INTERNAL STRESSES AROUND UNIT DISLOCATIONS

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Abstract

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According to the theory of dislocations, the atomic planes in real crystals are only approximately parallel to one another and may terminate inside the crystal (an edge dislocation) or join with one another, forming a single helical surface (a screw dislocation). Each dislocation causes a local distortion of the lattice and is a source of internal stresses.

Many investigators have obtained convincing data showing that the macroscopic distortion of the lattice is in fact composed of local distortions of the lattice around individual dislocations. In particular, if edge dislocations are arranged in the form of a so-called vertical row (Fig. 1), the stresses caused by them (far from the row) mutually compensate one another, while the distortions add up in such a way that the row as a whole is equivalent to a symmetric boundary of blocks rotated relative to one another through an angle

$$\vartheta = \frac{b}{D}. \tag{1}$$

Fig. 1. Vertical row of edge dislocations (block boundary). Each dislocation is marked by the symbol $\perp$. The orientation of the lattice changes discontinuously; macroscopic stresses are absent.

Fig. 2. Horizontal row of edge dislocations (line of sliding). The orientation of the lattice is preserved; the upper half of the crystal is shifted relative to the lower one.

Here b is the Burgers vector of the dislocations, and D is the distance between them. The use of formula (1) and other similar expressions makes it possible to check the validity of the theoretical predictions concerning dislocations as sources of lattice distortion ([1]-[5]).

We shall show that a study of a **horizontal** (Fig. 2) row of edge dislocations makes it possible to verify the predictions of the theory concerning dislocations as sources of internal stresses. It should be noted that such a verification of the theory has not yet been possible. The results of optical studies of stresses in silicon crystals ([6]), rock salt ([7]), and corundum ([8]) confirm the correctness of the theoretical calculations of the stress distribution around macroscopic dislocations of the theory of elasticity, and not around those unit dislocations that appear in the dislocation theory of crystals.

To the article by V. L. Indenbom and G. E. Tomilovskaya, p. 723

Fig. 4. Dislocations and stresses in a corundum crystal. The specimen is cut along the basal plane. 52 ×.
a –revealing dislocations by the selective etching method; the vertical row of figures is a block boundary, the remaining rows are slip lines;
b –the same specimen in crossed polaroids; the slip lines are bordered by bands of stress.

To the article by L. P. Morozov and N. A. Krotov, p. 747

Fig. 2. Microphotographs of transverse sections of systems: **a** –perchlorovinyl (1) –SKN rubber (2); **b** –gutta-percha (1) –paraffin (2). 100 ×.

Let a horizontal row consist of dislocations parallel to the Z axis and having a Burgers vector parallel to the Y axis. The extra half-planes of the dislocations are perpendicular to the Y axis and are located along the positive half-axis X (Fig. 2). If the dislocations are spaced at a distance D, then the elastic deformation y_y^{I} of the upper half of the crystal will differ from the corresponding deformation y_y^{II} of its lower half by the total thickness b/D of the extra planes ending on a unit segment of the Y axis:

$$y_y^{\mathrm{II}} - y_y^{\mathrm{I}} = \frac{b}{D}. \tag{2}$$

Taking into account the zero equality of displacements parallel to the Z axis (plane-strain state) and the continuity of the components of the stress tensor acting on the YZ plane, one can calculate the jumps, caused by the dislocations, of the stresses Y_y and Z_z. In the case of an elastically isotropic crystal*

$$Y_y^{\mathrm{II}} - Y_y^{\mathrm{I}} = \frac{b}{D}\frac{E}{1-\mu^2}; \quad Z_z^{\mathrm{II}} - Z_z^{\mathrm{I}} = \frac{b}{D}\frac{\mu E}{1-\mu^2} \tag{3}$$

(E is Young' s modulus, μ is Poisson' s ratio). An analogous calculation for a row of edge dislocations parallel to the optical axis of corundum gives

$$Y_y^{\mathrm{II}} - Y_y^{\mathrm{I}} = \frac{b}{D} : (s_{11} - s_{13}^2/s_{23}); \quad Z_z^{\mathrm{II}} - Z_z^{\mathrm{I}} = \frac{b}{D} : (s_{13} - s_{11}s_{33}/s_{13}). \tag{4}$$

Thus, in the plane of a horizontal row of edge dislocations there occurs a jump in the magnitude of the normal stresses acting along the Burgers vector and along the dislocation lines. Comparison of this stress jump with the dislocation density found, for example, by the etching method can provide a direct check of the stress field around dislocations, just as the use of relation (1) provides a direct check of the geometrical properties of dislocations.

The formation of horizontal rows of edge dislocations is most likely to be expected in slip planes: a shear that begins at one edge of the slip plane and then propagates over the whole of this plane corresponds to the displacement, in the direction of slip, of successive dislocation lines. It is interesting that as early as I. V. Obreimov and L. V. Shubnikov ([9]), observing by an optical method the stresses in an NaCl crystal on both sides of the slip plane, noted that the stresses were visible when viewed both perpendicular and parallel to the slip direction**. In other words, Obreimov and Shubnikov observed (in our notation) both a jump of the stresses Y_y and a jump of the stresses Z_z.

However, an NaCl crystal is inconvenient for the quantitative study of slip lines, since it has mutually perpendicular slip directions, as a result of which deformation bands may be superposed on the slip line.

Apparently, it was precisely this circumstance that led to the difficulties noted in the comment on the paper by N. M. Melankholin and V. R. Regel ([11]). It is considerably easier to distinguish slip lines in a corundum crystal, where the slip directions are parallel to the faces of the prism and form an angle of 60° with one another (Fig. 3).

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* A relation equivalent to the first formula (3) was first indicated by Naem ([14]) in interpreting the results of an optical study of slip lines in AgCl crystals.

** It should be noted that the critical remarks by Anseles and Kuznetsov ([10]) concerning an error allegedly made in the calculations of Obreimov and Shubnikov are based on a misunderstanding.

Fig. 4a shows a plate of synthetic corundum (leucosapphire), cut along the basal plane and etched by the method of ([12]), which proved sufficiently sensitive for revealing individual dislocations. The vertical row of etch figures corresponds to a block boundary; the horizontal and inclined rows correspond to slip lines (the slip planes are perpendicular, while the slip directions are parallel, to the plane of the drawing). All three systems of slip lines are visible.

Fig. 3. Structure of corundum in projection onto the basal plane. Black circles —aluminum ions; white circles—oxygen ions. Thick lines indicate the possible Burgers vectors—minimal translation vectors in slip directions $[1\bar{1}00]$, $[01\bar{1}0]$, $[\bar{1}010]$. The slip planes $(11\bar{2}0)$, $(\bar{2}110)$, and $(1\bar{2}10)$ are perpendicular to the plane of the drawing.

Fig. 4b represents the same specimen in crossed polaroids. Both surfaces of the plate are wetted with immersion liquid. The slip lines are bordered by bands of stress which, as can be verified with a compensator or a sensitive plate, have different signs. The block boundary, as was to be expected, produces no macroscopic stresses.

Table 1

D, in μ	Stress jump, in kg/mm^2	Stress jump, in kg/mm^2
	theor.	exper.
19	1.6	1.6
21	1.5	1.8
22	1.4	1.7

It should be especially noted that local stress concentrations are formed at places where slip lines intersect one another or intersect a block boundary.

Let us assume that each etch figure corresponds to the emergence of a single edge dislocation, having a Burgers vector equal in magnitude to the minimal translation vector along the slip direction:

$$b = 2\sqrt{3}\,a \sin\frac{\alpha}{2} = 8.32\,\text{Å}, \tag{5}$$

where $a = 5.2\,\text{Å}$ is the edge, and $\alpha = 55°17'$ is the plane angle of the rhombohedron—the elementary cell of the corundum lattice. Then, taking into account the values of the elastic constants of corundum ([13]), $s_{11} = 2.84 \cdot 10^{-13}$, $s_{13} = -0.47 \cdot 10^{-13}$, $s_{33} = 2.21 \cdot 10^{-13}$ cm^2/dyne, we can, using (4), calculate from the observed density of etch figures in the slip lines the theoretical value of the stress jump. This calculated value may then be compared with the experimental value found by the optical method.

The corresponding optical stress constant was determined by us specially for this purpose and was found to be (with an error of the order of 10%) equal to $2.1 \cdot 10^{-7}$ cm^2/kg. The results are compared in Table 1. Allowing for possible errors due to the counting of overlapping etch figures, as well as for the insufficient accuracy of the constants, the agreement between the theoretical and experimental data must be regarded as quite satisfactory.

Thus, in our case each etch figure in a slip line does indeed correspond to the emergence of an edge dislocation with a unit Burgers vector, and all dislocations in the slip plane prove to be of the same sign.

It also follows from the agreement between theory and experiment that the ordinary theory of elasticity is apparently fully applicable to the calculation of internal stresses around individual dislocations, at least at distances of the order of several microns from the dislocation lines.

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