FATIGUE CRACK GROWTH IN HIGH TEMPERATURE ALLOYS

W. Hoffelner

Metallurgical Laboratory
BBC Brown, Boveri & Company, Limited
CH-5401 Baden
Switzerland

Summary

Fatigue crack growth rates of a variety of high temperature alloys (nickel-base, iron-base) were investigated at various temperatures in air and in vacuum. Growth rates at intermediate cyclic stress intensity ranges and also in the threshold regime were studied. At room temperature the materials show a rather similar crack growth behaviour in the Paris regime. Differences in the threshold regime can be explained by microstructural and closure effects. Even in vacuum the temperature dependence of the Paris regime cannot be explained only by the temperature dependence of Young's modulus. Oxidation predominantly influences the crack growth rates at elevated temperatures. Under these conditions crack growth in the threshold regime is governed by mechanisms similar to those at room temperature but additionally by oxidation induced mechanisms (crack branching, oxide induced closure) occur.

Introduction

Fracture mechanics considerations for high-temperature alloys are of importance for different reasons. On the one hand it was shown that preexisting defects (e.g. ceramic inclusions, casting pores) govern the fatigue properties of powder metallurgy (P/M) superalloys e.g. (1-3) and cast alloys (4). On the other hand especially for engine discs there is the tendency to employ fracture mechanics calculations for a better use of the remaining life.

As far as subcritical crack growth is concerned fatigue crack growth, creep crack growth, environmental degradation and their interactions are of importance. Since a discussion of all the related problems is far beyond the limits of this paper we will confine ourselves to mainly cycle-dependent fatigue crack growth including environmental effects. Creep and creep-fatigue interactions will only be touched on occasionally. In order to obtain an overview over this class of materials a variety of microstructures was taken into consideration: cast alloys, cast and wrought, P/M-material.

It is the aim of the present paper to point out fatigue crack growth characteristics for these alloys. This will be done with new data as well as with data from the literature.

Materials and Testing Facilities

As already mentioned in the introduction a variety of materials was investigated. These alloys are shown together with their typical chemical composition and the production technology in Table I. The tests were performed with hydraulic closed loop machines. The temperature range was room temperature up to 850° C. As environment air and vacuum (about $2 \cdot 10^{-5}$ torr) were used. The main part of the experiments was performed with DCB-type samples in load control. The testing frequencies ranged from one Hz up to about 40 Hz. In order to study the applicability of the K-concept a few tests were done with small center cracked plates of a width of 11.5 mm and 4.5 mm in thickness. At room temperature also tests at 20 kHz were done in the threshold regime with an ultrasonic fatigue machine which is discussed more detailed in (5).

General Considerations

Fatigue crack growth data are usually plotted as fatigue crack growth rates, $\Delta a/\Delta N$, as a function of the cyclic stress intensity range, ΔK . Such curves typically consist of three areas: The threshold regime ($\Delta K \approx \Delta K_{IC}$) at low stress intensity ranges, a part where the curve can be described by a power law $\Delta a/\Delta N = C(\Delta K)^n$ (Paris regime) and at very high stress intensity ranges the region where the cyclic fracture toughness, ΔK_{IC} , is reached. In this region usually high plastic deformation occurs. Therefore the crack growth rates can become specimen dependent because the requirements for K_{IC} -testing normally are not met. Due to all these uncertainties the ΔK_{IC} -part of the curve will not be treated here. The validity of the K-concept for high temperature alloys was studied for Udimet 700 (6), Hastelloy X (7) and for IN 738 LC (8). From these investigations follows that for intermediate ΔK -values the ΔK -concept can be used up to 850° C satisfactorily.

Table 1: Typical chemical compositions and production technology of the alloys

	C	Al	Ti	Nb	Ta	Cr	Mo	W	B	Zr	Ni	Fe
IN 738 LC	0,17	3,4	3,4	0,9	1,7	16	1,7	2,6	-	0,1	bal.	-
IN 939	0,15	1,9	3,7	1,0	1,4	22,4	-	2,0	0,009	0,1	bal.	-
IN 100	0,18	5,5	4,7	-	-	10,0	3,0	-	0,014	0,06	60	-
A-286	0,04	0,33	2,04	-	-	14,86	1,24	-	0,005	-	25,29	bal.
IN 718	0,03	0,50	0,98	5,25	-	18,3	2,95	-	0,003	-	52,8	bal.
IN 901	0,05	0,25	2,78	-	-	12,6	5,76	-	0,002	-	42,5	bal.
RA 333	0,08	-	-	-	-	25,5	3,2	3,2	-	-	46	18
Hastelloy-X	0,1	-	-	-	-	22,0	9,0	0,6	-	-	bal.	18,5
IN 800-H	0,08	0,35	0,35	-	-	21,0	-	-	-	-	32,5	bal.
IN 617	0,07	1,0	-	-	-	22,0	9,0	-	-	-	54,0	-
MA 6000	0,05	4,5	2,5	-	2	15	2	4	0,01	0,15	bal.	-
Astroloy	0,023	4,0	3,5	-	-	15	5	-	0,024	0,01	bal.	-

Many models and parametrizations are discussed in the literature for prediction or representation of the $\Delta a/\Delta N$ -curves ranging from pure curve fitting techniques (9) to more physically based crack growth laws e.g. (10, 11, 12). Although almost all these models are successful for particular materials and particular testing conditions until now no general formula for representation of the whole $\Delta a/\Delta N$ -K curve exists. Such relationships frequently include Young's modulus and yield strength terms. The importance of the elastic modulus was first demonstrated by Pearson (13). Particular good agreement was found by Speidel (14) for data gathered in vacuum using the relationship

$$\Delta a / \Delta N = 1.7 \cdot 10^{-6} \left(\frac{\Delta K}{E} \right)^{3.5} \quad (1)$$

An extended investigation of seven high temperature alloys (15) showed that at elevated temperature in air the change in Young's modulus alone cannot account for the temperature dependence of the crack growth rates. In the threshold regime the situation is more complicated as microstructural and mean stress effects become important. Typical $\Delta a/\Delta N$ -curves for the materials tested during the course of our investigation are shown in Figures 1 and 2. In the following, characteristics of such curves will be discussed together with literature results. Since we are interested in cycle dependent fatigue, effects of frequency had to be eliminated as far as possible. It was shown several times (4, 16, 17) that above about 0,5 Hz no pronounced frequency effects occur. Therefore no investigations performed at less than 0.1 Hz were taken into consideration.

Fatigue Crack Growth at Intermediate Stress Intensity Ranges

It can be seen from Figure 1 that at room temperature the crack growth rates form a relatively narrow scatterband at medium stress intensity ranges. For comparison also the parametrization equation 1 is given as a straight line. This line fits the data surprisingly well, demonstrating that in this case Young's modulus is in fact a correlating parameter. The temperature dependence of the crack growth rates at intermediate ΔK (30 MN.m^{-3/2}) is shown in Figure 3. The scatterband represents values of various literature investigations (6, 15, 18-25, 27) as well as new results. Data measured in inert atmosphere were plotted for comparison. Qualitatively the behaviour in vacuum is comparable to that in air proving the inadequacy of Young's modulus correlation. Although the air values fall into a reasonable scatterband at this particular ΔK one should not forget that larger differences between the alloys can occur at lower ΔK according to Figure 2. Since it can be assumed that these differences are due to environment and microstructure we will illustrate these points further.

As a typical example crack growth rates of two different modifications (grain size ASTM 2-3 and ASTM 6) of the alloy IN 901 are shown in Figure 3. It can be seen that the curves come together at ΔK of about 30 MN.m^{-3/2}. Below this value enhanced crack growth in air occurs. At intermediate ΔK values only an insignificant influence of grain size could be found. The expected tendency exists that the fine grained version leads to slightly higher growth rates. The slope of the Paris regime for the vacuum experiments can reasonably well be approximated by 3.5 similar to room temperature results. However, the temperature dependence cannot be explained with Young's modulus only, as already mentioned. This means that equation (1) must be modified with a temperature dependent term to become a useful

Figure 1 - Fatigue crack growth rates of various high-temperature alloys at room temperature.

Figure 2 - Fatigue crack growth rates of various high-temperature alloys at 850°C

Figure 3 - The influence of temperature on fatigue crack growth rates of several high temperature alloys at $\Delta K = 30 \text{ MNm}^{-3/2}$ (literature (6, 15, 18-25, 27, 36)).

Figure 4 - The influence of microstructure and environment on fatigue crack growth rates of IN 901 at 550°C. (fine grain ASTM 6-7, coarse grain ASTM 3).

estimate for crack growth behaviour in vacuum. Such developments are underway (26) and it seems that a temperature dependent modification of equation (1) covers a wide range of temperatures. As far as oxidation is concerned it appears from Figure 4 that below a certain growth rate enhanced fatigue crack growth takes place. It can be assumed that below these rates the oxidation rate is higher than the crack growth rate leading to enhanced crack growth. For the alloy IN 901 it can be seen from Figure 5 that oxidation enhanced crack growth is characterized by intergranular fracture. A similar behaviour was reported in literature also for other high temperature alloys (15). It is however not possible to generalize these observations and to assume that the presence of oxidizing conditions necessarily changes the fracture mode from transcrystalline to intercrystalline. This will be revealed by the experiments on the cast alloys and the oxide dispersed superalloys (ODS). Figure 6 shows the effect of environment on the crack growth rates of IN 738 LC and IN 939 at 850° C. The growth rates in vacuum are only slightly lower than those in air and the slopes of the curves are comparable. The fracture path is transgranular in both environments. Crack branching in air tends to reduce the effect of environment on the crack growth rates as discussed in (4). Transgranular fracture at 850° C in air was also found for IN 100, MA 6000 and for wrought Udimet 700 (15). The question of fracture morphology as a result of microstructure and environment has not as yet been solved completely. Taking into account results on P/M disc-alloys (28), other literature data (15, 16) and our own results it seems that fine grains and low amount of γ' (lower strength) tend to promote intergranular fracture. Coarse grained alloys with high content of γ' -forming elements (Al, Ti) promote transgranular fracture at intermediate ΔK -ranges. Also the Cr-content could contribute to the resistance to crack growth under oxidizing conditions (34). Further experiments are necessary to provide a quantitative explanation of this behaviour.

The Threshold Regime

A Comparison of different threshold values for different high temperature alloys measured in air showed no clear temperature dependance at R-ratios of 0.1. A very high spread of the threshold values of different high temperature alloys was found. In contrast to the Paris regime a pronounced effect of the R-ratio on the threshold values is typical for these materials as illustrated in Figure 7 for the cast alloys IN 738 LC and IN 939. The reason for this behaviour is due to many factors:

metallurgical effects:	crack branching
	grain size
	γ' -size and distribution
	corrosion attack
	crystallographic crack growth
closure effects:	roughness induced
	oxidation induced

Crack branching was shown to be the essential mechanism to explain differences in the threshold behaviour of the cast alloys IN 738 LC and IN 939 between air and vacuum (4). Since cracks do not branch in vacuum, lower thresholds were measured there (Figure 7). The effect of grain size is illustrated in Figure 4 taking IN 901 as an example. These results show the expected behaviour that the finer grained materials have the lower threshold value. Grain size is important together with crystallographic crack

a) $\Delta K = 13,2 \text{ MNm}^{-3/2}$

b) $\Delta K = 26,5 \text{ MNm}^{-3/2}$

Figure 5 - Morphology of fatigue cracks in IN 901 at 550°C in air.

Figure 6 - The influence of enviroment on fatigue crack growth rates of IN 738 LC and IN 939 at 850°C.

propagation as found for many high temperature alloys at low and at intermediate temperatures. Observations exist (29) that under these circumstances the cracks grow fast through one grain stopping at the grain boundary and propagating further after a certain reinitiation time. This is also consistent with other investigations on Nimonic 901 (35). Since under usual testing conditions the propagation rates are integrated over many grains a relatively high threshold is measured. In a situation however, where the dimension of one grain becomes important (short cracks, fatigue life of smooth samples) much lower thresholds are measured. Although this is not the only reason for the differences between short and long cracks it can contribute to the differences. Another reason for high threshold values may be closure effects. In case of cracks growing along crystallographic planes mainly roughness induced closure effects are expected due to the very rough fracture surface (30). The extremely steep drop to a threshold value occurring for many alloys under oxidizing conditions (see for example Figures 2 and 6) seems to be rather an effect of oxide-induced closure (31). To minimize the closure effects threshold measurements frequently are performed at R-ratios higher than 0.1. At high temperatures such a procedure might cause problems because at very high R-ratios creep crack growth under the acting mean load becomes important (32). In order to avoid closure and environmental effects as far as possible threshold values in vacuum measured at $R = 0.5$ were plotted in Figure 8 as a function of the temperature. At room temperature values measured in air were taken since investigation with 20 kHz (32) and also vacuum experiments (4) did not show a significant influence of environment at ambient temperature. Although only limited data are available it seems that the threshold values show a behaviour similar to the crack growth rates in vacuum shown in Figure 3. Only a small influence up to about 550° C can be measured. This effect increases with increasing temperature. The scatterband could be caused by grain size and other microstructural effects. Further experiments should be made to verify this behaviour of the threshold values.

Concluding Remarks

The mainly cycle dependent fatigue crack growth behaviour of nickel-based and iron-based superalloys was investigated at various temperatures in air and vacuum environment. Although a variety of materials (solid solution strengthened, γ' -hardening, cast, wrought, P/M) was investigated many similarities were found.

At room temperature the data can be successfully parametrized by $\Delta K/E$ in the Paris regime. At elevated temperatures, however, also in inert atmospheres the temperature dependence of E alone is not able to explain the temperature dependence of the crack growth rates. At these temperatures the crack growth rates in air are enhanced by oxidation. Microstructural features and also R-ratio influence the Paris regime very little in the whole temperature range. In contrast to this the threshold behaviour is extremely R-ratio and microstructure dependent, especially depending on the grain size. Larger grains led to higher thresholds than finer grains. The high scatter of the threshold values in air at $R = 0.1$ is probably due to microstructural and closure effects which can be magnified by the presence of oxidizing conditions. These effects (e.g. crack branching) are responsible for the fact that vacuum tests sometimes show lower threshold values than measured in air.

Figure 7 - The influence of load ratio ($R=K_{\min}/K_{\max}$), temperature, and environment on the fatigue crack growth threshold (ΔK_0) value of IN 738 LC and IN 939 (replotted from literature (36)).

Figure 8 - The influence of temperature on the threshold values of fatigue crack growth of high temperature alloys (literature (33)). The symbols represent vacuum data.

Acknowledgement

The author would like to thank Prof. M. O. Speidel for helpful discussions and H. Baldinger and W. Meixner for their skilled technical assistance. The work was performed within the frame of COST-50 and partly financed by the Swiss Federal Government.

Literature

- (1) J.M. Hyzak and I.M. Bernstein, "The Effect of Defects on the Fatigue Crack Initiation Process in Two P/M Superalloys" Part I, II Met.Trans. 13A (1982) pp. 33-52
- (2) D.A. Jablonski, "The Effect of Ceramic Inclusions on the Low Cycle Fatigue Life of Low Carbon Astroloy Subjected to Hot Isostatic Pressing", Mat.Sci.Eng. 48 (1981) pp. 189-198
- (3) W. Betz, W. Track, "Living with Defects in Powder Metallurgy Materials", Powder Met.Int. 13(4) (1981) pp. 195-198
- (4) W. Hoffelner, "High-Cycle Fatigue Life of the Cast Nickel-Base Superalloys IN 738 LC and IN 939", Met.Trans. 13A (1982) pp. 1245-1255
- (5) W. Hoffelner, P. Gudmundson, "A Fracture Mechanics Analysis of Ultrasonic Fatigue", Eng.Fract.Mech. 16(3) (1982) pp. 365-371
- (6) K. Sadananda, P. Shahinian, "A Fracture Mechanics Approach to High Temperature Fatigue Crack Growth in Udimet 700", Eng.Fract.Mech. 11 (1979) pp. 73-86
- (7) J. Huang, R.M. Pelloux, "Low Cycle Fatigue Crack Propagation in Hastelloy-X at 25° C and at 760° C", Met.Trans. 11A (1980) pp. 899-094
- (8) W. Hoffelner, M.Y. Nazmy, C. Wüthrich, "Life Time Prediction of Gas Turbine Components (Turbine Blades)", Report BBC Research Centre KLR 83-208 C, Baden, Switzerland 1983
- (9) C.G. Annis, R.M. Wallace, D.L. Sims, "An Interpolative Model for Elevated Temperature Fatigue Crack Propagation", Air Force Materials Laboratory Report, AFML-TR-76-176, (1976)
- (10) F.A. McLintock, "On the Plasticity of the Growth of Fatigue Cracks", pp. 65-102 in Fracture of Solids, D.C. Drucker, ed., Inter.Sci.Publ, 1963
- (11) S. Purushothaman and J.K. Tien, "Generalized Theory of Fatigue Crack Propagation Parts I, II" Mat.Sci.Eng. 34 (1978) pp. 241-246 and pp. 247-262
- (12) S.D. Antolovich, A. Saxena, G.R. Chanani, "A Model for Fatigue Crack Propagation", Eng.Fract.Mech. 7 (1975) pp. 649-652
- (13) S. Pearson, "Fatigue Crack Propagation in Metals" Nature 211 (1966) p. 1077
- (14) M.O. Speidel, "Fatigue Crack Growth at High Temperatures" pp. 207-255 in High Temperature Materials in Gas Turbines, P.R. Sahm, M.O. Speidel eds., Elsevier Sci.Publ.Comp. 1974
- (15) P. Shahinian, "Fatigue Crack Growth Characteristics of High-Temperature Alloys", Metals Tech. (Nov. 1978) pp. 372-380
- (16) F. Gabrielli, R.M. Pelloux, "Effect of Environment on Fatigue and Creep Crack Growth in Inconel X-750 at Elevated Temperature", MIT-Report 12-40-81 1981
- (17) H.F. Merrick, S. Floreen, "The Effects of Microstructure on Elevated Temperature Crack Growth in Nickel-Base Alloys" Met.Trans. 9A (1978) pp. 231-236

- (18) T. Weerasooriya, J.P. Strizak, "Crack Propagation in Hastelloy X", Report ORNL/TM-7255, Oak Ridge National Laboratory 1980
- (19) R.M. Pelloux, "Creep-Fatigue Environment Interactions in Superalloys", Report AFOSR TR-81-0450 (1981)
- (20) B.A. Cowles, D.L. Sims, J.R. Warren, R.V. Miner Jr., "Cyclic Behaviour of Turbine Disk Alloys at 650° C" Trans ASME 102 (1980) pp. 356-363
- (21) R.B. Scarlin, "Fatigue Crack Growth in a Cast Ni-base Alloy" Mat.Sci.Eng. 21 (1975) pp. 139-147
- (22) M. Clavel, A. Pineau, "Frequency and Wave-Form Effects on the Fatigue Crack Growth Behaviour of Alloy 718 at 298 K and 823 K" Met.Trans. 9A (1978) pp. 471-486
- (23) C.W. Brown, M.A. Hicks, "Fatigue Growth of Surface Cracks in Nickel-based Superalloys" Int.J.Fat. Apr. (1982) pp. 73-81
- (24) L.A. James, "Fatigue Crack Growth in Inconel 718 Weldments at Elevated Temperatures" Weld.Res.Suppl. 57 (1978) pp. 17s-23s
- (25) B.A. Cowles, D.L. Sims, J.R. Warren, "Evaluation of the Cyclic Behaviour of Turbine Disk Alloys" NASA Report CR-159409 Oct. 1978
- (26) M.O. Speidel, W. Hoffelner, unpublished results
- (27) D.A. Jablonski, J.V. Carisella, R.M. Pelloux, "Fatigue Crack Propagation at Elevated Temperatures in Solid Solution Strengthened Superalloys" Met.Trans. 8A (1977) pp. 1893-1900
- (28) J. Gayda, R.V. Miner, "The Fracture Morphology of Nickel-Base superalloys Tested in Fatigue and CreepFatigue at 650° C" NASA Techn. Memorandum 81740, Apr. 1981
- (29) Private communication: R. Stickler, University of Vienna, 1983
- (30) J.M. McCarver, R.O. Ritchie, "Fatigue Crack Propagation for Long and Short Cracks in René 95 Nickel-base Superalloy" Mat.Sci.Eng. 55 (1982) pp. 63-67
- (31) S. Suresh, G.F. Zamiski, R.O. Ritchie, "Oxide-induced Crack Closure: An Explanation for Near-Threshold Corrosion Fatigue Crack Growth Behaviour" Met.Trans. 12A (1981) p. 1435-1443
- (32) W. Hoffelner, "Ultrasonic Fatigue of High Temperature Alloys" in Int.Conf. on Fat. and Corr.Fat. up to Ultrasonic Frequencies, J.M. Wells et. al eds., Champion PA, AIME 1982.
- (33) P.W. Liaw, T.R. Leax, W.A. Logsdon, "Near-Threshold Fatigue Crack Growth Behaviour in Metals" Acta metall. 31 (10) (1983) pp. 1581-1587
- (34) G.H. Gessinger, W. Hoffelner, "High Temperature Low-Cycle-Fatigue of P/M Superalloys" in Powder Metallurgy of Superalloys, G.H. Gessinger ed., in press
- (35) K.N. Melton, R.B. Scarlin, W. Hoffelner, "Crack Initiation and Propagation in a Forged Nickel-base Alloy under High Mechanical and Thermal Loading" pp. 659-671 in High Temperature Alloys for Gas Turbines 1982, R. Brunetaud et al. eds., D. Reidel Publ. Comp., Dortrecht 1982
- (36) S.R. Holdsworth, W. Hoffelner, "Fracture Mechanics and Crack Growth in Fatigue" pp. 345-368 in High Temperature Alloys for Gas Turbines 1982, R. Brunetaud et al.eds., D. Reidel Publ. Comp., Dortrecht 1982
- (37) M.O. Speidel, A. Pineau, "Fatigue of High-Temperature Alloys for Gas Turbines" pp. 469-512 in High Temperature Alloys for Gas Turbines, D. Coutsouradis et al.eds., Appl. Sci. Publ., London 1978