CREEP CRACK GROWTH IN A 1% CRMOV STEEL AND A 32% NI 20% CR ALLOY

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For two high-temperature materials, the creep crack growth behaviour was determined in twenty-five laboratories within the EGF Working Party "Measurement of Crack Growth under High Temperature Conditons". It is shown that the most satifactory correlations of creep crack growth rates are achieved with the creep fracture mechanics parameter C*. Little influence of specimen shape and size is observed and it is demonstrated that the specimen deformation and cracking behaviour occurs under plane-stress conditions.

OBJECTIVE

The overall objective of the Round Robin was to evaluate the consistency of the results gathered from different laboratories and to establish the ability of the field parameters to correlate creep crack growth rates in a range of testpiece geometries and ultimately in service components. To assist in this exercise, two laboratories performed analytical and numerical investigations to produce an appropriate unified evaluation procedure.

With this objective, twenty-five laboratories, (see Tab. 1) proceeded to conduct a Creep Crack Growth Round Robin to make an intercomparison of data generated in different laboratories according to agreed procedures. A compilation and a detailed discussion of the results is contained in the Final Report of the Round Robin investigations [1].

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TABLE 1 - Participants in the Round-Robin Investigation

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Materials and Specimens

The materials investigated were a 1% Cr steel (21 CrMoNiV 5 7) supplied by Buderus Edelstahlwerke, Wetzlar, which was tested at 550°C, and a 32% Ni 20% Cr alloy (Alloy 800 H) supplied by Vereinigte Edelstahlwerke, Kapfenberg, which was tested at 800°C. A detailed description of the materials and their properties can be found in [1].

Most of the results were gathered using 25 mm thick and 50 mm wide compact tension testpieces (CT25/50), but other specimen types (i.e. single-edge notched three-point bend, SENB3, single-egde notched tension, SENT, and centre-notched tension, CN) and sizes (i.e. thicknesses from 5 to 63 mm) were also tested. Except for a few, the specimens were fatigue precracked and provided with side grooves.

DISCUSSION OF RESULTS

Unified Correlations

The collations of the creep crack growth rate results from the Round Robin programme 'as-determined' by the individual partners gave poor correlations with K and, at least for the 1% CrMoV steel, with the C* parameter. The situation was not much improved with K when a single assessment procedure was adopted, suggesting that the linear elastic expression is not a satifactory parameter for describing creep crack growth data over a wide range of cracking rates in 1% CrMoV at 550°C and Alloy 800 H at 800°C.

When a unified evaluation procedure for determining C* was employed much improved correlations were obtained as shown in Figs. 1 and 2. The formula used for C* was

$$C^* = \eta_c \frac{F \dot{V}_c}{B_N b}, \qquad (1)$$

where F is load, $\dot{V}_c$ is the load line displacement ratedue to creep alone, B_N is the net section thickness between side grooves, b is the uncracked ligament and

η_c = (2+0.52 b/W)n/(n+1) for CT specimens of width W
η_c = 2n/(n+1) for SENB specimens
η_c = n/(n+1) for SENT specimens.

The observations, that correlation with C* was improved when a standard analysis of the data is carried out, implies that an appreciable cause of the initial scatter was due to the application of different methods of data assessment. For example in the present study, $\dot{a}$ and $\dot{V}$ were derived from the crack length and displacement versus time records using a range of techniques. These included manual, cubic spline and seven-point polynomial curve fitting routines. In the calculation of C*, some participants used analytical estimates, which are very sensitive to the choice of n in the creep law, and others utilised total load line displacement rate rather than that due to creep alone (i.e. $\dot{V}_c$). Similarly some used gross rather than net section thickness. The degree of scatter was particularly exaggerated with the C* estimates derived using theoretical representations of $\eta_c\dot{V}_c$ according to Ref. [2] in Eq. (1). The most consistent interpretations were obtained using the preferred standard evaluation route of the unified analysis.

A detailed examination of the experimental data has revealed the presence of 'tails' during the initial period of cracking when a decreasing or an approximately constant displacement rate prevails. These tails can occupy a significant proportion of the overall lifetime. The linear region on Fig. 2 corresponds with a progressively accelerating displacement rate and is associated with having achieved a steady-state distribution of stress and damage ahead of a crack tip. An approximate expression for describing this behaviour has been given by Nikbin, Smith and Webster [3] as

$$\dot{a} = 3 \cdot C^{*0,85}/\varepsilon_f^* \qquad (2)$$

with $\dot{a}$ in mm/h, C^* in $MJ/(m^2h)$ and ε_f^* is creep ductility appropriate to the state of stress at the crack tip. This is taken as the uniaxial creep ductility ε_f for plane stress conditions and $\varepsilon_f/50$ for plane strain. The predictions of this expression, for an average uniaxial creep ductility ε_f of 0.32 for alloy 800 H and of $\varepsilon_f = 0.15$ for the 1% CrMoV steel, are shown in Figs. 1 and 2. It is apparent that good agreement is obtained when plane stress is assumed consistent with a numerical calculation shown in Fig. 3. There, results of finite-element calculations using two material laws are shown in comparison to experimental results.

Transient Crack Growth in 1 % CrMoV steel

It is claimed that the early cracking behaviour can be attributed to the combined effects of primary creep deformation, the development of a creep damage zone around the crack tip and redistribution of stress during the transition from the initial elastic to the steady-state creep conditions. An indication of the redistribution time can be obtained from [4]

$$t_1 = \frac{G}{(n+1)C^*} \qquad (3)$$

where G is the elastic strain energy release rate. Since this formula is considered to provide an upper estimate of t_1, stress redistribution should be essentially complete for $t > t_1$. In the case of the 1% CrMoV steel, t_1 is tyically around 10 h and for alloy 800H only a few minutes. Strictly speaking C^* is only valid for values of $t > t_1$.

It has been found that elimination of data points with $t < t_1$ still leaves most of the 'tails'. These tails can be attributed to successive damage accumulation in front of the crack tip and can be described by slightly different equations [3-6]. For most tests, it has been found that the build-up of damage occupies about the first 0.5 mm of the crack extension.

Initial Cracking Rates

It can be argued that the inital cracking rates for the 1 % CrMoV steel, with a transition time of typically 10 h or more before stress redistribution has had time to occur, should be described by K. The correlations for all tested specimens are shown in Fig. 4. It can be seen that all the data can be described satisfact- orily by the same equation taken from [7] independent of size and geometry of the specimen.

CONCLUSIONS

Experiments and analyses have been performed on a 1% CrMoV steel and on Alloy 800 H which have shown that creep crack growth in these materials is described most satisfactorily by the creep fracture mechanics parameter C* using a unified evaluation procedure. Recommendations have been made about how to obtain the most reliable estimates of C* from experimental measurements. These have been supported by numerical computations.

It has been demonstrated that all the cracking took place under plane stress conditions. Increased scatter, due to 'tails' in the early stages of cracking in the 1% CrMoV steel, has been shown to be caused mainly by the progressive build up of damage at the crack tip until a steady state distribution is reached. This can take a significant part of the life of a specimen and can be important in practical applications. However, little evident of a 'tail' was noticed for Alloy 800 H due to the rather short transition time of several minutes.

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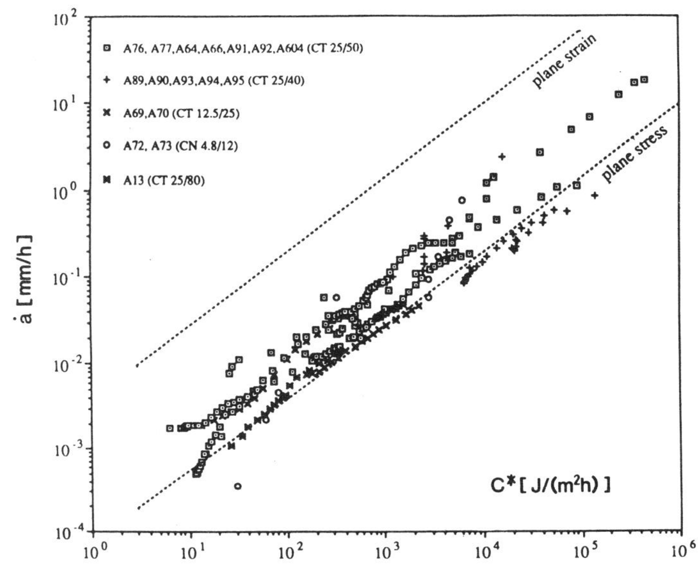

Figure 1 - Crack growth rate ȧ vs. C* integral for Alloy 800 H at 800°C.

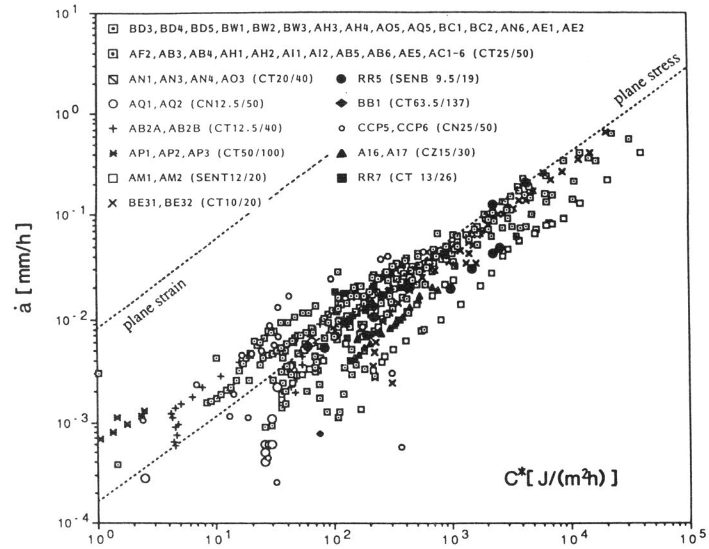

Figure 2 - Crack growth rate ȧ in 1% CrMoV steel at 550°C as a function of C*.

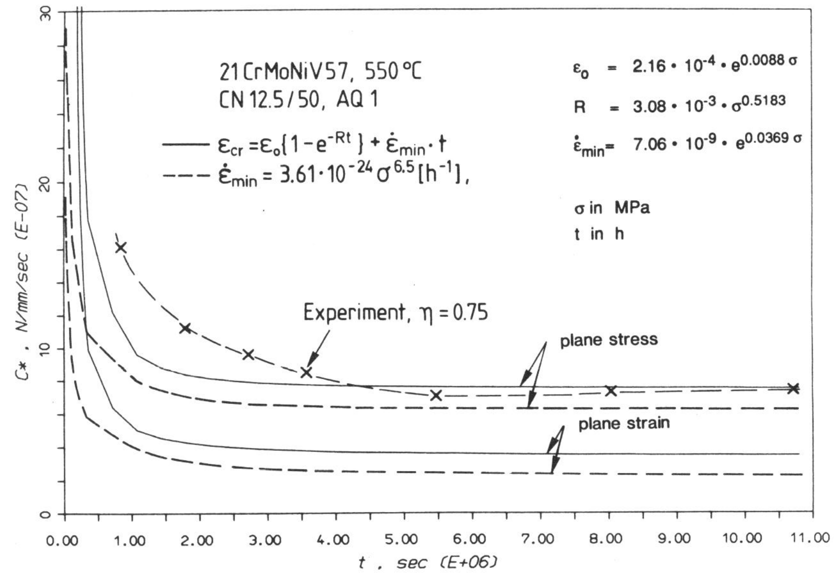

Figure 3 - Experimental and numerical values of C* for CN12.5/50 specimen AQ1, from [7]

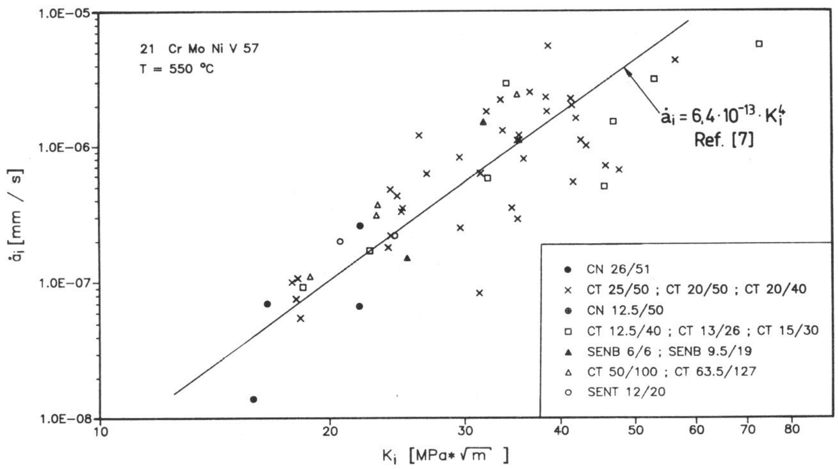

Figure 4 - Initial crack growth rate vs. initial stress intensity factor