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Abstract

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ON THE NUCLEATION AND DEVELOPMENT OF CRACKS IN DEFORMED CRYSTALS

The study of the regularities of fracture under tension of metallic single crystals shows that, in the course of deformation and destruction, two stages should be distinguished: A—a more or less slow process of formation and gradual growth of equilibrium* embryonic cracks at places with a high concentration of stresses, caused by inhomogeneities of plastic deformation, and B—a rapid process of propagation, through the entire cross section of the crystal, of a single crack that has lost equilibrium (or of several closely located and coalescing cracks) ([1,2]).

At stage A, the principal role belongs to the applied **shearing** stress τ, under whose action, in the course of plastic flow, various deformation inhomogeneities arise, including incomplete shears (dislocation pile-ups), and the corresponding local stress concentrations. The first stage in the appearance of the nucleus of a microcrack may then be represented as the coalescence of several dislocations and the formation of a hollow core. As τ increases, deformation inhomogeneities and stress concentrations increase, and on the basis of the embryonic core there appears an equilibrium microcrack, incorporating into its cavity a certain number of dislocations belonging to the pile-up and leading to a partial relaxation of the high stresses in the adjacent region. The growth of the crack at this stage, accompanied by the arrival of new dislocations, may be associated with various mechanisms. The crack may develop in the region of the head of the dislocation pile-up (in the zone of the greatest concentration of tensile stresses) at some angle Θ to the slip plane containing the pile-up ([1–6]). In this case, alternation of microscopic cleavages along different planes is possible—then the initial crack has wavy outlines, whereas subsequently it develops in the preferred cleavage plane. If the active slip plane is at the same time a cleavage plane, then already at the initial stages the angle Θ is small, and later the crack passes wholly into the named common plane (for zinc this is the basal plane (0001)) ([1]). It is possible, as V. N. Rozhanskii indicates, that the microcrack lies from the very beginning in the plane (0001); the cause of its appearance may then be characterized as the loss of elastic stability of the material on the compressed side of the dislocation pile-up ([7]).

In considering particular variants of the model described, one usually proceeds from the simplest form of a dislocation pile-up in one plane, arrested by a single obstacle. It is more probable, however, that in the process of inhomogeneous shear formation the highest stress concentrations are created by several (or even many) rows of dislocations of one sign lying in nearby parallel planes (or by an excess of dislocations of one sign over the number of dislocations of the other sign within some, more or less broad, system of closely spaced slip planes), and the arrangement of individual dislocations in the pile-ups may be quite varied and deter–

* By equilibrium cracks we mean those cracks which, under the given conditions, cannot propagate avalanche-like through the whole crystal (i.e., stable, "non-dangerous" cracks); such cracks may, however, persist after removal of the load.

is determined by the presence of numerous obstacles in the slip planes. In the literature on dislocation schemes for the nucleation of microcracks one can find descriptions only of individual particular models of this kind ([8]).

The approximate quantitative analysis of the development of an equilibrium crack described by us ([1,2]) is not, in essence, restricted to a particular model; on the contrary, it has a universal character and remains valid independently of the specific model. Indeed, it can be shown that the near elastic-stress field in the region of localization of an incomplete shear, of extent s, for any specific structure of this defect is characterized by the quantity $\tau(s/r)^{1/2}$. When, in the region of such a defect, a crack of magnitude c arises, having any outline and orientation, the elastic stresses of the near field under consideration will, generally speaking, be relieved along the entire interval s and over distances of order c away from this interval (increasing with crack growth in proportion to its length c)* . It follows from this that the energy gain upon opening of the crack will be proportional to the logarithm of c, and therefore, as we have shown earlier ([1,2]), the crack will be an equilibrium one, and its maximum dimensions may reach the value

$$c_{\max} = \beta\tau^2 L^2/G\sigma; \tag{1}$$

here L is the maximum dimension $s_{\max}$ of the region of localization of deformation inhomogeneities, i.e., the cross section of a single crystal (or a grain of a polycrystal); G is the shear modulus; σ is the specific free surface energy; β is a dimensionless coefficient, of order of magnitude close to 1.

The onset of stage B is determined by the applied normal stress p: the crystal ruptures when p reaches the value at which the most dangerous crack loses equilibrium, i.e. ([9,2]) at

$$p_c = \alpha(E\sigma/c_{\max})^{1/2} = \alpha'(G\sigma/c_{\max})^{1/2}. \tag{2}$$

Fig. 2

Figure 1: Fig. 2

Fig. 3

Figure 2: Fig. 3

The combination of conditions (1) and (2) leads to the formulation of the condition for constancy of the product $p_c \tau_c = \text{const}$ in brittle rupture of differently oriented single crystals, which is well confirmed by experiments ([2,11–14]).

In the present work, relation (1) was for the first time tested independently of relation (2). For this purpose the pattern of development of brittle-fracture cracks in crystals was studied with increasing shear and normal stresses applied to the active slip plane. The objects of investigation were amalgamated zinc single crystals (purity 99.99%) of 1 mm diameter and 10 mm length with different angles of inclination χ_0 of the basal plane to the specimen axis**. The specimens were subjected to uniaxial tension at room temperature with a constant rate of $\simeq 12\% \, \text{min}^{-1}$. At a given degree of deformation ε (in the interval from 1% up to values corresponding to rupture of the crystal) the stretching was stopped; the magnitude of the tensile stress was then recorded. The shear and normal stresses were determined from the relations $\tau = P_0 \sin \chi_0 \cos \chi$, $p = P_0 \sin \chi_0 \sin \chi$, where P_0 is the tensile stress referred to the initial cross section; χ is the angle of inclination of the basal plane to the specimen axis at the given value of deformation. Longitudinal sections were prepared from the stretched specimens, with the plane of the section perpendicular to the (0001) plane of the single crystal. After polishing, the sections were etched with a 10% aqueous solution of nitric acid and examined under a microscope. On all the sections investigated, internal cracks located in the (0001) plane were found; for each section the length of the largest crack $c_{\max}$ was measured.

* The conclusion that the energy gain is independent of the angle Θ agrees with Stroh' s result ([4]), obtained by him for a specific dislocation model.

** We have previously shown that amalgamated zinc single crystals exhibit a sharp loss of plasticity and strength already at room temperature ([10,11]). At the same time, in amalgamated zinc single crystals it is possible to observe very distinctly the appearance of brittle-fracture cracks already at early stages of their development.

To verify the dependence $c_{\max}(\tau)$, a series of specimens with the same $\chi_0 = 21^\circ$ was tested. As was shown earlier ([11]), in such specimens the limiting amount of plastic shear preceding brittle separation is considerably greater than in single crystals with large angles χ_0. For crystals with $\chi_0 = 21^\circ$, it proved possible to use a range of

Fig. 2

Fig. 3

deformations from 1 to 22%, which corresponds to an increase of τ from approximately 50 to 200 G/mm^2. Microphotographs of several polished sections are given in Fig. 1 (see the inset, p. 56); the obtained quantitative dependence of $c_{\max}$ on τ is presented in Fig. 2, which shows that $c_{\max}$, in full agreement with formula (1), increases linearly with increasing τ^2; the slope is $c_{\max}/\tau^2 = 0.22 \cdot 10^{-4}$ mm^5/G^2 $= 0.22 \cdot 10^{-15}$ cm^5/dyn^2.

Let $G = 3 \cdot 10^{11}$ dyn/cm^2, $\sigma_{\text{Zn-Hg}} = 200$ erg/cm^2 ([2,13]), and $\bar{L}_0 = 1$ mm. Then, according to (1), $c_{\max}/\tau^2 = \beta \cdot 0.17 \cdot 10^{-15}$ cm^5/dyn^2; hence $\beta \approx 1.3$.

To verify Griffith' s relation (2), part of the specimens with different χ_0, from 16 to 67°, was brought to rupture; in this process p_c was determined. The stretching of another part of identical specimens was stopped when the stress reached 90–95% of the rupture stress; from their polished sections the value $c_{\max}$ was measured. Fig. 3 shows that p_c increases linearly with increasing $c_{\max}^{-1/2}$.* The slope is about 100 G/mm$^{1/2}$ $= 0.32 \cdot 10^7$ dyn/cm$^{3/2}$. For the same values of G and σ, according to (2), we have $p_c c_{\max}^{1/2} = \alpha' \cdot 0.77 \cdot 10^7$ dyn/cm$^{3/2}$; hence $\alpha' \approx 0.4$.

Thus, the results obtained well confirm the universal character of the approximate quantitative scheme proposed by us earlier ([1,2]) for analyzing the development of a brittle-fracture crack.

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* Strictly speaking, relation (2) is valid only for $c \ll L$; in this connection the line $p_c = p_c(c_{\max}^{-1/2})$ cannot pass through the origin: when a crack reaches a size equal to the length of the slip plane L, the specimen falls apart at $p = 0$.

Note: Figure translations are in progress. See original paper for figures.