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ON THE MECHANISMS OF FATIGUE CRACK GROWTH

P. Neumann, H. Vehoff and H. Fuhlrott*

INTRODUCTION

In recent papers [1] an experimental technique was described to produce a well defined crack geometry by stage II crack growth in fcc single crystals. This technique was used now for studying further details of stage II growth in bcc metals, environmental effects on stage I and stage II growth, and cyclic brittle crack growth.

EXPERIMENTAL

Most mechanical tests were performed in a servo hydraulic testing system with advanced controlling capabilities. Especially for studying the initiation of brittle fracture it is necessary to control the machine with the instantaneous plastic elongation as a controlling variable: Since the elastic compliance of the specimen changes during crack growth, it is necessary to measure the compliance during unloading in the tensile phase of every cycle by an appropriate analog circuitry. From the elastic compliance signal, the load signal, and the total elongation signal the plastic elongation signal was obtained by analog multiplication and subtraction. This plastic elongation signal cannot be used as the controlling signal without further precautions since unavoidable fluctuations would cause fast unloading along the elastic slope of the specimen. With a strongly non-linear analog control the problem could be solved as described elsewhere [2]. The speed of this analog processing is far beyond that of a digital process controller.

In order to avoid inhomogeneous hardening at the notch, all crystals were cyclically hardened before notching. The absolute value of the mean load was always less than 5% of the load amplitude (depending on details of the control). The specimen could be tested in laboratory air or in a vacuum of 1.3mPa (10^{-5} Torr).

For direct observation of the crack tip during cycling inside a scanning electron microscope (Cambridge Instruments, Mark IIA) we modified a 2KN tensile stage from Cambridge Instruments for cyclic operation as described elsewhere [3].

STAGE II CRACK GROWTH IN BCC METALS

An experimental technique was described in [1] by which plane fracture surfaces and straight crack fronts can be produced in fcc single crystals by stage II crack growth. By choosing specimens of appropriate orientation and shape (cf. Figure 1) undisturbed plastic zones can be observed

*Max-Planck-Institut fuer Eisenforschung GmbH, 4000 Duesseldorf, FRG.

on the side faces of the specimen. In copper single crystals all kinds of details of the alternating slip model for stage II crack growth could be verified experimentally [1].

In order to find out whether in bcc metals the same mechanism is operative, similar experiments were carried out with bcc Fe3%Si single crystals. The slip geometry is different from that of fcc metals since there is no well defined crystallographic slip plane. Accordingly the conditions for obtaining plane fracture surfaces and straight crack fronts are not as stringent as in the case of copper. Figure 1 shows one of the possible configurations. In this special case the Burgers Vectors of the active slip systems are almost parallel to one of the side faces of the crystal. Observation of the crack tip on this side face revealed most details: Figure 2 shows a sequence of SEM micrographs of such a crack tip during straining inside a SEM. The pictures of the sequence were taken at constant plastic strain intervals of 0.4µm. Only a small fraction of the tensile phase of one single cycle is covered by the sequence. The black triangle in every picture is the lower part of the V-shaped crack tip (cf. [1]). The crack propagates in each picture from the top to the bottom. The upper row (pictures 1 to 3) shows the strengthening of the slip line, which emanates from the crack tip apex and runs to the lower right. The corresponding motion of the right hand side of the crack towards the lower right is clearly visible, especially if the position of the crack tip apex is examined in relation to the lowest slip line on the left. In picture 3 work-hardening has stopped the activity on the old slip line and another slip line, which emanates from the crack tip apex and runs to the lower left, is activated and strengthens in pictures 4 to 6. Again the corresponding motion of the left hand side of the crack towards the lower left is visible. The process continues with the activation of another slip line running to the lower right and so on. To the knowledge of the authors this is the first time that the alternate slip mechanism, which was proposed for fatigue as early as 1967 [4], could be observed in such a direct way.

ENVIRONMENTAL EFFECTS ON STAGE I AND STAGE II CRACK GROWTH

Stage I crack growth is defined as being parallel to the most highly stressed slip plane. It is generally accepted that only one slip system is operative during stage I. On the other hand it is well known, that slip on one slip system alone cannot produce a steadily growing crack, because the to and fro motion of dislocations produce slip steps but also annihilate them on the way back. Even if cross slip is assumed to occur, monotonically growing cracks can only be formed, if a sub-surface cavity exists [5] or a highly improbable dislocation configuration [6] is assumed to be present. Therefore it is most likely that surface reactions with the environment deteriorate the new surface of each slip step such that the surface is not annihilated during back slip but a crack is formed. In such a way a steadily growing crack can develop [7]. In order to be able to check, whether reactions with the environment are a necessary condition for stage I crack growth, conditions must be established, under which stage I crack growth occurs reproducibly and independent of pre-existing soft zones parallel to slip planes, which were produced by cyclic softening. Such zones are well known in pure metals as persistent slip bands and are even more pronounced in hardened alloys [8].

Therefore it was tried to obtain stage I crack growth in copper single crystals not only after the initiation phase but also after stage II crack growth up to arbitrary crack lengths. In order to get well defined stage II crack growth, specimens of the shape shown in Figure 1 were used. In fcc crystals it is necessary only, to make the specimen axis parallel to (100) and the root of the notch parallel to (011). For such a geometry and for crack opening displacements δ in stage II of up to 50µm no K_I factors are known. Therefore the load P50(a) for obtaining a δ of 50µm at the crack length, a, was measured. All other loads at the same crack length are then given in units of P50(a). Stage I crack propagation was obtained in air at loads less than 0.25·P50(a) for all crack lengths a. Under these conditions the crack propagation rate was less than 10nm/cycle. Figure 3 shows the fracture surface under various conditions. Note that at 0.3nm/cycle there is strictly stage II growth in vacuum. When the environment is changed to air under the same loading conditions the crack growth rate is increased by a factor of ten but in spite of that the mode changes to stage I. Also in all other experiments with crack rates down to 0.1nm/cycle we observed only stage II growth in vacuum. This is taken as evidence that reactions with the environment are indeed essential for stage I crack growth in copper.

In the course of these investigations it was found, that also in the range of stage II crack growth rates, i.e. from 30nm/cycle to 1µm/cycle, the growth rate in vacuum was only one tenth of that in air under otherwise identical conditions (Figure 4). In the literature rewelding has often been discussed as a possible reason for such environmental effects on stage II growth rates [9], but rarely critical experiments were carried out, which unambiguously proved that rewelding really happens. We performed two experiments of this kind:

The crack was propagated almost throughout the specimen until the decrease of the remaining cross-section during one cycle was comparable to the remaining cross-section itself. Figure 5 shows the hysteresis loops (load versus plastic elongation) obtained with such small remaining cross-sections. Going through the points d,e,f,g,h in Figure 5 represents one cycle. The decrease of the load during the tensile phase (from points d to e) due to the decreasing cross-section is obvious. After crack closure, however, the load at h is larger than it was before closure at the end of the preceding tensile phase at e. This means, that the cross-section must be larger now, i.e. rewelding at the crack tip must have happened. Other effects cannot be made responsible for the fact that the load at h is higher than at e. At h the crack is sharp and at e the crack had a V-shaped tip. If the stress concentrations in this extremely fully plastic situation (very small cross-section of some 0.5mm²) are at all different for both crack geometries, they should be stronger at h than at e. Therefore the load, which is necessary for yielding, should be smaller at h than at e. In a similar way it can be shown that the Bauschinger effect and residual stresses would make a contribution of the wrong sign. In addition it was found, that the effect is weaker in air, which can also be regarded as evidence for rewelding.

Further it was found that the growth rate increased with the testing frequency in air and in vacuum. This cannot be explained by the time required for the formation of an oxide layer, which reduces rewelding, because this would give rise to an opposite frequency effect. Therefore the time intervals for opening and closing the crack were varied independently. The opening time was not found to be relevant for the crack growth rate. The growth rate decreases, however, with increasing compression time. Again, this is regarded as direct evidence for rewelding. These experiments are in progress and more quantitative details will be given at the time of the conference.

The existing results do allow one important conclusion: The well known poor visibility of striations in vacuum, which we also found at growth rates up to 30µm/cycle (1), is simply a consequence of the enhanced rewelding, which will most likely occur locally and thus destroy any previous structure. Therefore one cannot simply conclude from the absence of striations that the alternate slip mechanism was not responsible for the crack growth. This illustrates that one cannot deduce the growth mechanism from the mere appearance of fracture surfaces and without further critical experiments.

CYCLIC BRITTLE CRACK GROWTH

Flat facets on a fatigue fracture surface are very often identified with cleavage fracture just from their appearance although crack growth along the very narrow cyclically softened bands in age hardened alloys can look very much the same [8]. In order to examine whether cyclic i.e. non-catastrophic cleavage is possible, experiments were designed to study the controlled initiation of cleavage cracks during cyclic loading in single crystals of Fe3%Si.

The previously discussed stage II crack growth in Fe3%Si is only obtained at small enough crack opening rates $\dot{\delta}$ of less than 10µm/s. If $\dot{\delta}$ is increased stepwise, a twofold increase in $\dot{\delta}$ is sufficient to change the fracture surface from ideally ductile stage II crack growth along (110) planes (upper, dark part of Figure 6 with straight and parallel striations) to a stable cyclic cleavage mode along facets of (100) cleavage planes (serrated lower part of the fracture surface in Figure 6). The ratio of the plastic elongation amplitude and the crack growth per cycle is in the cleavage mode about a tenth of that during ductile growth, which undoubtedly identifies this growth mode as non-ductile. In this stable cyclic cleavage mode the crack advances in well defined increments in every cycle as can be seen from the dark lines on (100) facets in Figure 6. Further there is evidence from observations of the crack tips on the side surfaces, that there is some plasticity accompanying this type of fracture. If $\dot{\delta}$ is increased by another factor of two, catastrophic brittle failure occurs with much less accompanying plasticity. Catastrophic cleavage was found to be initiated at the beginning of the tension phase, when the crack was closed. After a V-shaped crack tip has formed at low plastic elongation rate, 500 times larger rates can be applied during the remaining tensile phase without brittle failure. More experiments are in progress to study the conditions for the onset of brittle fracture under even better defined conditions with ductile and cleavage fracture both on a (100) plane.

CONCLUSIONS

1) The alternating slip model of stage II crack growth was observed in situ during cycling inside a SEM in Fe3%Si crystals.
2) Stage I crack growth could be obtained in copper crystals at arbitrary crack lengths by sufficient load reduction at growth rates below 10nm/cycle.

3) Stage I growth was never observed in a vacuum of 1.3mPa (10^{-5} Torr) at growth rates down to 0.1nm/cycle.
4) Stage II crack growth is strongly environment dependent up to large growth rates (30µm/cycle) due to rewelding.
5) Direct evidence for rewelding was obtained from a load increase after compression and a reduction of growth rates by an increase of the compression times.
6) Cyclic cleavage could be initiated at arbitrary crack lengths in Fe3%Si crystals by an increase in the rate of loading.
7) The accompanying plasticity in the cyclic cleavage mode is less than 10% of the plasticity necessary for ductile fracture.

ACKNOWLEDGEMENTS

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REFERENCES

1. NEUMANN, P., Acta Met. 22, 1974, 1155 and 1167.
2. NEUMANN, P., to be published.
3. NEUMANN, P. and VEHOFF, H., to be published.
4. NEUMANN, P., F.f.Metallkunde, 58, 1967, 780.
5. MOTT, N. F., Acta Met. 6, 1958, 195.
6. McEVELY, A. J. and MACHLIN, E. S., Fracture, p.450, Wiley, New York, 1959.
7. THOMPSON, N., WADSWORTH, N. J. and LOUAT, N., Phil. Mag., 1, 1956, 113.
8. DUQUETTE, D. J. and GELL, M., Met. Trans., 2, 1971, 1325.
9. ACHTER, M. R., ASTM STP 415, 1967, 181.

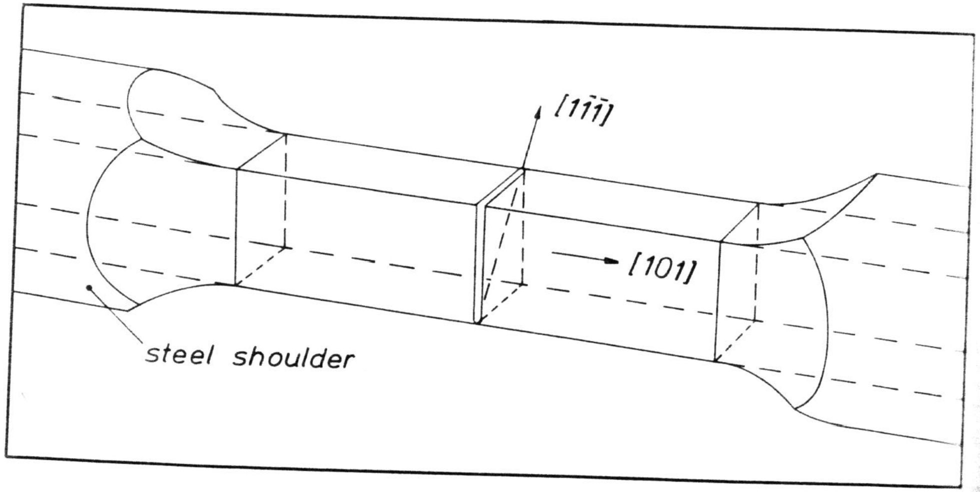

Figure 1 Shape and orientation of the Fe3%Si specimens. The copper specimens are of different orientation only: Specimen axis parallel to [100], root of the notch parallel to [011]. The size of the cross-section was 6 by 6 mm.

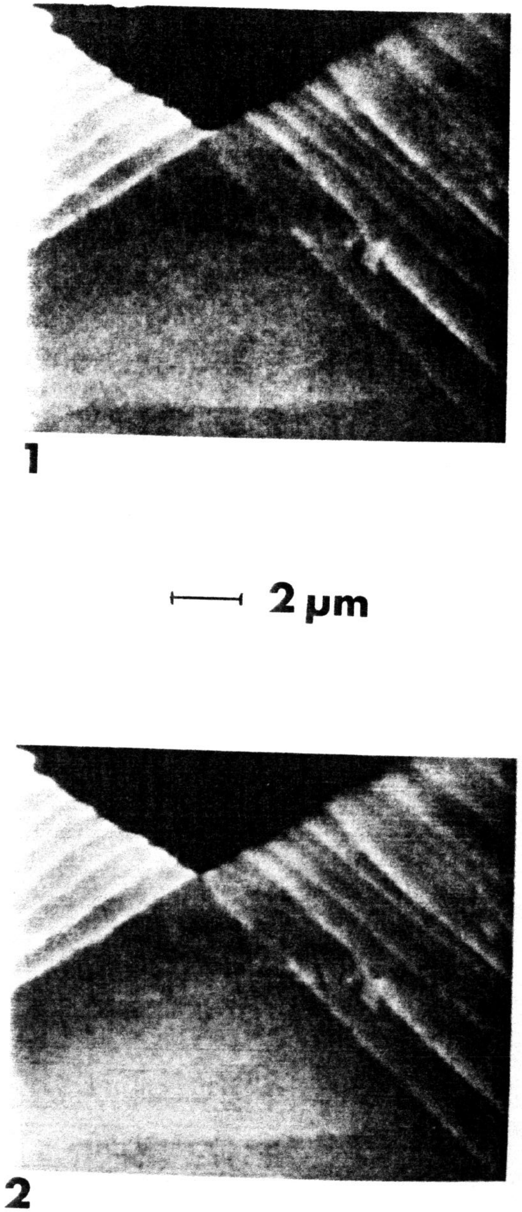

Figure 2 Sequence of SEM micrographs showing the alternate slip mechanism of stage II crack growth in an Fe3%Si single crystal. Angles are distorted due to the oblique scanning. For details see the text.

Figure 2 - continued

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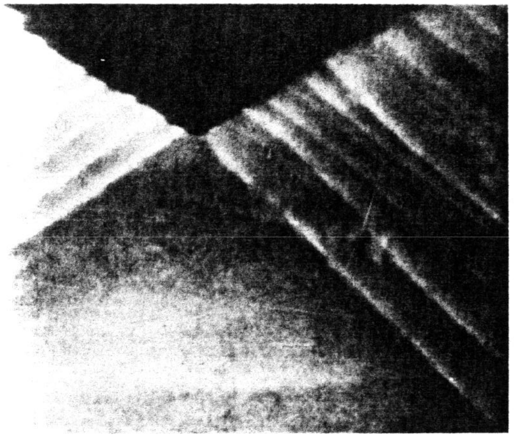

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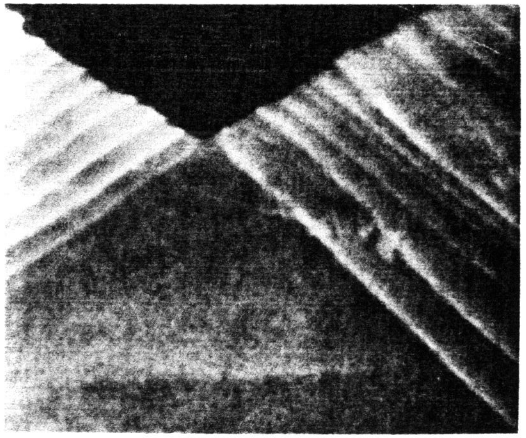

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Figure 2 - continued

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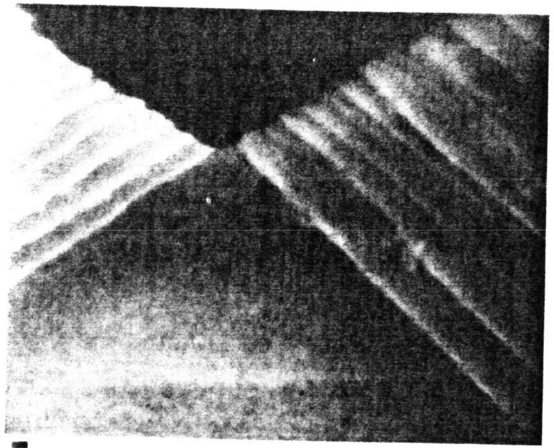

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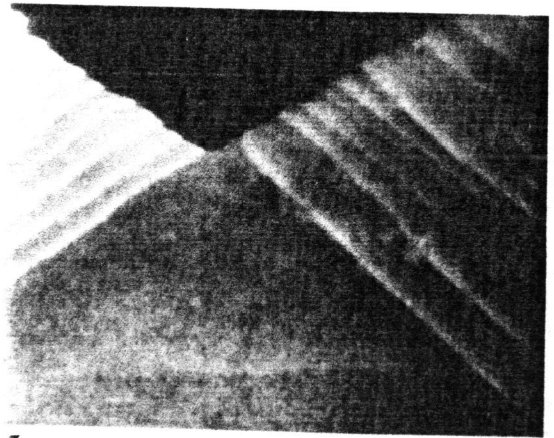

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Figure 2

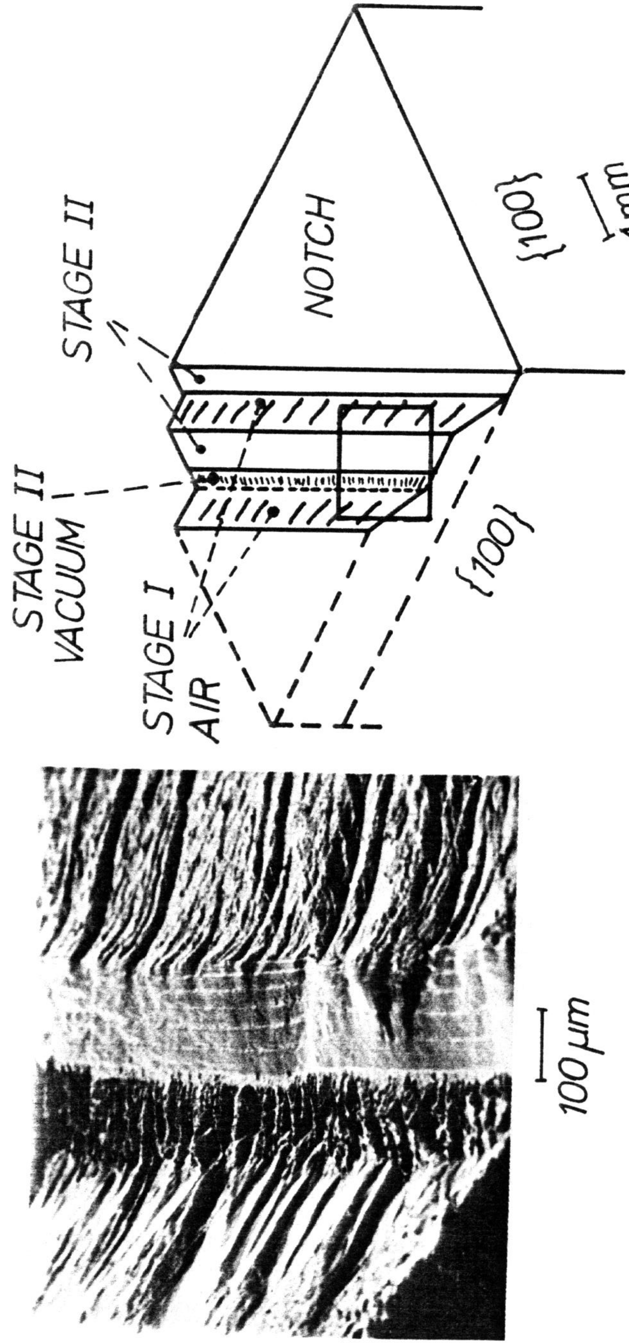

Figure 3 SEM micrograph of the framed area in the schematic drawing. Crack growth from the right to the left. Growth data: stage II growth starting from the notch, (not shown in the micrograph), stage I (air, P=0.2•P50, 3nm/cycle), stage II (air, P=P50, 50μm/cycle), stage II (vacuum, 0.2•P50, 0.3nm/cycle(!)), stage I (air, P=0.2•P50, 3nm/cycle).

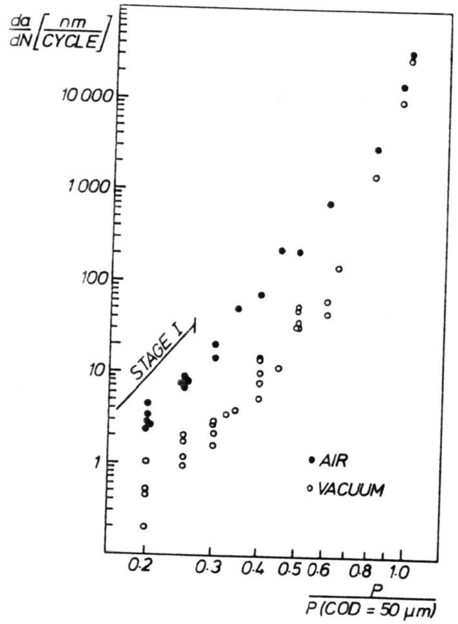

Figure 4 Crack growth data of [100] copper crystals in air and vacuum. Stage I growth is observed in the indicated load range only with air as environment.

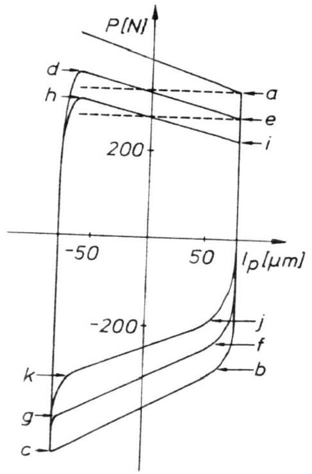

Figure 5 Hysteresis loops of copper specimen with very small remaining cross-section showing evidence of rewelding. For details see text.

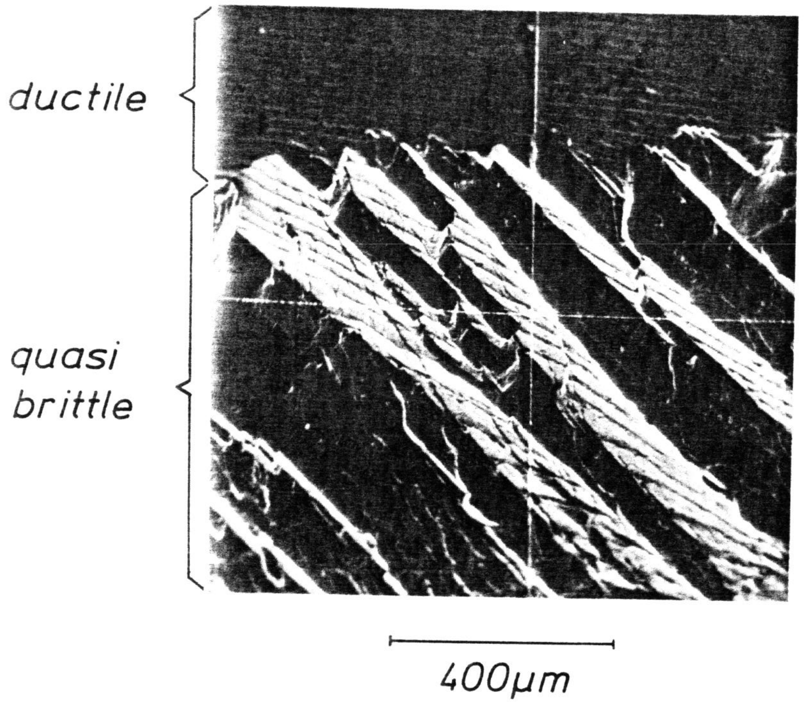

Figure 6 Fracture surface if a Fe3%Si crystal. Crack growth from the top to the bottom. Transition from ductile growth (upper part) to quasi brittle growth due to change of rate of loading. The crossing two white lines were made artificially for the purpose of quantitative stereo evaluation.