Creep Crack Growth Behaviour of an AISI 316 Steel Plate for Fast Reactor Structures

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Abstract

The paper presents and analyses creep crack growth data obtained at 550, 600 and 650°C in air with SENT and CT specimens on type 316 stainless steel plate for LMFBR applications. Crack initiation and crack growth are tentatively correlated to $K$, $\sigma_{net}$ and $J^\theta$ taking into account the constraint conditions due to specimen geometry. The validity of these parameters is discussed following the concept of transition time from small scale creep at the crack tip to extensive creep within the ligament.

Post exposure microstructural and fractographic investigations do evidence that grain deformation processes are mainly responsible for cavity evolution.

1. Introduction

Relevant research work is under way to provide creep crack growth (CCG) data for type 304 and 316 austenitic stainless steel widely used in LMFBR construction, so to support design, inspection and operation of those components whose integrity does assure plant safety and availability.

As the assessment of possible defects within structures and components operating in the creep range requires the evaluation of subcritical crack growth, CCG of base and weld material is being extensively investigated and summarised, MUSICO [1] in the temperature range 550 to 850°C both in air and liquid sodium environments, at constant load or constant displacement rate, adopting different specimen sizes and geometries (mainly DEN, SENT, CN, CT). The crack extension is currently measured against LEFM and EPFM mechanical driving parameters ($K$, $\sigma_{net}$ and $J^\theta$). The data are generally characterized by large scatter band, moreover these experimental differences in geometry and testing parameter do not allow their straightforward comparison.

Also, for high ductility materials, the generation of a comprehensive CCG database should rely on unified reduction of the measured values obtained in fully representative experimental conditions and thus requires the accurate identification of the mechanisms and microstructural processes responsible for crack initiation and propagation. The CCG engineering approach, re-
ported by SADANANDA and SHAHINIAN [2], mainly recalling the modified Paris law: \( \frac{da}{d\tau} \propto P^n \) where \( P \) is a suitable fracture mechanics parameter, have to comply, as close as possible, with the component geometry as well. Some preliminary indications show the main influence of the constraint existing at the crack tip on the correlation of CGG with mechanical parameters.

Following our previous paper, on room and high temperature creep behaviour of type 316 SS plate by D'ANGELO et al. [3], we present here on the same material the creep crack initiation and propagation results already accumulated as well as some introductory elements on acting micromechanisms.

2. Experimental and results

The material under investigation was sampled from a 50 mm thick plate of SA 240 type 316 modified austenitic stainless steel that was solution treated at 1060°C with an holding time of 55 min. and water quenched. Metallography shows that 6 ferrite is absent and the nominal diameter of average grain section is 125 \( \mu \)m. The chemical composition of the material is given in Table I.

CGG tests were carried out using two specimen geometries: Single Edge Notched Tension (SEN) with cross section 12 mm by 20 mm and Compact Tension (CT) 1 inch thick. All the specimens were sampled from the plate in the transverse direction (TL) with the base of the notch lying in the through thickness direction. Initial cracks were usually obtained by electro-discharge machining (EDM) with a tip radius of 0.15 mm. Moreover, in order to evaluate the effect of the crack sharpness, some specimens have been also fatigue pre-cracked. Crack length during testing was continuously monitored by the direct current drop potential method, specifically calibrated, with an error on the medium crack length measurement lower than 100 \( \mu \)m.

The CGG tests were performed in air at 550°C, 600°C and 650°C, using constant load tensile testing machines. On the CT type specimen the measurement of loading line displacements was made by means of two Linear Voltage Displacement Transducers (LVDT) which are coupled with extensometer arms pinned to the specimen; sensitivity of about 1 \( \mu \)m is thus obtained.

Testing conditions are summarized in Table II. For values of experimental parameter initially applied, like the net section stress \( \sigma_{net} \), ranging from 80 MPa to 120 MPa, no detectable amount of CGG was obtained in allowed time. Therefore, in this case, just crack incubation has been observed and most of representative tests are now in progress. Nevertheless, an attempt has been made to quantify parametric and constitutive equations which describe the crack behaviour in the creasing material.

Crack initiation times, as the time spent by the initial crack to grow of about one grain diameter, were correlated with net section stress applied:

\[
\tau_{I} = A \sigma_{net}^{-\delta}
\]

(1)
where $t_i$ is expressed in hour, $q_{\text{net}}$ in MPa, $A$ and $\delta$ are constants depending on temperature. At 650°C it has been obtained $A = 2.06 \times 10^{18}$ and $\delta = 7.8$.

For various temperatures, the rate of crack initiation has been described using an Arrheniustype dependence on temperature, hence

$$\frac{1}{t_i} = A' \exp(-Q/RT)$$

(2)

where $R$ is the perfect gas constant, $T$ is the temperature in °K and $Q$ is the crack initiation activation energy. For a $q_{\text{net}} = 119$ MPa the interpolation gave $A' = 5.23 \times 10^9$ and $Q = 210$ kJ/mole.

The correlation of the CCG rates with $q_{\text{net}}$, based on the experimental data, follows the well known equation:

$$\frac{da}{dt} = A'' q_{\text{net}}^m$$

(3)

where $\frac{da}{dt}$ is expressed in m/hour and $q_{\text{net}}$ in MPa with $A'' = 6.5 \times 10^{-29}$ and $m = 10.4$, when the net section stress is ranging from 200 MPa to 400 MPa, at a temperature of 650°C. These values are very close to the open literature data on type 304 austenitic stainless steel given by KRONHOLZ et al. [4].

Finally, an attempt has been made to calculate the transition time as proposed by RIEDEL and RICE [5] :

$$t_r = \frac{K^2(1-\nu^2)}{E(n+1)J^B}$$

(4)

where $(1-\nu^2)$ is reduced to 1 in plane stress conditions, $E$ is the modulus of elasticity, $\nu$ is the Poisson ration, $n$ is the Norton law stress exponent and $J^B$ is the energy rate line integral. At 550°C a value of $t_r = 1.6$ hour was computed assuming from the experimental data $E = 157,000$ MPa, $n = 6$, the stress intensity factor $K = 32.2$ MPa $\sqrt{m}$ and $J^B = 1.6 \times 10^{-1}$ J/m².

For higher temperature the transition time is even smaller than half an hour.

3. Discussion and conclusions

The applicability of Fracture Mechanics parameter to CCG phenomena essentially depends on two factors: the first one is linked with the possibility of describing the degree of constraint or the cracked component as well as of calculating this parameter on complex structures; the second factor is on the contrary strictly connected with the capability of such parameter to be correlated with CCG rates as determined by means of tests on specimens under uniaxial loading conditions.

Crack propagation in type 316 steel is mainly governed by deformation processes. Due to high ductility of the material, the relaxation will locally reduce the stress at crack tip; even before that, the intergranular damage produces a growth of the defect as suggested by CURBISHLEY and LLOYD [6]. The net section stress seems to be a correlation parameter more
in keeping with CGG process than the Linear Elastic Fracture Mechanics Stress Intensity Factor.

These considerations are also confirmed by the calculated values both of the crack initiation times and of transition times, previously reported. They support the trend of type 316 steel to pass rapidly from a small scale creep to a widespread creep on the ligament of the cracked specimen.

A characteristic CGG rate curve as function of the $\sigma_{net}$ is given in Fig. 1, together with open literature data: a good agreement of the various sources is evinced, considering the different testing temperatures.

The applicability of the energy rate line integral $J^0$ is limited because of the difficulties to be measured on actual components. Moreover, HARPER and ELLISON [7] questioned its use when applied to CGG propagation coupled with wide creep deformation not localized at crack tip. In this case, $J^0$ calculated on the base of loading line displacements will show the rate of energy dissipation not only at crack tip but also on the whole specimen for deformation. From this point of view, it is interesting to observe in Fig. 2 that for a crack still in the incubation period, the loading line displacement, and therefore proportionally $J^0$, are considerably increased without any crack propagation.

The activation energy for the crack initiation, calculated here in 210 kJ/mole, is in the same order of magnitude of the stationary creep activation energy given by MORRIS and HARRIES [8] and it is rather smaller than the creep crack propagation activation energy experimentally evaluated by KAWASAKI and HORIGUCHI [9].

As far as the effect of the precracking type is concerned, EDM and fatigue pre-cracked specimens were investigated; Fig. 3 and Fig. 4 show the longitudinal sections at midthickness in LA1 and LA2 tests performed at about the same experimental conditions. It may be noticed how the crack tip in fatigue precracked specimen has been blunted up to 0.1 mm radius before any propagation, while in the EDM cracked specimen the crack tip radius substantially held the initial value of 0.15 mm. Nevertheless, it should be stressed how the crack propagation length in both mentioned tests gave similar figures ranging from 0.1 to 0.2 mm. However, it is worth to mention the possible effect of oxide growth at the crack tip in covering the propagation if measured by drop potential monitoring, particularly with sharp notches in fatigue precracked specimens.

The fracture surface for test LA4, given in Fig. 5, shows typical intergranular patterns. At higher magnification, Fig. 6, it exhibits wedge type intergranular cracks as usually obtained with high matrix deformation and important grain boundary sliding.

The accumulated results on type 316 austenitic stainless steel, while testing is still underway, showed that CGG on SENT and CT specimens (preferable EDM notched) can be correlated at the best with $\sigma_{net}$. It has been found that the material is subjected at high $\sigma_{net}$ values to
creep crack initiation, with an activation energy equivalent to creep deformation. It could be therefore anticipated that for type 316 steel, CCG is just effective on extensively cavitated base metal, as well as in embrittled welded joints.

Acknowledgements

Part of the experimental CCG work has been performed at ERA, Technology Ltd., U.K., under an ENEL-CEP contract. The authors are indebted to Miss F. Bianchi for metallography and to Mr. A. Percivati for CCG tests conduction at ENEL-CEP Material Laboratory.

References


Tab. I - Chemical composition (% wt)

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<th>Tested material</th>
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<th>Si</th>
<th>Mn</th>
<th>P</th>
<th>S</th>
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<th>R</th>
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<th>Cu</th>
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Tab. II - C.C.G. testing conditions for type 316 steel.

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<th>ENEL Ref.</th>
<th>Temperature (°C)</th>
<th>Initial crack length (mm)</th>
<th>Initial K a amplitude (MPa m0.5)</th>
<th>Initial net (MPa)</th>
<th>Testing time (hours)</th>
<th>Specimen type</th>
<th>Precracking type</th>
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<td>4.00</td>
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<td>83</td>
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<td>LA3</td>
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<td>4.44</td>
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<td>CT</td>
<td>fatigue</td>
<td>in progress</td>
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Fig. 1 - C.C.G rate as function of literature data.
2. Lloyd - 625°C-SEN [Int.Conf.Sheffield, C213-80,1980]
3. Lloyd - 500°C-SEN [" " " " " "]
4. Solignac - 550°C-CT20 [CEA-R-5253, Saclay, F,1984]

Fig. 2 - Typical trends of crack growth (a) and loading line displacement (γLL) during CCG tests.

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Fig. 3 - CCG in fatigue precracked specimen.

Fig. 4 - CCG in EDM cracked specimen.

Fig. 5 - EDM fracture surface.

Fig. 6 - Intergranular microcracks at CCG fracture surface.