APPLICATION OF FRACTURE MECHANICS AND LOCAL APPROACH TO CREEP CRACK INITIATION AND GROWTH

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Creep crack initiation and growth tests have been performed in 316L type stainless steel at 575-650°C on different specimens geometries, especially on axisymmetrically cracked specimens. No unique correlation was found between the creep crack growth rates ($\dot{a}$) and global load- geometry parameters, K or C*. A simple model based on the crack tip stress-strain field obtained from viscoplastic fracture mechanics and on creep ductility exhaustion concept is shown to be unable to reproduce the experimental results. Another local approach based on computed intergranular damage is also presented.

INTRODUCTION

Macroscopic cracks can initiate and propagate in metallic parts during high temperature service under the combined influence of creep and fatigue damage. The need to reach higher service temperatures and stresses has led to lower safety margin. This is partly the reason why recently a large research effort has been devoted to the study of high temperature crack growth behavior. More specifically creep cracking, i.e. the initiation and the propagation of single macroscopic cracks under a sustained load at temperatures well within the creep regime is of interest for components subjected to slow varying loads (For a review, see e.g. |1,2,3|. When the data published in the literature are examined it is worth noting that most of the studies have been devoted to creep crack growth (CCG) and not creep crack initiation (CCI) although the latter stage can represent a large part of the life.

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Many studies have been made to correlate C.C.G. rates ($\dot{a}$) with a global parameter, such as the stress intensity factor, K, or the load geometry parameter C* derived from the extension of non linear fracture mechanics. The conditions under which one of these parameters is more appropriate to describe C.C.G. behavior can be established from the theoretical work by Riedel and Rice |4|. Recent experimental work on an Al alloy and an austenitic stainless steel has shown that there exists a number of limitations when this global approach is applied |2,3|. In the present study which is also devoted to type 316 stainless steel further tests were performed to determine eventual unique correlation between K and C*, and the C.C.G. rates or time to initiation, T_i. It is confirmed that these correlations do not hold for all the test conditions. The reasons for this situation are briefly discussed. Furthermore an attempt is made to apply a local approach to the problem. This local approach is based upon the knowledge of the stress-strain field ahead of the crack tip in conjunction with a damage function or a failure criterion. In the present case two criteria are used. One is based on ductility exhaustion concept. The other one relies upon quantitative intergranular damage measurements made on notched specimens |5,6|. To calculate the crack tip stress-strain field analytical solutions derived from Riedel and Rice work can be used. They involve a number of simplifying assumptions. This is the reason why in the present study numerical calculations using finite elements method were made. For this purpose tests were essentially carried out on axisymmetrically cracked specimens. This geometry was chosen because it does not require any assumption about the stress state conditions. In this presentation only a brief account of the numerical results in given.

EXPERIMENTAL PROCEDURES

The tests were carried out on one heat of type 316 L stainless steel which was studied previously |2,7|.

The details of the axisymmetrically cracked specimens are given in Fig. 1. They were fatigue precracked at room temperature. The crack length was measured by a potential drop technique |8|. Regular symmetric fatigue cracks were obtained. The final crack length was such that $a/b \simeq 0.45$. During precracking the maximum stress intensity factor was less then 15 MPa $\sqrt{m}$. Further tests on CT type specimens identical to those used in a previous study |2, 7| were also performed.

Creep tests were carried out at 575, 600, 650°C using a servo-hydraulic machine. An extensometer was attached to the axisymmetric specimen along the gage length L_o = 25 mm indicated in Fig. 1 to measure load line displacement, δ. From these tests two informations were obtained as shown schematically in Fig. 2, i.e. crack length a and δ versus time. Crack initiation time, T_i, was defined as the time necessary to initiate and propagate the initial fatigue crack over a distance X_c = 50 µm.

To calculate K the formula available from the literature |9| were used. To determine C* we used an expression which can be easily derived from the limit load analysis available for this type of specimen geometry |10|, i.e :

$$C^* = \frac{P\dot{\delta}}{2\pi R^2} \tag{1}$$

It is worth emphasizing that this definition for C* requires the experimental determination of the load line displacement rate $\dot{\delta}$. For CT type specimens we used the same definition as that employed in a previous study |2|.

RESULTATS AND DISCUSSION

The results of C.C.I. tests are given in Fig. 3. In this figure we have also included the results obtained on CT type and DEN type specimens |2,7|. It is observed that all the data points can be correlated with the following expression :

$$T_i \; C^{*0.65} = \text{cste} \tag{2}$$

It is worth emphasizing that this correlation applies to different specimen geometries tested at three temperatures and giving rise to a broad range of T_i values. Similar attempts in terms of K proved to be unsuccessful.

The results of C.C.G. rates measurements are reported in Fig. 4 and 5. In these figures only the results obtained on a number of axisymmetrically cracked specimens are given in order to illustrate the trends. In Fig. 4 it is observed that after an initial period where the C.C.G. rates increase rapidly the slope of the different $\dot{a}$ -K curves is reduced. From these results it can be concluded that there is no unique correlation using the global parameter K. Conversely in Fig. 5 where the data points are plotted versus C* a much better correlation is observed. A close examination to the results shows that for a given test there exists a regime of crack growth rates where $\dot{a}$ increases rapidly while C* remains almost constant. This regime is referred as stage I. After this initial period the C.C.G. rates increase almost linearly with C*. This regime is referred as stage II.

It was shown previously |2,3| that during stage II, especially at high C.C.G. rates, the apparent correlation between $\dot{a}$ and C* could simply be explained by the fact that the major component of the variation of $\dot{\delta}$ with crack length, which is used in the calculation of C*, is due to crack growth and not to the overall creep deformation. Therefore a quasi proportionnality between $\dot{a}$ and C* is predicted since it is essentially a correlation between

$\dot{a}$ and $\dot{a}$. For predictive purposes there remain strong limitations to the application of such a correlation.

This situation is not encountered at lower crack growth rates and, in particular, at crack initiation. It is felt that the correlation established in Fig. 3 can be very useful for practical applications. To model C.C.I., as stated in the introduction, a local approach based on simple models can be used. One of these models, the simplest one is based upon creep ductility exhaustion concept. The creep strain ε_c taking place at a critical distance X_c = 50 µm ahead from the crack tip can be assessed from the theoretical expressions derived by Riedel and Rice |4|. If only stationnary creep is used with the Norton law written as :

$$\dot{\varepsilon}_s = B\sigma^n \tag{3}$$

it is found that :

$$\varepsilon_c = B \left(\frac{C^*}{BI_n X_c}\right)^{\frac{n}{n+1}} \cdot T_i \tag{4}$$

where I_n is a numerical factor. In our material at 600°C n = 7.69 while at 650°C n = 9.09. Assuming that the creep ductility ε_c is constant, from this simple model it is inferred that the slope of the C*-T_i curve should be close to -1. This is somewhat different from the experimental value (≃ -0.65). Several reasons have been invoked to account for this difference. Among these the two most important are (i) crack blunting effect or more generally plasticity effect which are not taken into account in expression (4) and (ii) a metallurgical factor which is a reduction of creep ductility with strain rate. In particular creep tests on smooth and notched bars have clearly shown that the creep ductility of this material follows complex variations |6,11|. As a rule it is observed that the creep ductility of notched specimens is a decreasing function of time when time to failure is lower than about 1000 hours. This might explain qualitatively the trend in the results. For a more quantitative approach it would be necessary to take into account the effect of crack tip stress triaxiality on creep ductility.

This is one of the reasons why more recently an attempt was made to model crack initiation using a damage function which describes intergranular damage. The details of the procedure used to establish this damage function are given elsewhere |5,6|. This function is written as :

$$dD = A\Sigma^{\alpha}\, \varepsilon_{eq}^{\beta}\, d\varepsilon_{eq} \tag{5}$$

where A, α and β are numerical constants, ε_{eq}, is the equivalent Von Misès creep strain while Σ is the maximum principal stress. D is the relative proportion of cracked grain boundaries measured on unit area of a polished section.

Elastic-plastic and viscoplastic calculations were made on specimen geometry such as a/b = 0.437 submitted to an applied load P = 50530 N. These conditions correspond to a test in which the observed time to initiation was T_i = 600 hours.

The details of the computations are given elsewhere |12|. Here it is enough to say that the constitutive equations were the same as those used previously |2|. The details of the mesh size ahead of the crack tip are given in Fig.6. Computed intergranular damage, D, within the meshes located in front of the crack tip are shown in Fig.7 for various creep times. It is noticed that D increases rapidly within the first meshes. This is related to the distribution of Σ and ε_{eq}. Previous results obtained on extremely sharply notched specimens showed that intergranular damage at failure, D_c, reached a maximum value of 3% |6|. From the results shown in Fig.7, using T_i = 600 hours it is found that D = 3% over a critical distance $X_d \simeq$ 350 µm. This suggests that a criterion for crack initiation would be the achievement of critical intergranular damage (D_c = 3%) over the critical distance X_d = 350 µm. Further studies are necessary before reaching a definite conclusion. In particular they should take into account the statistical aspect of the problem.

CONCLUSIONS

(1) In 316L type stainless steel, there is no unique correlation between the creep crack growth rate, $\dot{a}$ and a global load geometry parameter, such as K or C*. At high crack growth rates the apparent correlation between $\dot{a}$ and C* is essentially due to crack growth.

(2) It is confirmed that there exists unique correlation between the creep crack initiation time, T_i and the initial value of C*.

(3) The differences between the experimental and the theoretical T_i-C* relation derived from theoretical singular field and a simple failure criterion based on critical strain over a characteristic distance are not yet fully understood. They could be related to the variation of creep ductility with strain rate.

(4) A damage function based on critical intergranular damage is shown to be able to correlate crack initiation provided that the calculations are made over a characteristic length.

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SYMBOLS USED

a = crack length (µm)

δ = load line displacement (µm)

$\dot{a}$ = creep crack growth rates (µm/h)

$\dot{\delta}$ = load line displacement rates (µm/h)

K = stress intensity factor (MPa $\sqrt{m}$)

C^* = load geometry parameter (N/mh)

T_i = time to initiation (h)

P = applied load (N)

R = crack radius (mm)

σ, ε = creep stress strain

Σ = maximum principal stress (MPa)

D = Intergranular damage function (%)

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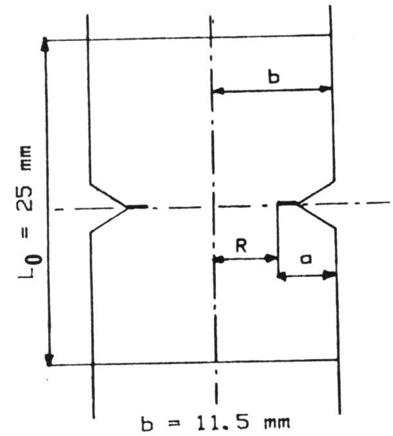

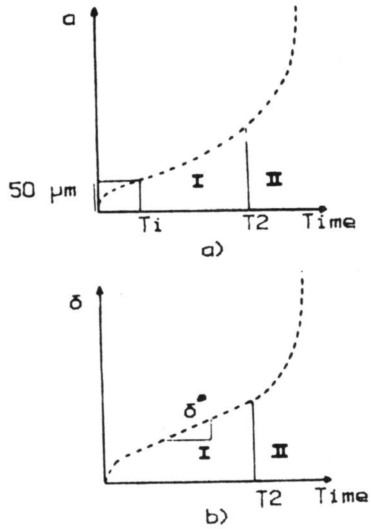

Fig. 1. Detail of axisymmetrically cracked specimens.

Fig. 2. Variation of crack length(a) and load line displacement (δ) versus (time).

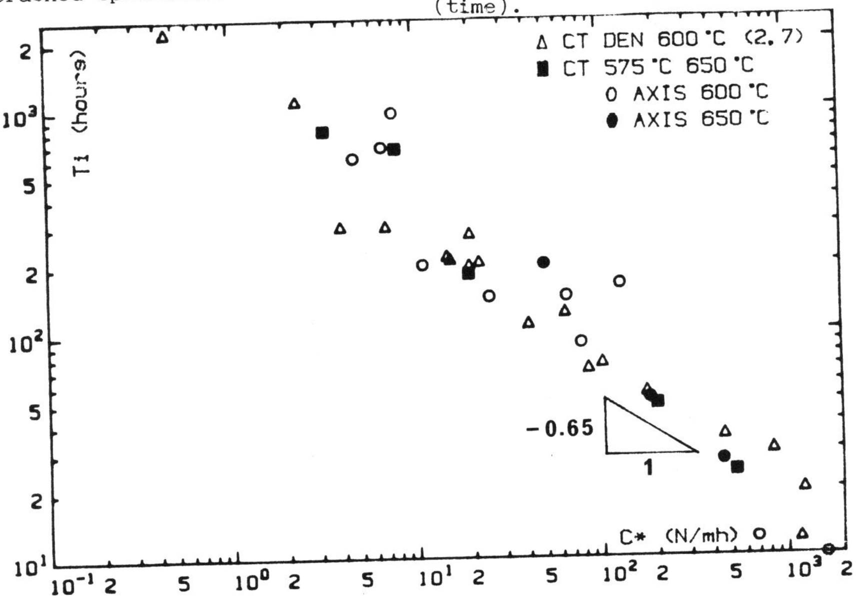

Fig. 3. Results of crack initiation tests

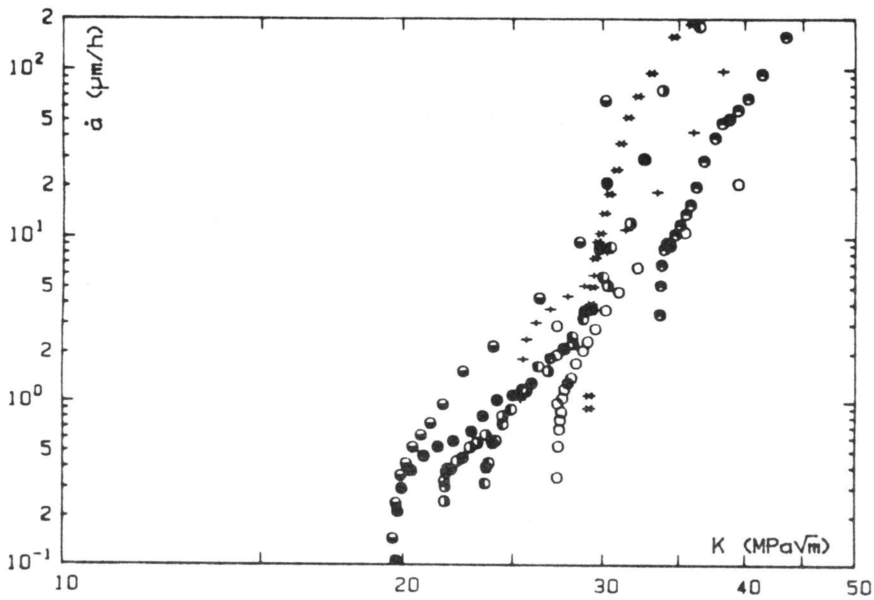

Fig. 4. Variation of crack growth rate (ȧ) versus K.

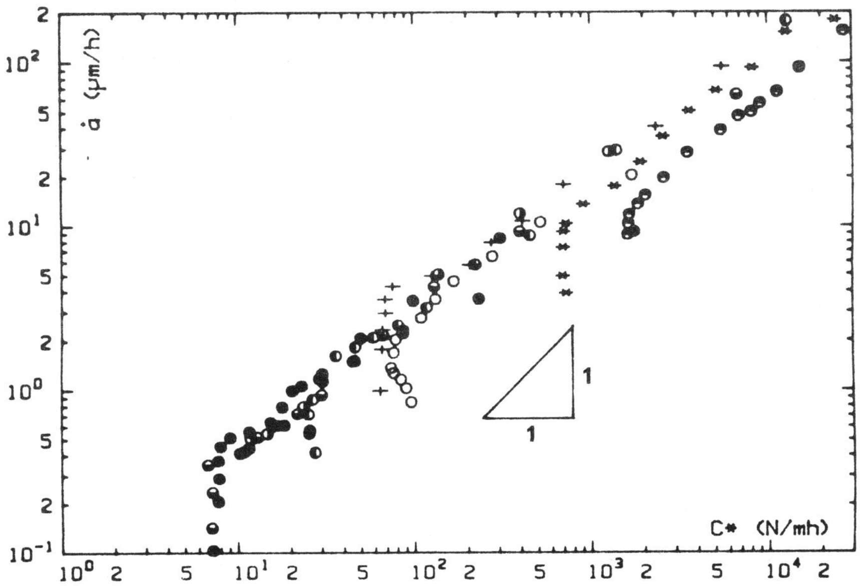

Fig. 5. Variation of crack growth rate (ȧ) as a function of C*.

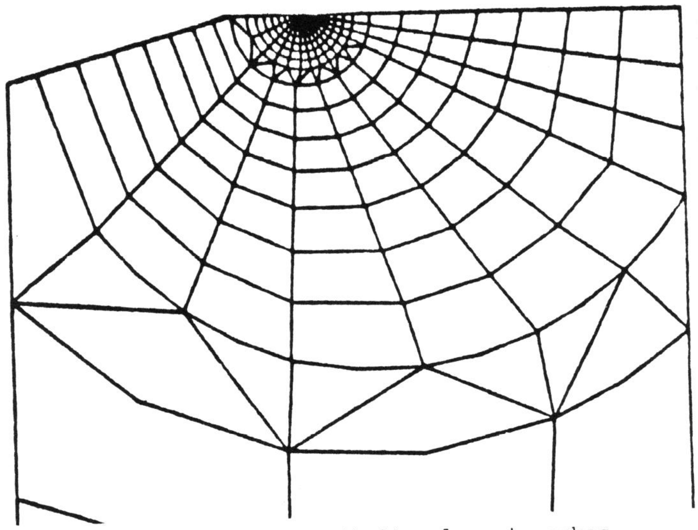

Fig. 6. Details of the finite element meshes.

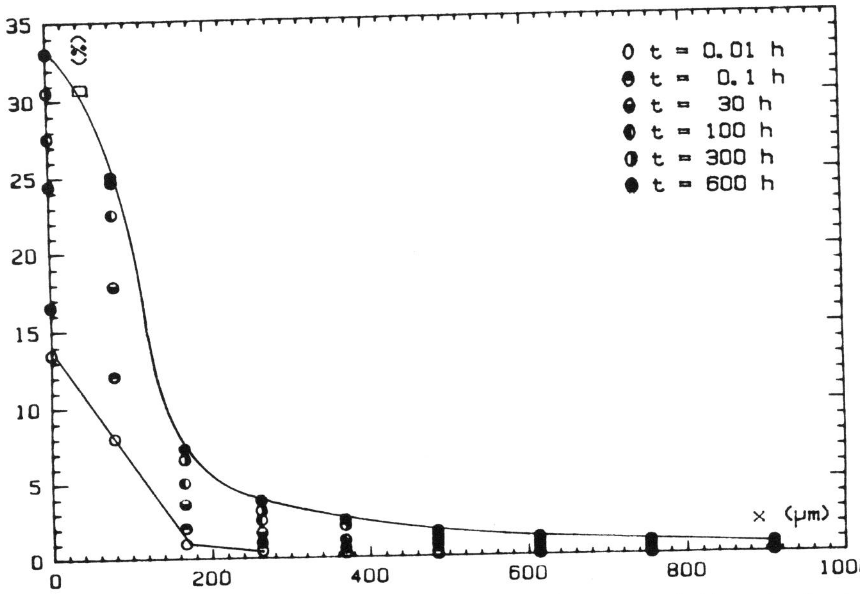

Fig. 7. Calculated intergranular damage (D) ahead of the strain tip (X) as a function of time.