

CHARACTERIZATION OF ENVIRONMENT-DEPENDENT FATIGUE CRACK GROWTH IN ALLOY 718 AT 650°C

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Abstract

The elevated temperature fatigue crack growth behavior in Alloy 718 when subjected to a loading frequency lower than the transitional frequency of this alloy, is viewed to be fully environment-dependent. In this process, the crack growth increment per loading cycle is assumed to be equal to the intergranular depth of oxygen diffusion taking place at the crack tip during the cycle effective oxidation time. In order to identify the trend of this diffusion depth an experimental program was carried out on compact tension specimens made of Alloy 718 at 650°C in which fatigue crack growth measurements were made for two different cyclic load conditions, one including hold time periods at minimum load level while the second includes these periods at the maximum load level. This work resulted in establishing a relationship correlating the intergranular depth of oxygen diffusion and the instantaneous value of the stress intensity factor K . This relationship when integrated over the cycle effective oxidation time results in a closed form solution describing the environment-dependent fatigue crack growth rate. A comparison is made between the results of this solution when applied to different loading frequencies and the corresponding experimental results. This comparison indicates the validity of the proposed model, in particular at low ΔK values and lower loading frequencies where environment fully dominates the crack growth process.

1. Introduction

Time dependent effects on high temperature fatigue crack growth behavior in high strength structural alloys are generally ascribed to phenomena involving creep and/or environmental degradation processes. The relative importance of these two processes depends, in general, on the strengthening characteristics of the material, exposure time, load level and temperature. Nickel base superalloys and in particular Alloy 718, are a class of alloys that have been designed as highly creep resistant materials. Brooks and Bridges [1], for example, have shown that Alloy 718 is microstructurally stable after 10,000 hours exposure at temperatures up to 600°C. Nicholas et al [2] and Pineau [3] have estimated the Riedel-Rice characteristic time for transition from small scale yielding to extensive creep in Alloy 718 at 650°C to be in the range of 20 hours to 3 years, which is much larger than the possible cyclic periods and hold time durations in practical applications of this material. Furthermore, several studies on Alloy 718 have demonstrated that high temperature fatigue crack growth rate (FCGR) are decreased several order of magnitude when results of vacuum tests are compared with those due to low frequency loading in air test conditions. These observations, supported by the success of LEFM in predicting the crack growth rates in nickel base superalloys at elevated temperature conditions, give strong support to the conclusion that time-dependent effects in these alloys are principally due to environment degradation. Extensive work was carried out in order to understand the damage mechanisms associated with this type of degradation process. Important results of this work are related to the fact that in Alloy 718, oxygen is the detrimental element in air [4] and that environmental effects are highly localized at the crack tip and are manifested through grain boundary oxidation and subsequent embrittlement. Attempts to predict fatigue crack growth behavior in Alloy 718 have thus relayed on assessing both the cyclic and the oxidation damage components; the linear sum of which was assumed to represent the total crack tip damage. The relative contribution of these two components to the total damage has been linked, based on the pioneering work of Coffin [5], to the loading frequency. While different interpretation could be given for the reason of this linkage, it is, however, established that the damage process ranges from purely cyclic-dependent at high frequency levels to fully environment-dependent at very low frequency levels. In the environment-dependent crack growth process, crack increment per cycle is assumed to be directly related to the intergranular depth of oxygen diffusion taking place during the cycle effective oxidation time [6]. The objective of this paper is to examine the limitation of this assumption. The first part of the paper describes the various stages of crack growth processes in relation to the loading frequency. The second part will attempt to determine the cycle effective oxidation time of a loading cycle and provide a quantitative relationship between oxygen diffusion rate and crack tip driving force. This relationship when integrated over the oxidation time yields an explicit crack growth rate expression that will be the focus of analysis and comparison with experimental data.

2. High-temperature Fatigue Crack Growth Stages in Alloy 718

The relationship between loading frequency and environmental effects on the acceleration of the fatigue crack growth rate in Alloy 718 can be explained in terms of the intergranular oxygen diffusion process in the crack tip region. One of the governing factors of this process is the grain boundary diffusivity of oxygen — a subject that has not been extensively studied in the particular case of Alloy 718. It is recognized, however, that intergranular oxygen diffusion depends on the stress and strain states along affected grain boundaries [4,7,8]. Therefore, the diffusivity, D_g , of a stressed grain boundary could be expressed as a function of the inelastic strain energy density, $f(w_p)$, of this boundary. This is written as an Arrhenius relationship in the form [6]

$$D_g = D \exp\left(\frac{[-Q_g - f(W_p)]}{RT}\right) \quad (1)$$

where D is a diffusivity constant, R is the gas constant and T is the temperature in Kelvin. Through this definition, the influence of loading frequency and the associated deformation mode on the magnitude of D_g and $f(W_p)$, consequently on the crack growth response can be interpreted qualitatively as follows. High frequency loading, which is generally characterized by high slip density and a homogeneous form of deformation, would result in both strain accommodation as well as stress relief along affected grain boundaries in the crack tip region. Hence variations $f(W_p)$ and consequently D_g tend to be minimal. In addition, the increase in slip density, generally, leads to an increase in the lateral volume diffusion across affected grain boundaries. These two combined effects would result in limited or no acceleration of intergranular oxygen diffusion rate. In this situation, the influence of crack tip oxidation is minimal and crack tip damage becomes generally dominated by cycle-dependent effects, giving rise to transgranular fracture mode. On the other hand, low frequency loading accompanied by low slip density would promote grain boundary stress concentration resulting in an increase in the magnitude of both $f(W_p)$ and D_g . This is magnified, particularly, if stress relieve by grain boundary sliding is not permitted, as in the case of the highly creep resistant Alloy 718. Furthermore, the decrease in slip density would limit the grain boundary lateral volume diffusion process. Here, the expected increase in the grain boundary diffusivity and associated increase in depth of grain boundary oxidation results in an increase in the crack tip damage due to environment effects giving rise to intergranular fracture process.

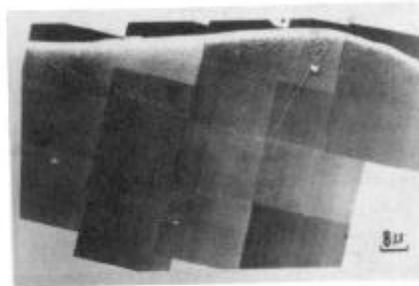


Fig.1 Slip line traces for high frequency (30 Hz) test at $\Delta K = 27 \text{ MPa}\sqrt{\text{m}}$ [9].

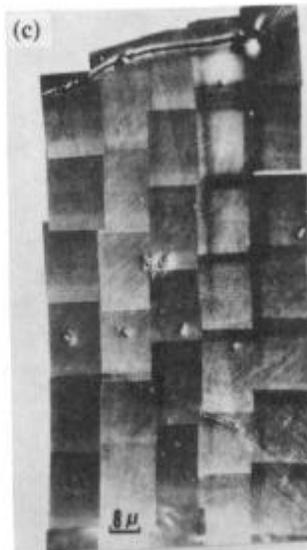
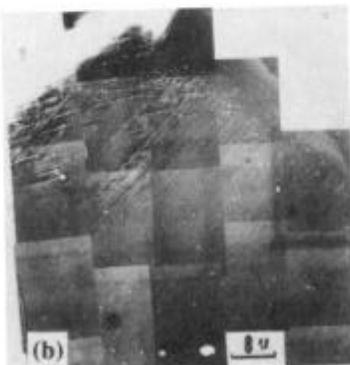
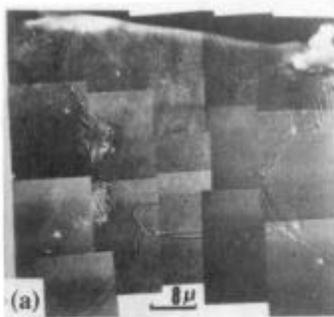


Fig.2 Slip line traces for low frequency tests at different ΔK levels (a) $21 \text{ MPa}\sqrt{\text{m}}$, (b) $27 \text{ MPa}\sqrt{\text{m}}$ and (c) $38 \text{ MPa}\sqrt{\text{m}}$ [9].

The relationship between slip density and fracture mode characteristics, as described above, has been investigated in the work of Ghonem et al [9] using compact tension specimens made of Alloy 718 with large grain size (50-120 μm) tested at 650°C in air. In this work fatigue fracture mode as well as slip line density were compared for loading frequency of 30 Hz vs that of 0.05 Hz which includes 300 seconds hold time at maximum load level. Slip line traces at and below the fracture surface of the test specimens were obtained using decoration technique described in [10]. These traces for different ΔK levels in both test conditions are shown in Figs.1 and 2. In Fig.1 which corresponds to a fully transgranular fracture mode, the degree of homogeneous deformation is evident by the high slip line density and the confinement of the reversed plastic zone to a narrow band near the fracture surface. This could be compared with Fig.2(b) which corresponds to a fully intergranular fracture and displays a lower slip line density as well as larger plastic zone size. One should also observe that in the three intergranular locations in Fig.2, the degree of slip homogeneity, measured as the inverse of the slip line interspacing, increases as ΔK increases.

On the basis of above argument the crack growth response of Alloy 718 with respect to loading frequency, f , has been divided, following the work of Pineau [11] and Nicholas and co-workers [12,13,14], into three distinctive types as shown in Figs.3 and 4. The first type is associated with high frequency loading in which the deformation mode is governed by a high degree of slip homogeneity. Cracking proceeds primarily in the matrix material (in contrast to the grain boundary), resulting in a predominantly transgranular fracture mode. The value of the frequency, f , required to produce this type of environment-independent, decreases as the magnitude of ΔK increases. This type of crack growth behavior is generally, predicted by the use of a Paris-type equation.

As loading rate decreases, the degree of slip homogeneity in the crack tip zone is lowered, resulting in a relative increase of the intergranular oxygen diffusion. Under this condition, the crack tip damage becomes a combination of oxidation and cycle-dependent components. In this type of response, the total crack tip damage has been described by many authors (15-20), using various models which are generally reduced to a form of the linear summation rule which can be expressed as:

$$\left(\frac{da}{dN}\right)_{\text{total}} = \left(\frac{da}{dN}\right)_{\text{cycle}} + \left(\frac{da}{dN}\right)_{\text{time-dependent}} \quad (2)$$

where, in the absence of sustained loading effects (hold times), the time-dependent term represents the contribution of oxidation to crack tip damage and could be written as a time integral in the form:

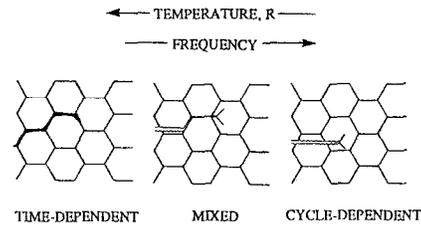


Fig.3 Schematic of fatigue crack growth mechanisms for time-dependent, mixed and cycle-dependent crack growth regimes in air environment [13].

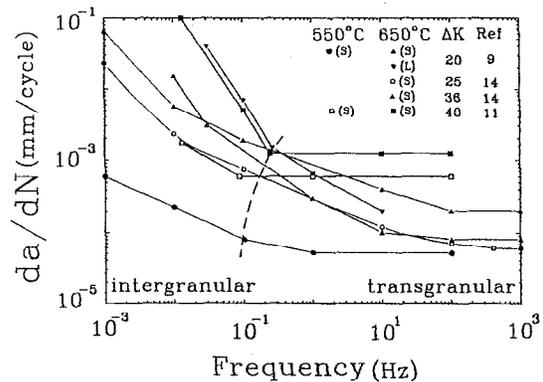


Fig.4 Effect of frequency on FCGR of Alloy 718 at $R = 0.1$ for different temperatures, grain sizes ($S \approx 20-50 \mu\text{m}$, $L \approx 150 \mu\text{m}$) and ΔK levels. Dashed line represents f_c .

of the linear summation rule which can be

$$\left(\frac{da}{dN}\right)_{\text{time-dependent}} = \int_{t_{ox}} \left(\frac{da}{dt}\right) dt \quad (3)$$

where t_{ox} represents the time period of the cycle during which the oxidation process is an active damage component. As shown in Figs.3 and 4, this oxidation enhanced process is characterized by a mixed transgranular/intergranular fracture mode. The degree of contribution of each of the cycle- and time-dependent terms in the above equation depends on both the frequency and ΔK values. For the same frequency, as ΔK increases the contribution of the cycle-dependent damage also increases, since increasing ΔK leads to an increase in the degree of slip homogeneity, see Fig.1. The increase in the cycle-dependent damage is measured by the increase in percentage of the transgranular features along the fracture surface. For the same ΔK value, however, the influence of the time-dependent damage increases as frequency decreases.

The third type of response mentioned above occurs for loading frequencies below a transitional level, f_c , where the crack tip damage becomes mainly an environment-dependent process in which crack growth is largely intergranular. The value of the transitional frequency, for a particular ΔK , in Alloy 718 was found to depend on both temperature and microstructure [9,11,14]. As shown in Fig.4, for a grain size of 50 - 100 μm

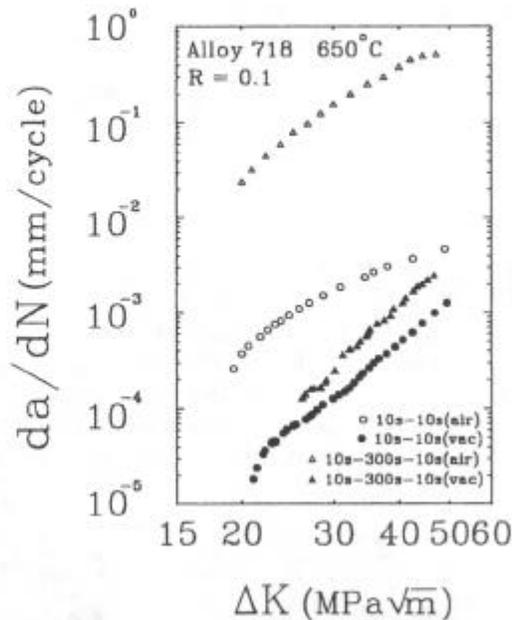


Fig.5 Comparison for FCGR in air and vacuum with and without hold time at maximum load level.

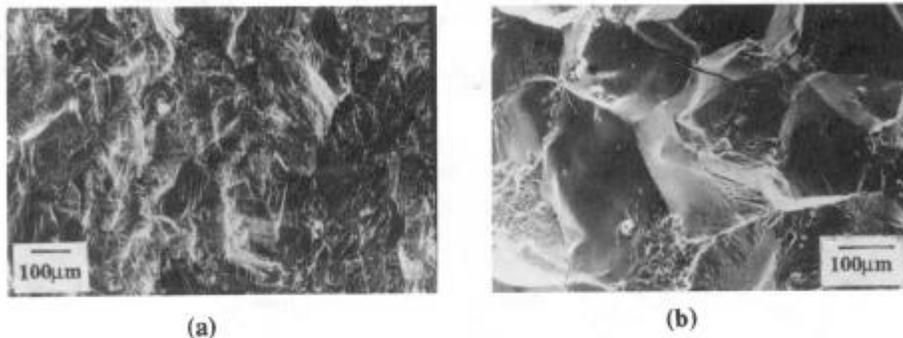


Fig.6 Scanning electron micrographs of fracture surfaces of Alloy 718 tested with 0.05 Hz at 650°C in (a) vacuum and (b) air environment.

and at 650°C and $\Delta K=40 \text{ MPa}\sqrt{\text{m}}$, the value of f_c is on the order of 0.1 Hz. The influence of loading cycle on crack growth behavior, where the test frequency is in the environment-dependent regime ($f < f_c$), is depicted in Fig.5 by comparing the FCGR of two CT-20 (ASTM E-647, $W=40\text{mm}$, $B=10\text{mm}$) specimens made of the same Alloy 718 and tested at 650°C and $f=0.05 \text{ Hz}$ in both air and vacuum (10^{-8} torr) conditions; labelled here cases A and B, respectively. The total number of cycles required to propagate the same crack length in both specimens was 5000 cycles and 65,000 cycles, in air and vacuum respectively. Furthermore, the fracture surfaces in both

cases are shown in Figs.6(a) and (b) where they exhibit fully intergranular fracture in the case of air testing and fully transgranular fracture in the case of vacuum testing.

If one makes the assumption that in the environment-dependent stage the crack growth increment per cycle is equal to the intergranular depth of oxygen diffusion, X , occurring during the cycle effective oxidation time, t_{ox} , then FCGR can be expressed as:

$$\begin{aligned} \left(\frac{da}{dN} \right)_{total} &= \int_{t_{ox}} \left(\frac{da}{dt} \right) dt \\ &= \int_{t_{ox}} \dot{X}(K, t) dt \end{aligned} \quad (4)$$

where \dot{X} is the instantaneous oxygen intergranular diffusion rate, generally it varies with time. The recognition that oxygen diffusion is an energy activated process, it could then be treated as a function of the instantaneous value of the stress intensity factor, K , acting on the crack tip during the cycle effective oxidation time. Testing the validity of this assumption (i.e. the ability of the above equation to describe the environment-dependent stage), requires the knowledge of both t_{ox} as well as the relationship between \dot{X} and K . These two requirements are not readily available in literature and therefore an attempt will be made here to determine these requirements experimentally.

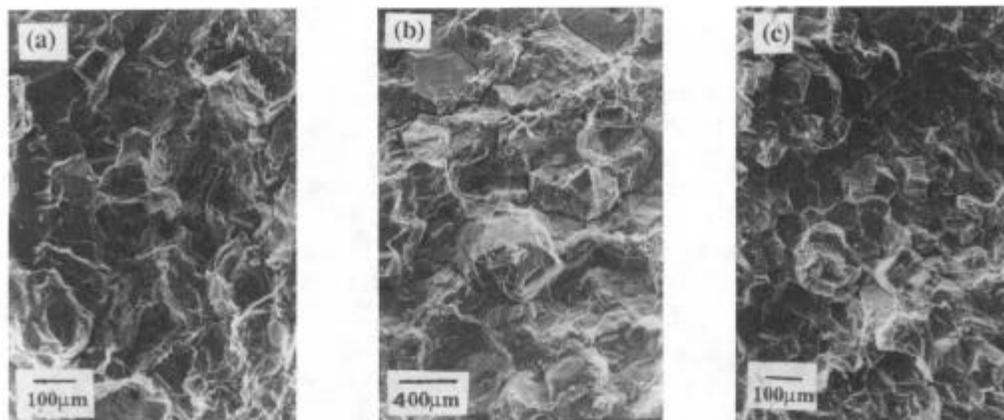


Fig.7 Scanning electron micrographs of fracture surfaces of Alloy 718 tested for 0.05 Hz at 650°C for (a) 25s-2.5s, (b) 25s-25s and (c) 90s-10s.

3. Cycle Effective Oxidation Time

The cycle effective time, t_{ox} , is defined here as being the period of the cycle during which the oxidation effects would take place. Several authors have assigned different measures to t_{ox} . For example, Achter et al [21] in their study of the effect of oxygen partial pressure on crack growth rate in Type 316 stainless steel at elevated temperature, proposed a calculation method which was based on the assumption that the time required for adsorbing a gas atom monolayer at the crack tip is equal to half of the tensile part of the cycle or a quarter of the whole cycle period. Wei and associates [22,23] in their attempt to predict environment assisted crack growth behavior in AISI 4340 steel, assumed that the value of the maximum load is the controlling factor for crack surface reaction rate. The cycle oxidation time in their work was assumed to be equal to half of the loading period plus half of the unloading period. Nicholas et al [2,16], in their work on Alloy 718, developed a model to predict purely time-dependent fatigue crack growth behavior by integrating the sustained load growth rate. They argued that the loading part of the cycle is

cycle is the part responsible for the environment-assisted effects. Similar conclusions were made by Floreen [24] in his work on grain boundary diffusion in nickel-base superalloys. Other investigators, specially Liu and McGowan [25], Rechet et al [19], Antolovich [26], Romanoski [27] and Saxena [18] employed, for different materials, different cycle period to represent t_{ox} . In the face of these various theoretical definitions of t_{ox} , an experimental attempt was made here to determine the effective oxidation time for Alloy 718 at temperature level of 650°C and for loading frequencies less than f_c .

The alloy used in this work was in the form of a rolled ring forging with conventional heat treatment. Compact tension specimens were machined from this alloy and subjected to fatigue crack growth tests conducted under constant load range, ΔP , and $R=0.1$ where R is the load ratio. These tests involved triangular waveforms with different cycle durations, all satisfying the condition that $f < f_c$. These durations include 27.5 seconds (25s-2.5s), 50 seconds (25s-25s), 100 seconds (25s-75s), and 100 seconds (90s-10s); labelled here as cases C, D, E and F, respectively. Fracture surfaces corresponding to all tests exhibited fully intergranular fracture features (confirming that $f < f_c$), as seen in Fig.7. Results in the form of fatigue crack growth rate, da/dN vs ΔK are plotted in Fig.8. They show the fatigue crack growth rate for case A to be identical to that of case H indicating that for the same frequency, varying the ratio of loading and unloading portions of the cycle will not influence the fatigue crack growth behavior. This result is contrary to results obtained, for example, by Coffin [28] in his work on fatigue crack initiation in OFHC copper at 400°C and on 304 stainless steel at 650°C. His results showed that in an asymmetric loading cycle, slow-fast loading is more damaging than fast-slow loading. Similar results have been obtained by other authors for crack growth behavior in various ductile materials [29]. However, their results have been interpreted as time-dependent crack growth caused by grain boundary cavitation. In this, cracking mechanism, cavity growth and consequently crack growth is aided by slow rate loading while unloading results in cavity healing with no contribution to the crack growth process. This cavity growth related mechanism is not operative in the highly creep resistant Alloy 718. In this alloy, for time-dependent environmental effects which are governed by temperature and load levels, both segments of the loading cycle should exert the same damage effects at the crack tip. Furthermore, cases C, D and E, which are of different frequencies but identical loading times, did not result in similar crack growth behavior so that an increase in the total cycle duration yields, as expected, an increase in the crack growth rate. The conclusion, based on these observations, is that for loading frequencies lower than the transitional frequency, the cycle effective oxidation time is equal to the total time of the loading cycle.

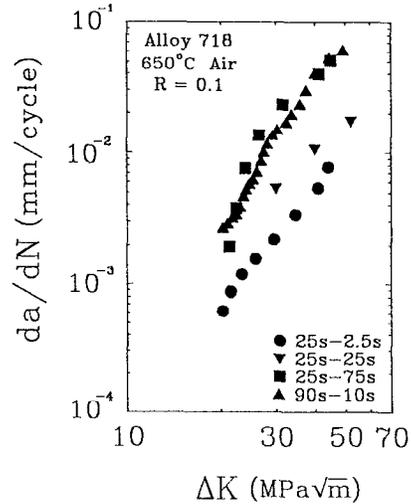


Fig.8 da/dN vs ΔK for different frequencies with different loading and unloading ratios.

4. Relationship between X and Stress Intensity Factor K

4.1 Concept

Work of Nicholas and Weerasooriya [2], has demonstrated that subjecting Alloy 718 to elevated temperature fatigue testing with a loading cycle having a frequency $f > f_c$, and an imposed hold time at the minimum load level for periods up to 1000 seconds did not result in any measurable acceleration in the crack growth rate when compared to the crack growth rate due to the base cycle without the hold time periods. By lowering the loading frequency to levels below the transitional frequency of the alloy, an accelerated intergranular crack growth was detected when addition of hold times at minimum load level were imposed for periods as small as 30 seconds, see Diboine et al [30]. This observed increase in the crack growth rate is interpreted in the work of Ghonem et al [9], as being a result of further intergranular oxidation taking place at

the crack tip during the hold time period, provided that the minimum load level does not contribute to the mechanical driving force of the crack tip. The crack growth rate under this type of loading calculated using a damage summation form similar to that of eq.(2) can be

$$\left(\frac{da}{dN}\right)_{\text{cyc-h-min}} = \left(\frac{da}{dN}\right)_{\text{cyc}} + \left(\frac{da}{dN}\right)_{\text{h}} \quad (5)$$

where the first term of the R.H.S. of the above equation represents the contribution to the crack growth rate due to the reversed part of the cycle and the second term is the contribution due to the hold time period. The above equation can be expressed in terms of (da/dt) as

$$\left(\frac{da}{dN}\right)_{\text{cyc-h-min}} = \int_{t_a} \left(\frac{da}{dt}\right) dt + \int_{t_h} \left(\frac{da}{dt}\right) dt \quad (6)$$

The term (da/dt) is viewed as being equal to the intergranular oxygen diffusion rate \dot{X} which, as mentioned before, is assumed here to be a function of the stress intensity factor, K. Eq.(6) could be rewritten as

$$\left(\frac{da}{dN}\right)_{\text{cyc-h-min}} = \left(\frac{da}{dN}\right)_{\text{cyc}} + \int_{t_h} \dot{X}(K, t) dt \quad (7)$$

As a first approximation, considering the average value of $\dot{X}(K, t)$ during the hold time period, i.e. $\langle \dot{X}(K) \rangle$, one could thus obtain this $\langle \dot{X}(K) \rangle$ from above equation

$$\langle \dot{X}(K) \rangle = \left(\frac{da}{dN}\right)_{\text{cyc-h-min}} - \left(\frac{da}{dN}\right)_{\text{cyc}} \quad (8)$$

This equation shows that the determination of \dot{X} at a certain value of K_h could then be achieved through the knowledge of $(da/dN)_{\text{cyc-h-min}}$ and $(da/dN)_{\text{cyc}}$, which permits a correlation between \dot{X} and K. The limitation in this approach, however, is the fact that K_h can not exceed the material threshold stress intensity factor which is estimated here as $8 \text{ MPa}\sqrt{\text{m}}$ (recall that the hold time load must not extend the crack during the hold time). A different approach is thus required to correlate \dot{X} with higher values of K. This can be achieved by considering the FCGR resulting from a low frequency cycle which include hold time periods at the maximum load level, $(da/dN)_{\text{cyc-h-max-air}}$, as being equal to the sum of the following components:

$$\left(\frac{da}{dN}\right)_{\text{cyc-h-max-air}} = \left(\frac{da}{dN}\right)_{\text{cyc-h-max-vac}} + \left(\frac{da}{dN}\right)_{\text{ox}} + \left(\begin{array}{c} \text{interaction} \\ \text{term} \end{array}\right) \quad (9)$$

where the first term in the R.H.S. of the equation is the FCGR due to an identical loading cycle performed in vacuum environment and $(da/dN)_{\text{ox}}$ is the increase in the crack growth in air environment due to grain boundary oxidation occurring during the entire cycle duration. The interaction term represents the mutual oxidation-creep effects which are assumed here to have a minimal influence on the total FCGR. This is based on the work of Ghonem et al [9] in which specimens made of Alloy 718 and subjected to predeformation have shown increased resistance to elevated temperature crack growth in air environment. The same results also obtained by Kendall et al [31] in their work of prestrained nickel-base alloy N901. One could then neglect the damaging effect of the interaction term, in particular at low and moderate values of K_{max} , where the small scale yielding condition is still appropriate in the crack tip region. Furthermore,

the oxidation term in eq.(9) can be expressed as the sum of two components. The first term is due to the oxidation effects during the reversible part of the cycle. The second term is due to oxidation effects associated with hold time period. This is written as

$$\begin{aligned} \left(\frac{da}{dN} \right)_{ox} &= \left(\frac{da}{dN} \right)_{cyc-air} + \left(\frac{da}{dN} \right)_{ox-h} \\ &= \left(\frac{da}{dN} \right)_{cyc-air} + \int_{t_h} \dot{X}(K, t) dt \end{aligned} \quad (10)$$

Following an argument similar to that proposed for eq.(6), the average value of $\dot{X}(K, t)$ during the hold time period, $\langle \dot{X}(K) \rangle$ can be obtained

$$\langle \dot{X}(K) \rangle = \left(\frac{da}{dN} \right)_{cyc-h-max-air} - \left(\frac{da}{dN} \right)_{cyc-h-max-vac} - \left(\frac{da}{dN} \right)_{cyc-air} \quad (11)$$

this type of correlation would then permit determining \dot{X} as function of K for values of K higher than that of the threshold level.

4.2 Experimental Requirements

An experimental program was carried out to provide data for solving eqs.(8) and (11). In this program two sets of crack growth experiments were performed on compact tension specimens made of the Alloy 718 previously described, at temperature level of 650°C and a stress ratio of 0.1. The first set of tests consisted of constant ΔK tests with and without hold time periods imposed at the minimum load level. Three different values of ΔK were selected; 30, 40, and 50 MPa \sqrt{m} . For each level of ΔK four different hold time periods were investigated 50, 100, 680 and 3600 seconds. The load level during the hold time period was determined such that the corresponding K_h value produced at the longest expected crack length ($a/w=0.7$) in the CT specimen would not exceed the K_{th} of this material. In this set of tests, the loading frequency was 0.02 Hz, which is lower than the transitional frequency f_c and thus insures that the crack growth process would be fully intergranular. Results of these tests in the form of the ratio $(da/dN)_h / (da/dN)_{cyc}$ vs t_h are shown in Fig.9. These results indicate that the crack growth rate is dependent on both t_h and the corresponding K value so that for the same value of ΔK the crack growth rate increases as t_h increases. There exists, however, a form of growth rate saturation as t_h approaches about 100 seconds; beyond this duration, oxidation passivation effects would prevent further oxygen diffusion at the crack tip as suggested by several authors, see for example the work of Andrieu et al [32]. Therefore, determination of the oxidation diffusion rate should rely on a time exposure of 100s or less; using time durations greater than this saturation time would lead

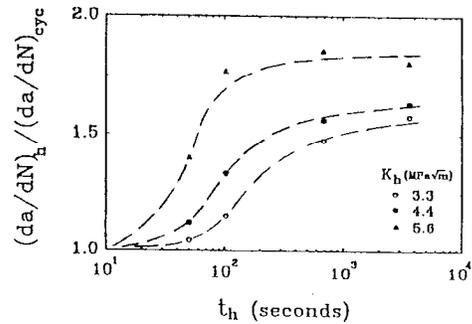


Fig.9 Effect of hold time at minimum load level on $(da/dN)_h / (da/dN)_{cyc}$ for different K_h levels.

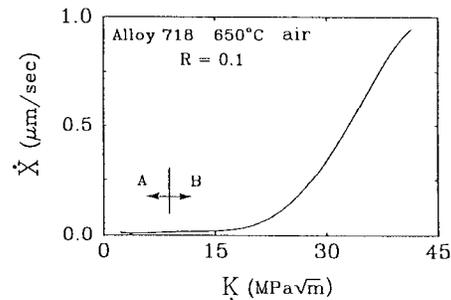


Fig.10 Intergranular oxygen diffusion rate \dot{X} vs stress intensity factor K obtained by (A) hold at minimum load tests and (B) hold at maximum load tests.

to underestimation of the diffusion depth. Fig.9 could then be used to provide the necessary input to eq.(8) yielding results in the form of \dot{X} vs K. These results are in region A of Fig.10.

The remaining requirements for solving equation were obtained by carrying out a second set of tests consisting of two constant ΔP crack growth experiments, in both air and vacuum (10^{-8} torr) environment. The loading cycle profile was identical for the two tests: a continuous cycle (no hold time) with a frequency of 0.05 Hz and an imposed hold time of 300 seconds at the maximum load level. FCGR vs ΔK data for these two tests are shown in Fig.6. These data were used to provide the terms $(da/dN)_{cyc-h-max-air}$ and $(da/dN)_{cyc-h-max-vac}$ in eq.(11). The term $(da/dN)_{cyc-air}$ was obtained from data corresponding to the data of 0.05 Hz tests in air results in form of \dot{X} vs K shown in region B of Fig.10. Now, by combining results of regions A and B in Fig.10, the relationship between \dot{X} and full range of K values can be established.

4.3 Analysis

From Fig.10, one can now establish a quantitative relationship between \dot{X} and K. This has been fitted to the form

$$\dot{X} = \frac{a}{1 + b e^{-cK}} \quad (12)$$

where a and b are related to upper and lower limits of the diffusion rate parameters, respectively, and c is a measure of the inflection of the \dot{X} vs K curve. Through nonlinear regression analysis the value a, b and c are calculated as 1.2×10^{-3} , 1696.96 and 0.2141, respectively.

In order to obtain an explicit expression of FCGR eq.(12) should then be substituted in eq.(4). For this purpose the stress intensity factor K can be expressed as function of time for triangle waveform loading in the following geometrical expression:

$$K = \begin{cases} (\alpha + \beta) \Delta K, & 0 \leq t \leq \frac{1}{2f} \\ (\alpha' - \beta) \Delta K, & \frac{1}{2f} \leq t \leq \frac{1}{f} \end{cases} \quad (13)$$

where ΔK is the applied stress-intensity factor range, R is the load ratio, $\alpha = R/(1-R)$, $\alpha' = (2-R)/(1-R)$ and $\beta = 2f$. An expression for the crack growth rate could then be obtained by substituting eq.(12) and eq.(13) into eq.(4), and integrating over the time range 0 to $1/f$, this yields

$$\frac{da}{dN} = Q \ln \left[\frac{(1 + \eta)(1 + \phi)}{(\zeta + \eta)(\zeta + \phi)} \right] \quad (14)$$

where

$$Q = \frac{a}{2cf\Delta K}, \quad \zeta = \exp(-c\Delta K)$$

$$\eta = \exp[-c(\alpha + 1)\Delta K], \quad \phi = \exp[-c(\alpha' - 1)\Delta K]$$

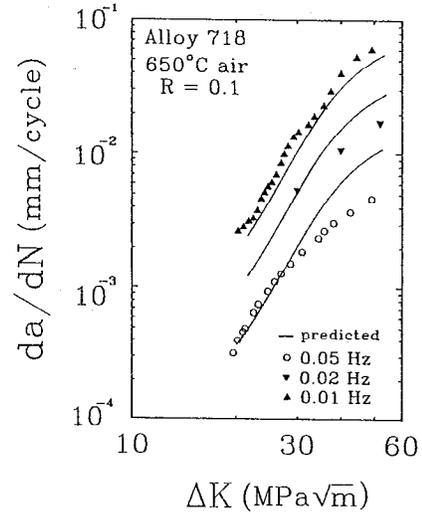


Fig.11 da/dN vs ΔK for both experimental and predicted results.

by applying series expansion on logarithmic function, eq.(14) can be rewritten as:

$$\frac{da}{dN} = \left[\sum_{m=1}^{\infty} G_m(f, R) \right] \Delta K^m \quad (15)$$

where

$$G_m = \frac{2ab}{f} \left[\frac{(-1)c^m}{(m+1)!} (A_{m+1} + B_{m+1}) \right]$$

$$A_{m+1} = \alpha^{m+1} - (\alpha + 1)^{m+1} + (\alpha' - 2)^{m+1} - (\alpha' - 1)^{m+1}$$

$$B_{m+1} = 2^{m+2}b [(\alpha + 1)^{m+1} - \alpha^{m+1} + (\alpha' - 1)^{m+1} - (\alpha' - 2)^{m+1}]$$

Eq.(14) is then used to predict fatigue crack growth rate data for three different loading frequencies, i.e. 0.01 Hz, 0.02 Hz and 0.05 Hz. Fig.13 shows a comparison between the predicted results and those experimentally generated. It could be observed that a general agreement exists between the values and trend of these two sets of data. A deviation however, is observed in the high ΔK regions for both 0.02 Hz and 0.05 Hz conditions. This deviation may be due to the fact that while eq.(14) treats the entire history of the crack growth process as time dependent, influence of very high ΔK values on saturating the time dependent effects, either due to decrease dependency of intergranular diffusion on K or due to increased cyclic damage to the extent that time available for diffusion is decreased, is not fully reflected in the model. One should observe however that for a very low frequency condition of 0.01 Hz, the theoretical and experimental results coincide in the full range of ΔK .

5. Conclusions

For cyclic loading frequencies lower than the transitional frequency of Alloy 718, the high temperature fatigue crack growth behavior is viewed here to be fully environment dependent process. Under this condition, the crack extension per cycle is assumed to be equal to the intergranular depth of oxygen diffusion taking place at the crack tip region during the cycle effective oxidation time. Two aspects have been particularly investigated - the cycle oxidation time and the grain boundary diffusion kinetics. The cycle oxidation time, t_{ox} , has been studied by comparing crack growth rates corresponding to symmetrical as well as asymmetrical waveform loadings with different frequencies. The subject of the diffusion kinetics of the grain boundary network in the crack tip region has been studied indirectly by performing a series of environmentally-controlled crack growth experiments. The conclusions of these two above mentioned studies can be summarized as:

1. The existence of an environment-dependent crack growth stage in Alloy 718 at elevated temperature conditions is supported by comparing results of crack growth tests in both air and vacuum conditions.
2. For creep-resistance high strength Alloy 718, the cycle oxidation time should be taken as the whole cycle duration including both the loading and unloading part of the cycle.
3. The grain boundary diffusivity in the crack tip region can not be treated as a material constant. It was found to depend on both the exposure time and the crack tip stress intensity factor; the higher the K level or the larger the exposure time t , the longer the intergranular depth of oxygen diffusion.
4. A closed form solution correlating the intergranular oxygen diffusion rate in the crack tip region with both t_{ox} and K is obtained. This solution when integrated, provides a measure of the crack growth rate. A reasonable agreement is observed, particularly in low ΔK region, when comparing results of this analytical solution with those experimentally generated.

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