The Growth of Small Cracks in the Single Crystal Superalloy CMSX-4 at 750 and 1000°C

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Abstract

Creep-fatigue crack growth experiments have been carried out on alloy CMSX-4 single edge notched specimens at 750°C (maximum root-temperature) and at 1000°C (maximum aerofoil temperature). At 750°C and below the cracks, controlled by the K-concept, followed a zigzag line due to alternate gliding on the {111} and {100} planes. An influence of the test atmosphere was observed, the tests in vacuum exhibiting higher crack growth rates than those measured in air during the early stages of testing. At high temperatures, such as 1000°C, the crack propagation was along the {100} planes.

Introduction

During the last decade, there has been a dramatic development in heavy-duty, stationary industrial gas turbines for electrical power generation. The application of cooling systems, the introduction of single crystal technology, both originally developed for aircraft engines, and the use of thermal barrier coatings have enabled gas temperatures as high as 1400°C to be reached. Figure 1 indicates the development line for turbine blading materials. One of the currently favoured blading materials is the second-generation single crystal alloy CMSX-4, with which long term material temperatures up to 950°C can be achieved. This alloy exhibits excellent creep resistance at these high temperatures due to hardening by γ' precipitates (more than 70% by volume) with optimized distribution and morphology and due to solid solution hardening by 3 wt.% Re addition. Single crystal alloys, however, exhibit a strongly anisotropic deformation behaviour.

Only limited information exists in the literature regarding the behaviour of technical cracks in single crystal Ni-base superalloys (mainly observed with CT specimens). Up to test temperatures of about 800 to 850°C, the crack propagation in polycrystalline Ni-superalloys is correlated with the stress intensity factor K_I. Numerical and finite element (FE) analyses have shown differences between the crack growth behaviour in isotropic and anisotropic Ni-superalloys /1, 2/. Therefore, the use of similar K_I values for the description of crack growth phenomena in anisotropic single crystalline Ni-superalloys is totally uncertain, because the role of gliding systems for the crack propagation is unknown. However, numerical estimations /3/ have demonstrated that similar K values are obtained for cracks perpendicular to the stress direction and cracks with deviations of up to 30° from the stress direction.

The typical temperature dependence of yield and rupture strength of superalloys and the different deformation mechanisms (dislocations and particles interactions) control the temperature dependence of the growth of macroscopic flaws. Only limited results for fatigue and creep-fatigue crack growth of small cracks in single crystal alloys, especially for CMSX-4, are available in the literature, e.g. /4-9/. Unfortunately, there are some results in the open literature that are contradictory and need clarification.

Estimation of allowable operational time and allowable numbers of operational cycles requires information concerning the initiation and growth of small flaws in the area of root (about 750°C) and the blade (about 1000°C). Therefore, a study of the growth of small cracks at 750 and 1000°C was undertaken for the alloy CMSX-4. The possible influence of test atmosphere was also studied, by comparing the results of crack growth tests in air with those from tests carried out in vacuum.

Experimental details

Material

Single crystalline superalloys, first developed for aero gas turbines, exhibit a significant improvement in creep and fatigue resistance, compared with conventionally cast superalloys, e.g. IN 738 LC, so that a potential increase of 80C in the materials operating temperature may be obtained. From the application in aero gas turbines, three classes of single crystalline superalloys may be denoted: the first generation with reduced grain boundary hardening alloy elements; the second generation with additions of about 3 wt.% Re; and the third generation alloys which have somewhat higher Re contents (about 5 to 6 wt%).

Alloy CMSX-4 is a typical second-generation single crystal material hardened by about 70 vol% γ' with solid solution strengthening of the γ channels in-between the cuboidal γ' particles due to the addition of 3 wt.% Re. The nominal chemical composition of CMSX-4 in wt% is as follows: Ni bal, 5.6 Al, 1.0 Ti, 6.5 Ta, 6.5 Cr, 0.6 Mo, 6.0 W, 9.0 Co, 3.0 Re, 0.1 Hf, less 0.002 C. The heat treatment consists of: 6 h solid solution treatment at 1305°C with controlled cooling rate, followed by a two stage ageing treatment of 6 h at 1140°C and 20 h at 871°C, both with controlled cooling rates. The material was as delivered as single crystalline plates, directionally solidified in the <001> direction and fully heat treated.

Small cracks in turbine blades were simulated by single edge notched specimens (SEN) with artificial edge or corner cracks. Figure 2 shows the details of the test pieces used for crack growth experiments.

Superalloys 2000
Edited by T.M. Pollock, R.D. Kissinger, R.R. Bowman, K.A. Green, M. McLean, S. Olson, and J.J. Schirra
TMS (The Minerals, Metals & Materials Society), 2000

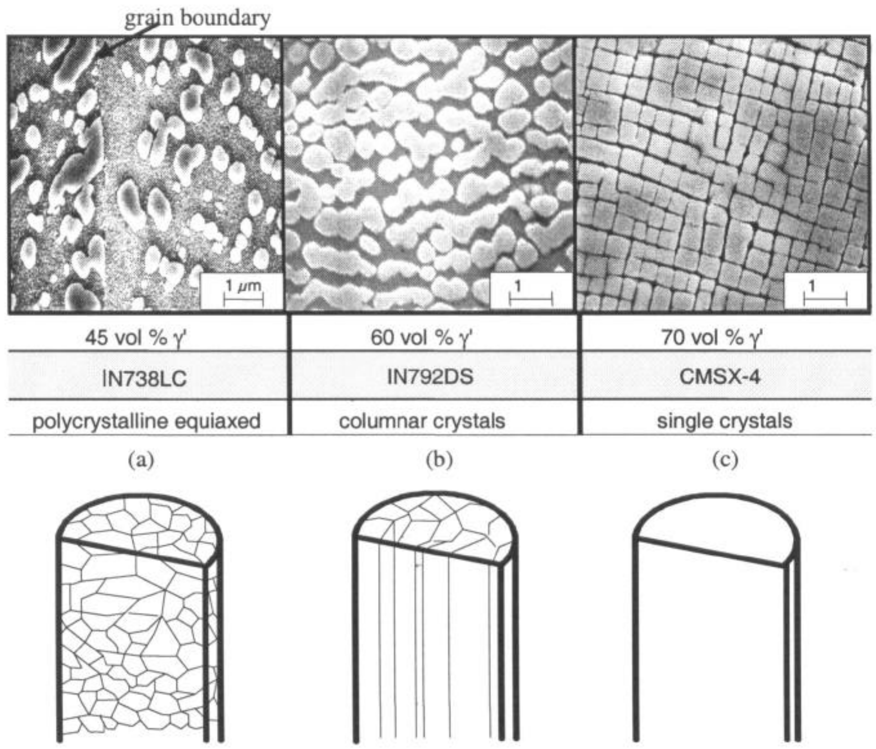


Figure 1
Schematic diagram of alloy development for superalloys: (a) polycrystalline, equiaxed conventionally cast IN 738 LC, (b) columnar grained directionally solidified IN 792 and (c) single crystalline alloy CMSX-4.

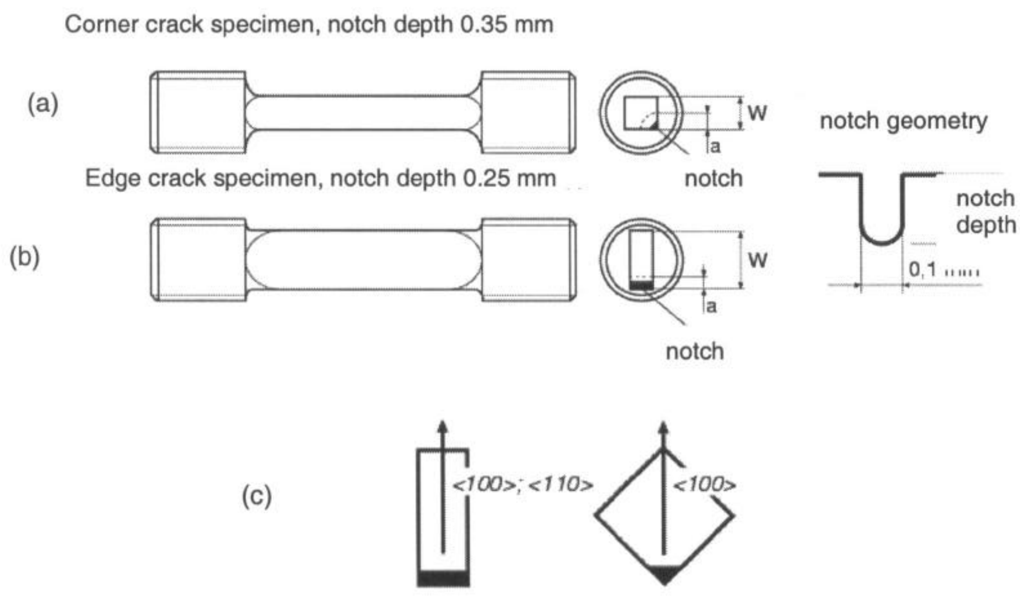
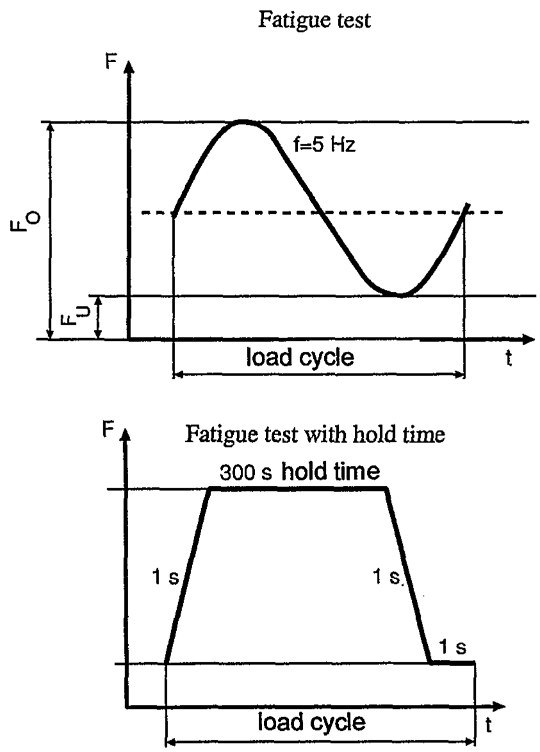


Figure 2
Geometries of SEN specimens; (a) corner crack, cross-section 8 x 8 mm, (b) edge crack, cross-section 4.5 x 12 mm, (c) secondary crack orientations



Load ratio $R = F_u/F_o = 0.1$

Figure 3
Schematic presentation of the loading cycles for fatigue tests with and without hold times

Some fracture mechanics equations

The K_I function for SEN specimen under stress is normally given by

$$K_I = \sigma \sqrt{\pi a \cdot c} \qquad (1)$$

where K_I is the stress intensity, σ the applied stress, a the crack length and c a correction factor that depends on the crack length a and the specimen geometry. For small cracks, $c = 1.12$ and for larger cracks, c is given by

$$1.99-0.41\left(^{a}/_{w}\right)+18.7\left(^{a}/_{w}\right)^{2}-38.5\left(^{a}/_{w}\right)^{3}+53.8\left(^{a}/_{w}\right)^{4} \qquad (2)$$

where w is the specimen width, for $0 \le (^{a}/_{w}) \le 0.6$.

SEN specimens with corner cracks exhibit a much more complicated stress distribution ahead of the crack tip than edge crack specimens and represents the realistic crack geometry within a component due its three-dimensionality. Mathematical results are given in /10, 11/, in which a square edged crack surface area is estimated. Thereby one differentiates between the stress intensity factors along the surface of the specimen and in the direction of 45°. Then the stress intensity factor across the whole crack surface may be estimated by

$$K_{I\,mean} = \frac{K_{I\,45^\circ} + K_{I\,surface}}{2} \qquad (3)$$

$$= \left(0.97 - 0.09\left(\frac{a}{w}\right)^2\right) \cdot K_{I\,surface} \qquad (4)$$

The approximations help in understanding the crack propagation of a corner edged crack. Further estimations as well as the experimental observations indicate that the K_I-concept may be used for both test temperatures.

Test conditions

The fatigue and creep fatigue experiments were carried out at 750 and 1000°C in air and in vacuum (ca 2.5×10^{-5} bar) using a servohydraulic test machine. The fatigue loading was sine-wave with a frequency of 5 Hz and the creep fatigue load cycle was trapezoid with a hold time of 300 s at maximum load and 1 s at minimum load, as shown in Figure 3. To obtain a sharp crack the notched specimens were subjected to a 10 Hz sine-wave load 50% above the ΔK value and a load ratio of 0.1 at the test temperature before the actual test. The induced cracks were 0.4 – 0.5 mm long in the edge crack specimens and 0.3 – 0.4 mm in the corner crack specimens. The test were stopped at a crack length to specimen width ratio of 0.5.

Results and discussion

Figure 4 shows the results of crack growth experiments carried out at 750 and 1000°C in air and in vacuum. At 750 and 1000°C, the fatigue crack growth behaviour of the specimens resulted in the expected functional behaviour ("Paris-Erdogan") of da/dN versus ΔK_I (the cyclic stress intensity factor). At 750°C, the threshold values were higher and the slope of the Paris equation not so steep compared to the values at 1000°C. The influence of crystal orientation seemed to be more marked at 750 than at 1000°C. Specimens with a <100> crack orientation came to a rapid fracture by a spontaneous change to the {111} sliding planes. The <110> crack orientation did not show this behaviour.

As a consequence, one may assume that the position of the crack front to the {111} sliding planes is responsible for the occurrence of the observed rapid fracture. At 1000°C and for low ΔK values and low crack growth rates, the crack growth behaviour may be understood as typical crack behaviour of small cracks. This behaviour could be explained by the crack closure because of plastic deformation at the crack tip, oxidation and depletion of the crack surface areas and the start of the γ' rafting process. These influences blunt the crack tip and the stress singularity decreases, so that the crack could be stopped or slowed down.

In CMSX-4, oxidation at the crack tip became more important at the higher test temperatures. The comparison of the crack growth experiments in air and vacuum showed that the oxidation process influences significantly the crack initiation point or the initial stage of crack growth (see Figure 4).

Fractographic examinations using scanning electron microscopy (SEM) indicate for CMSX-4 (high volume fraction of γ') a slightly different behaviour compared to equiaxed Ni alloys with γ' volume fractions below 50%.

The crack surface of CMSX-4 at 750°C in air followed at low ΔK-values the γ channels or the γ/γ' interface region. At high ΔK-values, a change in the crack surface growth to the {111} plane was observed and cutting of γ' precipitates occurred.

Metallographic examinations to reveal the mechanisms of crack growth (see Figure 5) showed that:

- at 750°C and low and medium ΔK values, the propagation followed a zigzag line by changing the orientation along the {111} and {100} planes, mainly following γ/γ' interface. At higher ΔK values sliding along the {111} planes by cutting of the γ' precipitate occurred.
- at 1000°C, crack propagation along the {100} planes was mainly observed, which means propagation by cross sliding on {100} planes. The cutting of γ' precipitates became difficult and the crack path followed the γ/γ' interface areas along {100} planes.

Figure 6 compares the fatigue and the creep-fatigue behaviour at 1000°C. The edge crack specimen showed the same threshold values for both types of test, but the creep crack curves did not exhibit any changes in the crack growth rate. Therefore one may expect that creep-fatigue is more influenced by the deformation at the crack tip than by oxidation. If K is used as the stress intensity factor controlling the creep crack behaviour, the fatigue and the creep-fatigue results lie in the same range. At high K values and crack growth rates, the differences between fatigue and creep-fatigue became more significant. Because of these observations, one may conclude that creep dominates the crack growth process at low ΔK or K values, and fatigue at high ΔK or K values.

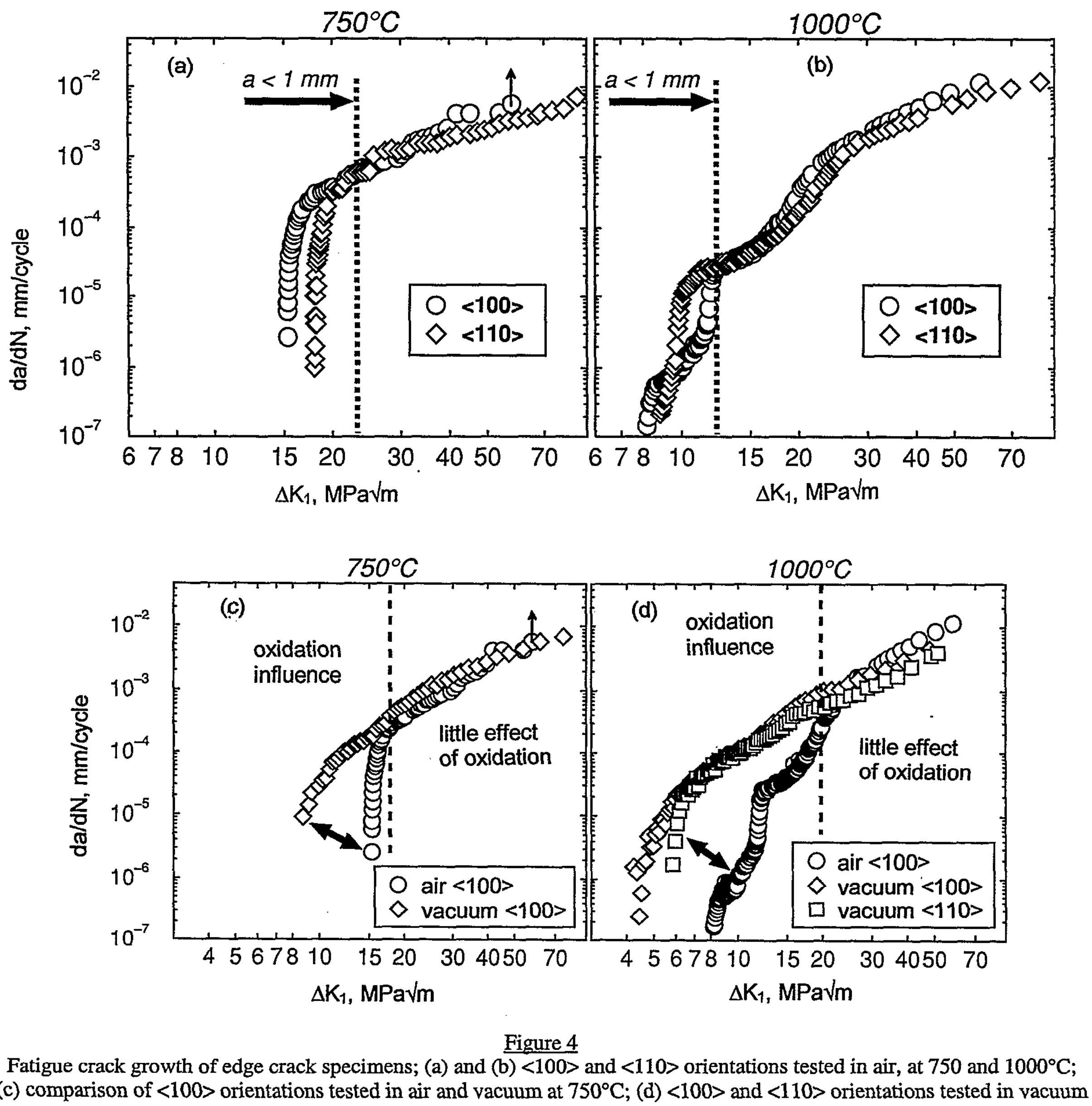


Figure 4

Fatigue crack growth of edge crack specimens; (a) and (b) <100> and <110> orientations tested in air, at 750 and 1000°C; (c) comparison of <100> orientations tested in air and vacuum at 750°C; (d) <100> and <110> orientations tested in vacuum at 1000°C compared with <100> orientation tested in air

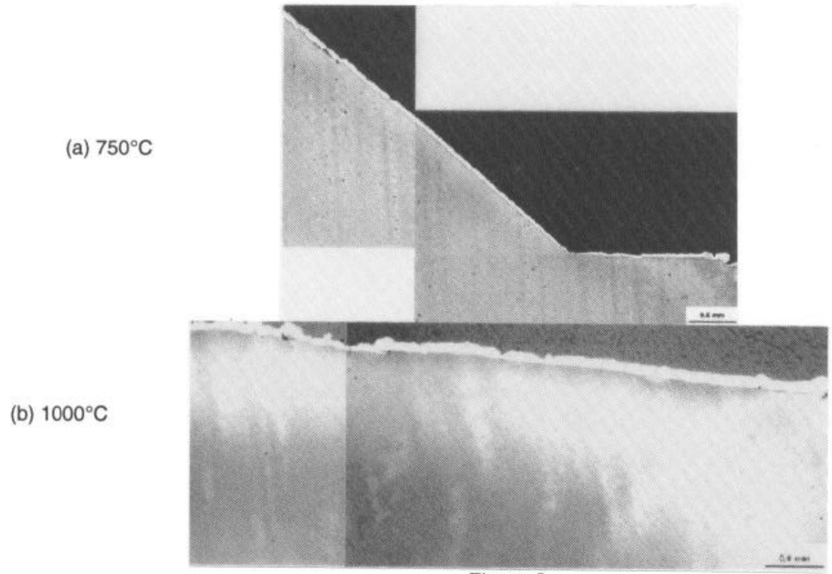


Figure 5
Crack surfaces of corner edge specimens of CMSX-4 at (a) 750 and (b) 1000°C in air

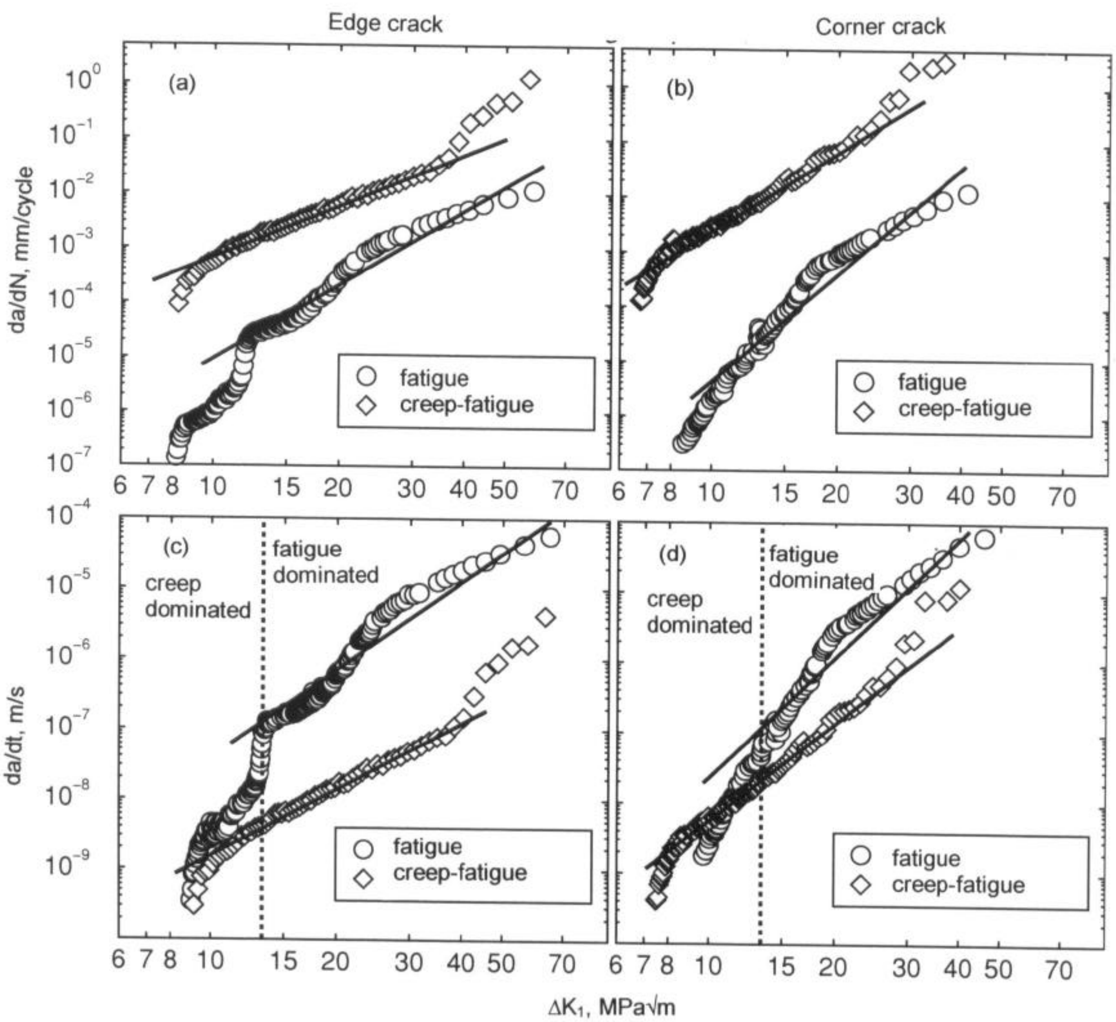


Figure 6
Fatigue and creep/fatigue crack growth of CMSX-4 specimens tested in air; (a) edge crack, ΔK fatigue parameter, (b) corner crack, ΔK fatigue parameter, (c) edge crack, K creep crack parameter, (d) corner crack, K creep crack parameter

Conclusions

At 750°C, the crack growth threshold values are higher and the slope of the Paris equation not so steep compared to the values at 1000°C.

The oxidation behaviour at the crack tip becomes more important at the higher test temperatures. The oxidation process influences significantly the crack initiation point and the initial stages of crack growth.

At 750°C and low/medium ΔK values, the crack follows a zigzag line by alternating between the {111} and {100} planes, mainly following γ/γ' interface. At higher ΔK values, sliding along the {111} planes by cutting the γ' precipitate occurs.

At 1000°C, crack propagation along the {100} planes is observed, which means propagation by cross sliding at {100} planes. The cutting of γ' precipitates becomes difficult and the crack path follows the γ/γ' interface areas along {100} planes.

At high K values and high crack growth rates, the differences between fatigue and creep-fatigue become more significant. It may be concluded that creep dominates the crack growth process at low ΔK or K values, and fatigue at high ΔK or K values.

Acknowledgement

The support of this work by Deutsche Forschungsgemeinschaft in the Priority Programme "Microstructure and Mechanical Properties of Metallic High Temperature Materials" is gratefully acknowledged. This was a preliminary investigation for the co-ordinated research work with the RWTH Aachen, SFB 561: "Thermomechanical behaviour of highly cooled, porous, multilayer structures for combined cycle plants."

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