INFLUENCE OF ENVIRONMENT ON MICROCRACK PROPAGATION IN HIGH TEMPERATURE LOW CYCLE FATIGUE

M. Reger and L. Rémy*

The growth of the dominant crack was monitored using a potential drop technique in plain low cycle fatigue specimens of MARM509 tested at 900°C in air and in vacuum. The crack growth rate was found to be correlated with cyclic J integral as adapted from finite element computations of Shih and Hutchinson. Oxidation was found to increase crack growth rates particularly at short crack lengths. Fatigue crack growth rates in air were in good agreement with values predicted from a summation of the rate in vacuum and of an oxidation contribution which was computed from oxidation kinetics data.

INTRODUCTION

High temperature low cycle fatigue damage in plain specimens can often be considered as the growth of a dominant microcrack up to a depth about 1 to 2 mm. This has been shown recently to hold in MARM509, a cast cobalt based superalloy, due to early crack initiation (1,2). A sound fatigue life prediction can be expected only if one is able to describe the growth of such a crack under various test conditions. A promising way is to use an elastic-plastic fracture mechanics approach. In particular the empirical extension of the J integral concept to cyclic loading which involved gross plasticity, has been shown to be successfull in correlating long crack and short crack results (3-5).

The present work was conducted to examine whether the dominant crack in a plain low cycle fatigue specimen could be described by a J integral approach using a single crack approximation. Fatigue crack growth rates were determined from a potential drop technique for tests in air and in vacuum to examine the influence of environment. The experimental method and the analysis used will be described first. The crack growth results will be then discussed and the influence of environment will be emphasized.

EXPERIMENTAL PROCEDURE

The composition of the heat of MARM509 used is (in wt pct) : 0.59C, 11Ni, 23.2Cr, 7W, 3.31Ta, 0.3Zr, 0.22Ti, 0.17Fe, bal.Co. Specimens were machined from cylindrical castings 20 mm in diameter after a heat treatment at 1230°C for 6 hrs. The average grain size is about 0.8 mm and the alloy is primarily a face centered cubic matrix with a few pct of interdendritic MC and $M_{23}C_6$ carbides (1,2).

* both authors are with the Centre des Matériaux, Ecole des Mines de Paris, ERA CNRS 767, B.P. 87, 91003 EVRY CEDEX.

Low cycle fatigue tests were carried out on cylindrical specimens 8 mm in diameter and 12 mm in gauge length using total axial strain control. The tests were conducted on a modified screw-driven tensile testing machine using push-pull strain cycling with a total strain rate $\dot{\varepsilon}_t$ about 7 $10^{-4}s^{-1}$ (frequency of a few 10^{-2} Hz). The tests reported here were made at 900°C using a radiation furnace in air and under a vacuum of 2 10^{-3} Pa.

An a.c. potential drop technique was used to monitor the growth of the dominant crack. Stabilized current input is applied to both sides of the specimens. The potential drop was measured across specimen gauge length using spot welded leads and after amplification was continuously recorded throughout the test.

PRELIMINARY ANALYSIS

Crack Length Determination

The growth of the dominant crack in a plain fatigue specimen was continuously monitored using a potential drop technique. The length of this crack was calculated from a two dimension analysis which assumes a single crack situation (6). Two ways were used to examine the validity of this approximation. First, specimens tested under various conditions, were broken to measure the final crack depth. Secondly tests were periodically interrupted and plastic replicas of the specimen surface were taken. The length of the dominant crack on specimen surface was so directly determined. This length was converted into a crack depth using a relationship which had been obtained from broken or sectionned specimens. Both kinds of results are compared with calculated crack depths in Fig. 1. The potential drop analysis is shown to properly monitor the depth of the dominant crack in a single specimen throughout testing. However, the analysis used yields crack depths about 13 pct lower than experimental ones. Similar conclusions were recently obtained by Gangloff (7).

J-Integral application to fatigue crack propagation

The range of stress intensity factor ΔK has proved to be a powerful correlating parameter in fatigue crack propagation under small scale yielding conditions and it enables to handle crack growth data from different specimen geometries. For cracked members where gross plasticity may occur under monotonic loading, it has been necessary to introduce energy parameters such as the line integral J (8) instead of K. As Paris did previously with K (9), recent experimental works tried to extend J integral to cyclic loading. Though no theoretical justifications for such an extension are yet available, this approach has proved to be successful for deep cracks (3) and more recently for short cracks (4,5).

However under numerous circumstances short cracks were observed to grow faster than long cracks even when no gross plasticity occurs (4,5,10). Therefore short cracks behave as if they were longer than their physical size by a length l_o (4,5). A material constant length l_o was empirically found to rationalize differences between short and long crack data (4,5). Results in various materials suggest that this empirical constant is related to a microstructural unit such as the grain size (10).

In the present case, an approximate analysis for cyclic J was made assuming a single semi-circular edge crack in an semi-infinite medium (most results refer to crack depths in the range 0.2 to 1.5 mm). The plane stress computation of Shih and Hutchinson for a center crack panel in the limiting case of infinitesimal crack length was taken as a basis (11). The solution for the elastic-plastic condition was taken as the sum of solutions for the elastic and fully plastic cases respectively. The correction for geometry effects in the fully plastic case was crudely approximated to that in the elastic case, since no complete solution was available.

The expression for cyclic J integral was deduced from the monotonic loading case where stress and strain were replaced by stress and strain ranges. This is actually the hypothesis used by Rice in order to describe cyclic plasticity at the tip of a fatigue crack in the small scale yielding case (12). Following Rice we also assume that the relationship between stress and strain could be described by the cyclic work hardening law as obtained from stabilized hysteresis loops. However in the case of fully reversed strain cycling it is necessary to account for the crack closure phenomenon. The fatigue crack in push pull conditions was assumed to open as soon as the stress becomes tensile. As pointed by Shih and Hutchinson (11) the elastic contribution should correspond to an effective crack length which incorporates Irwin's plastic zone size correction. For plastic zone sizes smaller than the grain size, which is the case with numerical value used here, plasticity should spread over the entire grain size. Therefore the plastic zone size should be taken as l_o, the grain size. This is similar to the empirical constant used by previous authors; however here this constant applies only to the elastic contribution. From these hypotheses the following expression can be derived for the cyclic J integral, in push-pull conditions :

$$J_{cycl.} = 0.51(\sigma^2_{max}/E).\pi.(a+l_o) + 0.51\ \sigma_{max}.\Delta\varepsilon_p.a.f(n) \quad \text{.........} \quad (1)$$

with $f(n) = 3.85\sqrt{n}\,(1-\frac{1}{n}) + \pi.\frac{1}{n}$

where a is the crack length, σ_{max} is the maximum tensile stress, $\Delta\varepsilon_p$ is the plastic strain amplitude and n is the cyclic work hardening exponent in the relationship $\Delta\varepsilon_p = k\ \sigma^n_{max}$ (with k a constant).

RESULTS AND DISCUSSION

The growth of the dominant fatigue crack was studied for different plastic strain ranges between 2 10^{-4} and 8 10^{-3} at 900°C in air. The crack growth rates are reported as a function of cyclic J in Fig. 2 with no plastic zone size correction (putting $l_o = 0$ in Eq. 1). There is a fairly good correlation of data relative to different strain ranges by J_{cycl}. In spite of the rather large uncertainty of data at shorter crack lengths (i.e. corresponding to lower values of J_{cycl}. for a given strain range), short cracks exhibit faster kinetics at high strain amplitudes. This overall behaviour is consistent with previous experiments on short cracks (4,5,10).

The crack growth curves are plotted as a function of cyclic J including the plastic zone size correction in Fig. 3 (i.e. putting l_o equal to the grain size of 0.8 mm in Eq. 1). Some improvement is obtained over Fig. 2 and crack growth rates agree with a power relationship :

$$da/dN = C\ J^m_{cycl.} \quad \text{.........} \quad (2)$$

where C and m are two constants.

Most data are within a factor of 1.5 of the least square fit line of Fig. 3. However some deviations at short crack length remain, when results published at room temperature on other materials showed no significant deviation anymore using a l_o correction (4,5). So such faster kinetics of short crack lengths could be due to an environmental influence.

In order to investigate the validity of this hypothesis, some tests were carried out in vacuum at the same temperature. Crack growth rate results in vacuum are reported with those in air as a function of cyclic J including the l_o correction in Fig. 4. Crack growth rates in vacuum are significantly lower than in air. However the two strain ranges are now in good agreement with Eq.2 with an exponent m about 1.7, higher than in air (m about 1.4). Similar trends in the comparison of vacuum and air results have been obtained in steels by Skelton (13).

Therefore oxidation enhances crack growth under high strain amplitude cycling and this influence is stronger at shorter crack length. For instance, under a plastic strain range of 0.3 pct the crack growth rate in air is near 3 μm/cycle for crack lengths near 0.2 to 0.4 mm, when the corresponding rate in vacuum should be about 0.4 μm/cycle. On the contrary, the rate in air is only twice that in vacuum for a crack length around 1.2 mm. Though very useful to correlate different strain range data, the elastic-plastic fracture mechanics approach cannot completely account for high strain crack growth rate results when there is a strong environmental interaction.

The simplest way to rationalize an influence of oxidation on fatigue crack propagation is to consider crack advance as resulting from the summation of two contributions, a purely mechanical one and another due to oxide cracking at the crack tip. Thus one may write :

$$da/dN = (da/dN)_{mech.} + (da/dN)_{oxid.} \qquad (3)$$

The study of crack propagation path in air actually showed that cracking propagates preferentially along MC carbides which are more rapidly oxidized than the surrounding matrix (2). The crack advance due to oxidation at the crack tip was recently modelized from oxidation kinetics measurements at specimen surface exposed in air at 900°C under a cyclic stress or under no stress (14). Preferential oxidation along the carbides was found to be stress dependent and general matrix oxidation to be strain range dependent. Therefore the oxidation term in Eq. 3 is a function of the volume fraction of carbides, cycle period, maximum tensile stress and plastic strain range, and is supposed, for sake of simplicity, to be independent of crack length. Thus it is constant for given strain range and cycle period.

The mechanical contribution to crack advance was simply taken as that corresponding to the crack growth rate equation in vacuum. Eq. 3 is thus a real prediction of crack growth rate in air since it depends only on vacuum experiments and on physical measurements of oxidation kinetics and microstructural features.

Experimental crack growth rates in air and predicted values are in good agreement as shown in Fig. 5 within a factor of two. However the behaviour observed at the lowest strain amplitude suggests that the oxidation contribution could be slightly dependent upon crack depth.

CONCLUSIONS

The crack growth rate of the dominant crack in high strain low cycle fatigue of plain specimens of MARM 509 was determined from a potential drop technique. Results from different tests were found to correlate with cyclic J integral including a plastic zone size correction equal to the grain size.

Crack growth rates were lower in vacuum than in air. Deviations from the cyclic J integral rationale are observed at short crack lengths in air, at high strains. They were attributed to oxidation.

Crack growth rates in air were in good agreement with values predicted from vacuum data and from an oxidation model based on measurements of oxidation kinetics.

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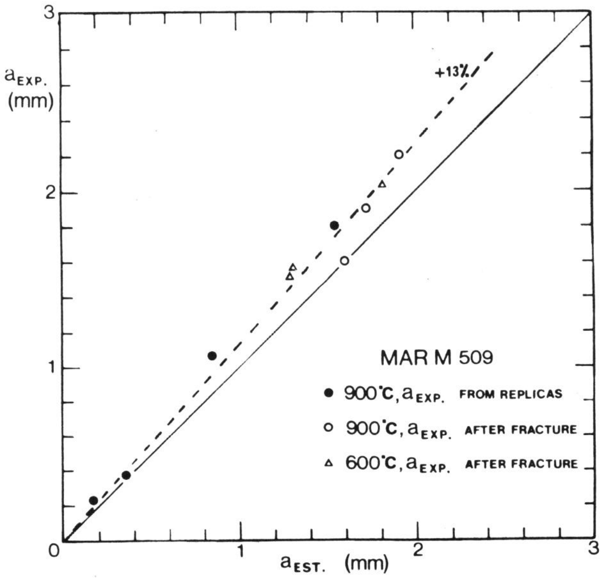
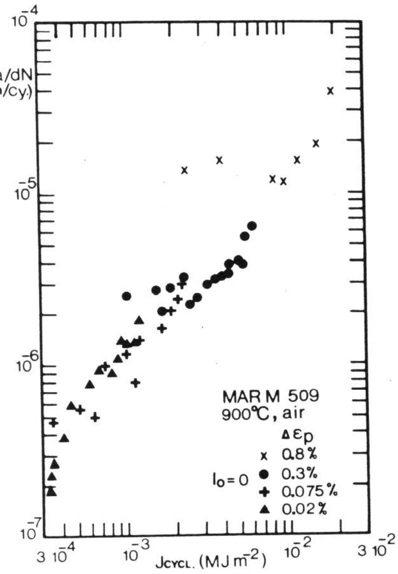


Figure 1 - Comparison of calculated (a_{est})and experimental (a_{exp})**cracks** depth on specimen, for several test conditions.



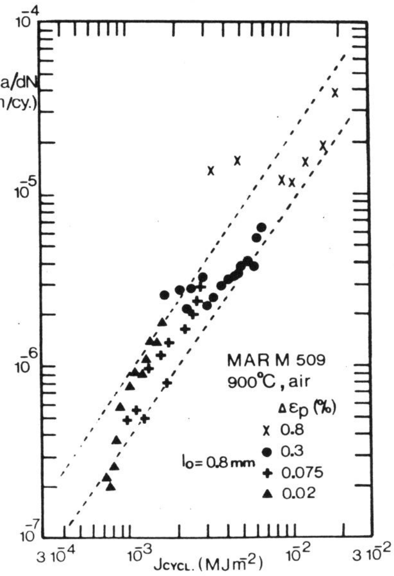


Figure 2 (left) and 3 (right) - Variation of crack growth rates as a function of J_{cycl} for various strain levels - l_o = 0(left), l_o = 0.8 mm(right)

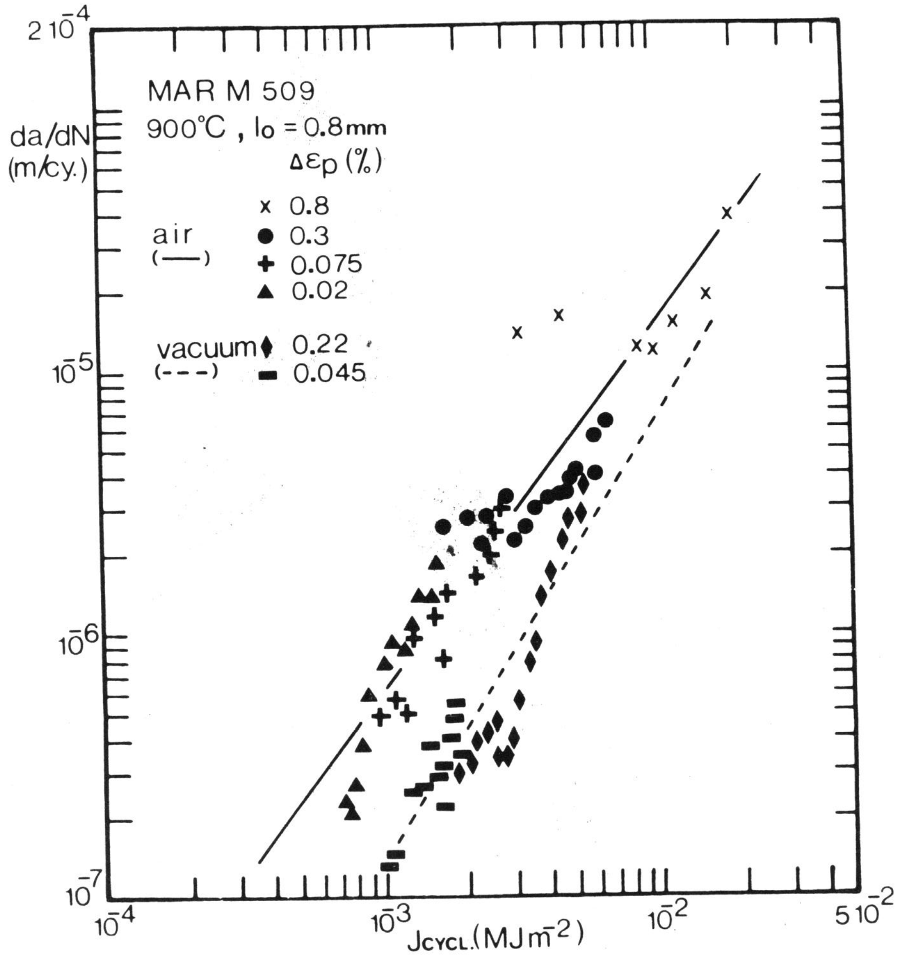


Figure 4 - Variation of crack growth rates as a function of J_{cycl} for various strain levels : comparison between results in air and in vacuum - (l_o = 0.8 mm).

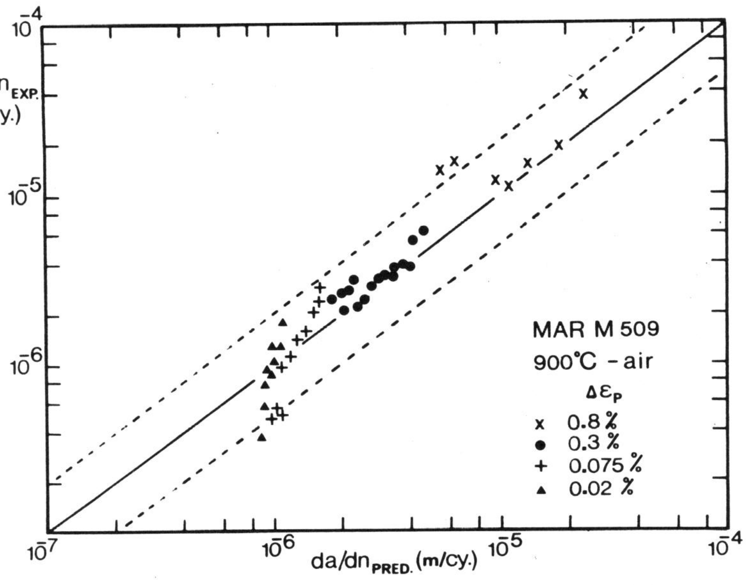


Figure 5 - Comparison of experimental (da/dN_{exp}) and predicted (da/dN_{pred}) crack growth rates, in air.