Creep crack initiation and propagation on type 316 stainless steel

D.D'Angelo & V.Regis
ENEL, Centro Ricerca Termica e Nucleare, Milano, Italy

1. INTRODUCTION

Although in elevated temperature design, current codes and practices disregard fracture properties and subcritical crack growth of structural materials for power plants, recently attention has been paid very extensively to the evaluation of possible defects induced during fabrication and construction stages or generated by the operation of the components. It has been widely acknowledged for instance that, in geometric and metallurgical singularities, defects might initiate, and are therefore expected to propagate, unless particular conditions are met which will result in crack arrest.

The investigations carried out on type 316 stainless steel plate for high temperature applications allowed to assess creep crack growth (CCG) properties of base material with respect to initiation and propagation phases. It has been found (Bell 1986) that for high ductility materials, as type 316 stainless steel, incubation time is an overwhelming fraction of total propagation time of cracks in structures operating in the creep range.

In order to avoid an over estimation of the different metallurgical and mechanical contributions to creep crack growth, which would cause undue safety margins, and to permit the development of more confident defect assessments, it is necessary to reach:

i) improved experimental testing methodologies,
ii) implemented data processing techniques,
iii) a suitable selection of proper Fracture Mechanics parameters.

Jointly with the presentation of additional CGG results (D'Angelo 1985) we will try here to analyse and to discuss, into some extend, experimental and theoretical issues strictly connected to CCG initiation and propagation, with a view to identify the application range of different Fracture Mechanic parameters, as supported by a detailed investigation of fractography and microstructural feature of the crack tip region. Possible mechanisms of defect initiation and propagation will be consequently introduced.

2. EXPERIMENTAL

Type 316 modified stainless steel of a 50 mm thick plate, solution treated for 55 minutes at 1060° C and water quentched, follows fair closely chemical specifications and exhibits zero δ ferrite content with a medium grain size of 123 µm.

The main results of tensile, thoughness, room and elevated temperature creep, creep-rupture and relaxation properties, already reported (D'Angelo 1983), showed a very ductile behaviour of the material in the as-received condition, although sensitizing heat heatment (100 h at 750 °C) induced a noticeable drop of upper shelf impact energy (about 50% at room temperature). Furthermore, at high strains (>0.5%) low cycle fatigue data in the 550-650°C range suggested the opportunity to adopt more conservative reference curve.

We will report here more extensively on CCG testing at 550, 600, 650 °C with net applied stresses, σ net, ranging from 80 to 350 MPa. Single Edge Notched in Tension (SENT) with a 12x20 mm cross section and 1" Compact Tension (CT) specimens, machined in TL orientation with the notch in the thickness direction, were used; initial crack has been obtained both by electrodischarge machining (EDM) with a notch radius of 0.15 mm or by fatigue precracking at room temperature accordingly to ASTM E 399-83. As it can be observed in the experimental matrix (Tab. 1) single CCG tests have lasted from 10 hours to 5.000 hours.

Table 1. CCG experimental matrix and testing conditions

Test Ref.	Temperature °C	Initial crack length mm	Net stress MPa	LEFM K_I MPa$\sqrt{m}$	C^* J/m^{2}s	Testing time hours	Initiation time hours	Specimen type	Precrac.
CTA3	550	26.53	339	35	$1.01\cdot10^{-1}$	1653	$\geqslant$1653	CT	FATIGUE
CTB4	"	26.79	397	41	$1.08\cdot10^{-1}$	1600	500	"	"
CTC5	"	25.27	333	38	$1.22\cdot10^{-3}$	7753	3362	"	"
CTD6	"	25.84	561	60	$2.93\cdot10^{-1}$	217	44	"	"
CTX4	"	26.04	563	60	$2.73\cdot10^{-1}$	194	$\simeq$ 50	"	"
CTX5	"	25.18	333	36	$5.62\cdot10^{-1}$	1295	$\geqslant$1295	"	"
CTX6	"	25.3	339	37	$1.70\cdot10^{-1}$	2810	1000	"	"
BHS 100	600	25.50	306	33	-	1504	520	"	"
BHS 101	"	27.00	365	37	-	195	10	"	"
LA1	650	3.95	83	10	-	4777	3270	SENT	"
LA2	"	4.00	82	11	-	4777	$\geqslant$ 4777	"	EDM
LA3	"	4.53	154	20	-	1917	817	"	"
LA4	"	4.00	218	27	-	7	$\simeq$ 1.5	"	"
LA5	"	4.997	133	19	-	4423	2071	"	FATIGUE

Crack growth monitoring during the test is assured by direct current potential drop method (PD). The sensitivity of the measurements, close to 30 µV/mm, allows to discriminate crack length increments of 0.03 mm. In fact, taking into account the contribution of the crack tip blunting, of the oxyde bridging fractured surfaces and of the thumb nail crack propagation front, we assumed that an error of 100 µm on the measurement of the crack length should be tolerated.

During the tests, we also monitored the loading line displacement (V_{ll}) by means of a double extensometer with 5 µm sensitivity inductive gauges. Usually the specimen is hold at testing temperature for 8 to 24 hours before applying the load, looking for PD stabilization: this,in order to avoid initial drop in PD measurements often reported on fatigue precracked specimens (Hollstein 1985).

CCG data are processed initially by eye in order to identify univocally the initiation of crack propagation and then by computer to obtain, from PD vs time measurements, crack length, crack growth rate and fracture mechanics parameters.

3. RESULTS AND DISCUSSION

Experimental results will be considered with regard to crack initiation and crack propagation issues: typical PD and V_{ll} behaviours versus time are given (Fig. 1), they should be considered as an updating of the results of an on-going programme (D'Angelo 1986).

3.1 Crack initiation

The very istant where crack propagation do initiate is generally difficult to detect in Fig. 1 type curves. Though many criteria have been proposed, no complete experimental support is till now available (Schwalbe 1984). This difficulty may be explained by possible PD changes, with no crack length increase, in correspondance of crack tip blunting in high ductility materials or correlated to resistivity changes induced by strain or damage ahead the crack tip or also following long term ageing and microstructural evolution. It should be remarked from the analysis of V_{ll} vs time curve that its behaviour can be rationalised on the basis of three main contributions: instantaneous plastic strain, creep strain and creep crack growth followed by crack opening. A model has been proposed (Harper 1977) describing V_{ll} in the three different stages:

i. stress redistribution and crack tip blunting, with a decreasing V_{ll} rate;
ii. stress redistribution and crack initiation, with constant V_{ll} rate;
iii. predominant crack gowth, with increasing V_{ll} rate.

We have therefore assumed for crack growth initiation time, the instant where V_{ll} rate exhibit a minumum or the inflexion point of V_{ll} curves vs time, generally also coincident with that of PD. At correspondent value of PD has been associated the medium crack length ($\bar{a}_0$) as measured in 12 locations at specimen fracture surface (it should be stressed that these measurements also include the plastic deformation at the crack tip). Calibration curve is therefore obtained relating linearly the initiation ($\bar{a}$ o, PD at turning point) values with final medium crack length and PD at the test interruption.

Initiation times, t_i, have been correlated (Fig. 2) for the different testing temperatures with σ net and C^* line integral of energy rate, which provided at 550 °C:

$$t_i = 5.33 \times 10^{-19} \cdot \sigma^{-6.522}$$

where t_i is expressed in hours and σ in MPa.

Quantitative metallographic investigations on tested specimens provided better understanding of crack blunting and an evaluation of effective crack initiation time, thus substantiating the predictive model.

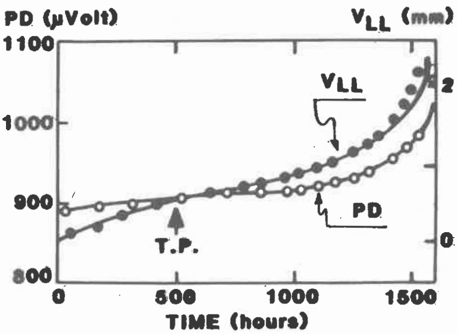


Figure 1. PD and V_{ll} behaviours at 550°C (test CTB4)

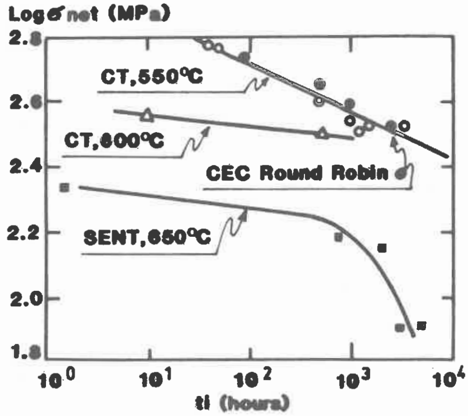


Figure 2. CCG Initiation times

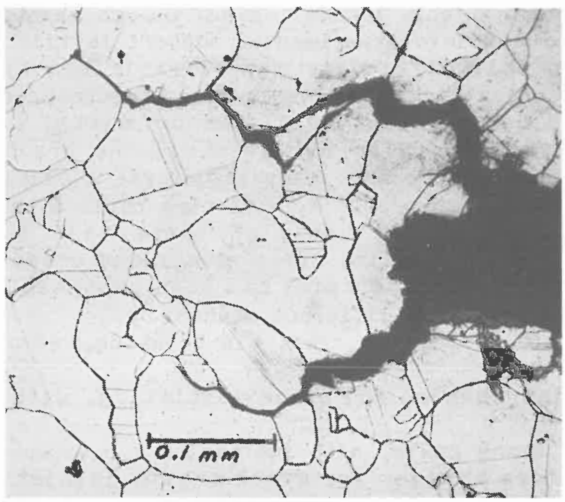


Figure 3. Intergranular crack propagation patterns at blunted notch

SENT specimens, respectively with EDM and fatigue notch, both showed a 0.13 mm increase of notch radius, and an intergranular crack growth from 0.20 to 0.46 mm in agreement with analytical PD calibration curves. CTX5 test on a CT specimen fatigue precracked has been interrupted very close to PD and V_{ll} turning point, as defined from the CTX6 test still running. Optical metallography (Fig. 3) gave for the blunted crack tip a measurement of the radius of 0.03 mm against the initial of 0.25 μm, and maximum integranular crack growth of 0.25 mm. These observations jointly with additional data do support the consistance of to crack initiation at minimum V_{ll} rate.

Let's assume now an increase of plastic time-indipendent crack tip radius (Shih 1979):

(3) $$\Delta r_p = 0.5 \cdot d_n \frac{K_I^2}{E \sigma_o}$$

in plane stress conditions, with σ_o material flow stress, K_I LEFM stress intensity factor, E Young's modulus and d_n a dimentionless function of the strain hardening exponent and of yield strain.

Crack tip blunting due to creep, could be deduced, by means of the elastic analogy, from:

$$(4) \qquad \Delta r \cong d_n \cdot \frac{J}{\sigma_o}$$

where J is the energy line integral of the Elasto-Plastic Fracture Mechanics, and conseguently:

$$(5) \qquad \dot{\Delta} r \cong d'_n \cdot \frac{C^*}{\sigma_i}$$

where σ_i is assumed to be the internal stress (Threadgill 1971) and C^* follows the equation (Koterazawa 1977):

$$(6) \qquad C^* = \frac{2F}{B(w-a)} \cdot \frac{dV_{11}}{dt}$$

By integration of equation (5) $(t=0;\dot{V}_{11}=0)$ at creep crack initiation time t_i we should have a creep blunting given by:

$$(7) \qquad \Delta r_c = \frac{2 \cdot F}{B \cdot (w-a)} \cdot (V_{11})_{t_i} \cdot 0.5 \frac{d'_n}{\sigma_i}$$

Δr_p and Δr_c bluntings are the largest at crack initiation since when crack propagates it will unload the crack tip.

With all the values specific of our test conditions CTX5, for $d'_n=1$ (limit condition of non-hardening) and the V_{11} measured value of 0.13 mm we get:

$$\Delta r_{tot} = \Delta r_p + \Delta r_c = 0.0175 \text{ mm}$$

which appear consistent, though underestimated, with respect to the measured value of 0.03 mm.

3.2 Creep crack growth

For type 316 stainless steel we calculated transition time t_{tr} following Riedel and Rice equation (Riedel 1980):

$$t_{tr} = \frac{K^2 (1-\gamma^2)}{E (n+1) C^*}$$

giving the time required to shift from elastic conditions to creep. At temperatures in the range 500°C to 650°C two hours transition times are obtained, thus showing that comparatively long incubation times make difficult to consider K as a sound parameter for describing stress/strain at crack tip and therefore as a parameter well correlated with CCG rate.

These considerations are consistent with the evaluation of the region (Riedel 1983) where creep strain is dominating.

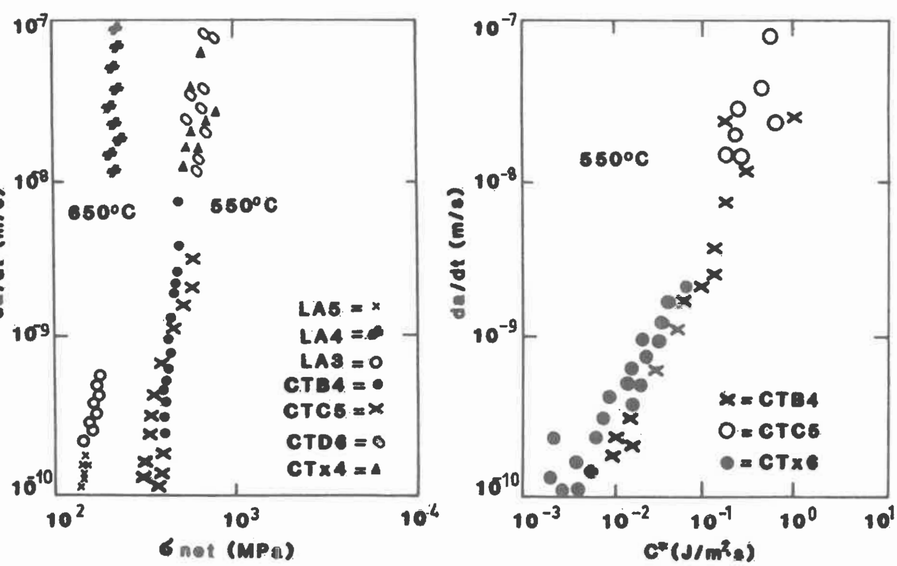
$$r_{creep} = K_I^2 (E.B.t)^{2/n-1} . F_c(\vartheta)$$

$F_c(\vartheta) = 0.5$ for $\vartheta = 0$ in plane stress and for $n=6$. B and n follow the Norton low: $\dot{\varepsilon} = B\sigma^n$ and are defined from creep testing. With our typical experimental values at 650°C: $r_{creep} \cong 55$ mm, close to the crack length and to the specimen ligament dimensions.

For very ductile materials stresses redistribution at crack tip is the predominant factor with respect to crack propagation (Musicco 1983), also σ net calculated on the ligament could be used instead of C* for correlation with creep crack growth. However, transient C^* (Ellison 1985), quite similar to that calculated by means of equation (6) based con V_{11} measurement in ductile steel as type 316 at high stresses, better consider plastic deformation and primary creep at crack tip. We have therefore correlated CCG rate with C^* and σ net (Figg. 4 and 5). Least squares interpolation at 550 °C provided for CCG rate in mm/hr, where σ net is expressed in MPa and C* in J/m². s:

$$\frac{da}{dt} = 1.29.10^{-29} \; \sigma_{net}^{\;8.837} \qquad\qquad \frac{da}{dt} = 1.49.10^{-4} \; C^{*1.142}$$

Micromechanisms governing crack initiation and propagation may be rationalised according to Ashby fracture maps (Ashby 1978). We have to consider here $T/Tm \cong 0.5$ and σ^{net}/E ranging from 6.5 10^{-4} to 4.5 10^{-3}: thus making it possible intergranular creep fracture for long exposition times, transgranular creep cracking at medium term exposures and lastly relatively fast ductile fracture. CCG testing on type 316 specimens will develop stress redistribution at the crack tip, strain accumulation and intergranular cavities nucleation and proceed at first with low CCG rate (Stage I) followed by an acceleration (Stage II) as showed in Figg. 4 and 5.



Figures 4, 5. Type 316 CCG rates vs σ_{net} and C*

Low initially applied stresses have been here experienced to produce almost the same CCG rate as higher stresses (Huthmann 1987): they will just differ by the initiation time, which is needed to bring crack tip strained region up to a critical value of deformation to fracture as well as the remaining ligament material subjected to a gradient of damage by creep strain, cavities and microstructural damage. Moreover, just metallographic and fractographic investigations may discriminate between different behaviours through which CCG evolves (Airoldi 1987).

4. CONCLUSION

Initiation times and CCG behaviour of type 316 stainless steel have been obtained at different temperatures. CCG rate has been correlated with K_I, σ net and C^*, paying a great attention to experimental data reduction in order to decrease data scatter and improve evaluation of rest life on operating defected components.

Testing on SENT and CT specimens, EDM notched and fatigue precracked, made it possible to propose a model well supported by the experimental data to predict crack tip blunting and therefore to measure crack initiation times.

An extension of metallographic work is in progress in order to better understand micromechanisms involved.

ACKNOWLEDGEMNTS

The Authors are grateful to ENEL for permitting publication of this paper.

The experimental work has been performed at ERA Technology Ltd, at Istituto Ricerche Breda and at ENEL - Thermal and Nuclear Research Center.

Special appreciation is due to Miss P. Bianchi and to Mr. A. Percivati respectively for metallography and for CCG tests conduction at ENEL - CRTN - Materials Laboratory.

REFERENCES

Airoldi,G. Bedodi, S.D'Angelo, D. 1987 The small scale creep zone in 0.5Cr 0.5Mo 0.25V steel during creep crack growth tests. Third Int. Conf. on Creep and Fracture of Engineering Materials and Structures. University College of Swansea, U.K.

Ashby, M.F. Fields, R.J. Weerasooriya, T. 1978. Fracture mechanisms in two austenitic and one ferritic steels.
University of Cambridge, Department of Engineering, Report n. CUED/C/MATS/ Tr. 47 U.k.

Bell, K. 1986. The characterization of creep crack initiation from pre-existing defects using fracture mechanics concepts.
COST 501 - ERA Report n. 86 -0166 R, Leatherhead, U.K.

D'Angelo, D. Ragazzoni, S. Regis, V. 1983. Room and high temperature creep behaviour of an AISI 316 steel plate for fast reactor structures.
Paper L 12/6 - Proc. of 7 the Int. Conf SMIRT, Chicago, USA

D'Angelo, D. Regis, V. 1985. Creep crack Growth Behaviour on type 316 SS plate for fast breeder reactor components. Paper F2G1/6, Proc. of 8th Int. Conf. SMIRT, Bruxelles, B.

D'Angelo, D. Regis, V. 1986. Relevant mechanical characteritics of an AISI 316 steel plate for high temperature structural integrity evaluation. Proc. of 2nd Int. Seminar on Standard and structural analysis in elevated temperature applications for reactor technology. Venice, I.

Ellison, E.G. Musicco, G.G. Pineau, A. 1985. State of the art report on Fracture Mechanics (Fracture in the creep range). CEC Study Contract No RAP-053-UK Vol. I Report, Vol. 2-3 Appendices.University of Bristol, Dep. of Mechanical Engineering, U.K.

Harper, M.P., Ellison, E.G. 1977. The use of the C^* parameter in predicting creep propagation rates. Journal of strain analysis, Vol. 12 No. 3: 167-179

Hollstein, T. Kienzler, R. 1985. Numerical Simulation of Creep Crack Growth Experiments. COST 505 Project D22. Fraunhofer Institut for Werkstoffmechanik. Annal Progress Report. Freiburg, W.G.

Huthmann, H. Aubert, M. Curbishley,I. D'Angelo, D. 1987 Round Robin Tests on the Creep Crack Growth Behaviour of 316 L Mod. Stainless Steel. CEC - Study Contract n. RA.071.D-Final Report (to be published).

Koterazawa, R. Mori, T. 1977. Applicability of Fracture Mechanics Parameters to crack propagation under creep conditions, Trans. ASME, J. of Eng. Mater, Technology, 99:298-305.

Musicco, G.G. 1983. A review on the creep crack growth on AISI 304 and 316 stainless steels. CEE -WGCS Activity Group 3 Contract Report, JRC Ispra, SA/1.07.01.83.12. I.

Riedel, H. Rice, J.R. 1980. Tensile Crack in Creeping Solids. Fracture Mechanics Twelfth Conference, ASTM STP 700 : 112-130.

Riedel, H. 1983. The use and the limitations of C^* in creep crack growth testing. Proc. of Int. Symposium on Fracture Mechanics, Science Press, Beiging, China, pp. 987 - 1018.

Schwable, K. H., D. Hellmann 1984: Correlation of stable crack growth with the J-integral and the crack tip opening displacement, effects of geometry, size, and material, Progress report on an investigation of the elastic-plastic R-curve methodology, GKSS 84/E/37, GKSS Research Centre, Geesthacht.

Shih, C.F. 1979. Relation ship between the J-integral and crack opening displacement for stationary and extending cracks. General Electric, Report n. 79 CR. 0075.

Threadgill, P.L., Wilshire, B. 1971. Mechanisms of transient and steady state creep in a γ' hardned austenitic steel. Proc. of 2nd Int. Conf. on Strength of Metals and alloys, ASM.