LOW CYCLE FATIGUE PROPAGATION OF MICROCRACKS IN TWO SUPERALLOYS

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ABSTRACT

The growth of the dominant microcrack was monitored in low cycle fatigue specimens using a potential drop technique. Two cast superalloys with a fairly large grain size were studied : a moderate strength cobalt-based alloy MAR-M509 at 600°C and a high strength nickel based alloy IN 100 at 1000°C under vacuum. Fatigue crack growth rates results showed a good correlation with cyclic J integral for the first alloy but are poorly correlated for the second one. A good agreement with data from long cracks in CT specimens was obtained introducing a plastic zone correction equal to the grain size.

KEYWORDS

Fatigue crack propagation, nickel based alloy, cobalt based alloy, high temperature fatigue, short cracks.

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 MAR-M509, a cast cobalt based superalloy, due to early crack initiation (Reuchet and Rémy 1979,1983). 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.

The range of stress intensity factor ΔK has proved to be a powerful correlating parameter in fatigue crack propagation under small scale yielding conditions by facilitating the treatment of crack growth data from different specimen geometries. For cracked members where gross plasticity may occur under monotonic loading, it has been necessary to introduce characterising parameters such as the line integral J (Rice, 1968) instead of K. As Paris did previously with K (Paris, 1964), 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 (Dowling and Begley, 1976) and more recently for short cracks (Dowling 1977, El Haddad and coworkers. 1979).

Short cracks have been reported to propagate faster than long cracks even in the regime of linear elastic fracture mechanics (LEFM), (Dowling 1977, El Haddad and coworkers 1979, Taylor and Knott 1981).

Therefore the present work was undertaken to clarify the behaviour of the dominant microcrack in fully yielded low cycle fatigue (LCF) specimens. As grain size was thought to be an important parameter, two cast superalloys with a fairly large grain size, used as vane and blade materials, were investigated : a cobalt based superalloy MAR-M509 of moderate strength and a high strength nickel based alloy IN 100. As oxidation can have a very large effect on crack growth rate at high temperature (Reger and Rémy 1982), care was taken to avoid any environmental influence. So MAR-M509 and IN 100 were studied in air at 600°C and in vacuum at 1000°C respectively. Fatigue crack growth rate (FCGR) results from short cracks on fully yielded LCF specimens are so presented and compared with long crack results from LEFM tests. The existence of a "short crack" specific behaviour is then discussed.

MATERIALS AND EXPERIMENTAL PROCEDURE

Two alloys were used in this study MAR-M509 and IN 100. The composition of the heats used is (in wt pct) for MAR-M509 : 0.59C, 11Ni, 23.2Cr, 7W, 3.31Ta, 0.3Zr, 0.22Ti, 0.17Fe, bal.Co; for IN 100 : 0.18C, 14.7Co, 10.3Cr, 3.15Mo, 1.01V, 4.60Ti, 5.68Al, 0.15Fe, bal.Ni. Specimens were taken from cylindrical castings of 20 mm in diameter. The as cast cobalt base alloy was heat-treated at 1230°C for 6 hours before machining. The average grain size is about 0.8 mm for the casting conditions used. The corresponding microstructure is that of a face centered cubic (fcc) matrix with a few pct of interdendritic carbides.

The as cast nickel based alloy was heat treated at 1150°C during 3hrs before machining, to simulate the coating heat treatment. The average grain size is about a few millemeters in the casting conditions used. The microstructure is that of a fcc matrix, which is hardened by a high volume fraction of fine γ' precipitates, with interdendritic MC carbides and large areas of blocky γ'.

Low cycle fatigue tests were carried out on cylindrical specimens 8 mm in diameter and 12 mm in gauge length using axial, total strain control. Specimens were heated by a radiation furnace. The tests were conducted on a modified screw-driven tensile testing machine for most tests and on an electrohydraulic machine for the higher frequencies. The load was continuously recorded as well as stress strain hysteresis loops occasionally. Saw tooth total strain cycling fully reversed was carried out at a constant strain rate $\dot{\varepsilon}_t$ about $7\ 10^{-4}\ s^{-1}$ and $2\ 10^{-4}\ s^{-1}$ for the cobalt and nickel base alloy respectively. This corresponds to a reference frequency of a few 10^{-2} Hz.

Saw tooth strain cycles were also applied to IN 100 under vacuum at 1000°C using a frequency thirty times higher, i.e. about 1 - 2 Hz. A vacuum chamber, mounted on the testing machine, enables a vacuum of $2\ 10^{-3}$ Pa to be maintained throughout the test.

Low cycle fatigue tests on MAR-M509 at 600°C in air and on IN 100 at 1000°C in vacuum are presented as well as LEFM tests on CT specimens (with dimension W=32 mm and thickness B=6 mm).The latter were carried out at the frequency of 20 Hz in air at constant amplitude loading with R=0.1, on an electrohydraulic machine.

The two alloys MAR-M509 (at 600°C in air) and IN 100 (at 1000°C in vacuum) have different strengths and so different cyclic hardening characteristics. This can be shown using the classical power relationship for the cyclic stress-strain curve :

$$\Delta\sigma = \sigma'(\Delta\varepsilon_p)^{n'}$$

where $\Delta\sigma$ and $\Delta\varepsilon_p$ are the peak stress and plastic strain amplitude respectively, as determined from hysteresis loops. The cyclic work hardening exponent n' is 0.121 and 0.254 for MAR-M509 and IN 100 respectively while the corresponding values of the cyclic hardening coefficient σ' are 1650 and 2810 MPa.

A potential drop technique was used to monitor the growth of the crack. This technique was applied on CT specimens and on the cyclindrical LCF specimens for the dominant crack. As shown previously (Reger and Rémy , 1982 a) the potential drop technique can monitor properly the depth of the dominant crack in a single LCF specimen throughout testing. So the development of microcracks could be studied in the depth range 0.2 to 1.2 mm.

DEFINITION OF J INTEGRAL FOR CYCLIC LOADING

In the case of short LCF cracks, an approximate analysis for cyclic J was made assuming a single semi-circular edge crack in a semi-infinite medium (most results refer to crack depths in the range 0.2 to 1.2 mm). The plane stress computation of Shih and Hutchinson (1976) for a center crack panel in the limiting case of infinitesimal crack length was used as a basis. 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 (1967) in order to describe cyclic plasticity at the tip of a fatigue crack in the small scale yielding case. Following Rice we also assumed that the relationship between stress and strain could be described by the cyclic work hardening law as deduced 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.

From these hypotheses the cyclic J integral can be written as the sum of its elastic and plastic components :

$$J_{cycl.} = J_{cycl,e} + J_{cycl,p} \qquad (1)$$

$$\text{with } J_{cycl,e} = 0.51\ (\sigma^2_{max}/E)\ \pi a \qquad (2a)$$

$$J_{cycl,p} = 0.51\ \sigma_{max}\Delta\varepsilon_p\ f(n)\ a \qquad (2b)$$

$$\text{and with } f(n) = 3.85\ n^{1/2}\ (1 - 1/n) + \pi/n$$

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).

In the case of the long crack loaded in the LEFM range (CT specimens) cyclic J was taken as $(\Delta K)^2/E$ as usually (Dowling, 1977).

TEST RESULTS

The FCGR from smooth LCF specimens of MAR-M509 tested in air at 600°C are reported as a function of cyclic J integral in Fig. 1. Cracks loaded under a given strain range exhibit a nearly smooth variation of FCGR with cyclic J (or crack length). In addition all the strain ranges correspond to a single FCGR-cyclic J curve within the experimental scatter. Further these short crack results and the crack growth curve determined from LEFM tests on CT specimens, as shown in Fig. 1, are in pretty good agreement. FCGR of short crack is only higher than that of long cracks for the lowest strain range ($\Delta\varepsilon_p = 4.5\ 10^{-4}$) : the FCGR is about three times that observed on the CT specimen. So for this material short cracks in a elastic-plastic regime behave nearly like long cracks in the linear elastic regime.

The FCGR from LCF specimens of the nickel based superalloy IN 100 tested in vacuum at 1000°C and at two frequency ranges, are reported as a function of cyclic J integral in Fig. 2. For comparison results obtained at a frequency of 20 Hz in air on CT specimens are included in this figure. As a matter of fact, for frequencies higher than 1Hz, the FCGR is actually independent upon frequency so that crack growth in air is similar to that observed in vacuum (Reger and Rémy, 1982 b). So results from CT specimens at this frequency can be taken as representative of vacuum results.

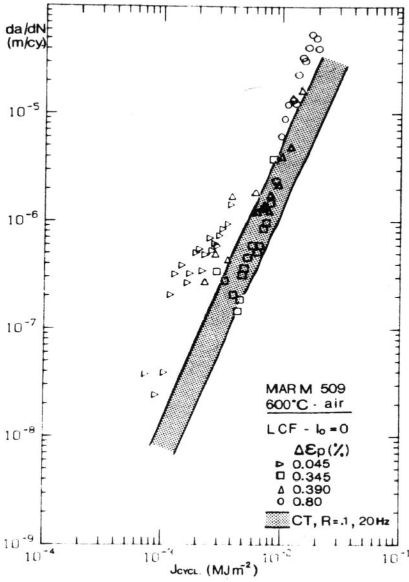

Fig. 1 : Variation of FCGR (da/dn) with the cyclic J integral ($J_{cycl.}$) for MAR-M509 at 600°C in air.

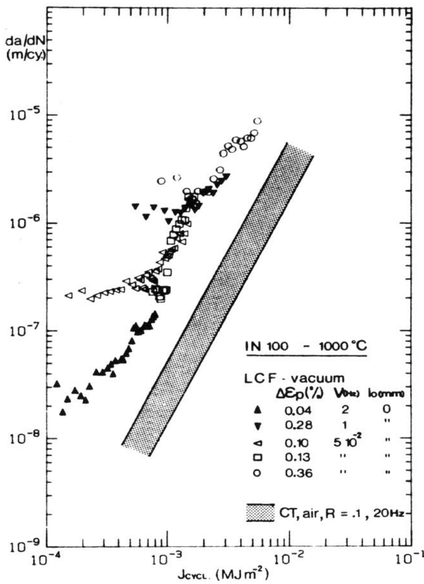

Fig. 2 : Variation of FCGR (da/dn) with the cyclic J integral ($J_{cycl.}$) for IN 100 at 1000°C in vacuum.

Strong differences are observed with respect to the first material. Firstly for a given strain range, FCGR exhibits a near plateau regime where the growth rate is quite independent upon crack length. At higher crack length the FCGR- crack length relationship has an exponent about two. As a consequence the FCGR results for the different strain levels cannot be correlated properly by the cyclic J integral.

In addition, the rate of these short cracks is an order of magnitude higher than that observed for long cracks in CT specimens (loaded in the linear elastic range) and even two order of magnitude for short cracks in the near plateau regime.

DISCUSSION

Therefore the present work on superalloys of fairly large grain size has shown two different behaviours : in the cast cobalt based superalloy there is a good agreement between short cracks in the elastic-plastic regime and long cracks in the linear elastic regime while in the nickel based superalloy the first kind of cracks propagates more rapidly than the second ones.

Under numerous circumstances short cracks were observed to grow faster than long cracks even when no gross plasticity occurs (Dowling 1977, El Haddad and coworkers, 1979). Therefore short cracks behave as if they were longer than their physical size. A material constant length l_o was empirically found to rationalize differences between short and long crack data. Taylor and Knott (1981) analysed results in various materials and they suggested that this empirical constant was related to a microstructural unit such as the grain size.

However in their analysis of the center cracked panel under monotonic loading Shih and Hutchinson (1976) emphasized the fact that the elastic contribution of the J integral should include a crack length correction. An effective crack length should be used by adding Irwin's plastic zone size correction to the physical crack length. However, for plastic zone sizes smaller than the grain size, plasticity should spread over the entire grain size. Numerical calculations show that this is precisely the case here. Therefore the plastic zone size should be taken as l_o, the grain size. Equation 1 has to be used with $J_{cycl,\ e}$ given by the following expression, instead of Eq. 2a :

$$J_{cycl,\ e} = 0.51\ (\sigma^2_{max}/E)\ \pi\ (a + l_o) \qquad (3)$$

but with the plastic contribution $J_{cycl,p}$ being inchanged and given by Eq. 2b.

This is similar to the empirical constant used by previous authors, however this constant applies only to the elastic contribution. It is not obtained by fitting to experiments but from an a priori hypothesis, it can be measured from metallographic observations of specimens.

Therefore grain size measurements were carried out on both materials. In the case of MAR-M509 a grain size of 0.8 mm was determined which is fairly uniform over a specimen cross section. For the nickel base alloy IN 100 the grain size is much larger in the range 2 to 3 mm. So as these are only a few grains in specimen section, the grain size was measured on every specimen. The FCGR results are reported as a function of cyclic J integral including this plastic zone size correction in Fig.3 for MAR-M509 (with the average

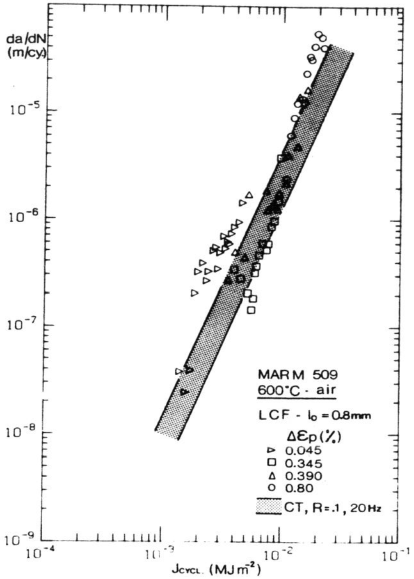

Fig. 3 : Variation of FCGR (da/dn) with the cyclic J integral ($J_{cycl.}$) corrected for plastic zone size. MAR-M509 at 600°C in air.

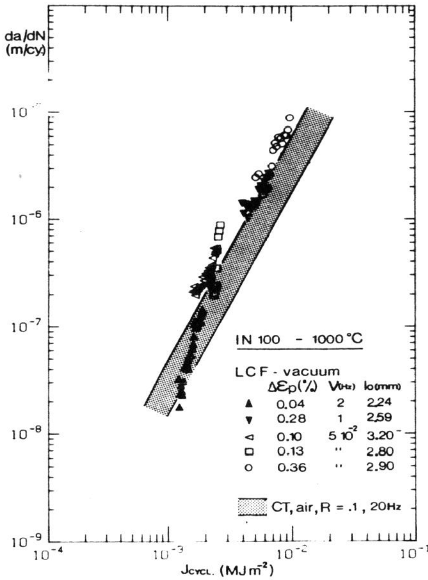

Fig. 4 : Variation of FCGR (da/dn) with the J integral ($J_{cycl.}$) corrected for plastic zone size. IN100 at 1000°C in vacuum.

grain size) and in Fig.4 for IN 100 (with the grain size of every specimen).

The agreement between results from LCF specimens and those from CT specimens which was fairly good without correction for MAR-M509 at 600°C is slightly improved. This is mainly visible for the lowest plastic strain range. This follows from the fact that the plastic contribution of J Integral is always larger than the elastic part except for the lowest strain level.

As can be expected, the effect of the plastic zone size is extremely important for the short crack results in the nickel based alloy. Firstly for a given plastic strain range, the near plateau regime has disappeared. Further within the experimental scatter results from different specimens can now be described by a single Paris-type equation :

$$da/dN = C \ (J_{cycl.})^m$$

where C and m are two constants (m is about 2.5 for IN 100 and 2.2 for MAR-M509). FCGR results from LCF specimens are in fairly good agreement with results from CT specimens. This drastic influence of the plastic zone correction, on short crack FCGR curves in this alloy, results from the large grain size and also from the fact that the elastic component of cyclic J given by Eq 2a is always higher than the plastic component given by Eq 2b.

Thus a plastic zone size correction taken as the grain size is very effective in bringing together FCGR results from short cracks in fully yielded specimens and from long cracks in LEFM CT specimens. From the present

results short cracks in LCF specimens propagate at a much higher rate than long cracks only when the elastic component of cyclic J integral is predominant over the plastic component. This effect is especially important for crack lengths smaller than the grain size.

CONCLUSIONS

The present investigation of the microcrack behaviour in fully yielded LCF specimens of two cast superalloys have shown that microcracks can propagate much faster than long cracks in CT specimens (as in the nickel alloy IN 100) or nearly at the same rate (as in the cobalt alloy MAR-M509). FCGR from microcracks in fully yielded specimens are in good agreement with those from CT specimens when one introduces a plastic zone size correction in the elastic component of cyclic J integral (the plastic zone size being taken as a grain size).

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