Effect of Cracks on Lifetime in Fatigue/Creep Conditions

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Abstract

The effect of cracks on lifetime in conditions of interaction of high cycle fatigue with high temperature creep was studied in the case of pressure vessel steel 2.25Cr-1Mo at 600 ^{o}C. The effect of cracks strongly depends on the ratio of cyclic and static stress components. While there is no negative effect of cracks for pure creep and for very small cyclic stress components, the effect of cracks is drastically negative for higher cyclic stress components. The explanation is based on simultaneous action of two competitive processes, namely stress relaxation due to high temperature and crack resharpening due to cyclic stress component.

1. Introduction

Cracks in mechanically loaded machine parts and structural units are always detrimental. The degree of the negative effect of cracks on the lifetime critically depends not only on the geometry of the loaded body and external forces, but also on the stress-time history, temperature and material. For example, while the effect of macroscopic cracks on the lifetime in the case of symmetrical cyclic loading (pure fatigue) is drastically negative at all temperatures, the effect of the same cracks in the case of static loading at high temperatures (i.e. in pure creep conditions) can be very weak or even nil.

For the case of pure fatigue there is always a way how to evaluate the lifetime of pre-cracked bodies. Namely, it is the fracture mechanics which offers this possibility. This way is not always easy, as - especially at high temperatures - the parameters of non-linear fracture mechanics must be used.

The case of pure creep of pre-cracked bodies is far more complicated. In dependence on the loading conditions (temperature, stress) and especially on the type and structural state of the material two extreme cases can occur:

(i) The crack continuously grows. The stress concentration at the crack tip leads to an intensive cavitation and thus to the crack growth. The crack propagation rate is controlled by the material and by the stress state at the crack tip.

(ii) The pre-cracked body behaves as if there were no crack at all. The stress concentration at the crack tip relaxes so quickly that the crack manifests itself only by decrease of the cross-section.

The first case is usually identified with creep-brittle behaviour the second case with creep-ductile behaviour. The crack growth in the case of creep-brittle materials can be - - at least into a certain extent - described by means of elastic-plastic fracture mechanics |1|. Creep-ductile behaviour of pre-cracked bodies can be more or less ideal. Fully ideal creep-ductile behaviour covers the cases in which the crack does not propagate at all and the body is fractured either by ductile or cavitational mode. Less ideal is the case when there is some crack propagation, but only by the end of the lifetime and after a substantional creep deformation. A more detailed description of crack behaviour in creep-ductile materials can be found in the paper by Sklenička et al. |2| for the case of Cu-2.5%Al alloy and in the work by Kozák |3| for the case of an austenitic steel.

This paper is devoted to the study of crack behaviour in the conditions of interaction of high cycle fatigue with high temperature creep. Suitable parameter for the quantification of the cyclic and static stress components is the R-ratio defined as

$$R = \sigma_{min} / \sigma_{max}, \qquad (1)$$

where σ_{min} and σ_{max} are the extreme values of stress in the stress cycle. Thus pure fatigue is characterized by $R = -1$ and pure creep by $R = 1$. At constant maximum stress we can expect a dependence of crack behaviour on the R-ratio.

2. Experimental procedure

For the experiments the bainitic steel of the type 2.25Cr-1Mo was used. Its chemical composition is given in Table 1. The steel was cooled in air from 940 °C and then tempered at 720 °C for 2 h. The basic mechanical properties are listed in Table 2.

Table 1. Chemical composition of the steel 2.25Cr-1Mo

Element (wt.%)									
C	Mn	Si	P	S	Cr	Ni	Mo	Cu	Al
0.14	0.55	0.25	0.018	0.009	2.38	0.33	0.95	0.08	0.015

Table 2. Basic mechanical properties at room temperature

$\sigma_{0,2}$ (MPa)	σ_{UTS} (MPa)	Reduction in area (%)
285	465	77.7

All the experiments were performed at 600 °C in air. The shape of the specimens is shown in Fig. 1. The central part of the specimen is a sheet with crack starter. The specimens were pre-cracked by room temperature symmetrical cycling in such a way that two sets of specimens were obtained, namely specimens with total crack length 2 ℓ_u = 4 mm and specimens with total crack length 2 ℓ_u = 8 mm. All the cracks were produced in such a way that the stress intensity factor amplitude by the end of the pre-cracking procedure was the same, namely slightly above the threshold value.

High temperature fatigue/creep loading was performed on a Fractronic 7801 resonant pulsator, which was modified in such a way that the total specimen elongation was continuously measured. The tests were performed for the maximum initial net stress σ_{max} = = 200 MPa for a large range of R. The frequency of the cyclic stress component was about 120 Hz.

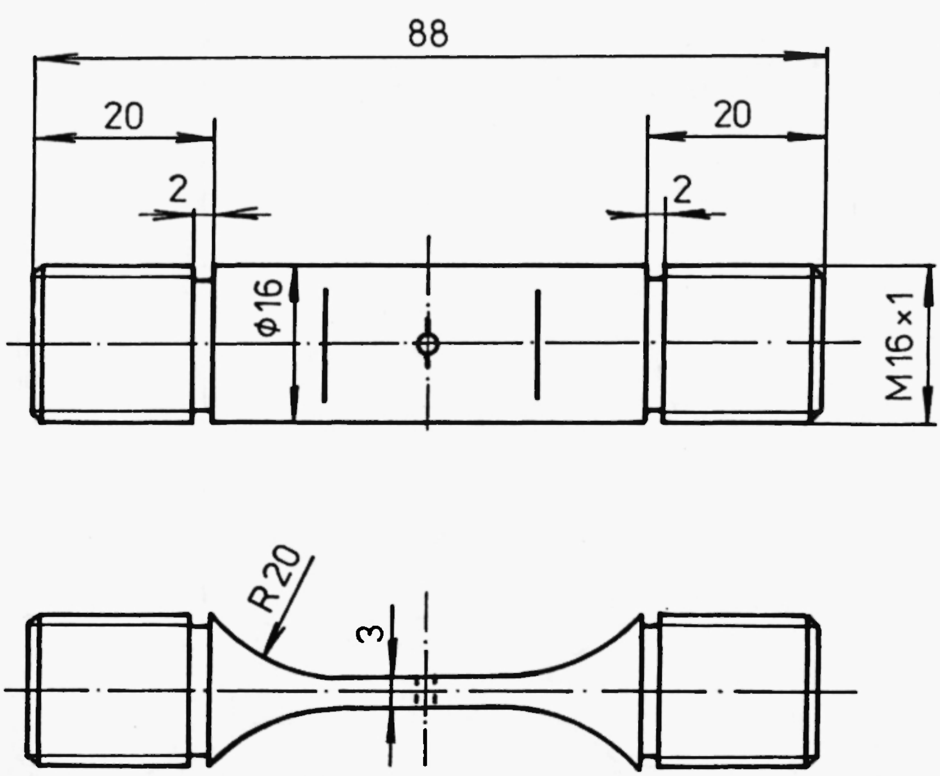

Fig. 1: Shape and dimensions of the specimens.

3. Experimental results

An example of the time dependence of the specimen elongation is presented in Fig. 2. Here the curves for pre-cracked specimens under pure creep (R = 1) and under two fatigue/creep conditions (R = 0.9 and R = 0.8) clearly document that both the time dependence and the total elongation at final fracture (last points on the curves) strongly depend on the R-ratio.

The main result is shown in Fig. 3. The lifetime is plotted here in dependence on the R-ratio. The smooth specimen data presented in the diagram are taken from our earlier work |4|.

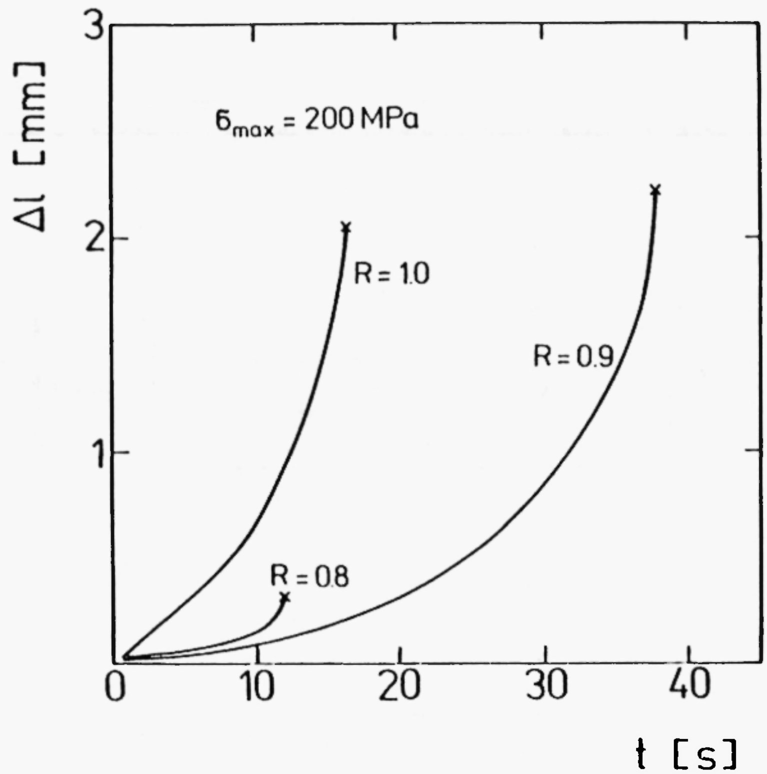

Fig. 2: Time dependence of the specimen elongation for three values of R-ratio. Specimens with the initial crack length $2\ell_u$ = 4 mm.

They were obtained on cylindrical specimens. It can be clearly seen that there is no negative effect of cracks in the case of pure creep. At R = 1 there is also no crack propagation at all. The specimens undergo ductile fracture as if there were no cracks. Thus the steel behaves as an ideally creep-ductile material. The increasing amount of cyclic stress component leads first to an increase of the lifetime. This is true for smooth specimens for R-values from 1 to 0.4 and for pre-cracked specimens with the initial crack length $2\ell_o$ = 4 mm for R-values from 1 to 0.9. For specimens with the initial crack length $2\ell_o$ = 8 mm, a slightly longer lifetime was observed in comparison to smooth specimens and specimens with shorter initial cracks. For lower

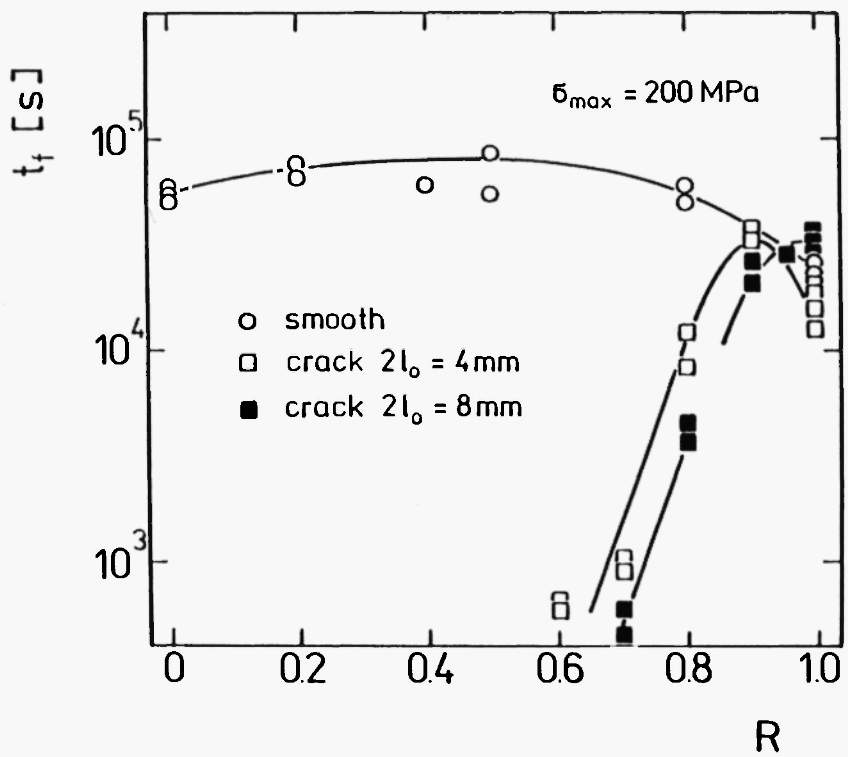

Fig. 3: Lifetime t_f vs. stress ratio R for smooth and pre-cracked specimens.

R-values (i.e. for higher cyclic stress components) the lifetime generally decreases. The important point is that the cracks have a drastic negative effect on the lifetime for $R < 0.9$.

The macroscopic type of fracture depends also strongly on the R-ratio. Irrespective of the specimen geometry, the fracture is of ductile type for R- values higher than the R-value corresponding to the maximum on the t_f vs. R curve (Fig. 3). Thus the ductile fracture occurs for $R > 0.4$ for smooth specimens and for $R > 0.9$ for pre-cracked specimens (Fig. 4). On the other hand the failure for R-values on the left side with respect to the maximum is due to fatigue crack propagation. For smooth specimens the cracks are first nucleated and then propagated.

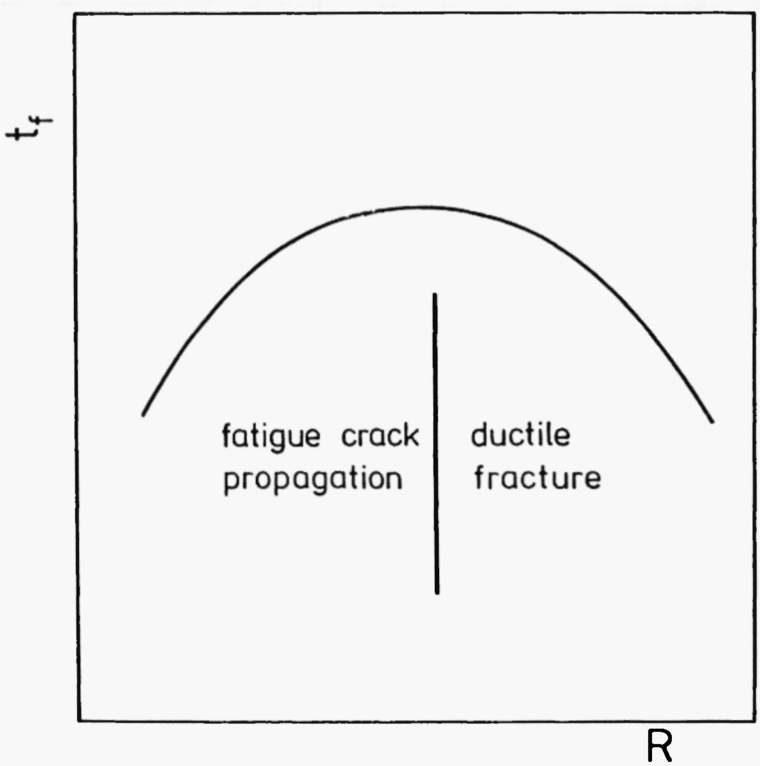

Fig. 4: Scheme of macroscopic fracture types.

Quantitatively the fracture can be best characterized by the equivalent strain to fracture, ε_{eq}, defined as

$$\varepsilon_{eq} = \ell n(A_0/A), \qquad (2)$$

where A_0 is the initial net cross section and A is the smallest cross section at the moment of final failure (i.e. fatigue fracture surface + ductile fracture surface). The dependence of the ε_{eq} on the R-ratio is shown in Fig. 5. The equivalent strain to fracture is relatively low in the range of failure by fatigue crack propagation (see Fig. 4) and sharply increases in the range of ductile fracture. The value of ε_{eq} for pre-cracked specimens is considerably lower than that for smooth specimens. The fracture is always of transcrystalline type.

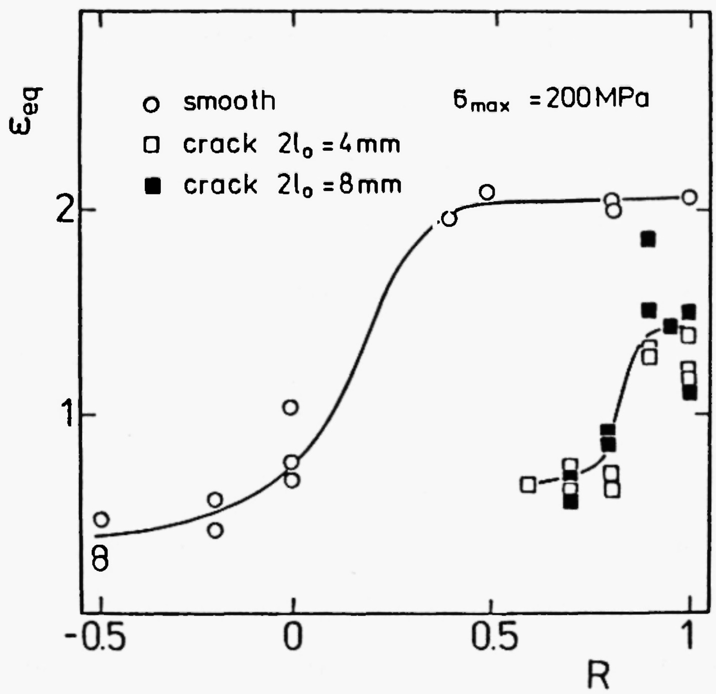

Fig. 5: Equivalent strain to fracture ε_{eq} vs. stress ratio R for smooth and for pre-cracked specimens.

4. Discussion

The results show that in the case of pure creep there is no negative effect of cracks on the lifetime. It even seems that the pre-cracked specimens with the longer initial cracks at R = 1 have a slightly longer lifetime that the smooth specimens. This can be probably related with the triaxiality of stress distribution in the remaining cross-section ahead of the crack tip. There are data in the literature showing in the case of notched specimens under pure creep conditions that high degree of stress triaxiality makes plastic deformation difficult and increases thus the lifetime |5 to 8|. Basically, the crack behaviour under pure creep conditions is determined by the rate of relaxation of stress concentration ahead of the crack tip. In our case the stress relaxation is - due to high temperature - - so quick that the crack does not represent any stress concentrator. Kozák |3| computed by finite element method the stress relaxation ahead of the cracks in 18Cr/10Ni steel subjected to pure creep at 700 ^{o}C. This steel behaves at the given conditions similarly to our steel, i.e. as creep-ductile material. The computation showed that the stress concentration is almost fully relaxed out within a fraction of the total lifetime. The remaining stress increase ahead of the blunted crack is not high enough to propagate the crack but probably causes the above mentioned triaxiality effect in the remaining cross section.

A small cyclic stress component (for R between 0.9 and 1) does not change this basic picture. In this range the lifetime increases with increasing cyclic stress component. The explanation here is similar to the case of smooth specimens. Namely, the stress concentration at the crack tip is too small to be important and thus the effect of cyclic stress component on the increase of the resistance of the whole bulk of material to unidirectional plastic deformation prevails |4|. Any further increase of the cyclic stress component ($R < 0.9$) effectively hinders the stress relaxation ahead of the crack tip owing to crack resharpening. This is due to the fact that in every stress cycle the crack tip is blunted during the increase of the stress but again resharpened during the decrease of the stress. Thus the stress relaxation is too slow and the crack can propagate right from the beginning of the cycling. The controlling quantity here is the cyclic stress state at the crack tip.

Fig. 3 also shows that lifetime depends only weakly on the crack length. The curves for $2\ell_0 = 4$ mm and $2\ell_0 = 8$ mm lie close to one another. We shall try to explain this first qualitatively and then also semi-quantitatively within the frame of linear fracture mechanics. All the experiments were performed at constant maximum initial net stress. Thus in the case of pure creep and in the whole region of ductile fracture (i.e. for $R > 0.9$, see Fig. 4) there is no reason for the ℓ_0-dependence of the lifetime, as the lifetime is controlled only by the net stress. Let us assume that the crack propagation rate in the region of fatigue crack propagation (i.e. for $R < 0.9$, see Fig. 4) is controlled by the stress intensity factor. The stress intensity factor is proportional to the remote stress,

σ_{rem} , and not to the net stress, σ_{net}. The relation between these two quantities is

$$\sigma_{rem} = \sigma_{net} (1 - \ell_o/b), \qquad (3)$$

where b is width of the specimens. At a constant σ_{net} the value of σ_{rem} decreases with increasing crack length. And this the main point of the qualitative explanation. σ_{rem} is higher for specimens with shorter initial crack. Thus at the moment of equal crack sizes the crack propagation rate in the specimen with shorter initial crack is higher than that in the specimen with longer initial crack. This compensation then leads to the experimentally found insensitivity of the lifetime to the initial crack size.

More rigorous treatment can be based on the correlation between the crack propagation rate $d\ell/dN$ and the strain energy density factor range ΔS of the type (see e.g. |9|)

$$\frac{d\ell}{dN} = B \ \Delta S^m, \qquad (4)$$

where m and B are materials constants.

For Mode I of loading, the strain energy density factor S is given as (see e.g. |10|)

$$S = a_{11} \ K_I^2, \qquad (5)$$

where K_I is corresponding stress intensity factor and a_{11} is a known function of elastic constants and polar angle φ . It holds

$$\Delta S = S_{max} - S_{min}, \tag{6}$$

where $S_{max} = a_{11}K^2_{max}$ and $S_{min} = a_{11}K^2_{min}$, i.e.

$$\Delta S = a_{11}\, K^2_{max}\, (1 - R^2). \tag{7}$$

The value K_{max} can be expressed as

$$K_{max} = (\sigma_{rem})_{max} \sqrt{\pi \ell}\, f(\ell/b) = \sigma_{max}(1 - \ell_o/b) \sqrt{\pi \ell}\, f(\ell/b). \tag{8}$$

For a sheet with central crack of the length 2ℓ it holds (e.g. |11|)

$$f(\ell/b) = \frac{1 - \ell/b + 0{,}326\, \ell^2/b^2}{\sqrt{1 - \ell/b}} . \tag{9}$$

Substituting eqns. (6), (7) and (8) into eqn. (4) we get

$$d\ell/dN = C\, .\, F(\ell), \tag{10}$$

where

$$C = Ba^m_{11}\, (1 - R^2)^m\, \sigma^{2m}_{max}\, (1 - \ell_o/b)^{2m} \tag{11}$$

and

$$F(\ell) = \left\{ \sqrt{\pi \ell}\, f(\ell/b) \right\}^{2m} \tag{12}$$

The lifetime can be obtained by integrating eqn. (10) from the initial crack length ℓ_o to the final crack length ℓ_f

$$N(\ell_o, \ell_f, R) = \frac{1}{C} \int_{\ell_o}^{\ell_f} \frac{d\ell}{F(\ell)} = \frac{1}{C} \cdot I \qquad (13)$$

At the given experimental conditions the quantity 1/C reflects the fact that the net stress is constant. The quantity I then covers the geometrical factors. The effect of these two quantities on the lifetime are contradictory. While I is a decreasing function of ℓ_o, 1/C is an increasing function of ℓ_o. Fig. 6 shows this point graphically. The weak dependence of the lifetime on the initial crack length ℓ_o (Fig. 3) can be then related to the mutual compensation of the two contradictory factors.

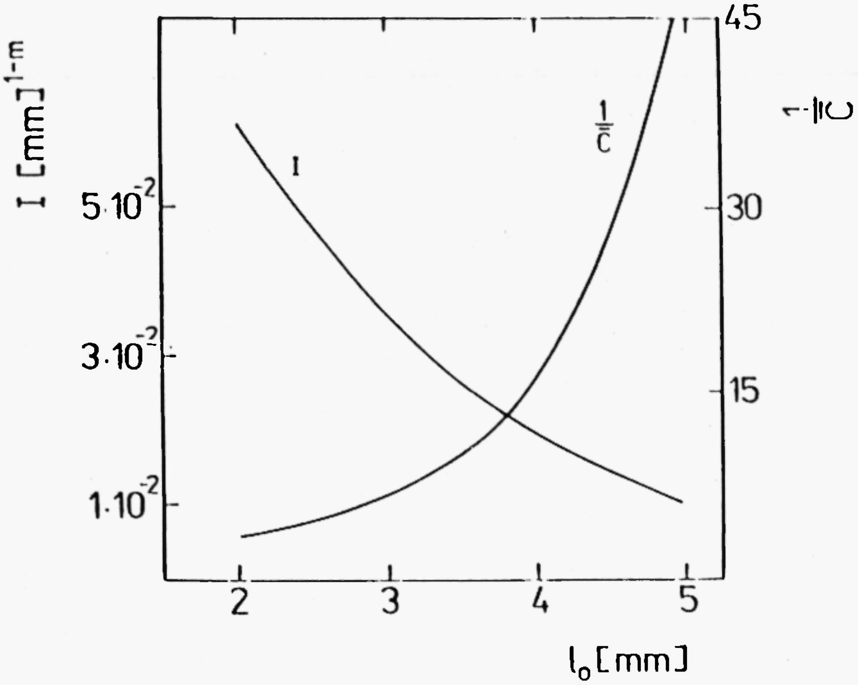

Fig. 6: Quantity I (eq. 13) and normalized values of C (eq. 11) in dependence on the initial crack length (for m = 2). $C = C/Ba_{11}^{m}\ (1-R^2)^m\ \sigma_{max}^{2m}$.

Conclusions

(i) Effect of cracks on the lifetime in fatigue/creep conditions depends on the ratio of cyclic and static stress components. For pure creep the cracks have no negative effect. With the increasing cyclic stress component (exactly for $R < 0.9$ in our case of maximum initial net stress 200 MPa) the effect of cracks on the lifetime becomes drastically negative.

(ii) The mechanisms of failure is ductile fracture for pure creep and for very low cyclic stress components ($R > 0.9$) and fatigue crack propagation for higher values of the cyclic stress components ($R > 0.9$).

(iii) The explanation of the effect of R-ratio on the crack behaviour lies in two competitive processes:

(a) relaxation of stress concentration due to high temperature - this process prevails for $R > 0.9$

(b) resharpening of crack and formation of new stress concentration due to cyclic stress component - this process prevails for $R < 0.9$.

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