Dislocation Processes and Plastic Work in Dependence on Crack Velocity in Ionic Crystals

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For a propagating crack in an elastic-plastic material a considerable portion of the total energy required is due to the plastic work in a region around the moving crack tip, even in case of the comparatively brittle ionic crystal, such as LiF and NaCl. In a number of investigations the generation and propagation of dislocations by running cracks in these crystals was observed (1 to 11). In this paper, results of experiments are described, in which plastic-zone size and the density of dislocations within the plastic zone around the crack tip were measured in dependence on crack velocity in LiF crystals. From this, the plastic portion of fracture energy in dependence on crack velocity was calculated.

The dislocation processes expected

If a crack propagates in a NaCl-type crystal within the usual cleavage plane $\{100\}$, dislocations gliding on the six $\{110\}$ slip planes 1g to 6g (fig. 1) can be activated to move by the shear stresses around the crack tip. The main plastic processes occuring in case of a constant crack velocity are schematically shown in fig. 2 (for details see (9) and (11), where these processes are described on the basis of the quasistatic (9), (11) and also the relativistic shear-stress distribution (11) around the crack tip). The dislocations are pushed ahead of the crack tip within the plastic zone of size d, and they are carried along by the crack, thereby leaving long trails of dislocation lines parallel to the crack surface. The orientation of the remaining dislocations produced during fracture is given by the intersection line of a plane parallel to the crack surface, in a distance d/2, with the slip planes 1g to 6g of the individual dislocations. According to estimations in (11), d/2 will be about 1 μm at $v_c = 10^4$ cm/s, and about 0,08 μm at 10^5 cm/s, resp. (see also (8)).

Experimental details

Berg-Barrett, Lang and Barth-Hosemann X-ray techniques were first used to detect the dislocations produced during crack propagation. However, the best results were obtained by applying a sensitive etch method adapted to reveal dislocations with a depth between 0,1 and 1 μm in the high crack-velocity region. The dislocations parallel to the fracture surface can be etched, as was found out by the experiments, because of the following effect (see fig. 3). The dislocation, in fig. 3a parallel to the surface, is moved under the action of image forces towards the fracture surface. Bow-outs are formed (fig. 3b) between pinning points such as impurities or intersecting dislocations. In fig. 3c the bow-outs have reached the surface, rearrangement of the half loops formed occurs and the emergence points of these half loops are then etched in the etchant, where a large number of etch pits mark the position of the originally straight dislocation line of fig. 3a. Flat-bottomed etch pits will form, if the etch pits are deeper than the deepest point of the half loops. It can be shown that the distance d/2 of the originally straight dislocations from the surface, i.e. half the plastic zone size, can be determined by $d/2 = d_s(1+v_1/v_n)$, where d_s is the depth of the (flat-bottomed) etch pits, v_1 is the solution rate of the surface layer removed by etching, v_n the solution rate along a dislocation line perpendicular to the surface. Crystals were etched with destilled water containing different molar concentrations of $FeCl_3$ (12), mostly near 10^{-3} molar concentration. Etch periods were between 10 to 50 seconds and etching was started immediately after fracture (within 1/10 s). Measurements necessary to determine d were done with an interference microscope. The flat-

bottomed etch pits very often can hardly be seen in the eyepice of an optical microscope - only on the photoplates at higher magnifications - if their depth is below 0,5 μm. This will be the reason, why Gilman and others (1), (3) did not observe a continuous production of dislocations at high crack velocities. Crack velocities were measured in a way described in (3). Crystals having a size of about 2 x 10 x 15 mm^3 were cleaved along a $\{100\}$ face with a chisel. Prior to cleavage the crystals were annealed at 630 °C for 200 h. The optically clear crystals (VEB Carl Zeiss, Jena) then have a dislocation density of about $10^4/cm^2$. The critical resolved shear stress is 0,265kg/mm^2.

Results

A typical example of an etched cleavage surface is shown in fig. 4. Crack-propagation direction is marked by the arrow. The dark etch pit bands D indicating discontinuous dislocation generation are deformation zones already described by Gilman (1), (3). They are due to the action of flexural waves (13). In the upper right of fig. 4, also in the left part, we observe long rows of etch pits lying in two perpendicular $\langle 100 \rangle$ directions. These etch pits mark the continuously produced dislocations on slip planes 1g to 4g. Originally individual dislocations very often have a length up to a few mm. In the middle part of the photo no rows of etch pits can be observed in the region between deformation zones, but a careful examination at higher magnification shows the rows of flat bottomed etch pits having much less a depth there (see fig. 5). The deep etch pits E in fig. 5 correspond to grown-in dislocations. From micrographs like fig. 5, N, the linear density of dislocations along crack front and also, from the depth of the flat-bottomed etch pits, the plastic zone size d was determined. From these data the plastic energy per unit area of fracture surface is estimated: $W = \sum_{i=1}^{6} \tau_i b d_i N_i$, with τ_i = shear stress necessary to move the dislocations along with the crack on each of the six slip planes - these values are derived from experimental curves (14) describing dislocation dynamics ($\tau \approx$ - A/lnv_c/v_o) -; b = Burgers vector. Results are shown in fig. 6. N, d and W are plotted against crack velocity. For N and d experimental error is indicated. The values of these quantities aproach zero at crack velocity near 2 to 3 $\cdot$ 10^5 cm/s. For comparison, the surface energy of LiF is about 340 ergs/cm^2. According to fig. 6 the following relations can be derived: N$\sim$ - ln v_c/v_o and d$\sim$ (ln $v_c/v_o)^2$. From this it follows that W$\sim$ (ln $v_c/v_o)^2$. Contrary to the observations of Gilman and others (3) and to the estimation of Burns and Webb (8), dislocations are carried along with the crack beyond a velocity of 6 $\cdot$ 10^4 and 9 $\cdot$ 10^3 cm/s, resp.

References

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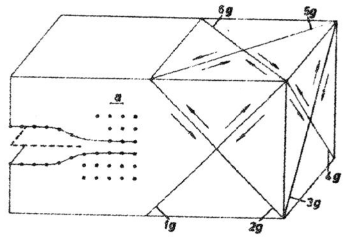

Fig. 1

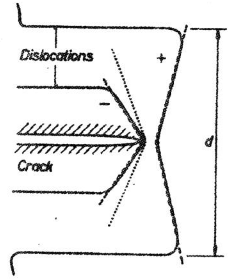

Fig. 2

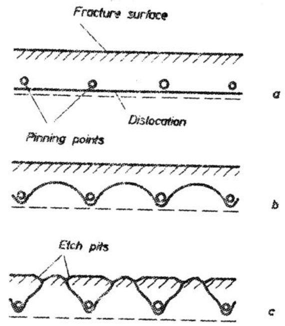

Fig. 3

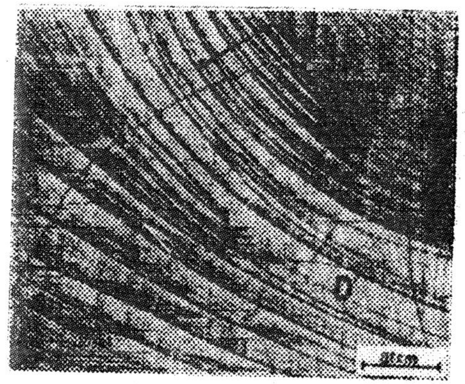

Fig. 4

Fig. 5

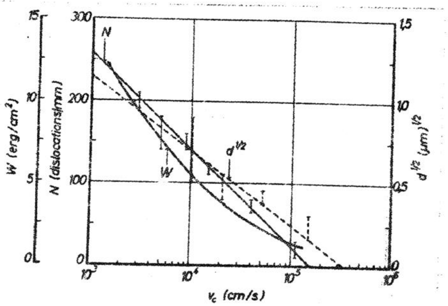

Fig. 6