22. Critical Experiments on the Nature of Fatigue in Crystalline Materials

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ABSTRACT

To investigate the role of cross slip in the failure of crystals by fatigue, tests have been made on LiF, NaCl, AgCl, and KRS-5 crystals. The first two do not exhibit easy cross slip, while the latter two do. The fatigue results and metallographic observations indicate that easy cross slip is required for ordinary fatigue failure. The Cottrell-Hull mechanism does not operate in the experiments, while the Mott mechanism may.

Introduction

In recent years, significant advances have been made toward an understanding of the mechanism of fatigue of metals. These advances have been experimental as well as theoretical in nature; a comprehensive review of the subject will be found elsewhere.[1] The principal experimental discovery has been that extrusion [2] and intrusion [3] of material along slip bands takes place during cyclic loading. Since fatigue cracks usually start in slip bands, it is thought that the cracks result either from the growth of intrusions [3] or from voids left behind extrusions.[4] Another important finding is that fatigue failures occur at temperatures as low as 4.2°K,[5] thereby clarifying the theoretical picture by eliminating thermally activated processes such as oxidation or vacancy diffusion as essential to the fatigue mechanism, although these processes may play a modifying role at higher temperatures.

Hull [6] examined the surface structure of slip bands in copper at 293°, 90°, 60°, and 4.2°K and found that the shape and size of the extrusions do not appear to be affected by the temperature at which the specimens are fatigued. These extrusions and intrusions consist of thin ribbons or crevices about 0.05 μ thick. Their height or depth varies widely, but the maximum value is of the order of 1.5 μ. Since these effects were observed at 1% of the estimated life, Hull considers that the major part of the fatigue process is devoted to the propagation of cracks.

The principal theoretical efforts aimed at explaining crack formation have been the dislocation mechanisms of Cottrell and Hull,[3] and Mott.[4] (See Chapters 18 and 21 of this volume by Thompson and Backofen, respectively, for descriptions of these mechanisms.) In Cottrell and Hull's model, an interaction of edge dislocations moving on two slip systems is required, while in Mott's model the cross slip of a screw dislocation is the characteristic feature.

A principal difference between the Cottrell-Hull and Mott mechanisms is that cross slip is essential to one but not to the other. By testing crystals that do exhibit cross slip, as well as those that do not, it should be possible to assess the relative validity of the two mechanisms, especially if the crystals are so oriented as to fulfill the requirements of the Cottrell-Hull mechanism. If cross slip is essential to fatigue failure, then crystals that do not cross slip should exhibit superior fatigue strength, as compared to those crystals that do cross slip. Experiments that bear on this important point have been made by Hull,[6] Honeycombe,[7] and Ebner and Backofen,[8] but they have not served to differentiate between the two mechanisms as clearly as might be desired. The aim of the present investigation was to carry out critical fatigue experiments along these lines.

Experimental Procedure

The materials selected were single crystals of LiF, NaCl, AgCl, and KRS-5 (44 mole % TlBr + 56 mole % TlI). All of these crystals are cubic; the first two do not exhibit easy cross slip at room temperature, while the latter two do. (In the former two, there are no two slip planes that have a common slip direction, while in the latter two crystals, there are such cross-slip combinations.) The LiF is reputed to be quite stable,[9] whereas the NaCl is subject to embrittlement, which is a function of time and the environment. The AgCl and KRS-5 both exhibit a high degree of plasticity and do not become embrittled with time.

All crystals were obtained from the Harshaw Chemical Company of

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Cleveland. The LiF and NaCl were obtained as blanks cleaved to the desired orientation. The AgCl was obtained as a single crystal, 3 in. in diameter by 1 in. thick. The orientation of the crystal was determined by X-ray analysis, and specimens of desired orientation were machined from the large crystal. The KRS-5 specimens were sawed from a crystal too small for a selection of orientation to be made, and so the as-received orientation was used. Pertinent data concerning these crystals are given in Table 1.

TABLE 1. Pertinent Data for Crystals Tested

Crystal	LiF	NaCl	AgCl	KRS-5
Crystal structure type	NaCl	NaCl	NaCl	CsCl
Slip direction	[110]	[110]	[110]	[100]
Slip plane	(1$\bar{1}$0)	(1$\bar{1}$0)	—*	(011)
Cleavage plane	(100)	(100)	—	—
Easy cross slip	No	No	Yes	Yes
Melting point (°F)	1592	1472	872	772

* Deforms by pencil glide.[10]

Specimens were tested in bending, and a configuration found suitable. (Approximate dimensions of the crystals are indicated in Fig. 1.) The stress concentration factor for this configuration is 1.02; it will be neglected in this chapter.

In shaping the crystals, it was found that hand grinding with metallurgical papers was the most satisfactory method. Final polish prior to annealing was done with cloths impregnated with 600 Alundum. Specimens were annealed in still air at a temperature equal to three fourths of the melting point in °F. The time of anneal was 3 hr, and this was followed by a furnace cool to room temperature. The specimens were then checked in a polarizing microscope and found to be stress free after this anneal. The LiF, NaCl, and AgCl were then chemically polished in solutions of NH$_4$OH, water, and sodium thiosulfate, respectively. The KRS-5 crystals were not polished after the anneal since no suitable solvent[11] was known. The crystals were then stored in a desiccator until tested.

Tests

The specimens were tested as cantilever beams under conditions of fully reversed alternating bending at 1800 cpm in Sonntag SF-2 sheet-bending fatigue machines. The deflection of the free end of the cantilever was measured using a 30-power microscope and stroboscopic illumination. The usual procedure was to start the test at some low stress level and gradually to increase the stress level in small steps, allowing at least 10^5 cycles between steps. A typical loading history is shown in Fig. 2. In order to prevent transient overloading, the starting amplitude of the free end was limited by means of a tapered snubbing block, which allowed the amplitude to be increased gradually to the full value. In one case, the initial stress was made quite high in order to determine if a higher than usual stress could be applied without causing fracture.

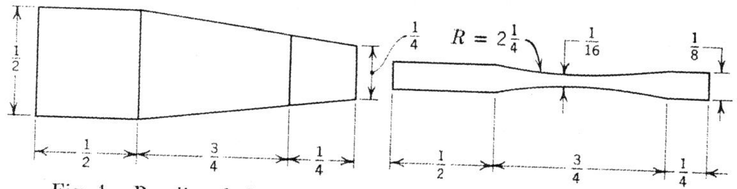


Fig. 1. Bending fatigue specimen. All dimensions are in inches.

The chief difficulty in testing LiF and NaCl crystals, resulting from their brittle nature, was in adjusting the grips to hold the specimens securely without crushing the specimens. The AgCl crystal and the metal of the grips caused a corrosive combination that was made less severe by spraying all metal contact surfaces with a plastic. The KRS-5 crystals presented no unusual difficulties beyond the fact that the dust is highly toxic if inhaled.

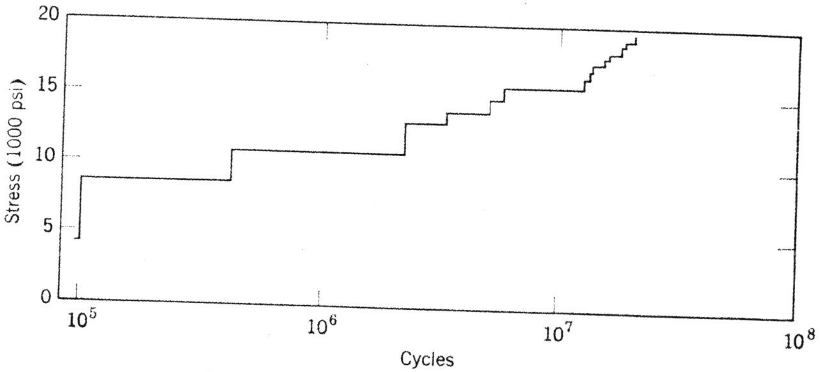


Fig. 2. Stress-cycle history of LiF crystal.

The orientations of the crystals were varied to achieve tests of various mechanisms. The Cottrell-Hull mechanism requires two intersecting slip systems, with slip directions out from the crystal surface. The orientation ideally suited to produce this slip in bending is that which has the crystal axis along the [100] direction and the plane containing the crystal axis and the axis of bending the (001) plane. In this case, the two slip planes (101) and ($\bar{1}$01) were at 45° to the tensile direction, and the two slip directions, [$\bar{1}$01] and [101], were out from the surface at the ideal 45° angle. This orientation was chosen for some of the tests to optimize the likelihood of the Cottrell-Hull mode of extrusion-intrusion development in the crystals that do not exhibit cross slip. It should be remarked that the other two potential slip planes in this orientation will be less likely to operate than those mentioned above, because the crystal dimension in the other two potential slip planes along the slip direction is much larger.[12]

Static load-deflection data were also obtained in the fatigue machine by means of a calibrated spring and dial, which are ordinarily used for applying mean loads. The readings obtained not only were of use in determining the yield strength, but they also served as a check on the dynamic calibration of the fatigue machine.

After failure, the specimens were subjected to metallurgical examination.

TABLE 2. Results of Bending Fatigue Tests

Material	Yield Stress (psi)	Spec. No.	Orientation	S_{max} (psi)	Cumulative No. Cycles	No. Cycles of S_{max}	Remarks
LiF	2500	22	(001) [100]	19,500	20×10^6	—	Failed on increase of load
	2500	23	(001) [100]	22,400	36×10^6	—	Failed on increase of load
	2500	27	(001) [100]	27,500	17×10^6	10^5	Failed in fatigue
	2500	28	(001) [100]	20,800	18×10^6	—	Failed on increase of load
	—	25	(001) [110]	27,700	28×10^6	—	Failed in static loading at 34,000 psi
NaCl	400	1	(001) [100]	5,220	28×10^4	—	Failed on increase of load
	400	2	(001) [100]	2,300	5×10^3	—	Failed on increase of load (poor surface)
	400	3	(001) [100]	6,000	32×10^4	—	Failed on increase of load
	400	4	(001) [100]	9,000	63×10^4	—	Failed on increase of load
	400	5	(001) [100]	8,120	20×10^6	2×10^6	Failed in fatigue
	400	6	(001) [100]	4,100	—	—	Failed on loading
AgCl	800	1	—	3,600	19×10^4	5×10^4	Failed in fatigue
	800	2	(001) [100]	2,600	30×10^4	8×10^4	Failed in fatigue
	800	3	(001) [100]	2,050	1.1×10^6	6×10^5	Failed in fatigue
	800	4	(001) [100]	1,400	8.4×10^6	4×10^5	Failed owing to corrosion at grip
	800	5	(001) [100]	1,820	5.5×10^5	5.5×10^5	Failed in fatigue
KRS-5	5000	1	(001) [100]*	7,020	10×10^6	3×10^3	Failed in fatigue
	5000	2	(001) [100]*	7,680	27×10^4	27×10^4	Failed in fatigue
	5000	3	(001) [100]*	4,800	15×10^6	2×10^6	Failed in fatigue
	5000	4	(001) [100]*	5,010	1.4×10^6	1.4×10^6	Failed in fatigue

* Approximately

Results

Fatigue Test Data

The results of the fatigue tests are tabulated in Table 2. In general, the LiF and NaCl crystals failed during a step increase of the applied load, although one crystal of each species failed after a large number of load cycles at the peak stress level. In no case was a fatigue failure obtained in either LiF or NaCl at stress levels up to at least *ten times* the static yield strength and beyond 10^7 cycles. (It may be of interest to note that the NaCl crystal that failed in fatigue endured approximately one week of testing while exposed to the laboratory environment.) All of the AgCl crystals except one and all of the KRS-5 crystals failed in fatigue. Many of these crystals failed at stress levels below twice the yield strength and in less than 10^6 cycles.

In Table 2, both the cumulative total of cycles applied and the number of load cycles at the peak stress level are listed. Inasmuch as the load histories of the specimens varied, it seemed better not to plot the results as *S–N* curves, although the few constant load tests for AgCl and KRS-5 crystals indicate that typical *S–N* curves would result for these materials. The LiF crystal, which was tested initially at 17,000 psi showed no apparent difference as compared to the remainder of the LiF specimens which were subjected to an initial stress of the order of 2000 psi. However, about 1000 cycles of gradually increasing stress level were involved in the transient period at the start of the test.

Static Test Data

Also listed in Table 2 are the average yield strengths of the crystals as determined for at least two crystals of each species by static loading in the fatigue apparatus prior to the start of the fatigue tests. Under static applied load in the fatigue apparatus above that necessary for yield, all the crystals except KRS-5 deformed continuously; that is, there was little, if any, strain hardening. Deviation from linearity on the load-deflection plot was taken as the yield strength of the material, and it is expected that the values are somewhat higher than would be found in tension tests of the same material. A few additional LiF specimens were tested to failure statically without significant increase in stress level above the yield strength, although the crystals exhibited considerable ductility. One LiF specimen was tested statically after reaching the 27,000-psi level. The crystal deformed plastically a slight amount before final failure occurred at 35,000 psi.

Typical results of the deflection measurements during both static and dynamic loading are shown in Figs. 3 to 6. It is noteworthy that only the AgCl crystals did not yield linear load-dynamic deflection curves. The other three materials were linear up to failure, thereby permitting the use of elastic theory in computing the bending stresses.

Optical Examination

After fracture, the specimens were examined under a microscope for features, such as extrusions or intrusions, related to the process of fatigue. In addition, an attempt was made to study the behavior of dislocations at various stages during a fatigue test of LiF by utilizing the etching techniques developed by Gilman and Johnston.[13] It was soon found, however, that at any stress level approaching the final fracture level the density of dislocation etch pits becomes so great over the entire region of maximum stress that little can be learned through this approach. An example of this effect of an auxiliary LiF specimen that was notched is shown in Fig. 7. The light area is a region of low stress, free of dislocations. The barrier to slip provided by an intersecting system is evident.

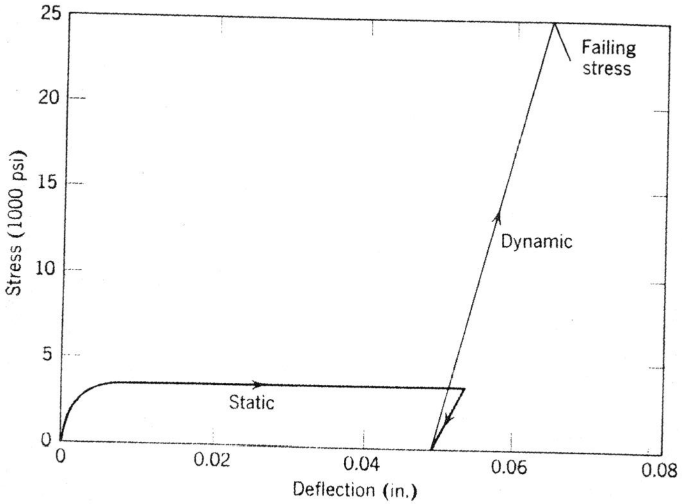


Fig. 3. Static and dynamic stress-deflection history of LiF crystal.

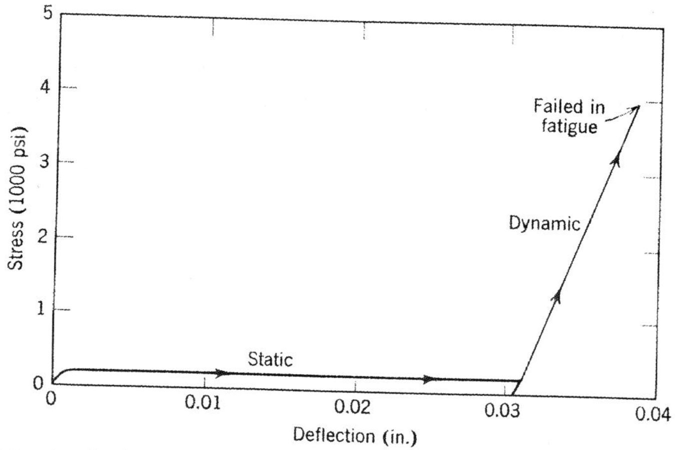


Fig. 4. Static and dynamic stress-deflection history of NaCl crystal.

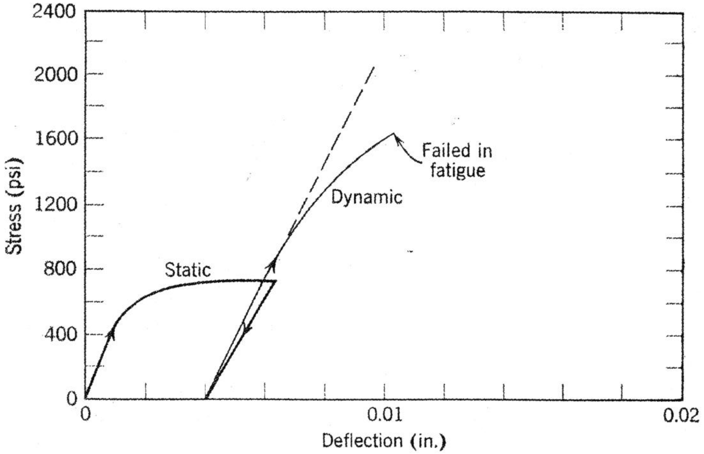


Fig. 5. Static and dynamic stress-deflection history of AgCl crystal.

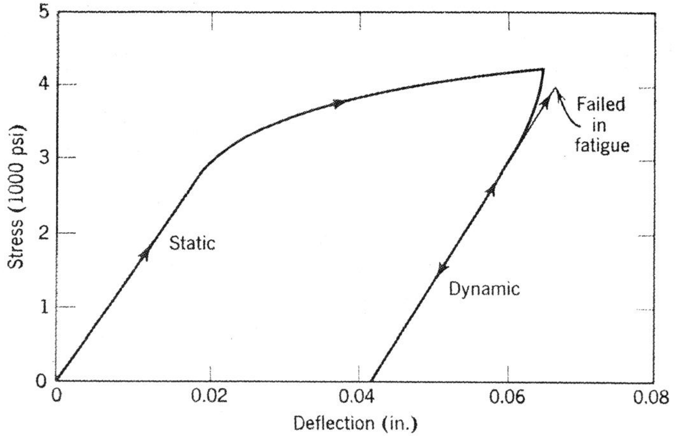


Fig. 6. Static and dynamic stress-deflection history of KRS-5 crystal.

Examination of the LiF specimens revealed that the surface at 500X was devoid of the large slip steps that are developed upon static loading. An example of a statically loaded surface is shown in Fig. 8*a*. The straightness of these slip bands is in contrast to the slip markings found in the AgCl and KRS-5 crystals shown in Figs. 8*b* and 8*c*. Some evidence of large-scale slip was found in the vicinity of the fracture region of LiF crystals tested under cyclic loading, but it may have been caused by the final static fracture of the specimen. The fact that etching revealed so many pits indicates that the entire surface was covered with fine slip lines.

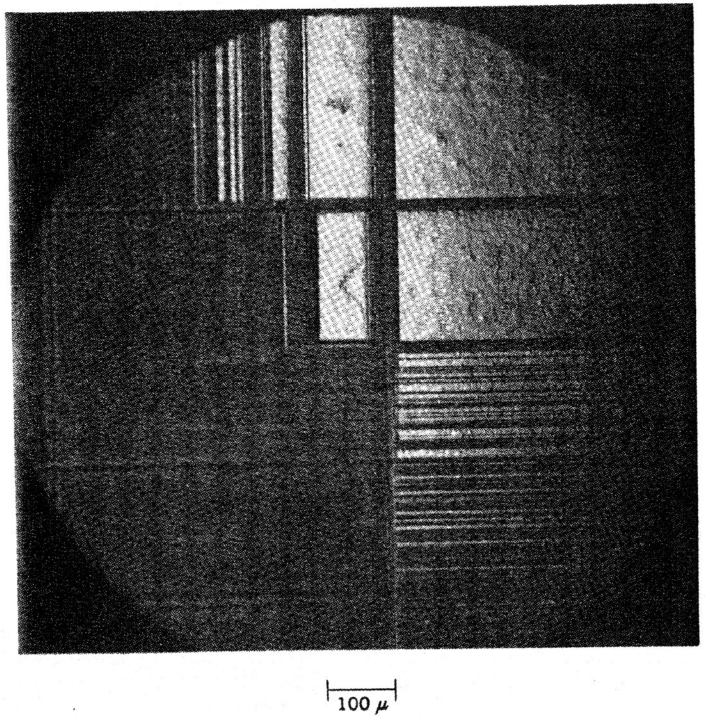


Fig. 7. Dislocation etch pits on LiF surface.

For both the LiF and the NaCl crystals, the main fracture occurred along the {100} cleavage plane. Most of the fractures appeared to originate at the surface; however, as Fig. 9*a* indicates, the initial fracture in the LiF specimen that failed in fatigue seemed to start beneath the surface, and a possible extrusion is evident (Fig. 9*b*). However, in an auxiliary test of an LiF specimen that involved a different specimen configuration, although a fatigue failure resulted, the fracture appeared to start at the surface. Examination of the surface of the NaCl crystals revealed that they were quite similar in appearance to the LiF specimens. In no case were extrusions or intrusions found.

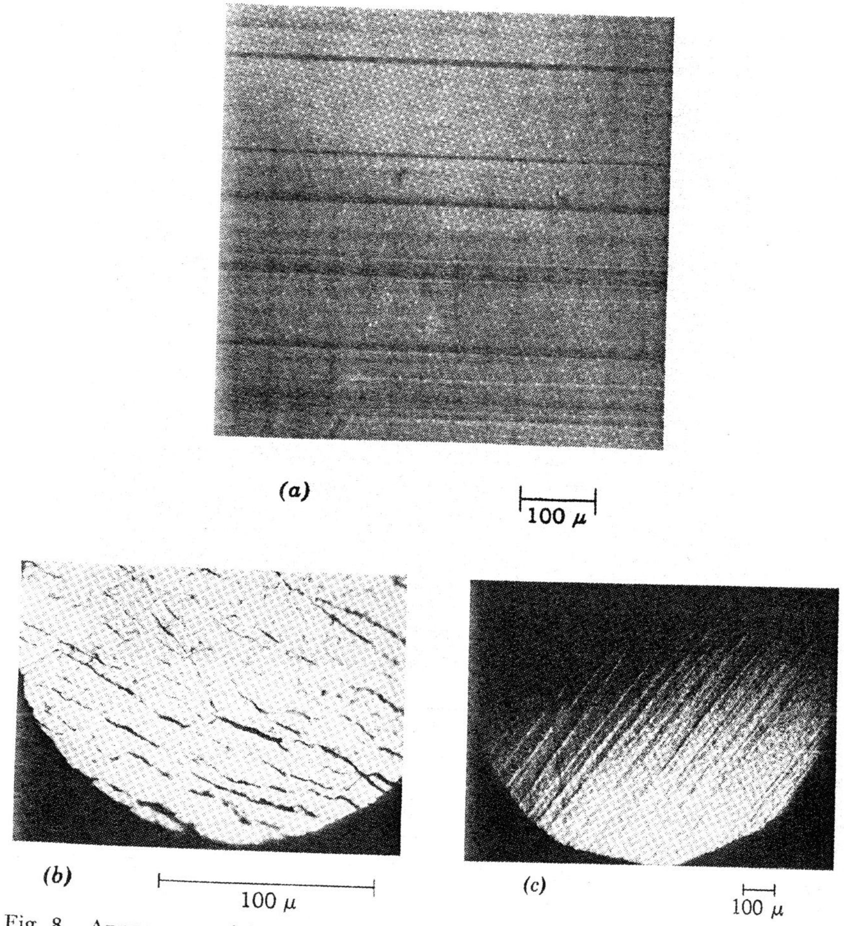


Fig. 8. Appearance of slip bands after static bending in crystals of (*a*) LiF, (*b*) AgCl, and (*c*) KRS-5.

The surfaces of the AgCl crystals were highly distorted as a result of cyclic loading. Extrusions were found only at the lowest stress level; these are shown in Fig. 10. They are quite similar in appearance to those presented by Forsyth.[14] If extrusions were present at the higher stress levels, their presence was masked by the general rumpled appearance of the surface. Figure 11 shows an example of this effect. The bright regions are approximately 10 μ below the average surface level and run more or less perpendicular to the direction of tension-compression. For these specimens, it was difficult to decide where cracks had initiated, but the bright areas include a network of fine lines. Examination of the side face of the AgCl crystals revealed that slip occurred in but one of the two equally stressed <110> slip directions lying in this side face. This effect is shown in Fig. 12.

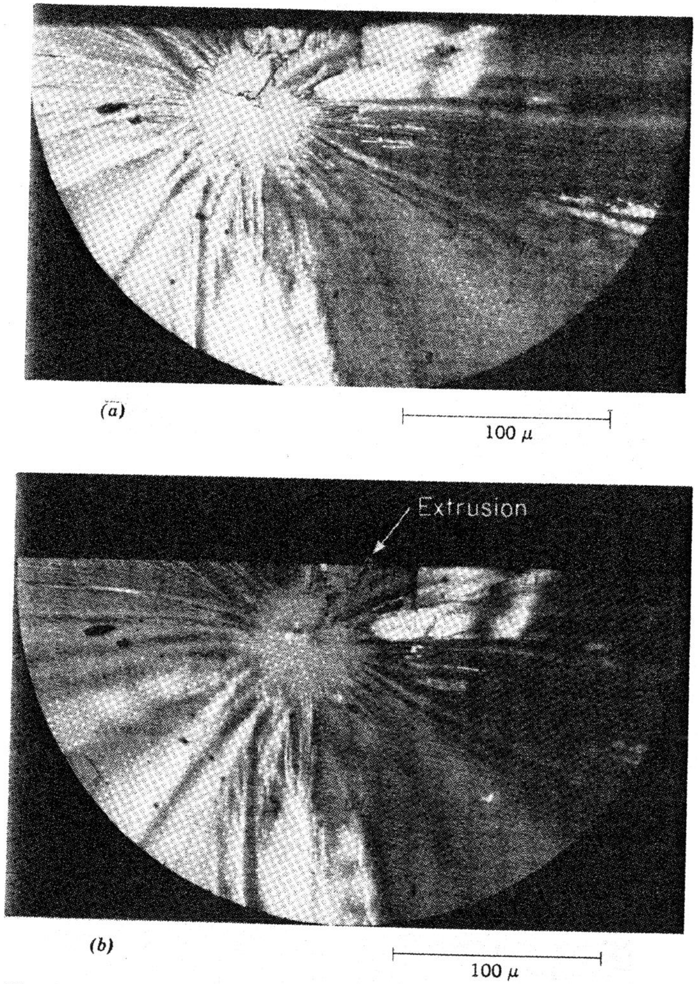


Fig. 9. Fatigue fracture nucleus in cross section of LiF crystal. (*a*) Focused on nucleus. (*b*) Focused on extrusion.

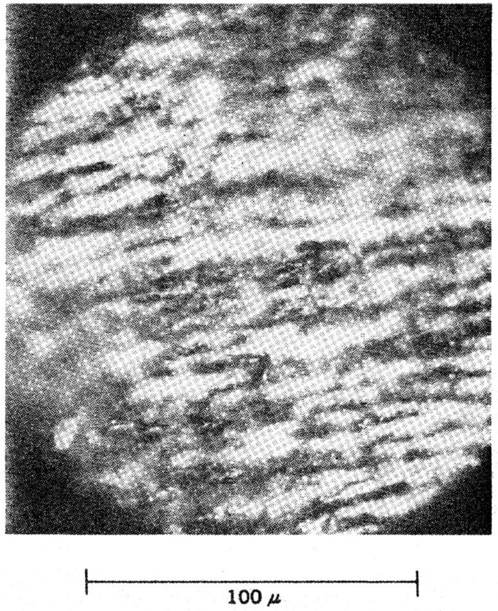


Fig. 10. Extrusions on AgCl crystal, $s_{\max} = 1400$ psi.

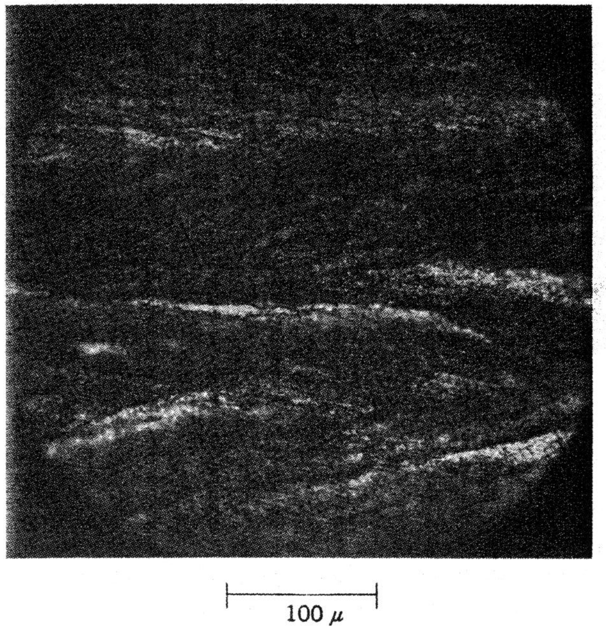


Fig. 11. AgCl at $s_{\max} = 2600$ psi.

Figures 13 and 14 are views of the top and side faces of a KRS-5 crystal. Again in contrast to the LiF, there is ample evidence of slip taking place and of the wavy nature of the slip lines. In this crystal, the fracture follows the slip plane of highest resolved shear stress and then runs more or less perpendicular to the direction of maximum tensile stress. In Fig. 13, a secondary slip system is operative which is responsible for the majority of the surface markings. The view of the side face (Fig. 14) shows the distorted nature of the surface where the principal slip direction emerges. Figure 15 shows extrusions that appear in the spearhead of a crack in KRS-5 along the trace of a slip band. Figure 16 is a lower magnification photomicrograph showing the crack and slip band.

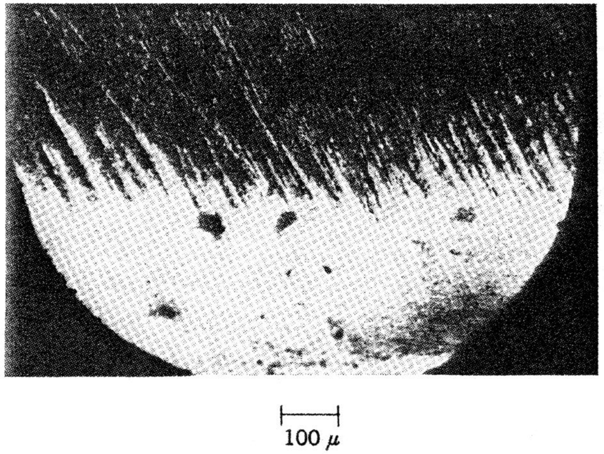


Fig. 12. Side face of AgCl crystal.

Discussion of Results

It is apparent from the results presented in Table 2 that the fatigue behavior of LiF and NaCl differs markedly from that of AgCl and KRS-5. The fatigue behavior of AgCl and KRS-5 resembles that normally obtained with cubic metals. However, LiF and NaCl more nearly resemble the behavior of single crystalline zinc.[15] Inasmuch as the obvious difference between the two groups is that cross slip is easy in the easy-to-fatigue specimens and quite difficult in the difficult-to-fatigue specimens, it is reasonable to expect that a causal relationship may exist between the two sets of observations. This suggestion is, however, worthy of much further experimental evaluation.

The intersecting slip mechanism of Cottrell and Hull has *not* operated

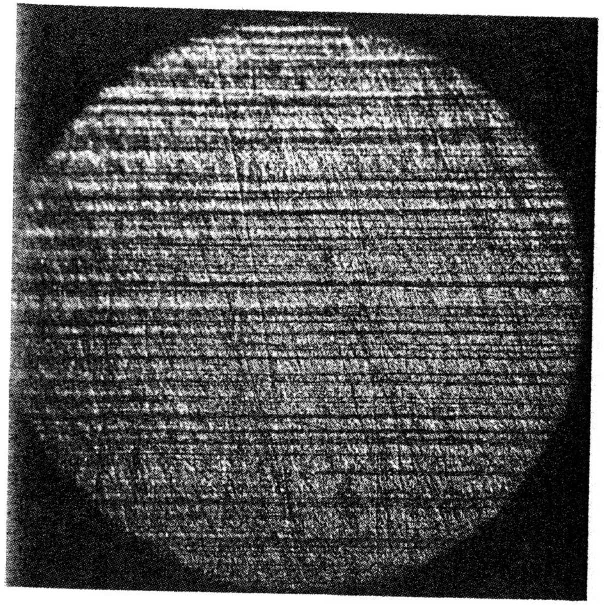


Fig. 13. Top face of KRS-5 crystal.

Fig. 14. Side face of KRS-5 crystal.

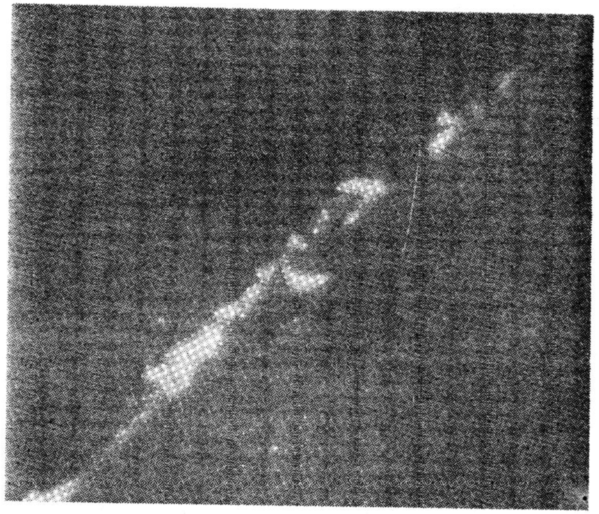


Fig. 15. KRS-5 extrusions.

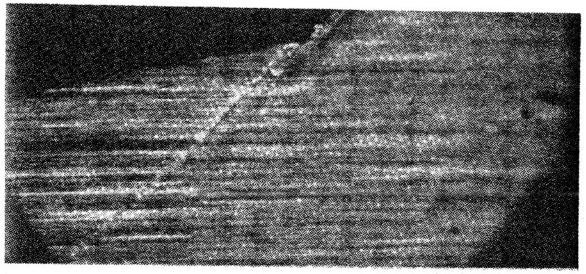


Fig. 16. KRS-5 extrusions. Specimen bent into plastic range prior to cyclic loading.

to produce easy fatigue in our experiments. Although conditions were extremely favorable for the operation of the C-H (Cottrell-Hull) mechanism in both the LiF and NaCl crystals (the slip planes were in the ideal C-H orientation), fatigue failure was extremely difficult to obtain. It appears reasonable to conclude that the C-H mechanism is not important in the production of fatigue failure. Some indications were obtained in preliminary tests with LiF, in which the orientation was varied, which implied that the fatigue resistance of the C-H orientation was less than that of two other orientations of the crystals relative to the applied stress

((001) [110] and (110) [$1\bar{1}0$]). However, it is not believed that the C-H mechanism is responsible for the single fatigue failure that occurred in this orientation. As shown in Fig. 9*a*, the fatigue fracture in one specimen appears to have nucleated *below* the specimen surface, contrary to the prediction of the C-H mechanism. (The fatigue crack in the other LiF specimen appeared to be surface-nucleated.) Furthermore, only one extrusion was discovered in examination of LiF and NaCl crystals despite careful observation. This case will be discussed later and is more nearly in agreement with the Mott mechanism than the C-H mechanism.

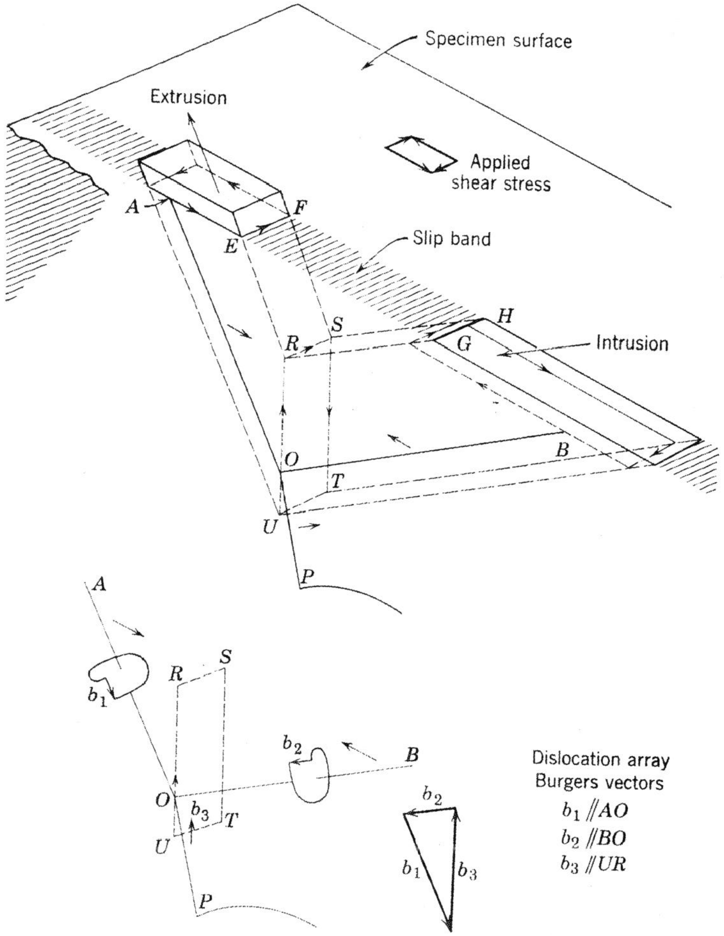
Several other observations were made of differences in behavior between the group LiF and NaCl as compared to AgCl and KRS-5. Slip bands were observed on the surfaces of both AgCl and KRS-5 (Figs. 10 to 14). However, they were not observable as steps on polished surfaces of LiF and NaCl in regions somewhat removed from the fracture surface. This is not to say that slip bands are not produced in alternating stressing in LiF and NaCl, because they are, as evidenced by the sequence of dislocation etch patterns shown in Fig. 7. However, the slip bands revealed by the dislocation etch pits do not form marked surface offsets. It is reasonable to conclude that the net slip per band is limited. These observations are completely consistent with the relative absence of cross slip in LiF and NaCl and the ease of cross slip in AgCl and KRS-5. As Seeger [16] and Mott [17] have noted, cross slip in Stage III of f.c.c. stress-strain behavior is responsible for the development of long slip bands, each slip band comprising a marked amount of slip per band. Prior to Stage III glide, there is a predominance of fine slip, uniformly distributed, with a relative absence of cross slip.

This consequence of cross slip may be the important factor in the development of extrusions and intrusions along slip bands and ultimate fatigue failure. To check this suggestion, the specimens were examined for the existence of extrusions and intrusions. Extrusions were found in both AgCl and KRS-5, as evidenced by Figs. 10 and 15. However, only one extrusion was found in examination of LiF and NaCl. (This case will be discussed later.) Incidentally, beautiful examples of extrusions and intrusions in AgCl have been reported by Forsyth.[14] On the basis of his observations of extrusions in AgCl, Forsyth was able to show that Mott's cross-slip mechanism for extrusion, which requires the existence of a crack beneath the extrusion, does not operate in AgCl. It is apparent from the observation of parallel extrusions and intrusions that the two are connected and that the flow of matter is achieved through the cooperation of neighboring intrusion-extrusion pairs. It is not difficult to understand why cross slip should aid the process of intrusion-extrusion. It is possible to devise many mechanisms in which the passage of screw dislocations about a closed path produces connected intrusion-extrusions. Several such mechanisms are illustrated in Fig. 17. Most of the intrusion-extrusion mechanisms involve the production of spiral prismatic dislocations. A mechanism by which intrusion-extrusions can be formed in the same slip plane and in which the extent of the extrusion is unlimited is illustrated in Fig. 17*a*. In this case, there are two screw dislocations of Burgers vectors b_1 and b_2 common to the primary slip plane. These dislocations *AO* and *BO* move in opposite directions in the primary plane during a part of the first half-cycle of stress. They then cross slip along their respective cross-slip planes during the same half-cycle of stress so that their common point *O* moves in the same direction *RS*. Another dislocation having the Burgers vector $-(b_1 + b_2) = b_3$ joins these screw dislocations at the node *O* and oscillates about the point *P* in the plane *RSTU*, where it is pinned because the dislocation moves out of the plane *RSTU* at this point. Upon application of the other half-cycle of stress, the screw dislocations move so as to complete the circuit. The consequence of this dislocation motion is an intrusion along the direction b_2, an extrusion along the direction $-b_1$, and continual slip along the plane *RSTU* in the direction b_3. Actually, the slip b_3 occurs on successive atomic planes, spaced at a distance b_2 apart, with each successive cycle of stress. To be specific, for a face-centered cubic metal, the following parameters may apply:

$$
\begin{array}{llll}
b_1 = [10\bar{1}] & \qquad & \text{Plane } AOB & = [111] \\
b_2 = [0\bar{1}1] & \qquad & \text{Plane } EFSR & = [1\bar{1}1] \\
b_3 = [\bar{1}10] & \qquad & \text{Plane } GHSR & = [\bar{1}11]
\end{array}
$$

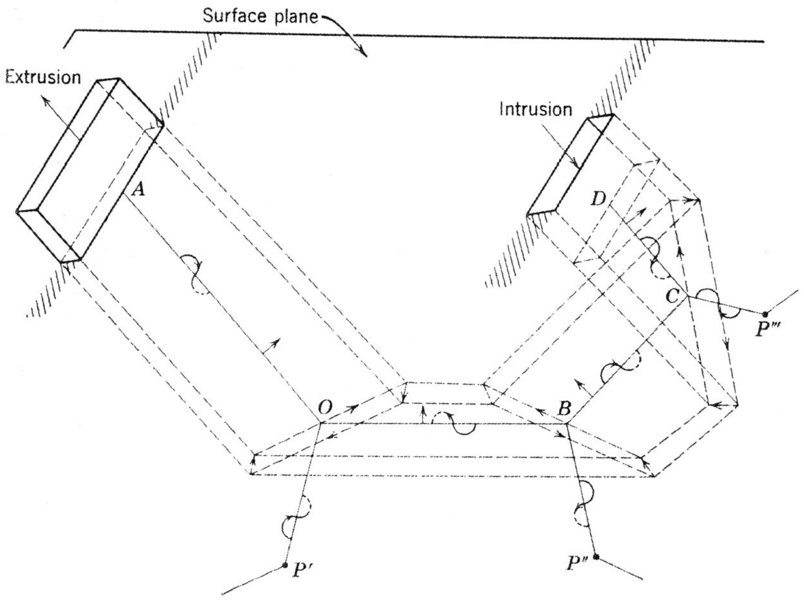
It is apparent therefore that in crystals in which cross slip occurs the production of intrusion-extrusion pairs should be much more facile than in crystals in which cross slip does not occur. This is in agreement with our observations.

It is interesting to note that our experimental results comprise a further basis for rejection of the speculation that vacancies are responsible for rapid fatigue failure. We have observed what we believe to be the production of excess lithium atoms at the surface of reverse-stress-cycled lithium fluoride. We find after cycling that a white deposit is developed on the surface which dissolves in alcohol. Inasmuch as LiF is insoluble in alcohol and lithium is soluble in alcohol and has a white color, we have tentatively identified the white deposit as lithium (more probably LiOH). If it is indeed lithium hydroxide, then the production of excess lithium at the surface must involve the simultaneous production of Li vacancies or F interstitials. The former is more likely to occur. Now, Keith and Gilman [18] have found evidence for the presence of point-defect clusters

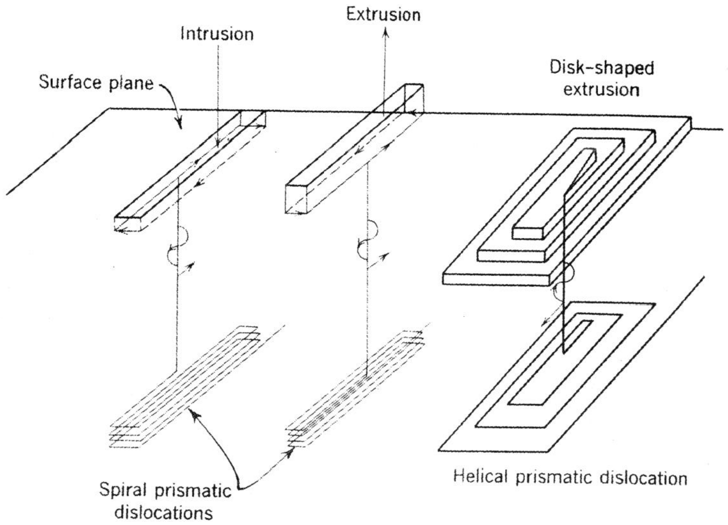


(*a*)

Fig. 17. Intrusion-extrusion mechanism by dislocation cross slip. (*a*) Intrusions and extrusions lying in same slip band, single-node nonstrain-hardening mechanism. (*b*) Extrusions and intrusions lying in parallel slip bands, triple-node nonstrain-hardening mechanism applicable to face-centered cubic structures. (All four slip planes active in cross-slip production of intrusion-extrusion tunnel.) (*c*) Extrusions and intrusions. Strain-hardening, zero-node mechanism. Length of extrusion or intrusion limited by build-up of spiral or helical prismatic dislocation at base.



(*b*)



(*c*)

Fig. 17. (*Continued*).

in reverse-stress-cycled LiF. Strangely, they cite this observation as evidence for the vacancy mechanism of fatigue as against the other mechanisms, despite the fact that they never cycled a LiF specimen to failure. With positive evidence for vacancy production and cluster formation in the absence of fatigue failure, it appears to be reasonable to conclude that vacancies are *not* exclusively involved in the fatigue failure mechanism.

Another observation which deserves discussion is that the static strengths of both NaCl and LiF have been enormously increased. This effect is not the same as the coaxing effect observed in metals or the effect obtained by stressing below the yield point, but it is the same as that observed in aluminum by Broom, Molineux, and Whittaker,[19] who gradually increased the stress in the range above the yield point and found that the endurance limit could be raised above the original tensile strength. It is believed that the marked strengthening found in LiF and NaCl is due to the fact that fatigue failure does not prematurely interrupt the strengthening process in these crystals.

There now seems to be little reason to doubt that easy cross slip is a prerequisite for easy fatigue failure. Our results, taken together with those of Fegredo and Greenough[15] on zinc and Wadsworth[20] on cadmium, can be interpreted most convincingly on this basis. It may be asked whether the effect of cross slip is mainly on the initiation of the fatigue crack or on the growth of the fatigue crack. There seems to be an ambiguity in the literature about the meaning of the terms crack initiation and crack propagation. It is said that the crack initiates in the first few percent of the fatigue life and propagates thereafter. There is no clear-cut demarcation between crack initiation and propagation. It is evident from observation of the formation of cracks that they develop along slip bands for a major fraction of the fatigue life of the specimen. For example, Thompson, Wadsworth, and Louat[21] show a crack that eventually caused the final failure lying in a slip band after 42% of the life. It seems reasonable to assume that the same process that initiated the smaller cracks in the slip bands operates during the growth of these cracks along slip bands. There is no difference in the appearance of these cracks other than their increased length. It is strange therefore to find implications in the literature that Head's[22] mechanism of crack propagation based on Orowan's model (which requires the strain hardening and fracture of volumes at the tips of cracks) can account for the major fraction of the fatigue life. There are several things about this model that cannot be reconciled with experiment: (1) Cracks do not deviate from slip planes until a major fraction of the fatigue life has transpired; (2) materials that do not fatigue readily show no saturation in strain hardening (Figs. 3 and 4), and some of those that do fatigue readily reveal such saturation,[8,23] contrary to the expectation based on Orowan's model; and (3) we have observed that the marked strain hardening exhibited by both NaCl and LiF did not aid fatigue in these materials more so than in AgCl, for example, which hardens to a much smaller degree. It is difficult to believe therefore that Head's mechanism is important to the crack growth that takes place during the major fraction of the fatigue life. It is far more reasonable to assume that crack growth requires the continuous extrusion-intrusion process to take place at the spearhead of a slip-band crack until the growth can occur more rapidly under the influence of the normal tensile stresses in the last few percent of the fatigue life. It is evident that more experimental evidence on the nature of crack growth is needed. It is suggested that transparent crystals are ideal materials for studying this phenomenon. It should be noted that the critical crack size at which failure becomes rapid in these crystals is small. The smooth area shown in Fig. 9*a* is produced at the end of the fatigue cycle in a very small fraction of the fatigue life. The critical crack size must be smaller than the diameter of the smooth region. The latter is 50 μ. It is estimated that the critical crack size is less than 10 μ.

It is interesting to note the fracture pattern associated with the single observation of an extrusion in fatigued LiF. Figures 9*a* and *b* reveal the pattern in the vicinity of the crack nucleus. There is a line that extends to the surface along a $<110>$ direction starting at the crack nucleus. At the intersection of this line with the surface, there appears to be an extrusion, which is best revealed in Fig. 9*b*. This pattern is just that to be expected if Mott's mechanism operated. That is, there seems to be an extrusion along a slip plane with an internal crack at the base of the extrusion. However, all other crack nuclei were found to exist at the surface. It is believed, therefore, that this is an isolated case.

The picture of the fatigue process which we believe emerges from our experiments is that crack initiation occurs by the formation of intrusions adjacent to extrusions. The proximity of intrusions and extrusions observed by Forsyth suggests that there is a connection between some intrusion and some other extrusion. In other words, the pile-up of material beneath an intrusion is believed to be transported to the region beneath an extrusion. Cross slip appears to be the easiest mode of slip that would yield such transport. This intrusion-extrusion process is believed to continue to occur at the spearheads of the cracks to cause propagation of these cracks along the slip bands. This process is believed to be interrupted during the last few percent of fatigue life. By this time, the crack has grown to a sufficient extent to propagate a local ductile fracture normal to the tensile stress. Hence, cross slip is believed to exert its effect on fatigue life, not through its effect on crack initiation but upon crack propagation. Actually, in our model, there is no transition point between crack initiation and propagation. It is believed to be a continual process of sequential intrusion-extrusions from the start of the fatigue cycle.

Conclusions

1. Easy cross slip is required for fatigue failure to occur in its normal fashion.
2. The Cottrell-Hull mechanism for the production of extrusions does not operate in these experiments and, in general, is not likely to operate.
3. The Mott mechanism may operate.
4. Vacancies are not involved directly in fatigue crack formation.
5. The Head mechanism of crack propagation is not consistent with the observations made in these experiments.

It is suggested that cross slip is not only important to crack initiation but is equally important in crack propagation and hence fatigue life. Although no evidence was obtained to disprove the Mott mechanism of crack initiation which requires cross slip, the simultaneous observation of extrusions and intrusions is believed to imply that extrusions do not require cracks beneath the surface. It is suggested that cross slip provides a means for the relatively easy transport of the piled-up material beneath an intrusion to the region beneath an extrusion.

It is further suggested that crack growth occurs by intrusion-extrusion at the spearheads and that cross slip is thereby required. It is believed that this is the process that takes time and is responsible for the fatigue life measured.

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