CREEP AND FATIGUE CRACK GROWTH IN SEVERAL CAST SUPERALLOYS

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Summary

Crack growth behavior in two cast nickel-base alloys, IN 713LC and IN 100, and a cobalt-base alloy, MAR-M509, under cyclic, static, and combined loads has been investigated at 927°C (1700°F) in air and in vacuum. The three alloys exhibited similar behavior in many respects. Fatigue crack growth rates generally were higher in air than in vacuum and also were raised by including a hold period at peak load. Two of the alloys at low stress intensities exhibited crack retardation and arrest. Most important, in vacuum the fatigue crack growth rates of the three alloys were essentially the same for a given ΔK.

Contrary to the environmental effect in fatigue, crack growth rate for static loads was lower in air than in vacuum. Also unlike in fatigue the creep crack growth rates in vacuum differed significantly among the three alloys. Thus creep crack growth was sensitive to microstructure whereas fatigue crack growth was not, excluding environmental effects. J-integral correlation with fatigue crack growth rates was comparable to the ΔK correlation.

Introduction

In the continuing effort to improve performance of high temperature components, the growth characteristics of small flaws that could exist in components have been receiving attention. Capability to predict this phase of the failure process, subcritical crack growth, would permit fuller utilization of component life. Also, understanding the behavior of flaws under service conditions should aid in the prevention of catastrophic failures.

Both time- and cycle-dependent processes may contribute to crack growth at high temperatures, the relative contributions would depend on the temperature, loading profile, and environment as well as the material. Of course even under cyclic loading, appreciable creep and environmental effects (time-dependent) can occur and could in fact control behavior. Components in the high temperature service environment frequently are subjected to all of these factors. For reliable predictions, knowledge of the behavior of the material under the various conditions separately and in combination is important.

Fatigue crack growth at high temperature has been the subject of many studies in recent years and a number of them (1-4) have been on superalloys. Also to a lesser extent investigations of creep crack growth behavior in superalloys (5-7) have been made to characterize behavior and to analyze correlation parameters for suitability for growth rate, and several have included combined static and cyclic loading in order to examine interactions of time- and cycle-dependent processes. Only a few such studies (8, 9) however have been conducted with cast alloys at the high temperatures encountered in service.

In this paper the crack growth behavior in two cast nickel-base alloys, IN 713LC and IN 100, and a cast cobalt-base alloy, MAR M509, under cyclic, static, and combined loads at 927°C (1700°F) in air and in vacuum is reported and analyzed in terms of growth rates. Correlations are made largely on basis of linear elastic stress intensity factor, K, and in some instances where load-line displacement data are available, on the basis of the non linear elastic-plastic parameter J-integral.

Experimental Details

The chemical compositions of the cast alloys IN 713LC, IN 100 and MAR-M509 are listed in Table 1 and the average grain diameters are approximately, 3-6, 1, and 1 mm respectively.

Table 1

Chemical Composition of Alloys, Weight Percent

Alloy	C	Ni	Cr	Mo	Cb&Ta	Co	W	Fe	V	Ti	Al	B	Zr
IN 713LC	0.06	76	11.4	4.0	1.99	0.1		0.05		0.60	5.75	0.015	0.13
IN 100	0.16	63	8.9	3.0		14.1		0.20	0.91	4.75	5.44	0.011	0.077
MAR-M509	0.62	10.5	23.4		3.47	54.	7.24	0.41		0.21		0.007	0.467

Compact tension specimens of nominal 3/4T dimensions except for 12.7 mm thickness and 0.635 mm deep side grooves were machined from the cast bars of nearly the same dimensions. These were fatigue precracked at room temperature prior to testing at 927°C (1700°F).

Static load tests were conducted in conventional constant load machines and cyclic load tests were conducted in electrohydraulic machines and for these, resistance heating and induction heating were employed, respectively. Load for the continuous cycling tests was applied in a triangular waveform at a frequency of 0.17 Hz at a R of 0.05 and for several tests, to include a creep effect, a dwell of one or five minutes was imposed at the peak load. Vacuum tests at 927°C were at a pressure of 1×10^{-6} torr. Crack length was measured periodically during the test with a traveling microscope. From plots of crack length versus cycles or time, crack growth rates were determined using a slope measuring device. Stress intensity factors were determined by expressions given elsewhere (2).

In several fatigue tests displacement was measured across the notch mouth using a strain gage extensometer with quartz contact rods, and load-displacement loops were recorded periodically at selected crack lengths. These data were used to determine the non linear parameter J-integral by a procedure developed previously (10) for load controlled fatigue. It involves measuring the area under the rising part of the load-displacement loop from minimum to maximum load, and calculating ΔJ from the Merkle-Corten equation (11),

$$\Delta J = 2\ (\alpha_1\ A + \alpha_2\ P\ \delta m)/Bb \qquad (1)$$

where A is the area, B the specimen thickness, b the length of uncracked ligament, P the peak load, and δm the displacement along the load line at maximum load. The constants α_1 and α_2 are the functions of a/W and are given in graphical form in Reference 11. It was shown previously (10) that values of ΔJ obtained by this estimation procedure are close to those determined by a more accurate but lengthy data reduction procedure. Comparison of fatigue crack growth rate correlations with ΔK and ΔJ are made to see which parameter is more suitable for predictive equations.

Results and Discussion

Crack Growth in IN 713LC

Fatigue crack growth rates in air and vacuum are given in Fig. 1 where the air curve represents three tests, each at a different load. Scatter in data among these tests appears to be large but since the higher load gives higher growth rates, there could be a load dependence. In previous work with IN 718 at 650°C (12) fatigue crack growth rate, for a given stress intensity factor, was found to decrease with an increase in load, a trend inverse to that in IN 713LC. This difference in response between the two alloys is probably due to a difference in the controlling mechanism in the basic crack growth process. Crack growth in IN 718 is very sensitive to an air environment in which the rate is greatly increased (13) and thus, if diffusion-controlled, would be dependent on time. Lengthening the time in test as a result of lower load could enhance environmental degradation and crack growth, as was observed with IN 718. In IN 713LC, crack growth rate was higher in air than in vacuum by about a factor of two, Fig. 1, but this difference is relatively small compared to that in IN 718. Here the crack growth process would probably be predominantly deformation-controlled and, if load-dependent, could be affected differently than in IN 718. Of course if the fracture mechanics parameter ΔK is to be truly applicable then the growth rate should not depend on load but only on stress intensity. Note that the air and vacuum curves tend to converge at high ΔK approaching the fracture stress intensity.

Inclusion of a 1-min. hold at the peak load of the cycle did not appear to significantly affect crack growth, Fig. 2. However, a 5-min. hold did increase growth rate except at low ΔK levels. Because both an air environment and creep can produce an increase in fatigue crack growth rate the cause of the hold-time effect cannot be identified here. It is of interest to note that air does not enhance creep crack growth in this alloy as will be seen later.

Crack growth rates in specimens under static load are presented in Fig. 3 and indicate a significantly higher threshold stress intensity for crack growth in creep than in fatigue. In the figure, crack growth under cyclic load is plotted on a time basis for comparison and in the stress intensity region where the curves overlap, crack growth is faster in fatigue. The transition at the high creep crack growth rates to a lower slope could be due to large crack tip deformation blunting the tip.

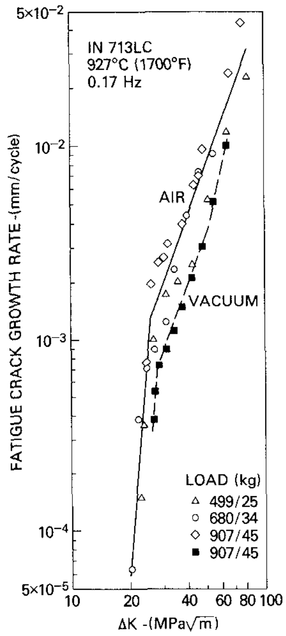

Figure 1 - Fatigue crack growth rates in IN 713LC at 927°C in air and vacuum.

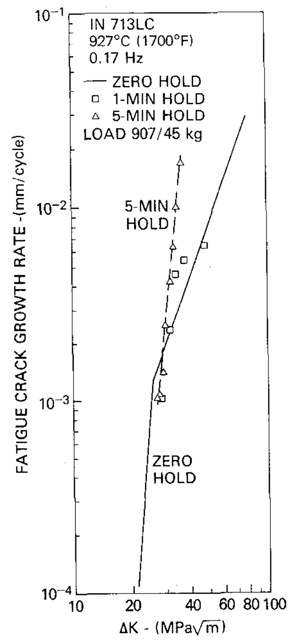

Figure 2 - Influence of hold time on fatigue crack growth in IN 713LC at 927°C in air.

Under static load the crack growth rates are lower in air than in vacuum, Fig. 3, contrary to the environmental effect under cyclic load where the rates in air are higher. In creep either crack branching promoted by air or load-bearing oxide formed at the crack tip could retard cracking. In contrast, in the fatigue cycle the tendency for crack branching would be less, or during unloading the crack tip oxide may fracture and disintegrate as soon as it is formed, permitting the crack growth enhancement mechanisms to then predominate.

Alloys in which crack growth is very sensitive to environment such as Inconel 718 and Inconel X-750 exhibit faster growth in air than in vacuum under static as well as cyclic load. For these alloys the operative mechanism might be surface absorption or short range grain boundary diffusion which enhances growth.

Crack Growth in IN 100

The fatigue crack growth behavior in IN 100 is represented in Fig. 4 which shows growth rate in air both continuously decreasing as well as increasing with increase in ΔK. At low ΔK, crack growth is retarded and virtually arrested. Several explanations can be advanced for the effect. In this range the stress intensity is sufficient for creep deformation to occur, causing relaxation of the stress field and blunting of the crack tip, or crack branching may take place, leading to crack arrest. At higher ΔK the stresses are sufficiently high to cause crack growth to occur at a rate fast enough to overshadow or minimize the relaxation effect and the crack branching. In vacuum the crack arrest phenomenon was not observed. Crack growth rates were higher in vacuum than in air at low ΔK but higher in air at high ΔK.

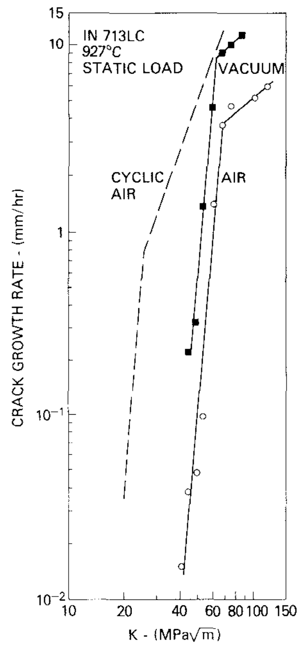

Figure 3 - Creep crack growth rates in IN 713LC at 927°C in air and vacuum.

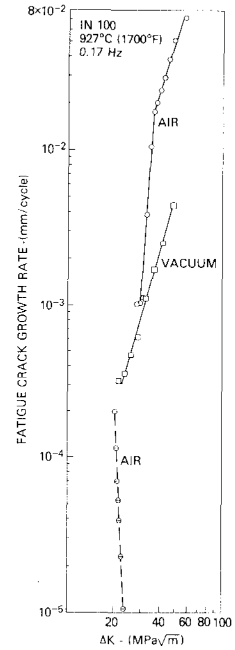

Figure 4 - Fatigue crack growth rates in IN 100 at 927°C in air and vacuum.

In contrast, creep crack growth in IN 100 was slower in air than in vacuum, as shown in Fig. 5. Higher stress intensities were required to initiate crack growth under static load than under cyclic load. Accordingly, growth rates were much lower under static load. These results agree with those on IN 713LC.

Crack Growth in MAR-M509

The fatigue crack growth behavior of this cobalt-base alloy resembles generally that of the two nickel-base alloys with respect to hold time and environmental effects, Fig. 6. Imposing a 1-min. hold at the maximuum load raised crack growth rate. Also in similarity with the two nickel alloys the rate in air for continuous cycling was generally higher than that in vacuum for a given ΔK. Since an air environment produces an increase in crack growth rate it is not possible here to identify the cause of the hold time effect on rate, whether it is creep or a longer exposure to the environment. Retardation of fatigue crack growth was also observed in MAR-M509. In one test crack growth initially decelerated up to a particular ΔK level and then accelerated to fracture. In another test begun at a lower starting ΔK, the crack grew at a decreasing rate until growth ceased, i.e., indiscernible over a period of several days.

Crack growth retardation and arrest were also observed in MAR-M509 under static load in air, Fig. 7. However if the starting K was increased slightly then crack growth accelerated until fracture occurred. As may be seen, these creep crack growth rates in air are lower than the fatigue rates plotted on a time basis. Contrary to the environmental effect in fatigue where crack growth rates were higher in air than in vacuum, in

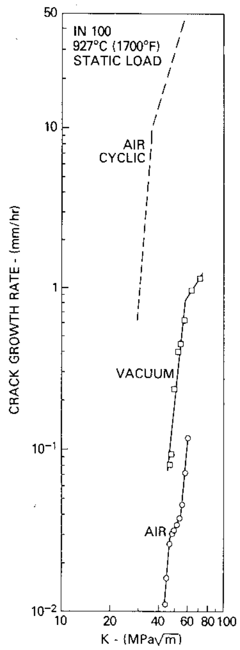

Figure 5 - Creep crack growth rates in IN 100 at 927^{O}C in air and vacuum.

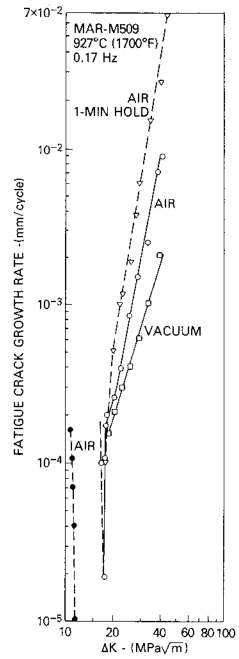

Figure 6 - Fatigue crack growth rates in MAR-M509 at 927^{O}C in air and vacuum.

creep the growth rates were lower in air. Somewhat different results were reported by Woodford and Bricknell (14) from conventional creep tests on MAR-M509 showing no effect on creep-rupture life of prior exposure in air at 1000^{O}C compared to exposure in vacuum, and they concluded that the alloy may be immune to oxygen damage. However, in cast nickel-base alloys they did find that prior exposure to air degraded creep rupture life. Again in contrast, our results show a beneficial effect of air, compared to vacuum, on creep crack growth. These data are not necessarily conflicting since the type of test differs, one includes the crack nucleation process whereas the other does not.

J-Integral Parameter Correlation for Fatigue

Because of the enhanced plasticity associated with high temperature crack growth and the expected inapplicability of linear elastic fracture mechanics to such behavior, the non-linear parameter ΔJ has been examined for suitability for correlations with fatigue crack growth rates in the cast alloys at 927^{O}C (1700^{O}F). The crack growth data for IN 100 and MAR-M509 alloys are plotted in Fig. 8 as a function of $\sqrt{\Delta J \cdot E}$, where E is the modulus of elasticity, to have the parameter in the same terms as ΔK for convenient comparison. In the linear elastic regime they should correspond. The IN 100 crack growth data represents continuous cycling and the MAR-M509 data represents a 1-min. hold cycle. Both sets of $\sqrt{\Delta J \cdot E}$ parameter based data points fall reasonably close to the corresponding ΔK based curves. No particular need for the ΔJ parameter is indicated by these data even at 927^{O}C and ΔK appears to be adequate. However, a complete evaluation requires a range of loads and crack sizes and specimens of several different geometries. It should be noted that in these alloys even though the temperature is high the crack ing was of a low ductile nature.

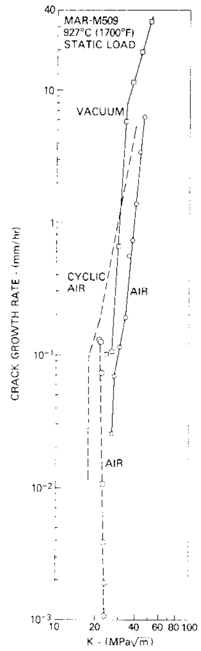

Figure 7 - Creep crack growth rates in MAR-M509 at 927^{o}C in air and vacuum.

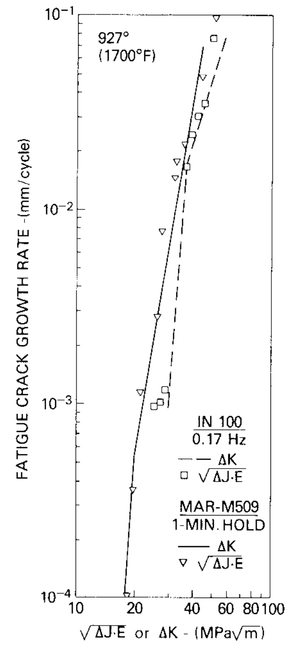

Figure 8 - Comparison of ΔJ and ΔK correlations with fatigue crack growth rates in IN 100 and MAR-M509.

Comparison of Fatigue Crack Growth Among Alloys

The rates of crack growth in air differed among the three alloys at low stress intensities where crack retardation occurred and at high stress intensities, Fig. 9. Essentially no difference in rates of the alloys was evident at intermediate stress intensities although the data for IN 100 did not cover the range well. In contrast, in vacuum the fatigue crack growth rates for IN 713LC, IN 100, and MAR-M509 were essentially equal for a given ΔK and can be represented by a single curve, Fig. 10. A similar lack of sensitivity of fatigue crack growth rate to microstructural and compositional differences among like alloys was observed previously with a group of stainless steels at room temperature (15). It is of interest to note that crack growth rates in alloys of different basic compositions at room temperature may be unified by normalizing with the modulus of elasticity (16). And in similar fashion changes in growth rate due to temperature may also be unified (13). There was no need for normalization here because the elastic moduli for the three alloys were very close.

Comparison of Creep Crack Growth Among Alloys

Creep crack growth rates, in contrast to fatigue rates, varied appreciably among the alloys, as shown in Fig. 11 for air. Alloy IN 100 exhibited the best resistance to creep crack growth and MAR-M509 the poorest resistance. Crack growth rates for the MAR-M509 were about a factor of 100 higher than the rates for the other two alloys and the indicated threshold K for growth is much lower for the cobalt-base alloy.

Differences in crack growth rates among the alloys were present also in vacuum and the relative order of resistance to crack growth of the

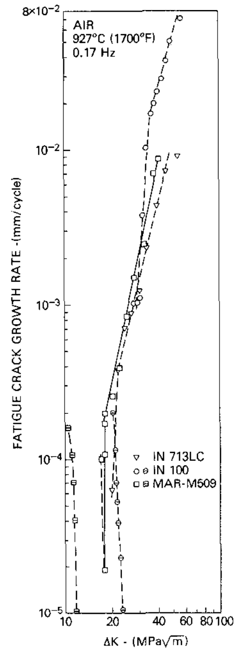

Figure 9 - Comparison of fatigue crack growth behavior in IN 713LC, IN 100, and MAR-M509 in air.

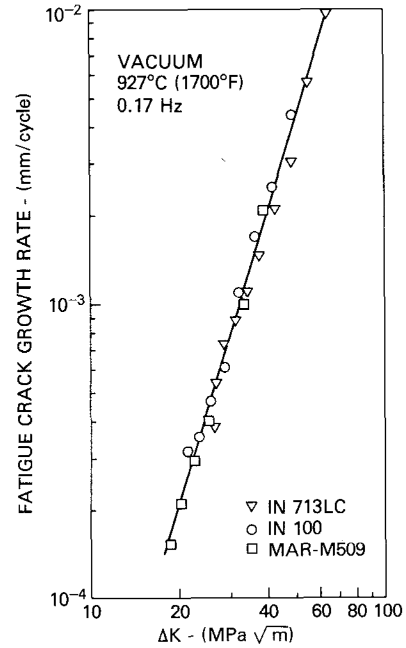

Figure 10 - Comparison of fatigue crack growth behavior in IN 713LC, IN 100 and MAR-M509 in vacuum.

alloys remained the same as in air, Fig. 12. The da/dN-K curve for each alloy in air was shifted to a higher stress intensity for a given growth rate, compared to vacuum, clearly showing that an air environment reduces creep crack growth rate whereas it increases growth rate in fatigue.

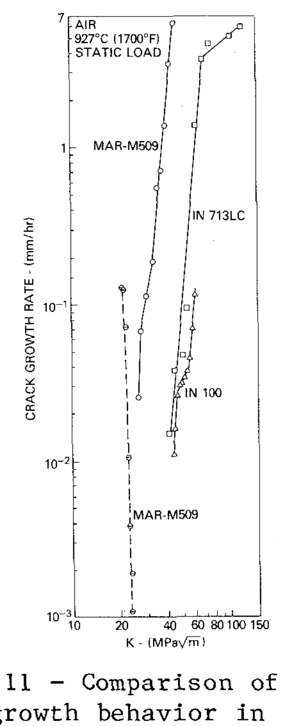

Figure 11 - Comparison of creep crack growth behavior in IN 713LC, IN 100, and MAR-M509 in air.

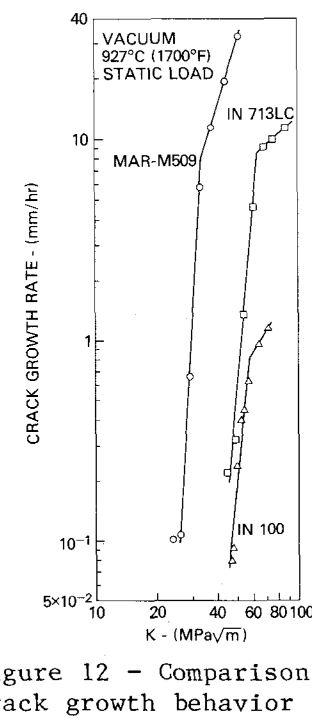

Figure 12 - Comparison of creep crack growth behavior in IN 713LC, IN 100, and MAR-M509 in vacuum.

It is important to note that crack growth in creep is much more sensitive to microstructure, including composition, than in fatigue independent of any environmental influence, as indicated by the large variation in crack growth rates among the alloys under static load and the absence of variation under cyclic load, Figs. 10 and 12. This difference in behavior must be related to the basic differences in the fatigue cracking and creep cracking processes. In fatigue, the constant reversal of deformation and strain hardening, restricted slip activity, and resharpening of the crack provides a more directed crack path with limited opportunities for divergence and consequently would be less affected by microstructure. On the other hand, in creep the crack has more freedom to find a path of lower resistance as provided by the microstructure, and microstructural differences among alloys would be reflected in the rate of crack growth. Also in creep the tendency for interdendritic and intergranular crack growth and crack branching is greater and differences in boundary phases and structural cell size among alloys would be expected to appear in their resistance to crack growth.

Conclusions

1. Fatigue crack growth rates in IN 713LC, IN 100, and MAR-M509 at 927°C are higher in air than in vacuum and are increased generally by hold time.

2. In contrast creep crack growth rates are lower in air than in vacuum and also are lower than fatigue growth rates on a time basis.

3. Retardation of crack growth may occur at low stress intensities in creep as well as in fatigue.

4. Parameters ΔJ and ΔK gave comparable correlations with fatigue crack growth rate.

5. Fatigue crack growth rates among the alloys, excluding environmental effects, were the same for a given stress intensity. On the other hand, creep crack growth rates varied significantly among the alloys.

6. The sensitivity of creep crack growth to microstructure and the apparent lack of sensitivity of fatigue crack growth are associated with the distinctly different crack growth processes in fatigue and creep.

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