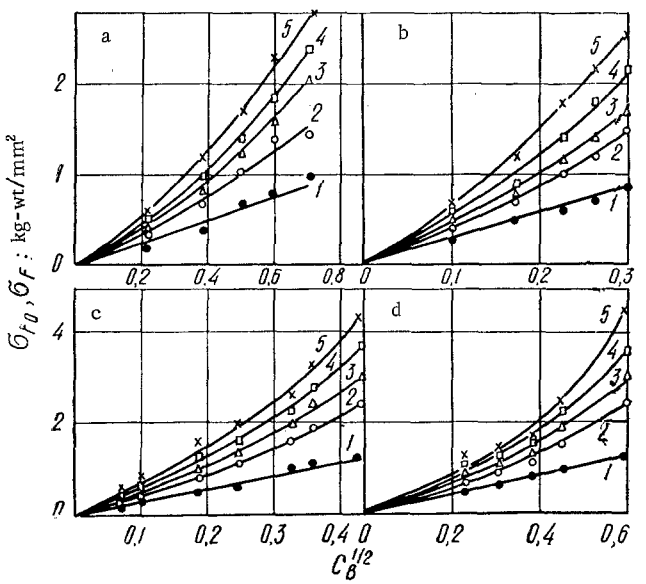

Fig. 1

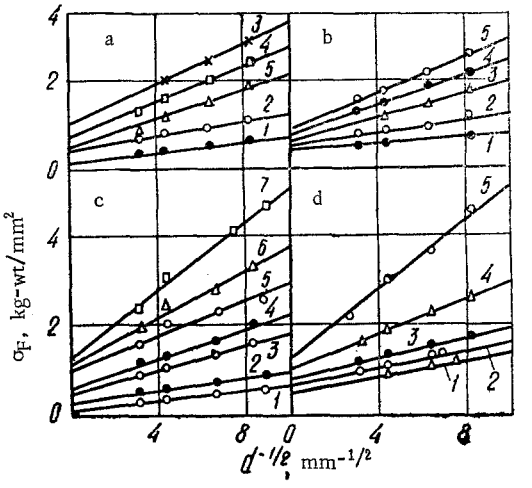

Fig. 2

Fig. 1. Concentration dependences of σ_F and σ_{F_0} on the following solid solutions: a) Cu–Ni; b) Cu–Ge; c) Cu–Al; d) Cu–Zn. 1) σ_{F_0}; 2) grain diameter of 100 μ; 3) 55 μ; 4) 25 μ; 5) 13-15 μ.

Fig. 2. Dependence of σ_F on the grain size in copper-based alloys. a: Cu–Ni. 1) 5.0; 2) 15.3; 3) 50.0; 4) 60.0; 5) 83.0 at.% Ni. b: Cu–Ge. 1) 1.0; 2) 3.0; 3) 5.0; 4) 7.0; 5) 9.0 at.% Ge. c: Cu–Al. 1) 0.5; 2) 1.1; 3) 3.5; 4) 6.0; 5) 10.5; 6) 12.5; 7) 18.0 at.% Al. d: Cu–Zn. 1) 5.0; 2) 10.0; 3) 15.0; 4) 20.0; 5) 25.0 at.% Zn.

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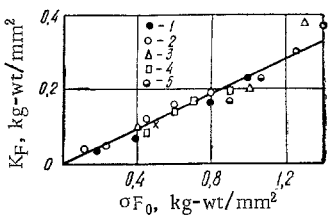

Fig. 3. Dependence of the parameter K_F on σ_{F_0}: 1) Cu–Ni; 2) Cu–Al; 3) Cu–Zn; 4) Cu–Ge; 5) Cu–Mn.

In solid solutions having a statistically disordered atomic distribution the increase in the resistance to dislocation motion (τ_f) which occurs upon alloying is due, according to current theory (see, e.g., [1, 2]), to the interaction of moving dislocations with atoms of the alloying elements. The functional relationship between τ_f, the concentration of the dissolved elements, and the atomic properties of the alloyed components has been calculated only for binary alloys; different theories have predicted different relationships. Mott and Nabarro [1] found a linear increase in τ_f with increasing concentration of the solid solution (c_B), while according to the theory of Fleischer [2] the quantity τ_f is proportional $c_B^{1/2}$.

We report here an experimental study of the effect of alloying on the resistance to dislocation motion in copper-based solid solutions for the case of a statistically disordered atomic distribution.

The Mott–Nabarro theories are usually supported by experimental data on the changes in the yield point of single crystals and polycrystalline samples which occur during alloying. However, these changes are due not only to the resistance to dislocation motion, but also to the strain hardening which occurs during the attainment of the macroscopic yield point (see, e.g., [3-5]). Nor is it always possible to compare the theoretical values of τ_f with the theoretical resistance to the beginning of plastic deformation (in particular, because of the short-range order in alloys). In order to study the effect of alloying on the value of τ_f in any solid solution, we used a method of determining τ_f which is based on the measurement of the friction force of the moving dislocation.

2. The resistance to dislocation motion which results from the interaction of the dislocation with atoms of the alloying elements should be the same in the forward and reverse directions. Accordingly, this resistance enters the stress τ_F due to friction forces (which includes all forces which change sign when the dislocation reverses direction). The value of τ_F, on the other hand, can be determined quite simply, e.g., from half-wave hysteresis, from the equation

$$\tau_F = \frac{1}{2}(\tau_1 - \tau_2), \tag{1}$$

where τ_1 is the stress at the instant at which the load is removed, and τ_2 is the stress at which reverse plastic deformation begins during the unloading.

From the half-wave hysteresis we can determine the stress σ_F due to friction of the moving dislocation in polycrystalline materials also. In this case, however, as will be shown below, the value of σ_F should depend on the grain size d because the deformation of one grain must match that of another.

When we apply an external stress σ_1, and the deformation is transferred from grain to grain, the position of any dislocation in a planar accumulation can be described by

$$\sigma_1 = \sigma_{F_0} + \sigma_K + D\sigma_{F_0} d^{-1/2}, \tag{2}$$

where σ_{F_0} and σ_K are the resistance to dislocation due to friction forces and to conservative forces, respectively, $D\sigma_{F_0} d^{-(1/2)}$ is the resistance to dislocation motion due to the grain boundaries, and D is a constant.

During unloading, the reverse motion of the dislocation begins when the external stress falls to a value

$$\sigma_2 = \sigma_K - \sigma_{F_0} - D\sigma_{F_0} d^{-1/2}. \tag{3}$$

It follows from Eqs. (2) and (3) that we have

$$\sigma_F = \frac{1}{2}(\sigma_1 - \sigma_2), \tag{4}$$

where

$$\sigma_F = \sigma_{F_0} + D\sigma_{F_0} d^{-1/2} = \sigma_{F_0} + K_F d^{-1/2}. \tag{5}$$

At low deformation temperatures the short-range order and the dislocation blocking should not contribute to τ_F or σ_F, for two reasons: first, in order to determine τ_F and σ_F we record the stress at the beginning of the dislocation motion in the direction opposite the motion during loading, i.e., in a plane in which

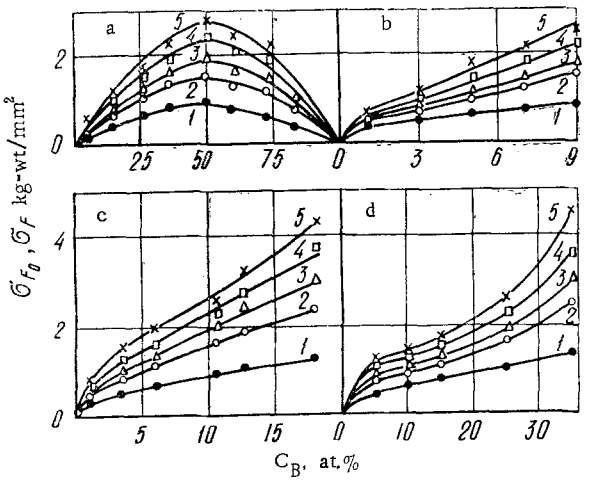

Fig.4. Dependences of σ_F and σ_{F_0} on $c_B^{1/2}$. The notation is the same as in Fig.1.

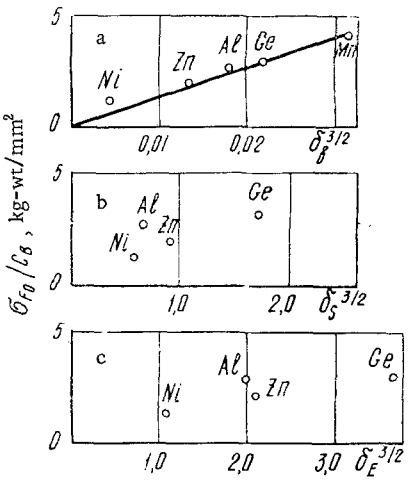

Fig.5. Relation between $\sigma_{F_0}/c_B^{1/2}$ and the parameters $\delta_b^{3/2}$ (a), $\delta_s^{3/2}$ (b), and $\delta_E^{3/2}$ (c).

the short-range order has been destroyed. Second, at low temperatures diffusive harding cannot occur. Under these conditions τ_F and σ_{F_0} thus constitute the resistance to dislocation motion in the case of a statistically disordered atomic distribution. Since τ_F and σ_{F_0} in pure metals are several orders of magnitude below the values in alloys, we can assume that τ_F and σ_{F_0} are due to the interaction of a moving dislocation with atoms of the alloying element.

3. We studied the solid substitutional solutions Cu–Ni, Cu–Al, Cu–Ge, Cu–Zn, and Cu–Mn having various grain sizes. The stress σ_F was determined from the half-wave hysteresis on the basis of Eq. (4). The samples were deformed by extension at room temperature at a rate of 10^{-5} sec^{-1}. The microscopic deformation was measured within $2 \cdot 10^{-6}$ by strain gauges. Stress σ_F was measured after a deformation of $\varepsilon \leq 10^{-3}$, since only up to $\varepsilon = 10^{-3}$ is σ_F independent of the degree of deformation; i.e., here, it is only in this range that the dislocations which arise during the deformation do not contribute to σ_F.

Figure 1 shows the changes observed in σ_F when copper is alloyed with nickel, zinc, aluminum, and germanium; we see that σ_F depends not only on the concentration of the alloying element but also on the grain size. The dependence of σ_F on d is described by $\sigma_F = \sigma_{F_0} + K_F d^{-(1/2)}$, where the parameter K_F depends linearly on σ_{F_0} (Fig.2). For all the alloys the experimental points conform to a common straight line in the plot of K_F vs σ_{F_0} (Fig.3). These results confirm the arguments based on the model for determining σ_F.

It should also be noted that in the alloys having a poorly defined short-range order the stress τ_f is equal to the resistance to the beginning of plastic deformation.

Accordingly, we can compare the theoretical values of τ_f with the stress σ_{F_0} due to the interaction of a moving dislocation with atoms of the alloying elements.

4. Figure 1 (curves 1) show the changes in σ_F when the copper is alloyed with various elements; we see from Fig.1 that the composition dependence of σ_{F_0} is nonlinear, i.e., cannot be described by the Mott–Nabarro theory. On the other hand, in a plot of σ_{F_0} vs $c_B^{1/2}$ this dependence turns out to be linear (Fig.4, curves 1). In other words, the changes in σ_{F_0} resulting from the alloying are qualitatively those predicted by the Fleischer theory, according to which we have

$$\tau_f = \frac{G\delta^{3/2}}{760} c_B^{1/2}, \qquad (6)$$

where $\delta = |[\delta_G/(1 + (|\delta_G|/2)] - \alpha\delta_b|$; $\delta_b = (1/a)(da/dc_B)$; $\delta_G = (1/G)(dG/dc_B)$; $\alpha > 16$ for an edge dislocation, $\alpha < 16$ for a screw dislocation, a is the lattice constant, and G is the shear modulus.

Figure 5 shows $\sigma_{F_0}/c_B^{1/2}$ as a function of the parameters $\delta_b^{3/2}$, $\delta_s^{3/2} = |[\delta_G/(i + (|\delta_G|/2)] - 3\delta_b|^{3/2}$, (the extreme values for screw dislocations), and $\delta_E^{3/2} = |[\delta_G/(1 + (|\delta_G|/2)] - 16\delta_b|^{3/2}$ (extreme values for edge

dislocations). In plots of $\sigma_{F_0}/c_B^{1/2}$ vs $\delta_s^{3/2}$ and $\sigma_{F_0}/c_B^{1/2}$ vs $\delta_E^{3/2}$ we do not find a common dependence. A good correspondence is observed only with the parameter $\delta_b^{3/2}$. Accordingly, the quantity σ_{F_0} is governed by the interaction of the stress field of moving dislocations with the distortion fields around dissolved atoms.

On the other hand, there is a good correlation between $\tau/c_B^{1/2}$ and the parameter δ_s for the macroscopic yield point τ of polycrystalline solid substitutional solutions based on copper and silver (and their single crystals) [2]. It was concluded in [2] on the basis of these results that the increase in the resistance to dislocation motion observed during alloying in these alloys is due not only to atomic mismatch (δ_b) but also to the mismatch between the elastic moduli of the components (δ_G). The difference observed between the descriptions of the macroscopic yield point and the resistance to dislocation motion (σ_{F_0}) can apparently be explained by assuming that the former is governed by the stress required for the propagation of a Luders–Chernov band front rather than the motion of an individual dislocation.

CONCLUSIONS

1. The stress σ_F due to friction forces has been determined in several copper-based solid solutions. Under the conditions of the procedure used to measure σ_F, on the basis of the half-wave hysteresis, the value of σ_F depends on the grain size d in a manner described by the Petch equation.

2. The stress σ_{F_0} due to friction forces in alloys having an infinite grain size is due to the interaction to moving dislocations with atoms of the alloying elements.

3. The value of σ_{F_0} changes in proportion to the square root of the concentration of the alloying element and is governed by the interaction of moving dislocations with the distortion field near the dissolved atoms.

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