ON THE INFLUENCE OF STRUCTURAL DAMAGE UPON THE FRACTURE TOUGHNESS OF METALS

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ABSTRACT

In most elastic-plastic materials, a propagating crack is preceded by plastic deformation occurring in a zone in front of the crack tip.This deformation can be regarded as a structural damage that facilitates the growth of the crack..If temperature and deformation speed are properly selected, no plastic zone of that kind will appear spontaneously. Then the size of the zone can be determined beforehand. For instance, measuring the fracture toughness at high strain rate in Mode I, a very sharp plastic zone near the crack is obtainable by steady-state loading in Mode III.

In this way, fracture toughness for some structural steels, brass, and aluminum alloys has been studied as influenced by

a) general yielding
b) local yielding in front of the crack
c) residual deformation due to loading history and
d) creep damage.

These experiments show that the fracture toughness decreases as a function of plastic strain, when this is comparatively small, then gradually assumes a constant value. If the yielding is limited to a small area near the crack tip, the plastic zone must be very sharp to produce any noticeable effect on the toughness.

The determination of the connection between toughness and plastic strain might become an instrument for giving approximate solutions to problems where linear fracture mechanics is not to be applied because the spontaneous plastic zone is not sufficiently small.

Introduction.

Continuum mechanics provides the theoretical basis required for the designer in his calculations in the field of mechanics of materials. Most practical applications of solids, however, deal with polycrystalline aggregates rather than with single crystals. Consequently, the physical properties of large regions of the solid, e.g. the grain boundaries, differ considerably from the properties of the undisturbed, continuous crystal grating. For a structure that is being designed, fracture mechanics supplies a method to take into account discontinuities in the form of defects in the material.

Often reliable results can be expected from the application of linear fracture mechanics to metals containing defects, which may seem remarkable as the theory behind it is virtually based on the concept of continuity. On the other hand, it sometimes happens that linear theory is not sufficient because properties of the material, type of loading, and dimensions of the structure do not always fulfil the requirements for that theory to be applicable.

In order to improve the theory, one may look for further information concerning the physical behaviour of defects. The interest of the designer is to prevent the defects from growing so much that the safety of the construction is at stake, whereas the physicist may inquire about the phenomena involved in crack growth so as to gain a better understanding of the mechanisms governing the crack propagation.

In most elastic-plastic materials, a propagating crack is preceded by plastic deformation occurring in a zone in front of the crack tip. This deformation can be regarded as a structural damage that facilitates the growth of the crack.

The scope of this paper is to investigate the interaction between plastic deformation applied in the region near to the crack tip before the crack begins to move and in such a way that the magnitude of the deformation can be selected arbitrarily.

Practical approach.

In the present study, the fracture toughness has been measured as influenced by different types and magnitudes of plastic deformation. The strain range is between 0 and 16.3 % and the following types of deformation have been tried.

1. General yielding
2. Local yielding in front of the crack
3. Residual deformation due to loading history
4. Creep deformation.

Fracture toughness in plane strain is denoted by K_{IC} and should be determined according to the ASTM requirements (ref. [1]). In this investigation, plane strain conditions are not always fulfilled. Hence, the values obtained are termed K_Q instead.

1. General yielding.

Experimental procedure.

Plane plate specimens were tensioned to strains between 0.7 and 16.3 %. From each of these plates, 6 samples were taken in the form of "compact tension (CT)" specimens, Fig. 1 & Fig. 2. The CT specimens were then run in fatigue, the load varying between 0 and 3000 N in tension. The crack starts at the bottom of the notch and proceeds perpendicular to the large surfaces of the specimen. After it has reached an appropriate length, the fatigue treatment is stopped, two edges are screwed onto the specimen, a "clip-gauge" is inserted between them, and the "crack opening displacement (COD)" is recorded vs. the applied force in a tensile test.

The experimental set-up is shown in Fig. 3. The specimens fracture as can be seen from Fig. 4. From the COD-force diagram the K_Q values are easily obtained.

Materials.

Five materials were tested with the following composition and properties.
Material no. 1. Domnarvet DOMEX 400, C 0.14 %, Si 0.44 %, Mn 1.54 %, P 0.025% S 0.024 %, N 0.012 %, Nb 0.031 %. Lower yield point 435 MN/m^2, upper yield point 446 MN/m^2, ultimate tensile stress 566 MN/m^2.
Material no. 2. Domnarvet DOMEX, obtained especially for this project, starting from DOMEX 420, rolling and heat treatment. Contents: Cu 0.7 %, Ni 0.7 %. Yield stress σ_s=550 MN/m^2, ultimate tensile stress σ_F=750 MN/m^2, normalised ferritic-pearlitic steel, brittle-ductile transition temperature $-20^\circ-\pm 0^0$C.
Material no. 3. UHB SIS 2225, C 0.25 %, Si 0.25 %, Mn 0.5 %, Cr 1.0 %, Mo 0.2 %, $\sigma_s \approx 490$ MN/m^2 , $\sigma_F \approx 700$ MN/m^2 .
Material no. 4. Brass, SM 1459-04 (SIS 5168), Cu 59 %, Zn 39 %, Pb 2 %.
Material no. 5. Aluminum, SM 6950-06 (SIS 4425), Zn 4.8 %, Mg 1.2 %, Mn 0.2 % Cr 0.2 %, Zr 0.1 %, the balance being Al. $\sigma_{0.2}$=320 MN/m^2, σ_F=360 MN/m^2 .

Evaluation.

The effective crack length, a, is measured between the plane through the axes of the two holes and the tip of the fatigue crack. The limit load P_Q is evaluated from the COD-P curve as shown in Fig. 5, using the method described in ref. [1]. If the thickness of the specimen is B and the width is W, then the following equation applies.

$$K_Q=(P_Q/BW^{1/2})\{29.6(a/W)^{1/2}-185.5(a/W)^{3/2}+655.7(a/W)^{5/2}-1017.0(a/W)^{7/2}+ +638.9(a/W)^{9/2}\} \qquad (1)$$

In order that the calculated K_Q values should be true material constants, $K_Q=K_{IC}$, two subsidiary conditions must be fulfilled, namely

a) plane strain conditions are dominating, i.e.

$$B\geq 2.5(K_{IC}/\sigma_s)^2 \quad \text{, and} \qquad (2)$$

b) the plastic zone in front of the crack must be kept so small that linear fracture mechanics applies, i. e.

$$a \geq 2.5(K_{IC}/\sigma_s)^2 \qquad (3)$$

These inequalities are always fulfilled for materials nos. 3 & 5. However, comparisons can be made between specimens of the same material, all of them being machined to the same dimensions, even if this is not the case.

Results.

The K_Q values have been plotted as functions of the plastic strain in Fig. 6. There is a marked decrease in fracture toughness for the three steels observable from a rather small amount of strain. For 2 & 3, K_Q may be lowered by 30 %. Material no. 1 regains a little of its toughness for larger strains. The whole toughness of the virginal material, however, will never reappear after deformation. The most probable explanation of the fact that part of the original toughness is returning might be the deformation hardening, which is realised as follows.

It has been proved (ref. [2]) that when the crack length, a, is fixed then

$$K_Q=C.\sigma_s \qquad (4)$$

in plane stress, C being a constant in the sense described below. A specimen of sufficient thickness will give

$$K_Q=K_{IC}=\text{constant}, \qquad (5)$$

independent of σ_s. For a specimen with B somewhere between the two extremes plane stress and plane strain, a slight dependence should be expected, say

$$K_Q=C.\sigma_s \quad , \; 0\leq\alpha\leq 1 \quad , \qquad (6)$$

where α stands for that proportion of the residual fracture surface that consists of shear lips, Fig. 7.

$$\alpha=1-x/B \qquad (7)$$

Hence, K_Q should be corrected as for the influence of the rise of yield stress caused by strain hardening. One obtains an upper limit for the correction putting $\alpha=1$. As expected, C differs from one degree of deformation to another and that difference in the corrected values K_Q^{*} may be attributed to structural damage due to plastic strain. K_Q^{*} vs. strain is plotted in Fig. 8.

Conclusions.

The fracture toughness of some structural steels strongly depends on the plastic strain obtained in uniaxial tension. It has been measured in the direction of deformation. This effect was also found by Kalish et al. (ref.[3]) although their experiments were carried out on steels that change their metallographic structure due to deformation. They also deal with material where K_Q decreases with increasing yield stress (cf. fig. 9).

The fracture toughness of brass and aluminum does not seem to be sensitive to deformation. Possibly, local recovery occurs close to the crack tip during the fatigue stage.

The work per unit volume in uniaxial tension is

$$w_o=\int\sigma d\varepsilon \quad , \qquad (8)$$

where the energy density w_o stored in the material can be regarded as an estimate of the damage. An additional damage is produced near the bottom of the notch before the crack can start growing: loading the specimen for the final fracture toughness test, a plastic zone of radius r_s forms, Fig. 10, where

$$r_s=(1/6\pi)(K_{IC}/\sigma_s)^2 \tag{9}$$

in plane strain (ref.[4], p. 84). An estimate of the additional energy density required to imply spontaneous crack growth would be

$$w_1=C_1 r_s{}^2 \sigma_s/a^2 \quad . \tag{10}$$

In equations (9) & (10), σ_s is the yield stress corresponding to the strain arrived at in uniform plastic deformation. a is the crack length and C_1 is a constant, chosen so that w_1=1 for w_o=0. Fig. 11 shows w_1 plotted against w_o. Evidently, after straining to 1.6 % (the least strain used) the energy density needed for crack propagation is only 20-25 % of that needed in virginal material.

2. Local yielding.

For a specimen equipped with a suitable notch, local yielding will take place near the notch when the specimen is loaded. ("Local" means "limited to the vicinity of the notch".) The maximum extension of such a zone would be of the same order of magnitude as r_s, eq. (9). If the deformation speed is high enough, there is no time for yielding. Also, if the specimen is first loaded very slowly until some yielding occurs (preloading) and then the loading speed is increased to the value where no yielding takes place (main loading) it may be assumed that size and shape of the original plastic zone are preserved throughout the loading. The size of the plastic zone can be determined by proper selection of the preloading. The fracture toughness as a function of this size will now be studied. That the deformation speed is high enough means that it must exceed the limiting speed of the dislocations, as those are responsible for the plastic flow.

Experimental procedure.

Preloading in torsion as well as in tension have been tried. The main load is always tensile. In the testing rig of a hydraulic testing machine an optional equipment is installed, so that the specimen can be subjected to a constant torque or a constant tensile force as long as the main loading lasts. The torsional equipment is seen in Fig. 12. Strain gauge dynamometers provide continuous registration of torque and force. The displacement of the piston and hence the deformation of the specimen across the notch is also measured.

Design of specimen.

A cylindrical specimen with a circumferential notch has been chosen, Fig. 13. The notch, however, must be made sharper. A fatigue crack (as in section 1) produces the best results. It must possess good circular symmetry, which is easy to obtain in bending rotation. The deflection amplitude is determined arbitrarily. Machined notches and notches, obtained by an electric spark have also been tried. Fig. 14 shows two fractured specimens.

Materials: Selection.

Material no. I. Fagersta, mild carbon steel SIS 1550.
Material no. II. Domnarvet, C-Mn-V steel, equal to no. 2, section 1.
Material no. III. Brass, SM 1658, SIS 5170.
Material no. IV. Aluminum, Sv. Al-kompaniet B 51 swp, SIS 4054.

Materials: Properties.

For all materials, the tensile and torsional testing diagrams of the steady state have been recorded. The yield stress in shear τ_s is greater than the corresponding value $\sigma_s/\sqrt{3}$, obtained from the tensile test.

Measuring values.

During main loading - for the determination of fracture toughness - the force is measured as a function of time. It is recorded on a Tektronix dual beam oscilloscope 549 with storage cathode-ray tube. A typical registration is shown in Fig. 15. The straight line there is the piston stroke.

Evaluation.

For the specimen of Fig. 16, the stress intensity factor, according to Bueckner (ref. [4] , p. 82), is

$$K_I=P/D^{3/2}(1.72D/d-1.27) \qquad (11)$$

The strain rate (ref. [5]) in front of the crack tip is estimated by

$$\dot{\varepsilon}\approx(1/\ell E)(K/\sigma_s)\dot{K} \qquad (12)$$

Here, ℓ is the width of the plastic zone, for instance the width of a glide band, Fig. 17. Assuming the crack length of 2a perpendicular to the stress direction far away from the tip, the far-away-stress being σ, the following expression, in principle, is valid for K

$$K=\sigma\sqrt{\pi a} \qquad (13)$$

Due to the geometry of the specimen, the constant may differ from $\sqrt{\pi}$, but the order of magnitude of K is still the same which is sufficient for this approximation. Inserting the experimental values for brass, one obtains $\dot{\varepsilon}\approx10^4\ s^{-1}$. In several tests, the strain rate is even greater. Above the limiting strain rate due to dislocations, which is approximately 2.10^4, no yielding is likely to occur, during the main loading.

Several possibilities exist for the description of the degree of plastic deformation. One of these is the use of the width of the plastic zone formed in the preloading stage, r, which requires knowledge of the connection between torque M and zone width r. For this, a paper by Walsh & Mackenzie (ref. [6]) has been used in all experiments. The function r=r(M) is correct only for an infinitely sharp notch, i.e. a crack. If the notch root radius is ρ, the error due to this discrepancy is of the order ρ/a in r.

Fig. 15, giving P for eq. (11), illustrates the behaviour of material no. I preloaded in tension. Force scale: 10 kN/cm, time scale: 0.1 ms/cm, displacement of piston 1.7 m/s .

Results.

Preloading in torsion. The apparent fracture toughness as a function of the relative width of the plastic zone r/a is drawn in Fig. 18 a.

Preloading in tension. Only material no. I has been studied. $K_Q = K_Q(r/a)$ is shown in Fig. 18 b. To find an r in this case one proceeds as follows. By means of the stress intensity factor K_I an effective stress is estimated

$$\sigma_e(r=r,\ \theta=0) = K_I(2.4/2\pi r)^{1/2} \tag{14}$$

Cf. Fig. 21. Using K_I according to eq. (11), one finds K_I/P for each specimen. If P_Q is the critical load as measured on the oscilloscope and σ_s the yield stress (ideally plastic material) one may write

$$\sigma_s = \sigma_e = K_I\{P_Q/P\}(2.4/2\pi r)^{1/2} \tag{15}$$

From eq. (15), r is easily solved.

3. Residual stresses.

If the torque, high enough to produce plastic flow, is removed before the main loading starts, the influence of residual stresses upon the fracture toughness can be investigated, using exactly the same procedure as in section 2. $K_Q(\emptyset)$ and $K_Q(r/a)$ are seen in Fig. 18 c and Fig. 18 d, respectively. The permanent angle of deformation $\emptyset$ is defined in Fig. 16, left.

2 & 3. Discussion.

The fracture toughness tends to decrease when the plastic zone size, obtained in torsion, is increased (Fig. 18 a). This effect does not appear, neither after the development of a plastic zone in tension, Fig. 18 b, nor because of residual stresses, Fig. 18 c & Fig. 18 d. The decrease in K_Q seems to set in even for a very small zone width r. The scattering of the test results is considerable, and probably this method is not sensitive enough to reveal the exact amount of yielding, if any, required to facilitate crack initiation or growth. Presumably, that should be done using a microscopical approach. The explanation why preloads in tension and torsion produce different results may be that if the preload is pure tension, the crack tip is blunted and the plastic zone much thicker and more diffuse (ref. [7], fig. B2) than in torsion which gives a thin zone near the plane of crack propagation. To check this and also that the zone size is in accordance with what is expected from theory, some microhardness measurements were made in the plane through the axis of the specimen and perpendicular to the fracture, for materials nos. I & II. One finds r≈1.2 mm in a case when 0.6 mm was expected. Probably, some defects exist from the beginning and they are so severe that local yielding occurs from the torque even further away from the notch than theoretically predicted. In most cases, however, r is the same as measured and calculated.

Because of spontaneous yielding, a larger region round the crack is subjected to unloading when the crack propagates. It is convenient to add the zone width r to the crack length a under those circumstances, for calculating K_{IC}. In this application, the material probably regards itself as purely elastic up to fracture, even where it has already been deformed plastically during preloading. Obviously, one should insert a, not a+r, into eq. (11).

As has already been mentioned, the material is unable to determine the loading history, i.e. whether the yielding was caused by tensile or shear stresses, etc. The only thing the material "feels" is how the yielding is localized and - possibly - the degree of deformation. If the test is made up by a torsional preloading stage and a combined torsional-tensile main loading stage, it will be justified to evaluate the K value as obtained in combining Mode I and Mode III:

$$K_{tot}=\sqrt{K_I^2+\{4/(\kappa+1)\}K_{III}^2} \quad , \qquad (16)$$

$$\kappa=\begin{cases}(3-\nu)/(1+\nu) & \text{plane stress}\\ 3-4\nu & \text{plane strain}\end{cases} \qquad (17)$$

K_I is obtained from eq. (11), whereas

$$K_{III}=\tau_o\sqrt{\pi a} \quad , \qquad (18)$$

a=notch depth, τ_o=shear stress on the cylindrical surface of specimen,

$$\tau_o=2M/\pi R^3 \quad , \qquad (19)$$

2R=initial diameter of specimen. K_{tot} as a function of r/a differs very little from K_Q as a function of r/a (Fig. 18 a) and is therefore not shown.

4. Creep damage.

There is a dual purpose with this investigation. Firstly, it will turn out if the decrease in fracture toughness, expected after yielding, is unaffected by recovery at the creep temperature, causing blunting of microcracks, release of strain energy stored in the material, etc. Secondly, the Kachanov creep fracture theory introduces creep damage D_c

$$D_c=(A-A_r)/A \quad , \qquad (20)$$

where A is the initial area and A_r the remaining load-carrying area. No reliable methods to measure D_c exist as yet.

Experimental procedure.

Conventional uniaxial creep tests were carried out on cylindrical specimens, Fig. 19. The stress was constant in each test, ranging between 138 and 278 MN/m^2 throughout the series. Temperatures were 575, 600, and 625 ^oC. The total residual strain was measured after each test. Cylindrical fracture mechanical specimens were sampled as shown in the figure. Fatigue cracks were produced in bending rotation..

Material.

The material tested was a martensitic heat resistant steel, Bofors ROP 46, containing C 0.20 %, Cr 12 %, Mo 0.5 %, V 0.4 %, and Nb 0.2 %.

Evaluation.

K_Q is calculated from the force causing instability using eq. (11). As before, D is the specimen diameter, d the diameter of the residual fracture surface. When the latter is elliptical, one defines

$$d=(\text{major axis}\times\text{minor axis})^{1/2} \qquad (21)$$

Results.

In Fig. 20, K_Q has been plotted vs. creep strain. The dependence turns out to be the same as in the previous sections. Evidently, even very small strains (0.16 % is the smallest investigated) affect the toughness considerably.There was no possibility to connect these observations with any microscopical traces of creep damage, e.g. microcracks, as this material is evidently not sufficiently brittle. The change in K_Q because of creep strain is so pronounced, however, that it might well be used as an aid in estimating D_c. The present material only covers small strains, hence further examinations are necessary.

5. Some fractographic observations.

For the discussion in section 2 it is most important that the size of the plastic zone is determined with some accuracy. Further observations have been carried out on the fracture surface, using scanning and transmission electron microscopy. As expected, a ductile zone is seen between the fatigue crack and and the brittle residual fracture surface. Its width coincides well with the calculated width. Strange enough, a narrow ductile region can be noticed even if no preloading has occurred. Probably, it developed during the fatigue stage. Preloading will completely destroy all the traces of yielding during fatigue.

6. Acknowledgments.

This investigation was carried out at the Division of Mechanics of Materials, Royal Institute of Technology, Stockholm. The author wishes to thank the head of the division, Prof. Janne Carlsson, who first suggested the study, for valuable advice. He also wishes to thank many of his collegues at the division and at the Swedish Institute for Metal Research for much help and encouragement.

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7. Complete report on this subject is found in ref. [8] .

Zusammenfassung.

ÜBER DEN EINFLUSS DER MATERIALBESCHÄDIGUNG AUF DIE BRUCHZÄHIGKEIT METALLISCHER WERKSTOFFE.

Für die meisten elastisch-plastischen Werkstoffe wird ein Riss, der sich verbreitet, von plastischer Verformung im Gebiet vor der Riss-Spitze vorangegangen. Man könne sich vorstellen, dass diese Verformung die Rissverbreitung erleichtert. Wenn aber Temperatur und Verformungsgeschwindigkeit passend gewählt werden, ist es möglich, diese spontane Plastizierung zu vermeiden. Dann lässt sich die plastische Zone im voraus beliebig gross machen. Beispielsweise erhält man eine solche Zone, die ganz scharf begrenzt ist, in Torsion gekerbter Rundproben. Die Bruchzähigkeit bestimmt man dann in einem gewöhnlichen

Zugversuch, wobei das Verdrehungsmoment konstant gehalten ist.

Auf diese Weise wird die Bruchzähigkeit einiger Stähle, Messings und Aluminiums studiert, als beeinflusst von

a) allgemeine Plastizierung

b) begrenzte Plastizierung an der Riss-Spitze

c) Restdeformation infolge der Belastungsgeschichte und

d) Kriechbeschädigung.

Diese Versuche deuten an, dass zunehmende Dehnung mit abnehmender Bruchzähigkeit verbunden ist. Vielleicht eignet sich dieses Verhältnis zum Gebrauch, um Näherungslösungen zu finden für Probleme mit Werkstoffen, die in besonderen Anwendungen eine so grosse plastische Zone entwickeln, dass sie mit linearer Bruchmechanik nicht behandelt werden können.

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Text to figures.

Fig. 1 Plane plate tensile specimen for general yielding test.

Fig. 2 CT-specimen with machined "Chevron" notch acting as a starter for fatigue crack.

Fig. 3 Tensile test measuring force vs. COD.

Fig. 4 Fracture surfaces of CT-specimens. First figure in each group denotes material.

Fig. 5 Plots obtained from apparatus of Fig. 3. Critical load P_Q denoted by an arrow.

Fig. 6 Apparent fracture toughness vs. plastic strain. No correction applied.

Fig. 7 Proportiom of shear lips on fracture surface,

Fig. 8 Comparison of K_Q and K_Q^*, the value obtained after strain hardening correction. Material no. 1.

Fig. 9 K_Q vs. yield stress. Principle behaviour. Similar behaviour observed if K_Q is plotted against specimen thickness.

Fig. 10 Assumed shape of spontaneously developed plastic zone.

Fig. 11 w_1 vs. w_o (see text).

Fig. 12 Torsional loading device.

Fig. 13 Specimen used in sections 2, 3, 4.

Fig. 14 Fracture surfaces of specimens used in local yielding experiments.

Fig. 15 Example of registration in high-speed tension.

Fig. 16 Cylindrical specimen, definition of variables.

Fig. 17 Model of glide mechanism.

Fig. 18 a. K_Q vs. r/a. Torsional preload.
b. K_Q vs. r/a. Tensile preload. Note zero suppression for K_Q.
c. K_Q vs. Ø. Residual stresses.
d. K_Q vs. r/a. Residual stresses.

Fig. 19 Creep specimen

Fig. 20 K_Q vs. total creep strain. Note zero suppression for K_Q.

Fig. 21 Definition of coordinates.

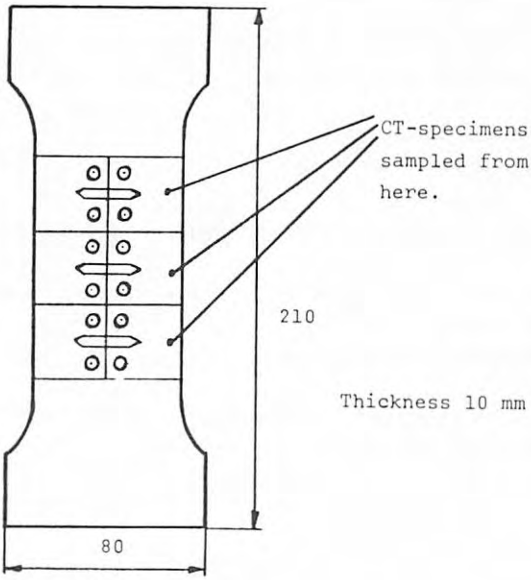
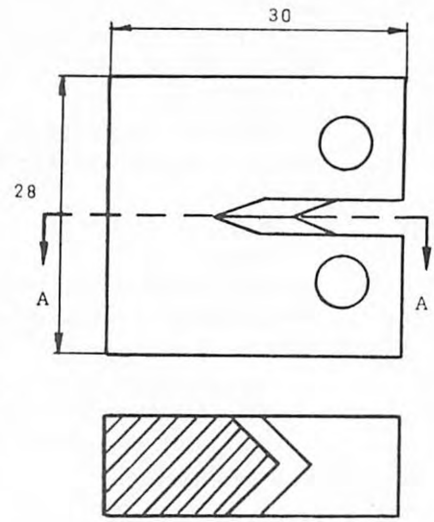


FIG. I



Section AA

FIG. 2

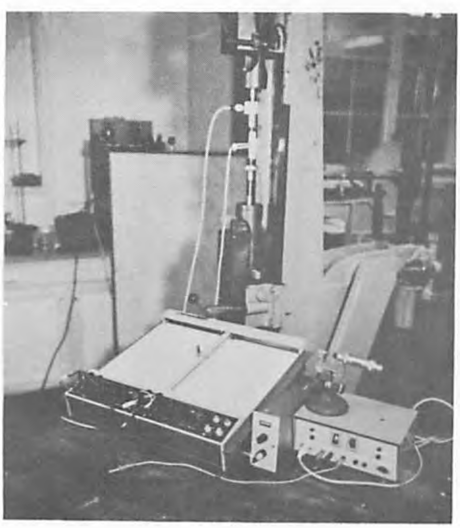


FIG. 3

FIG. 4

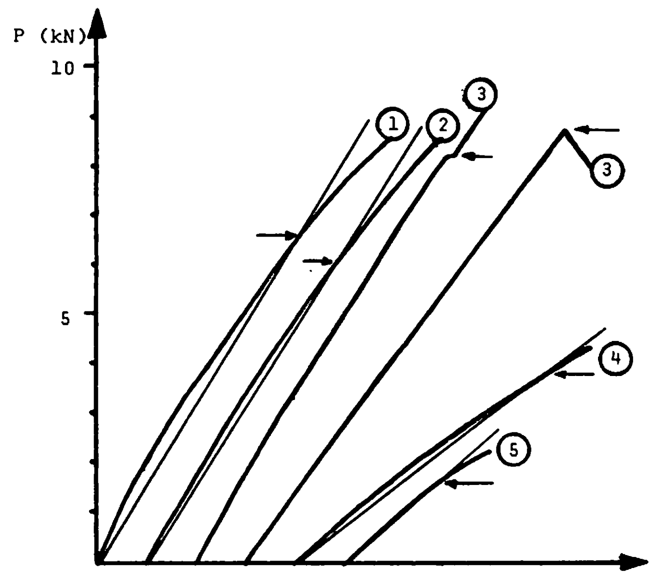


FIG. 5

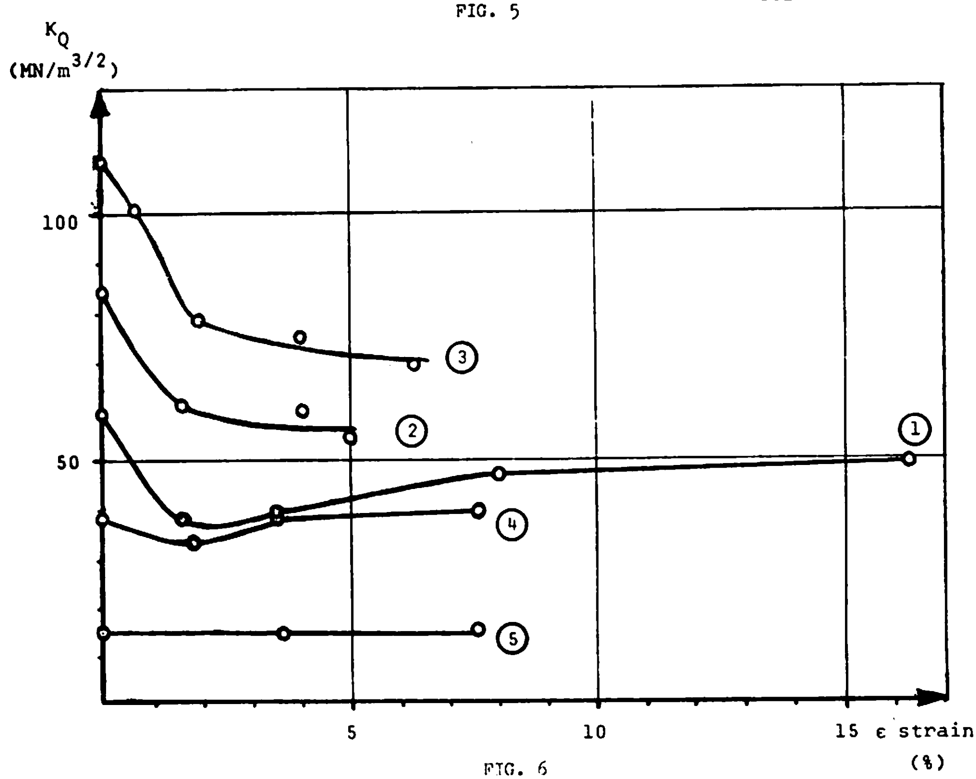


FIG. 6

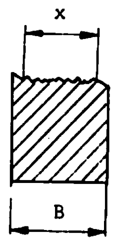


FIG. 7

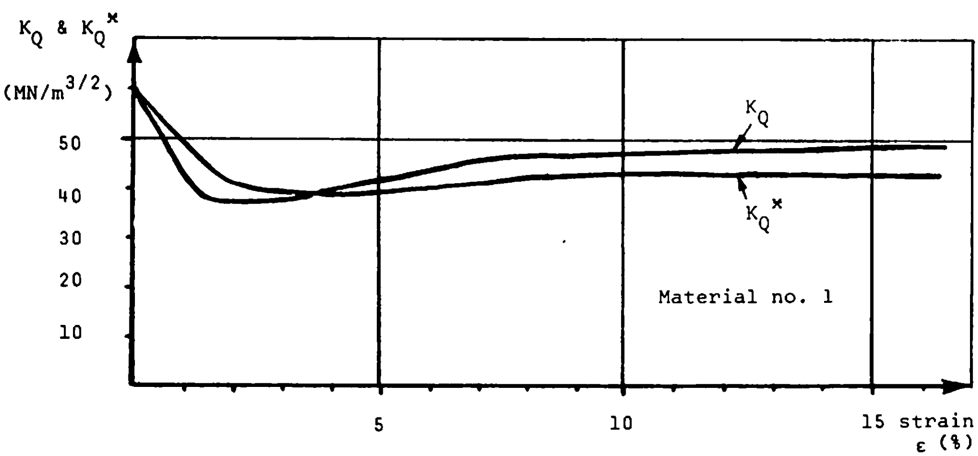


FIG. 8

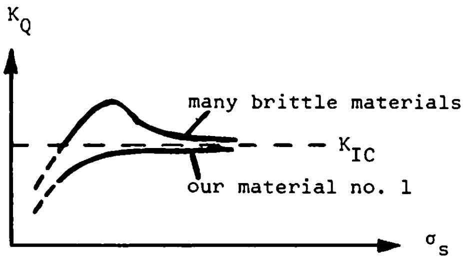


FIG. 9

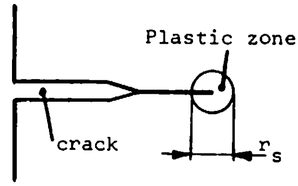


FIG. IO

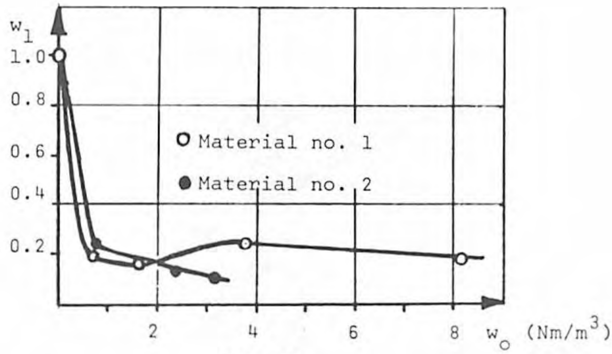


FIG. II

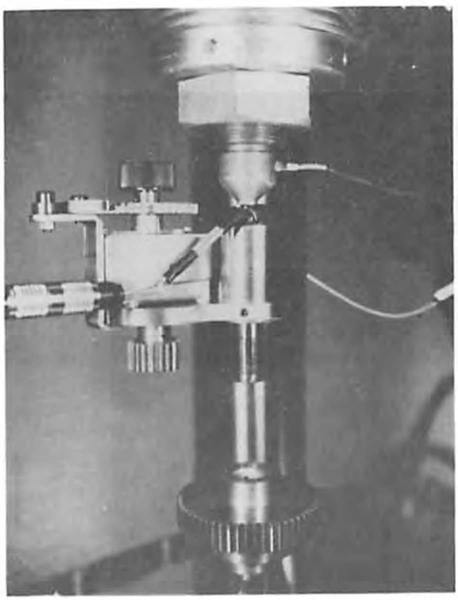


FIG. I2

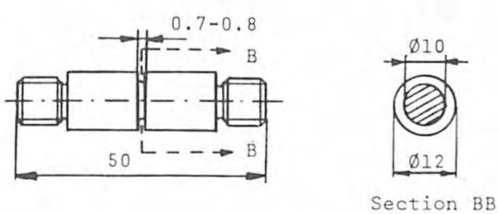


FIG. I3

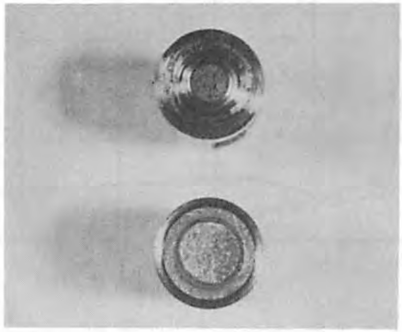


FIG. I4

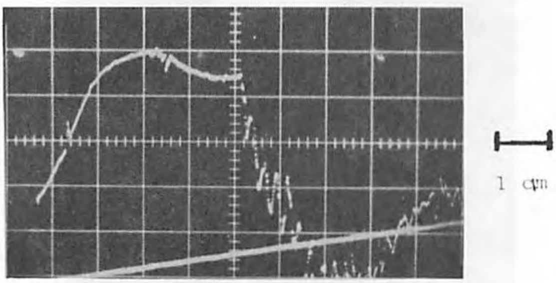


FIG. I5

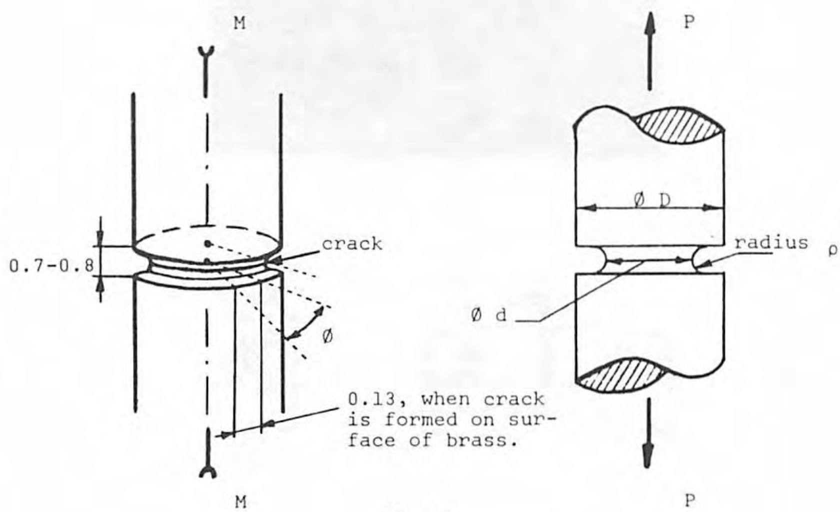


FIG. I6

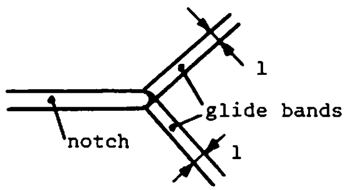


FIG. I7

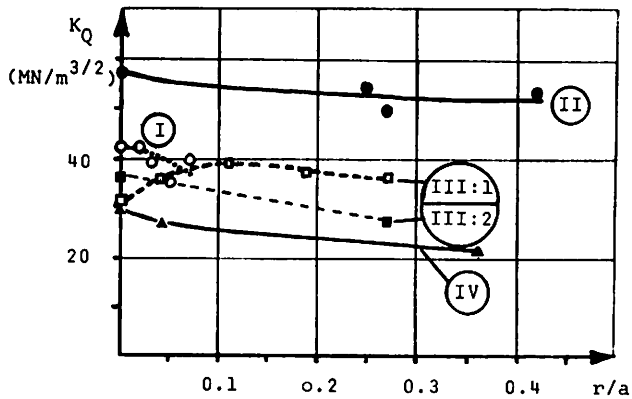


FIG. I8 - a

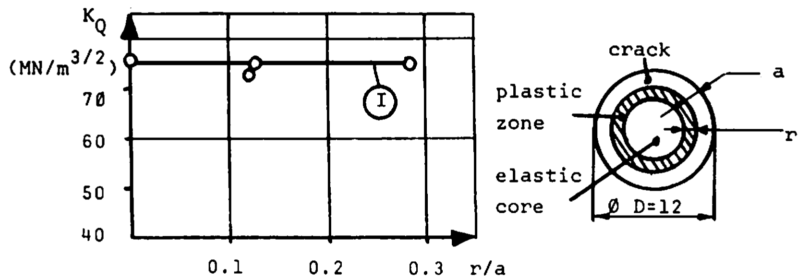


FIG. I8 - b

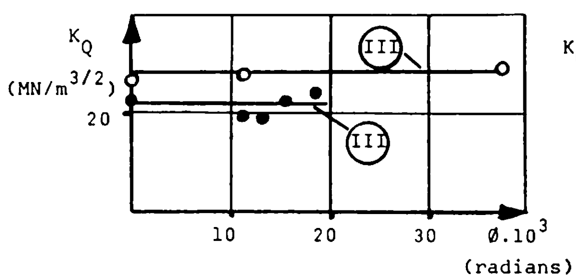


FIG. I8 - c

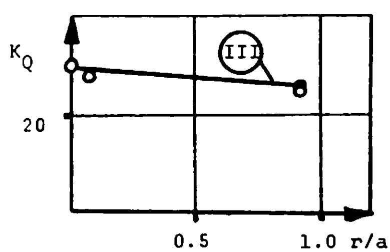


FIG. I8 - d

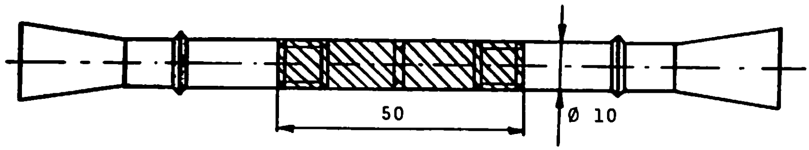


FIG. I9

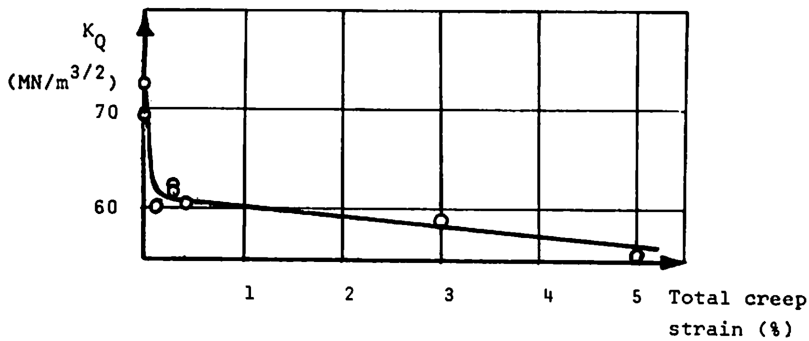


FIG. 20

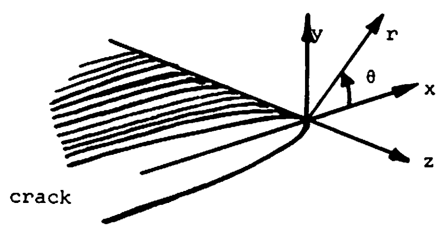


FIG. 2I

DISCUSSION

Q G. BARTHOLOME, Germany

Do you expect general yielding in a reactor pressure vessel as cause of structural damage ?

A N. G. OHLSON, Sweden

General yielding was selected as one type of structural damage in this investigation and the purpose was to determine its influence on K_Q. Pressure vessels are usually designed in such a way as to avoid yielding even after damage because of irradiation. If yielding does occur, however, it certainly affects K_Q as is reported in the paper. In nozzle regions of a vessel, for example, plastic strain is likely to take place.

Q J. H. BOWEN, U. K.

Can you suggest why the yielded material failed ductile, but the undamaged material failed brittle ?
(Corollary: Is it reasonable that if a specimen shows an increased area of ductile failure, it can have a lower (say 20%) Kc_Q).

A N. G. OHLSON, Sweden

One explanation could be that yielding generates a number of voids which helps the crack growth in the yielded region. If it is said of a fracture surface that it displays ductile failure, it means that a certain amount of plastic deformation has preceded the fracture. Presumably, plastic deformation during preloading produces the same effect.
The fact that the undamaged material fails brittle is just what should be expected at the high deformation velocity used in the main loading stage.
The results of this paper should not be interpreted in such a way that a specimen showing an increased area of ductile failure near to a fracture surface actually has a lower K_Q which, I think, would be contradicted by general experience. What can be stated from these results is merely that prestraining of the material might cause a decrease of K_Q.