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EFFECT OF STACKING-FAULT ENERGY ON THE DEVELOPMENT OF A DISLOCATION SUBSTRUCTURE, STRAIN HARDENING, AND PLASTICITY OF fcc SOLID SOLUTIONS

E. F. Dudarev, L. A. Kornienko, and G. P. Bakach

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INTRODUCTION

The stacking-fault energy γ is an important characteristic of fcc metals and solid solutions. Since it determines the splitting of a complete dislocation into partial dislocations, it plays a major role in processes associated with the interaction, cross slip, interaction, and climb of dislocations, i.e., it controls the development of the dislocation structure. The difference in the dislocation structure due to the different splitting of dislocations should manifest itself in the behavior of fcc metals and alloys under plastic deformation. The role of an individual dislocation as a carrier of plastic deformation can change during plastic flow with the appearance of collective modes of deformation. As will be shown below, however, even in the case of deep strains the flow stress depends on γ. The effect of other parameters of the alloy (degree of order, grain boundaries, their extent, etc.), it is true, are superimposed on this dependence. To a considerable extent the stacking-fault energy also determines the plasticity of fcc alloys at low deformation temperatures.

In this article we correlate the results of our own studies and data published in the literature on the formation of a dislocation structure in fcc solid solutions with different

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V. D. Kuznetsov Siberian Physicotechnical Institute, Tomsk State University. Translated from Izvestiya Vysshikh Uchebnykh Zavedenii, Fizika, No. 3, pp. 35-46, March, 1991.

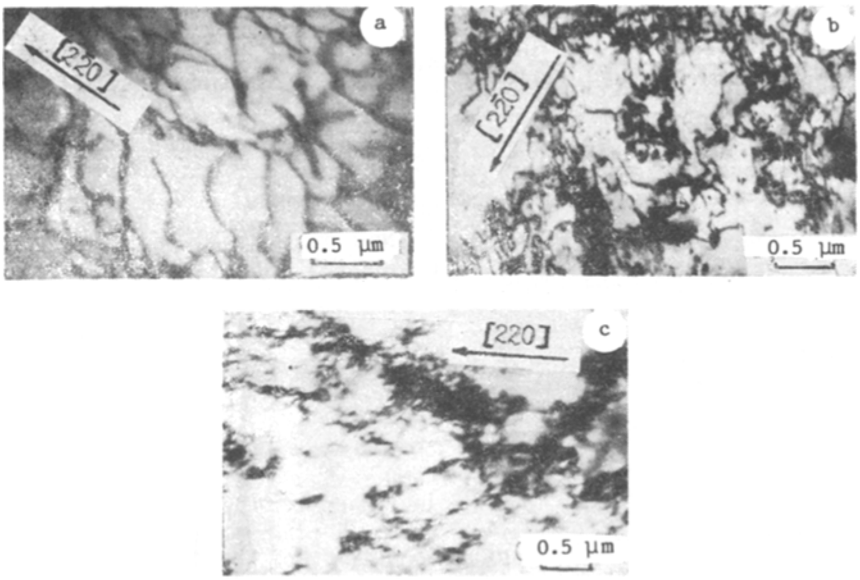


Fig. 1. Dislocation structure in single crystals of Cu–2.8 at.% Al alloy ([100] orientation) at tensile deformations of 10 (a), 30 (b), and 60% (c).

stacking-fault energies and consider the effect of γ on the strain hardening and plasticity of the given class of alloys. Considerable attention is focused on the role of grain boundaries in the formation of a dislocation substructure and strain hardening.

1. DEVELOPMENT OF A DISLOCATION SUBSTRUCTURE IN SINGLE CRYSTALS AND POLYCRYSTALS WITH VARIOUS STACKING-FAULT ENERGIES

Studies carried out to date make it possible to put together a fairly complete picture of the evolution of dislocation structures in fcc metals and alloys at deformation temperatures $T < 0.4\ T_m$ (T_m is the melting point). A uniform dislocation distribution is observed in single crystals with a high or intermediate stacking-fault energy regardless of their orientation in the initial stages of plastic deformation (Fig. 1a) [1]. An inhomogeneous dislocation distribution, when the fields of individual groups of dislocations are compensated to the greatest degree, becomes energetically advantageous when the dislocation density increases [2]. This results in the formation of chaotically distributed dislocation tangles (Fig. 1b), in which the dislocation density gradually rises. Cell walls form as the deformation proceeds (Fig. 1c). The dislocation density of both signs is the same in the cell boundaries and the cells are not disoriented.

The cellular structure develops during the deformation, as manifested in a change in its parameters (cell size, boundary width, and dislocation density at the boundary and inside the cell). It was found theoretically, by minimizing the total dislocation energy [3], and confirmed experimentally [4] that the average cell size d_c is inversely proportional to the shear stress

$$d_c = \frac{KGb}{\tau - \tau_0}, \tag{1}$$

where K is a constant, G is the shear modulus, b is the Burgers vector, and τ_0 is the flow stress.

Since the flow stress τ depends on the average dislocation density ρ as

$$\tau = \tau_0 + \alpha G b \rho^{1/2}, \tag{2}$$

the dependence of d_c on ρ is obtained in the form (α is a parameter characterized by the interdislocation interaction)

$$d_c = K\alpha^{-1}\rho^{-1/2}, \tag{3}$$

i.e., the cell size is determined by the average dislocation density. It decreases as the dislocation density increases and then reaches saturation. The limiting cell size is 0.2-0.5 μm [1]. The cells are made smaller by the formation of new boundaries inside the cells [5]. The smaller the cells, the smaller the probability that two or more dislocations, which can be the basis of a new boundary when blocked inside a cell, will meet inside the cells. This, according to Wilsdorf [6], can cause the cells to shrink at a slower rate and then to remain constant in size. The shrinkage of the cells is accompanied by a decrease in their wall width $h_c = cd_c$, where c is a constant [6].

The dislocation substructure evolves further as follows. Against the background of the cellular structure with critical parameters broken dislocation-disclination subboundaries, with disorientation angle 0.5-1° arise in the direction of the shear. The density of these boundaries decreases rapidly with the degree of deformation [1, 7]. In the end a disoriented band substructure is formed (Fig. 2a).

The further evolution of the dislocation substructure was studied systematically upon rolling and compressing single crystals [8]. It turned out that the preferred orientation of boundaries in the substructure is disrupted as the reduction increases. Individual broken subboundaries and dipole configurations of subboundaries become characteristic elements of the substructure: the distance between subboundaries decreases with the degree of plastic deformation. The diffraction contrast undergoes a jump on passing through a subboundary, indicating a discrete disorientation of the subboundary. The reflections in the microdiffraction photograph are split.

The internal structure of the bands is rather involved and evolves during plastic deformation. The existence of extinction contours inside the band is evidence of continuous disorientation and of excess dislocation density in the band [9, 10]. The disorientation angle in the band substructure, as determined by electron-microscopic and local x-ray diffractometric methods [10], is 2-8°.

An increase in the degree of deformation results in curvature of the deformation subboundaries. The band substructure loses its crystallographic directivity (Fig. 2b). According to [8, 11], this may be due to, first, the formation of a two-dimensional disoriented band substructure, which leads to an increase in the long-range stress fields inside the band. Second, it may occur because the long-range stress fields are made more complicated by the appearance of components of the dislocation-density tensor, which determine the transverse curvature of the band. In this case even without a second system of subboundaries the arising stress fields break up the microband into localized and disoriented microsegments. The resulting fragmented substructure is a substructure with multidimensional discrete and continuous disorientations, in which the disorientation angle increases with the plastic deformation, reaching 15-40° [8, 12-14].

Other dislocation substructures form in alloys with low stacking-fault energy ($\gamma < 20$ mJ/m^2) [15, 16]. In the early stages of development of plastic deformation we observe two-dimensional pileups of dislocations of gradually increasing density (Fig. 3a). The dislocations in the pileups are often split. Then a net structure forms with a high dislocation density (Fig. 3b). At first the net structure is uniform and then dislocation condensations appear, forming loose walls. Fine microtwins also appear against the background of this

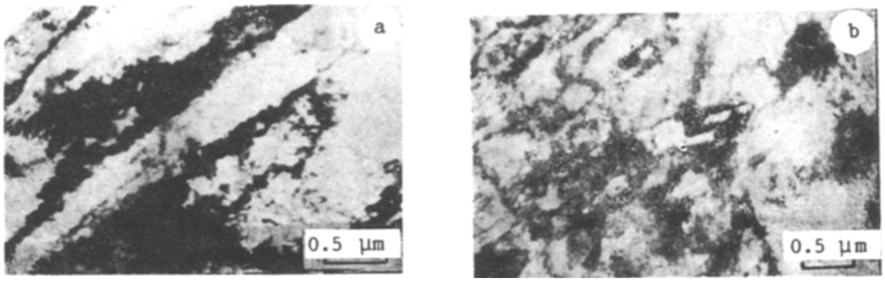


Fig. 2. Dislocation substructure in single crystals of austenitic stainless steel ((100)[100] orientation) at deformations of 65% (a) and 85% (b) due to rolling.

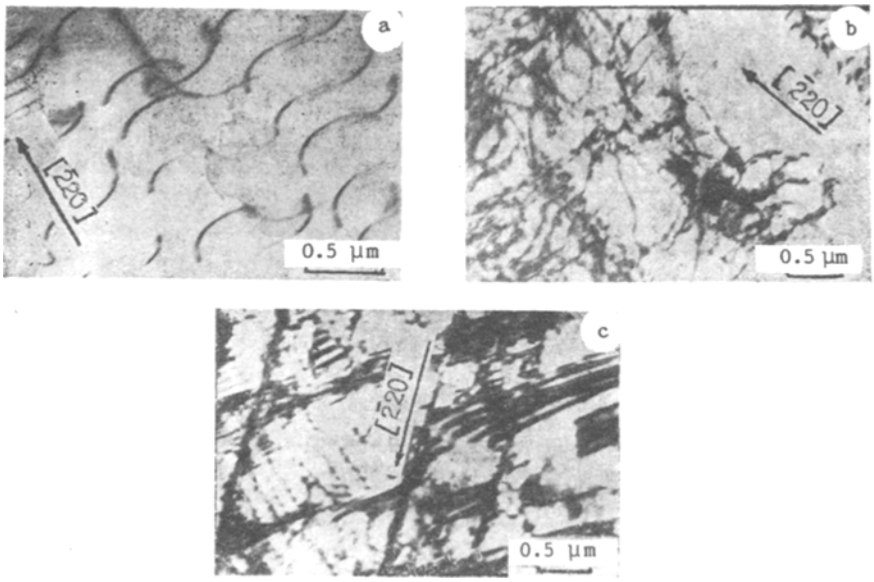


Fig. 3. Dislocation structure in single crystals of the Cu-13.8 at.% Al alloy ([100] orientation) under tensile deformations of 1 (a), 20 (b), and 50% (c).

substructure (Fig. 3c). Under deep deformations we observe disoriented bands with a continuous disorientation (this is indicated by the existence of extinction contours). Only low-angle boundaries are observed until necking begins and the disorientation of the microregions relative to these boundaries is less than 3-5°.

Alloys with intermediate stacking-fault energies ($20 < \gamma < 40$ mJ/m^2) have intermediate substructures which are a mixture of dislocation structures from the first and second groups of alloys [8, 16, 17]. A uniform distribution of split dislocations is observed in these alloys at different stages of plastic deformation and then a tangle structure is formed. A large number of stacking faults appear. But a developed cellular structure, characteristic of alloys with a high γ, does not form. Nor are deformation twins detected. The same substructures as in alloys with $\gamma > 40$ mJ/m^2 form at deep plastic deformations.

At any value of the stacking-fault energy a dislocation substructure in a single crystal oriented for multiple slip develops much more rapidly than when the orientation is for single slip, although consistently the same structures form [1, 8].

The same dislocation substructures form in polycrystals under plastic deformation as do in single crystals [1, 8]. All the substructures, however, begin to form at lower degrees of deformation and develop more rapidly. This is facilitated by the increased dislocation density due to the presence of grain boundaries and the multiple slip that develops from the very onset of deformation [1, 18].

The laws governing the variation of the dislocation substructure parameters with increasing degree of deformation and flow stresses are the same in single crystals and polycrystals. For example, the average cell size d_c in polycrystals and in single crystals decreases at first and then reaches saturation at $d_c \simeq 0.2$ μm (Fig. 4). As the grain size d decreases the dislocation density rises and, therefore, the cell size is reduced (Fig. 4a). The linear relation (1) holds between the flow stress and the reciprocal cell size (Fig. 5). The cell walls become thinner during deformation [14, 19].

The stacking-fault energy has the same effect on the development of the dislocation substructure in single crystals and polycrystals. A decrease in γ hinders the formation of tangle, cellular, band, and fragmented substructures [1, 20]. With decreasing γ the onset of the formation of cells moves into the region of large deformations and their average size grows (Fig. 4b) [20]. At the same time, the limiting cell size does not depend on γ and is roughly 0.2 μm. The effect of γ on the substructure development persists in the stage of deep plastic deformations. A fragmented (fine-grained) structure with large-angle boundaries (Fig. 6a) is formed in high-γ alloys before fracture. In low-γ alloys a fine platen-like twin structure evolves into a band substructure with disorientation angles of up to 30° (Fig. 6b).

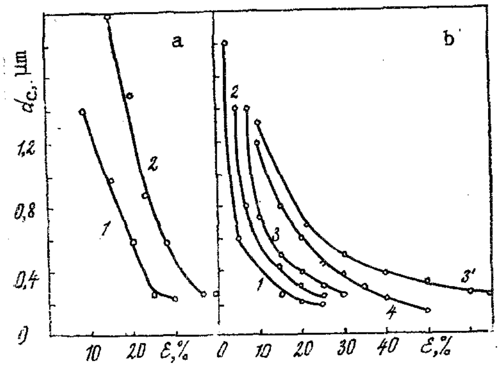


Fig. 4

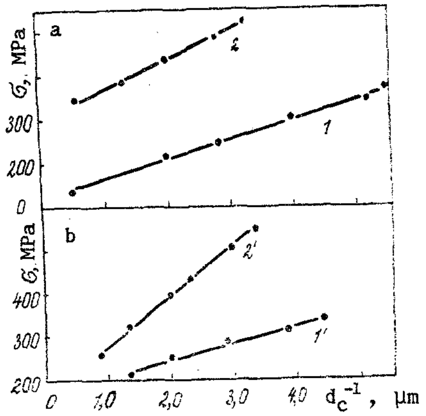


Fig. 5

Fig. 4. Cell diameter d_c versus degree of plastic deformation ε of polycrystals of: a) Cu–12 at.% Zn, d = 15 µm (1) and 100 µm (2), T = 293 K; b) Ni–Al with 2 at.% Al (1) and 8 at.% Al (2), Cu–Al with 2.8 at.% Al (3, 3') and 5.7 at.% Al (4), d = 25 µm, T = 293 K (1, 2, 3, 4) and 77 K (3').

Fig. 5. Applied stresses σ versus d_c in polycrystals of: a) Ni–Al with 2 (1) and 8 at.% (2), T = 293 K; b) Cu–Al with 2.8 at.% Al at 77 K (1) and 293 K (2), d = 25 µm.

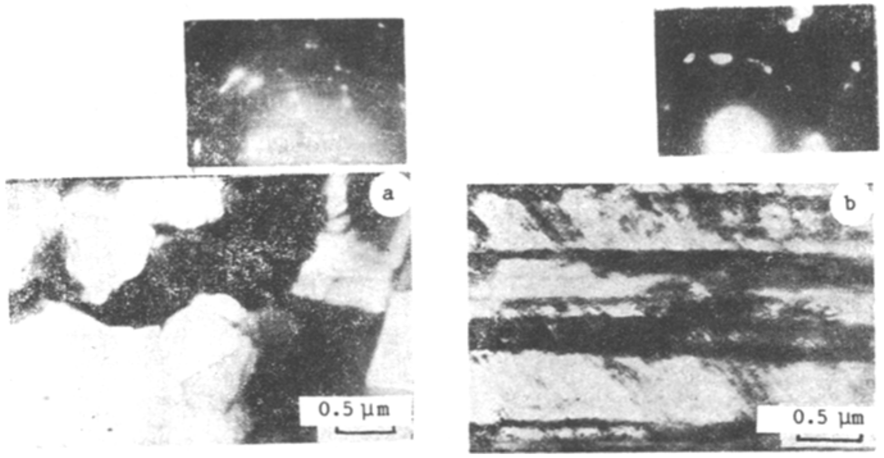


Fig. 6. Substructure in Cu–2.8 at.% (a) and Cu–13.8 at.% Al (b) alloys at ε_{tr} = 1.3.

2. INTERRELATIONSHIP BETWEEN STRAIN HARDENING PARAMETERS AND THE STACKING-FAULT ENERGY

In polycrystals of fcc metals and solid solutions, in contrast to single crystals, regardless of the value of the stacking-fault energy, the "true stress–relative strain" and "true stress–true strain" flow curves, as a rule, do not display distinct stages and the flow curves are smooth and "parabolic" [18, 21]. Nor is it possible to distinguish all stages on the curves of the hardening coefficient versus the degree of deformation and the flow stress.

In order to distinguish the stages in the development of the plastic deformation the flow curves are replotted in different coordinates, so that they are divided into a number of segments and dislocation substructures are assigned to them [7, 13, 17, 19]. One way of isolating the stages, which makes it possible to determine the specific form of the power law of accumulation of plastic deformation $\sigma = \sigma_y + A\varepsilon^n$, i.e., its parameters n and A, is by analyzing the flow curve in log–log coordinates (σ_y is the yield stress and ε is the relative elongation). This method is used extensively for pure metals and alloys [18, 22, 23]. To check that this equation can be satisfied for solid solutions with various γ, elsewhere [18] we replotted the flow curves in the coordinates $\ln(\sigma - \sigma_y)$ vs $\ln\varepsilon$. We found

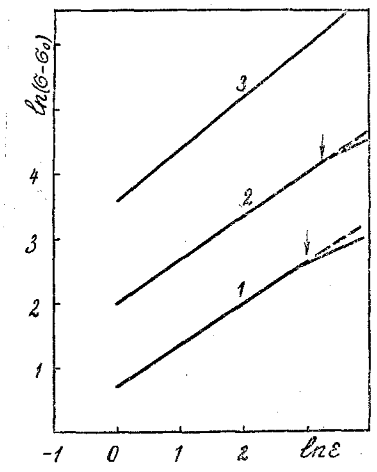


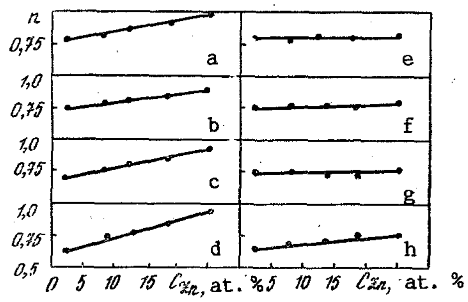
Fig. 7. σ–ε Flow curves of Cu–Zn polycrystals with 3 (1), 12 (2), and 25 at.% Al (3) in log–log coordinates: T = 293 K, d = 25 μm.

that in such coordinates up to the loss of stability against uniform elongation the flow curves consist of one or two linear segments (Fig. 7). Two stages are observed in alloys with high or intermediate stacking-fault energies. With decreasing stacking-fault energy and test temperature the transition to the second stage moves to higher degrees of plastic deformation. In low-γ alloys the flow curve is described by a single power law.

Comparison of the resulting transition points to the second stage of plastic flow with a dislocation substructure showed that when the deformations are close to the transition point the formation of a cellular structure ends, the cell size reaches the minimum value, and deformation subboundaries of the dislocation-disclination type appear. Since the formation of cellular and disoriented substructures is hindered when γ decreases, the appearance of new modes of deformation and, hence, the onset of the second stage moves to larger deformations. When describing the flow curve with the power law $\sigma = \sigma_y + A\varepsilon^n$ we can distinguish the stage associated with the formation of nondisoriented substructures.

In the stage of development of nondisoriented substructures parameters n and A of the power law, which characterize strain hardening, vary as the stacking-fault energy and deformation conditions vary. An interrelationship is observed between the rate of development of nondisoriented substructure and n and A. Regardless of the grain size and the temperature they increase as the stacking-fault energy decreases [24]. As an example Fig. 8 shows data for Cu–Zn alloys in which γ decreases strongly with rising concentration of the alloying element.

The dependence of n on γ becomes stronger when the temperature rises and the grain size increases. As the deformation temperature increases n decreases and its temperature dependence is more pronounced at high stacking-fault energies (Fig. 9).



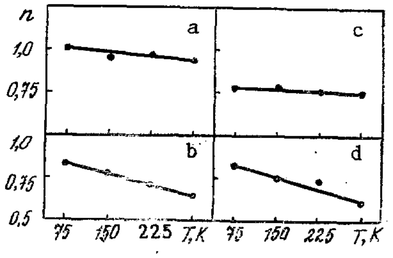


Fig. 8

Fig. 9

Fig. 8. Concentration dependence on the parameter n of Cu–Zn polycrystals; d = 100 μm (a, b, c, d) and 15 μm (e, f, g, h) deformation temperature 77 (a, e), 153 (b, f), 223 (c, g), and 293 K (e, h).

Fig. 9. Temperature dependence of the parameter n of Cu–Zn polycrystals with 25 (a, c) and 3 at.% Zn (b, d), d = 100 (a, b) and 15 μm (c, d).

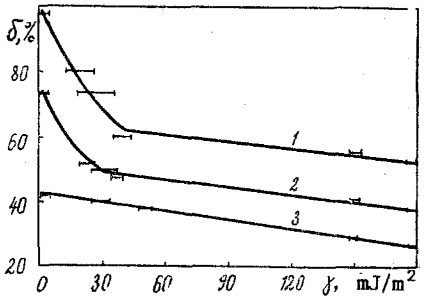


Fig. 10. Uniform elongation δ of polycrystals of fcc Cu–Zn, Cu–Al, and Ni–Al solid solutions versus the stacking-fault energy γ, deformation temperature 77 (1), 150 (2), and 293 K (3); d = 25 μm.

The variation of the strain-hardening parameter A as the stacking-fault energy and deformation temperature vary correlates well with the variation of n: A increases with n and vice versa. The variation of the grain size affects the value of A less than that of n [18].

The ultimate uniform elongation, which characterizes the onset of necking, is one of the flow-curve parameters, which reflect strain hardening at deep plastic deformations, when the role of an individual dislocation and its structure should decrease. The relation between the ultimate uniform elongation and the flow-curve parameters stems from the condition of loss of stability against uniform elongation

$$\sigma = d\sigma/de, \tag{4}$$

where σ is the flow stress, $d\sigma/de$ is the strain-hardening coefficient, and e is the true strain.

In [20] we verified that this condition can be satisfied for polycrystals of solid solutions with various γ. Necking occurs at $\sigma = d\sigma/de$, regardless of the composition of the alloy and the deformation temperature.

Figure 10 shows the dependence of the uniform elongation δ on the stacking-fault energy Cu–Zn, Cu–Al, and Ni–Al alloys at various deformation temperatures. Only a single dependence of δ on γ is observed at each deformation temperature: the ultimate uniform elongation δ decreases with increasing stacking-fault energy. In the region of high and intermediate values of γ ($\gamma > 40$ mJ/m^2) the dependence of δ is roughly the same at all deformation temperatures. In the region of low values of γ (below 40 mJ/m^2) the dependence of δ on the stacking-fault energy is more pronounced when the deformation temperature is lower. These data indicate that under the conditions of development of collective modes of deformation the effect of the structure of an individual dislocation on plastic flow stresses persists.

At the same time variation of the grain size from 15 to 100 μm has virtually no effect on the plasticity characteristics. This accords with the variation of the strain-hardening coefficient as the degree of deformation increases at different grain sizes. The effect of the grain size on the dislocation substructure parameters and, therefore, on the strain-hardening coefficient weakens with growing degree of deformation.

3. EFFECT OF THE STACKING-FAULT ENERGY ON THE GRAIN-BOUNDARY CONTRIBUTION TO THE FLOW STRESS

The contribution of grain boundaries to the flow stress σ of solid solutions with various stacking-fault energies was determined in [25]. This was done by using the Ashby model [26], according to which σ increases with decreasing grain diameter d as a result of the increase in the dislocation density ρ, and the additional dislocation density ρ_G in a polycrystal in comparison with a single crystal is due to the concurrent plastic deformation of grains without violation of continuity along the boundaries. In the model ρ_G is assumed to depend linearly on d^{-1}, which, as was determined in [27], corresponds to experimental data in the stage of formation of nondisoriented substructures. In this stage of deformation the flow stress of a polycrystal is determined by the scalar dislocation density ρ, i.e., $\sigma = \sigma_y + \alpha G b m \rho^{1/2}$, where σ_y is the polycrystal yield stress, α is a parameter that reflects the interdislocation interaction and depends on the nature of the dislocation distribution, G is the shear modulus, b is the Burgers vector, m is the Taylor coefficient, and $\rho = \rho_S + \rho_G$ (ρ_S is the scalar dislocation density in the "equivalent" single crystal).

In the Ashby model the contribution of grain boundaries to the flow stress can be determined if we write flow stress as

$$(\sigma-\sigma_y)^2 = (\alpha G b m \rho_s^{1/2})^2 + (\alpha G b m \rho_G^{1/2})^2 \tag{5}$$

and analyze the dependence of $(\sigma - \sigma_y)^2$ on d^{-1}. In [6] $\sigma_S = \alpha G b m \rho_S^{1/2}$ is the fraction of polycrystal hardening due to dislocation density that would have been accumulated in a corresponding single crystal and $\sigma_G = \alpha G b m \rho_G^{1/2}$ is the fraction of hardening due to the presence of grain boundaries.

Processing of the experimental data in [25] on the strain hardening of polycrystals showed that the linear dependence of $(\sigma - \sigma_y)^2$ on d^{-1} is satisfied as long as the substructure does not contain disorientation or twinning boundaries. Both flow-stress components σ_S and σ_G increase as the substructures develop (Fig. 11). At the outset of plastic flow the grain-boundary contribution σ_G exceeds, or is equal to, the contribution σ_S from randomly accumulated dislocations. The component σ_G grows at a slower rate with the degree of deformation, however, than σ_S does, and this is more pronounced at higher stacking-fault energy. At some degree of deformation, which increases as the stacking-fault energy and grain size decrease and the temperature is lowered, therefore, σ_S and σ_G become equal and then randomly accumulated dislocations begin to make the main contribution to the hardening (Fig. 12). The relative contribution of grain boundaries to the hardening of fcc solid solutions decreases with the development of the substructure, this decrease being greater at a higher stacking-fault energy and deformation temperature.

The absolute contribution of grain boundaries to the flow stress (σ_G) depends on the stacking-fault energy and deformation temperature, besides the degree of deformation and grain size. At all grain sizes and degrees of deformation σ_G increases abruptly when γ and T decrease (Fig. 13). By delaying the formation of a cellular structure, a lowering of the stacking-fault energy and deformation temperature promote an increase in the grain-boundary contribution to the flow stress. These same data indicate that accommodation adjustment of the grains occurs via the motion and multiplication of dislocations in the substructure being formed.

4. PLASTICITY AND DUCTILE FRACTURE

When single crystals and polycrystals undergo tension plastic flow does not end in uniform deformation. From the time when stability against uniform elongation is lost, plastic flow is localized in the neck. Here the dislocation substructure continues to develop and a fracture surface forms at a certain stage in its evolution.

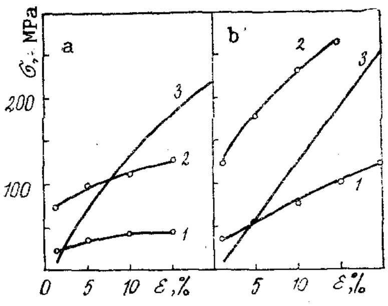


Fig. 11

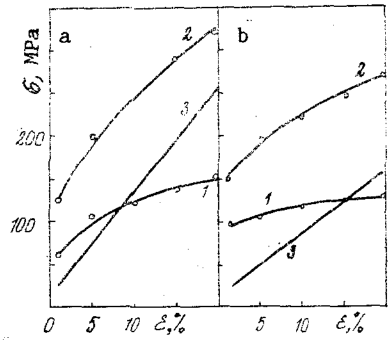


Fig. 12

Fig. 11. Effect of the degree of plastic deformation on the contributions from the geometrically necessary dislocations σ_G (1, 2) and randomly accumulated dislocations σ_S (3) to the flow stress of polycrystals of the alloys Cu–8 at.% Zn (a) and Cu–25 at.% Zn (b); d = 100 (1) and 15 μm (2); T = 293 K.

Fig. 12. Effect of the degree of plastic deformation on the contributions from geometrically necessary dislocations σ_G (1, 2) and randomly accumulated dislocations σ_S (3) to the flow stress of polycrystals of the alloys Cu–25 at deformation temperatures of 77 (a) and 293 K (b); d = 50 (1) d = 50 (1) and 15 μm (2).

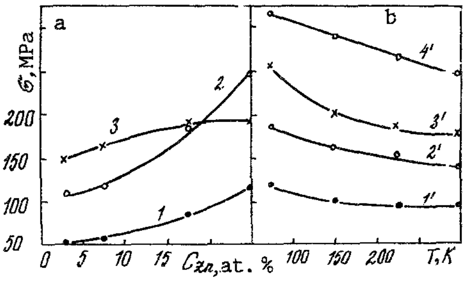


Fig. 13

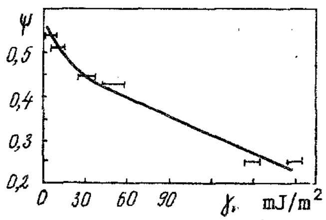


Fig. 14

Fig. 13. Concentration and temperature dependence of the contributions to the flow stress of polycrystals from geometrically necessary dislocations σ_G (1, 2, 4') and randomly accumulated dislocations σ_S (3) in the alloys: a) Cu–Zn with d = 100 (1) and 15 μm (2), T = 77 K; b) Cu–5.7 at.% Al with d = 100 (1'), 20 (2'), and 10 μm (4'), ε = 15%.

Fig. 14. Narrowing ψ in the neck versus the stacking-fault energy, d = 25 μm, T = 77 K.

The local plastic deformation in the neck can be evaluated from the relative narrowing ψ. According to our earlier data [20] the latter depends on the stacking-fault energy in exactly the same way as does uniform elongation, i.e., it decreases with the growth of γ (Fig. 14). An understanding of this effect of γ on ψ was obtained by examining the evolution of the substructure in the neck. Polycrystals of Cu–Al with a grain size of 25 μm were deformed at room temperature.

In the alloy Cu–2.8 at.% Al (γ = 40 mJ/m^2) a cellular structure with critical parameters and containing boundaries with a disorientation angle of 0.5–1° encompassing several cells, is observed just prior to necking. As deformation develops in the neck the number of these boundaries increases, the disorientation angle increases, and a band substructure forms. The crystallographic directivity of the subboundaries vanishes and a fragmented substructure forms with large-angle boundaries. At the same time of fracture the disorientation angles reach 30° with a fragment size of 0.5-1 μm.

In the alloy Cu–13.8 at.% Al the twinning structure formed by the time of loss of mechanical stability against uniform deformation continues to develop in the neck. As the degree of deformation increases band disoriented substructures are formed with discrete and continuous disorientations. The internal structure of the bands consists of twins from different systems. At the time of fracture the disoriented microvolumes (fragments) have a size of roughly 0.5 μm at disorientation angles of up to 40°.

Plastic deformation in the neck, therefore, develops by gradual accumulation of the disorientation of neighboring elements of the substructure, regardless of the stacking-fault energy. The same type of substructure, fragmented with large-angle boundaries, form in alloys with different values of γ by the time fracture occurs.

The degree of preparation for the formation of a fragmented substructure at the time of necking is different in alloys with different values of γ. Since the same substructure forms by the time fracture occurs, the local deformation requiring that substructure to form will be different. The deformation in the neck in this case is larger when the substructure formed by the time the mechanical stability is lost is further along in evolution from the critical (fragmented) substructure. It thus follows that as the stacking-fault energy decreases the relative narrowing in the neck should increase, as is indeed observed experimentally (see Fig. 14).

The presence of grain boundaries due to the additional dislocation density accelerates the formation of each successive dislocation substructure and by the time necking begins the degree of preparation for the formation of a fragmented substructure is higher in polycrystals than in single crystals. And since single crystals and polycrystals fracture when they have the same substructure, the local deformation in the neck is larger in single crystals than in polycrystals.

Since fracture occurs with the same type of substructure, the nature of the fracture is observed to be uniform. Regardless of the preceding evolution of the dislocation structure, it is ductile and transcrystallite, and the fracture surface has a characteristic pitted relief [6, 20]. Depending on the stacking-fault energy and the temperature, only the details of the overall picture of the fracture surface change, because of the morphological features of the fragmented substructure. The minimum size of the fracture elements and the average size of the fragments coincide, however.

Finally, we must point out one more fact. The flow curve in the neck is described by a power law with the same parameters as in the second stage of uniform deformation, associated with the formation of a disoriented substructure.

In conclusion, we note that the specific mechanisms of substructural hardening at different stages of plastic deformation require separate consideration as does the mechanism of ductile fracture along boundaries of fragments.

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