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FORMATION OF A SURFACE-LAYER SUBSTRUCTURE DUE TO FRICTION

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The mechanism by which a fragmented microstructure is formed when there are large degrees of deformation which occur on the surface of a metal during friction, is considered. It is shown that size of the fragments under these conditions is a minimum and is related to the dislocation density, reaching a critical value in an extremely deformed state, above which the interdislocation interaction forces are greater than the internal-friction forces.

Experimental investigations of the processes that occur during friction have shown that a thin surface layer occurs near the friction surface having a structure which differs considerably from the structure of the bulk material. The properties of the surface layer determine the characteristics of the friction and the wear resistance of the material to a considerable extent, and hence a study of the processes that occur in this layer, its structure, and the mechanisms by which it is formed are of paramount importance for understanding the tribotechnical properties of the material as a whole.

The purpose of this paper was to describe experimental and theoretical investigations of the structure of the surface volumes. To study the features of the friction-surface structure, we chose various materials, namely, 15NZMA martensite steel, 36NKhTYu austenite high-strength alloy in two structural states, copper, molybdenum, and 9KhS instrument steel. Structural investigations were carried out by electron microscopy and x-ray diffraction.

The main conclusion of electron-microscope investigations of the structure of the surface layer is that the structure formed during friction, consisting of finely dispersed fragments of material of the specimen with different inclusions of the contrabody material, in the final analysis is independent of its initial state and represents the maximum possible deformed state for the given conditions.

The thickness of the layer of fragmented structures (≈ 20 μm) also depends only slightly on the material investigated and the nature of its preliminary treatment.

The main structural features are common to all the materials investigated. The structure of the fragmented layer is characterized by the presence of strongly azimuthal disorientations and small dimensions of the components. The azimuthal disorientations are a few degrees within the limits of the former grain and reach several tens of degrees on the intergrain or twinning boundary. The dimensions of the regions with uniform orientation (subgrains) are of the order of 0.01-0.1 μm (Fig. 1a, b).

The phase composition of the fragmented layer is also independent of the preliminary treatment of the material. Thus, in austenite alloy a uniform structure is observed, which is the same both for tempered and for aged alloy, with reflections corresponding to austenite (Fig. 1c).

During deformation by friction, texturization of the surface layers occurs. In investigations of the microstructure of undeformed samples of 36NKhTYu alloy the distribution of the orientations of the grains gives almost equal probabilities that the orientations {110}, {211}, {123} will occur and somewhat lower probabilities that {111}, {310} and {100} will occur, but after friction the fragments are oriented so that the probability that the {110} orientation will occur becomes practically equal to unity. This indicates that rotation of the fragments occurs until some of the {110} planes coincide with the friction plane (a similar texture is observed in the case of deformation by rolling [2]).

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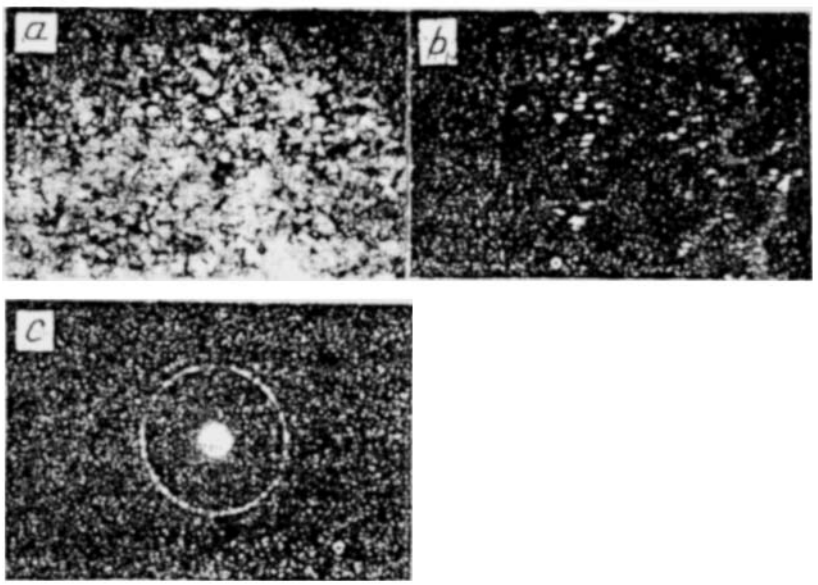

Fig. 1. Microstructure (a) (×44,00), dark-field image (b), and electron diffraction pattern of the surface of 36NKhTYu alloy (c).

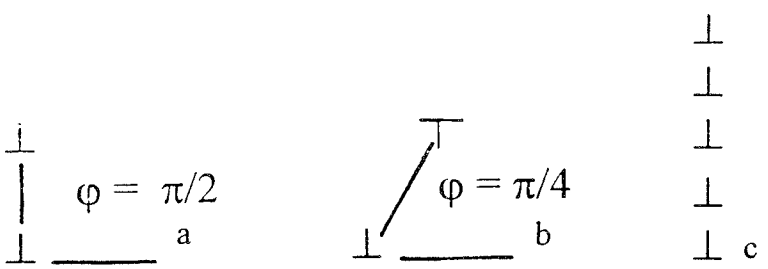

Fig. 2. Equilibrium configurations of edge dislocations with the same Burgers vectors (a) and opposite Burgers vectors (b). The small-angle disorientation boundary formed by dislocations (c).

The formation of a fragmented structure with component dimensions of the order of 0.01-0.1 μm is characteristic for all forms of plastic deformation, be it deformation of the surface layers due to the action of friction [1], rolling [2], deformation by a pressure + shear scheme in a Bridgman chamber [3], or active uniaxial deformation [4]. Any further plastic deformation does not lead to any greater reduction in the size of the substructure, which suggests that there is a characteristic parameter which determines the minimum possible fragment size. We will show below that this length is a fundamental characteristic of a material, defined solely by the lattice parameter, the shear modulus and the so-called friction stress of a nondislocation nature [5], i.e., the stress which must be produced in an ideal crystal in order that the motion of a rectilinear dislocation should begin. The formation of fragmented structures for large plastic deformations was considered in detail for the first time, although from a somewhat different point of view, in [6].

The formation of dislocation substructures begins at the initial stages of plastic deformation at dislocation densities when the mean forces of interaction between neighboring dislocations is less than the internal-friction forces. At these stages the interaction energy of the dislocations has no effect on the formation of dislocation structures. The latter have a purely kinetic nature and will not be considered here. Here, we are concerned with all types of undisoriented substructures including lattice and cell structures according to the classification given in [4]. A review of theoretical models of the formation of such structures can be found in [7].

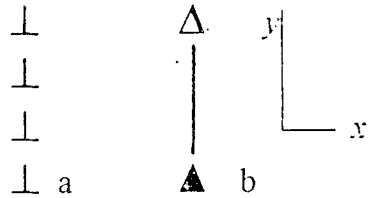

Fig. 3. Small-angle disorientation boundary (a) and its representation in the form of a disclination dipole (b).

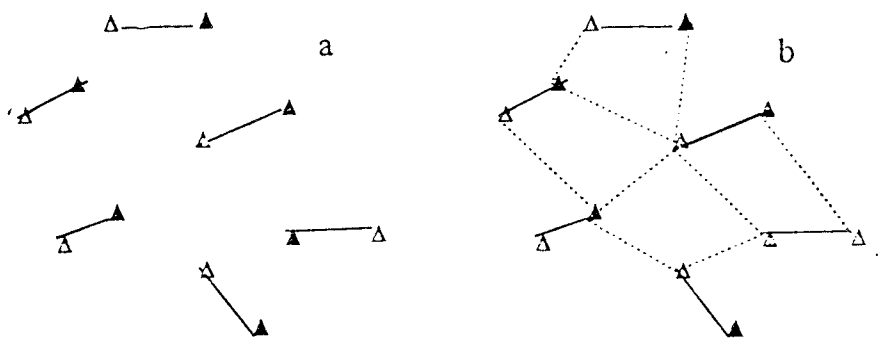

Fig. 4. Disclination structure formed as a result of the development of instability of the dislocation ensemble (a) and its rearrangement into a fragmented structure with discrete disorientations by completion of deficient small angle disorientation boundaries (represented by the dashed lines) (b).

When the dislocation density ρ reaches the critical value for which the interdislocation interaction forces become equal to the friction forces, the latter cease to maintain the dislocations in energetically unsuitable positions, and a dislocation structure begins to be formed so as to minimize the energy of the dislocation ensemble and, consequently, is independent of its prehistory. We will determine the characteristic length parameter which controls this process.

The value of the projection of the interaction forces of parallel dislocations onto their slip plane is equal to [8]

$$f = \frac{\mu b^2}{2\pi r} \cos\varphi \tag{1}$$

for screw dislocations and

$$f = \frac{\mu b^2}{2\pi (1 - \nu) r} \cos\varphi \cos 2\varphi \tag{2}$$

for edge dislocations. In (1) and (2) μ is the shear modulus, b is the value of the Burgers vector of unit dislocation, ν is Poisson's ratio, r is the distance between dislocations, and φ is the angle between the Burgers vector and the radius vector connecting the dislocations. The root mean square values (averaged over the angles) of these forces are equal to $\mu b^2/2\pi r \sqrt{2}$ and $\mu b^2/2\pi(1 - \nu)r$, respectively. For the characteristic value of Poisson's ratio $\nu = 0.3$ we obtain practically the same value of the mean force for both forms of dislocations:

$$\langle f \rangle \approx \frac{\mu b^2}{2\pi r \sqrt{2}}. \tag{3}$$

Spontaneous rearrangement of the dislocation system begins when the following condition is satisfied:

$$\langle f \rangle = \sigma_F b, \tag{4}$$

where σ_F is the friction stress of nondislocation nature [5]. From (3) and (4) we have the following expression for the critical distance between dislocations:

$$r_c^{(1)} = \frac{\mu b}{2\pi\sqrt{2}\sigma_F}. \tag{5}$$

This distance is reached for a mean scalar dislocation density

$$\rho_c^{(1)} = 1/(r_c^{(1)})^2 = \frac{8\pi^2\sigma_F}{\mu^2 b^2}. \tag{6}$$

Unfortunately, we do not have comprehensive information on the nature of the dislocation structure and the stages by which it develops in the alloys in question. Hence, to illustrate our model of the formation of the substructure we will use data for Ni_3Fe alloys, the sequence of changes of the substructures in which have been investigated in detail in [4, 5]. For these alloys in the unordered state we have $\sigma_F = 4 \times 10^7$ Pa [5, p. 182], $\mu = 7.5 \times 10^{10}$ Pa [9], and $b = 3.5 \times 10^{-10}$ m [9], whence

$$\rho_c^{(1)} = 1{,}8 \cdot 10^{10} \text{ cm}^{-2}, \tag{7}$$

which corresponds to a mean distance between dislocations $r = 0.07$ μm.

The nature of the structure formed can be illustrated most simply by using the example of edge dislocations. We know that parallel edge dislocations with equal and opposite Burgers vectors have equilibrium configurations, shown in Fig. 2a and b, respectively. Dislocations of opposite sign, which form dipole configuration (b), are annihilated in time by climb and in this sense do not form a stable structure. The accumulation of edge dislocations of the same sign in configurations of type (a) leads to the formation of "dislocation walls" — stable low-energy configurations (Fig. 2c), which are small-angle disorientation boundaries. Hence, when the dislocation density reaches the critical value (6), a spontaneous redistribution of the dislocations into small-angle disorientation boundaries begins in the material. Note that the estimate (7) for unordered Ni_3Fe alloys is practically identical with the experimentally observed critical dislocation density for which the formation of a substructure with disorientation begins in unordered Ni_3Fe alloys [4, p. 104].

The fundamentally important fact is that, at the first stage of the development of instability of the dislocation system, only "dislocation walls" of finite length, determined by the fluctuation characteristics of the initial dislocation ensemble, can be formed. We know that a semiinfinite dislocation wall can be treated as a partial disclination, whose axis coincides with the wall boundary [10]. Correspondingly, a finite dislocation wall is a disclination dipole (Fig. 3). Hence, the result of the development of instability in the dislocation ensemble is the formation of a disclination ensemble, represented schematically in Fig. 4a. Here, along the line joining the dislocation dipole (the x direction in Fig. 3b) there is a jump in the orientation of the crystal lattice. At the same time, for motion in the y direction the orientation of the lattice changes smoothly. The disclination structure described can consequently be referred to as a "substructure with continuous and discrete disorientations" in accordance with the classification in [4]. This mechanism of the formation of fragmented structures enables us to explain only the formation of boundaries of inclination and, in the case of "broken boundaries," wedge disclinations, since the stability of these boundaries is due to the low energy of the edge dislocations which are aligned in the wall. The formation of boundaries consisting of screw dislocations is, as pointed out in [10], one of the important and unsolved problems of the theory of disclination structures. A key to the solution of this problem is possibly the closed nature of dislocation lines in crystals. Ordering of the edge segments of closed dislocation loops, by virtue of purely geometrical constraints, should inevitably lead to some ordering in their screw segments also.

Since the mobility of disclinations in crystals is extremely small [11], a reduction in the energy of the disclination ensemble can only occur as a result of "completion" of deficient small-angle disorientation boundaries in the structure in Fig. 4; this is represented in Fig. 4b by the dashed lines. Due to the random arrangement of the disclinations, the number of dislocations required for such completion is equal to the number of dislocations which form the initial disclination structure. This indicates that the formation of a fragmented substructure with discrete disorientations is completed when the dislocation density reaches the value

$$\rho_c^{(2)} = 2\rho_c^{(1)}, \tag{8}$$

which is in complete agreement with experimental results [4, p. 104]. Consequently

$$r_c^{(2)} = \frac{\mu b}{4\pi\sigma_F}. \tag{9}$$

The characteristic length (9) determines not only the limiting scale of the fragmented substructure but also the critical dimensions of the small particles, thin layers (including surface layers of the bulk samples) and filamentary crystals, beneath which these objects are dislocation-free. In fact, per unit length of dislocation, situated parallel to the free boundary of the elastic continuum at a distance x_0 from it, there is a force

$$f = \frac{\mu b^2}{4\pi x_0}. \tag{10}$$

If this force is greater than the friction forces (4), the dislocations will be "ejected" from the crystal. The critical-thickness equation obtained from this has the same analytic expression as $r_c^{(2)}$. Small particles and filamentary crystals with dimensions less than $r_c^{(2)}$ should be entirely free of dislocation. As regards thin layers, we can only assert that the latter do not contain dislocations whose lines are parallel to the plane of the layer (or close to this orientation). Dislocations perpendicular to the plane of the layer are stable.

The dimensions of the fragments within the deformed state are of fundamental importance for determining the mechanisms by which deformation and fracture occur for large degrees of deformation. For example, at elevated temperatures in layers with submicron fragments, particularly at large disorientation angles, the initiation of a new plasticity channel is possible, namely, diffusion-controlled processes of grain-boundary slippage, which obviously, in fact, occurs in the surface layers [12]. A justified choice of the size of the fragments is also of importance for computer modelling of materials by the cell-automation method, which is one of the most promising methods of designing materials with specified properties [13].

REFERENCES

1. S. Yu. Tarasov and A. V. Kolubaev, Izv. Vyssl. Uchebn. Zaved. Fiz., No. 8, 9-12 (1991).
2. V. F. Sukhovarov, Discontinuous Separation of Phases in Alloys [in Russian], Nauka, Novosibirsk (1983).
3. V. A. Zhorin, A. V. Nefedov, V. A. Linskii, et al., Dokl. Akad. Nauk SSSR, **256**, No. 3, 598-600 (1981).
4. N. A. Koneva and É. V. Kozlov, Izv. Vyssh. Uchebn. Zaved. Fiz., **33**, No. 2, 89-106 (1990).
5. L. E. Popov, N. A. Koneva, and I. V. Tereshko, Deformation Hardening of Ordered Alloys [in Russian], Metallurgiya, Moscow (1979).
6. V. V. Rybin, Large Plastic Deformation and Fracture of Metals [in Russian], Metallurgiya, Moscow (1986).
7. E. C. Aifantis, "Computational material modeling," ASME, AD-Vol. 42/PVP-Vol. 294, 199-222 (1994).
8. L. D. Landau and E. M. Lifshits, Theory of Elasticity [in Russian], Nauka, Moscow (1987).
9. I. S. Grigor'ev and E. Z. Meilikov (eds.), Physical Constants. A Handbook [in Russian], Energoatomizdat, Moscow (1991).
10. V. I. Vladimirov and A. E. Romanov, Disclinations in Crystals [in Russian], Nauka, Leningrad (1986).
11. V. A. Likhachev, A. E. Volkov, and V. E. Shudegov, Continual Theory of Defects [in Russian], Izd. LGU, Leningrad (1986).
12. A. V. Kolubaev, V. L. Popov, O. V. Sizova, and S. Yu. Tarasov, in: Physical Mesomechanics and Computer Construction of Materials V. E. Panin (ed.) [in Russian], Nauka, Novosibirsk (1995), pp. 265-275.
13. S. G. Psakh'e, Yu. Khori, S. Yu. Korostelev, et al., Izv. Vyssh. Uchebn. Zaved. Fiz., **38**, No. 11, 58-69 (1995).