ON FATIGUE-CREEP-ENVIRONMENT INTERACTION AND THE FEASIBILITY OF FATIGUE MAPS

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ABSTRACT

The degradation of fatigue properties with increasing temperature and decreasing frequency is generally attributed to the fatigue-creep-environment interaction. A recently developed model which can be used to quantitatively predict the critical frequency below which intrinsic creep damage, i.e., intergranular cavitation, will occur, has been introduced in this paper. Based on this model, it is argued that for most laboratory tests the reported frequency dependence of fatigue life was mainly due to environmental degradation rather than creep damage. The vital role of the environment in determining the fatigue fracture mode is emphasized. Examples of fatigue cavity nucleation maps and fatigue crack propagation maps are included to explicitly display the various mechanisms involved in fatigue crack initiation and propagation.

KEY WORDS

High temperature, fatigue, creep, environment, cavitation, fatigue maps.

INTRODUCTION

It is now well established that frequency and hold times have a pronounced influence on fatigue behaviour of metals at elevated temperature (Coffin, 1980; Conway, Berling and Stentz, 1970, 1973; Manson, 1973; Solomon and Coffin, 1973; Woodford and Coffin, 1974). It is generally observed that fatigue life is reduced with decreasing frequency and increasing temperature; and the introduction of a hold period at peak stress (or strain) in the tensile going half cycle generally functions in the same way as a decrease in frequency (Plumtree and Yu, 1983; Sadananda and Shahinian, 1980). The reduction in fatigue life is generally attributed to the fatigue-creep-environment interaction. However, there is some uncertainty whether the creep (bulk cavitation) effect or the environmental effect is the dominant factor. Some researchers believed that the frequency dependence of fatigue life was mainly due to the creep effect. Others, (Coffin, 1972; Woodford and Coffin, 1974) among whom Coffin and his colleagues were leading figures, concluded that environmental degradation was the cause of the reduction in fatigue life. This uncertainty is understandable because the fatigue-creep-environment interaction is a complex subject. Essentially all of the reported studies appear to confirm that this effect depends on several factors which include material, temperature, frequency, stress (or strain) range and wave form. However, no attempt has previously been made to quantitatively explain this phenomenon from a mechanistic point of view. Starting with a brief overall review of the topic of intergranular cavitation in fatigue at elevated temperature, a quantitative model for fatigue cavitation is presented in this paper. In this model it is argued that the frequency dependence of intergranular cavitation is mainly the manifestation of the incubation time for cavity nucleation, which is the time for a cavity nucleus to grow to a critical size by vacancy clustering. It will be shown that this model provides a unifying approach to the subject of fatigue-creep-environment interaction.

INTERGRANULAR CAVITY FORMATION

It is generally agreed that creep damage in fatigue is due to the nucleation, growth and final linkage of cavities on grain boundaries under cyclic loading at elevated temperatures. In the past, research on this subject was mainly focused on two aspects: the micro-mechanisms of intergranular cavitation and the life behaviour of engineering materials.

For the first aspect, the studies were mainly performed on wrought pure metals and solid solution alloys (Gittins, 1968; Saegusa and Weertman, 1978; Singh and others, 1977; Skelton, 1966; Wigmore and Smith, 1971; Williams and Corti, 1968), although additional studies have been performed on more complex systems (Driver, 1971; Hill, 1967; Okazaki and others, 1983; Well and Sullivan, 1968). Tests were generally carried out at relatively high frequencies. These studies showed that in systems where grain boundary migration (GBM) is easy, that is, systems with virtually no second phases at grain boundaries, the boundaries align themselves at 45 degrees to the stress axis at an early stage of the fatigue test. Grain boundary serrations are developed as a result of the GBM, and these grain boundary irregularities serve as cavity nucleation sites (Saegusa and Weertman, 1978; Snowden, 1966). Statistical analysis showed that the cavitation on 45 degree boundaries exceeds the percentage of 45 degree boundaries in the total boundary population (Williams and Conti, 1968). These results strongly suggested that grain boundary sliding (GBS) is vital for cavity nucleation. Two critical experiments were performed that reinforce the evidence for G.B.S. controlled cavity nucleation. Skelton & Evans (1969) measured boundary offsets due to G.B.S. in the alloy Magnox AL80 and found the mean offset at fracture was equal to the mean cavity spacing. Williams and Corti (1968) found that in pure copper the cavity distribution peaked on boundaries oriented at 45 degrees to the stress axis in push/pull cycling, but on boundaries oriented at 0 and 90 degrees to the specimen axis in torsional cycling. One important discovery was that cavitation in fatigue at relatively high frequencies only took place when the frequency exceeded a certain limit, below which cavitation was absent (Skelton, 1966).

As far as the effect of frequency and hold time on fatigue life behaviour is concerned, laboratory tests were mainly performed on engineering

materials for high temperature applications. Among them, stainless steels and nickel base alloys were the most thoroughly studied materials (Coffin, 1977; Conway, Berling and Stentz, 1973; Sadananda and Shahinian, 1980; Solomon and Coffin, 1973; Woodford and Coffin, 1974). Most of the tests were carried out using relatively long cycle times. The outcome of this research has been recently reviewed by Coffin (1977, 1980). The most important discovery was that for fatigue tests in vacuum, intergranular cavitation and fracture only took place when the frequency was below a certain value; introducing an aggressive environment shifted the critical frequency to a higher value with a substantial reduction in fatigue life. Fatigue cavitation in these cases took place mainly on grain boundaries perpendicular to the stress axis (Coffin, 1980; Hwang and Pande, 1983; Okazaki and others, 1983; Sidey and Coffin, 1978). This suggests that the normal stress across the boundary was important for cavity nucleation and growth.

It was found that the introduction of a hold time at the peak tensile stress point generally reduced fatigue life. The extent of this reduction depended on the length of the hold time. When the hold time was relatively short, it functioned in the same way as a decrease in frequency (Conway and Berling, 1970; Conway, Berling and Stentz, 1973). In other words, for a given strain range, the same fracture time was obtained at the same value of the cycle time regardless of the wave form involved. On the other hand, a relatively long hold time led to grain boundary cavity formation, independent of the environment. The evidence for this arises from intergranular cracking in tensile strain hold tests in 20Cr-25Ni-Nb stainless steel at 750°C (Tomkins) and in a 1CrMov steel at 565°C (Ellison and Paterson, 1976) where little environmental effect can be expected. Correspondingly, a long period of hold time reduced fatigue life considerably more than balanced cycle fatigue tests at the same cycle time. An important discovery was that the reduction in fatigue life due to the hold time revealed a saturation effect (Conway, Berling and Stentz, 1973). An unanswered question, therefore, is whether such a saturation in loss of life occurs whenever longer hold time is employed.

In an attempt to clarify the seemingly controversial discoveries of the frequency dependence of fatigue cavitation mentioned above, Tang (1983) has recently suggested that a fatigue cavity can nucleate either at stress concentration sites at a grain boundary or at stress relaxed grain boundaries. For the former, the cyclic frequency must exceed a limit which is decided by the stress relaxation time, and for the latter, the time spent on the tensile going half cycle must exceed the incubation time for cavity nucleation. This incubation time is the time for a cavity embryo to reach a critical size which allows growth rather than being sintered out by vacancy back diffusion. Following the classical theory of heterogeneous nucleation of cavities by vacancy clustering (Raj and Ashby, 1975; Raj, 1978; Yoo and Trimkaus 1983), the correct estimate of the incubation time is given by

$$t_i = \frac{1}{6Z^2R_c} \tag{1}$$

where Z is the "Zeldovich factor" that describes back and forth fluctuations between the subcritical and supercritical regions (Zeldovich, 1942), and R_c is the reaction rate given by the probability of vacancy emission or absorption by the critical nuclei. This is often set equal to the probability of finding a vacancy on the surface of the critical nucleus times the jump frequency of the vacancy (Raj and Ashby, 1975). Substituting corresponding expressions for Z and R_c into Equation (1), one finds the incubation time is

$$t_i = \frac{\pi^2 KT}{2\sigma\Omega} \cdot \frac{r_c^3 F_v}{4C_o \delta D_b^o \exp(-Q_b/RT)} \quad (2)$$

where σ is the normal traction to the grain boundary, Ω is the atomic volume, r_c is the critical radius of a cavity for growth, F_v is a geometrical factor that depends upon the ratios of the specific surface and interfacial energy, C_o is the saturated vacancy concentration in the grain boundary, and δ, D_b, Q_b, R, K and T have their conventional meanings. It has been generally accepted (Taplin and Collins, 1978; Taplin, Topper and Tang, 1983; Veevers and Snowden, 1975; Weertman, 1974) that the critical radius r_c of a cavity for growth in balanced cycle fatigue is

$$r_c = \frac{4KT\gamma}{\sigma_o^2 \Omega} \quad (3)$$

where γ is the surface energy and σ_o is the stress amplitude. Substituting Equation (3) into Equation (2), one finds

$$t_i = \frac{8\pi^2 K^4 T^4 \gamma^3 F_v}{\sigma_o^7 \Omega^4 C_o \delta D_b^o \exp(-Q_b/RT)} \quad (4)$$

Therefore the critical frequency f_c below which the intrinsic creep damage would occur is given by

$$f_c = \frac{\sigma_o^7 \Omega^4 C_o \delta D_b^o \exp(-Q_b/RT)}{16\pi^2 K^4 T^4 \gamma^3 F_v} \quad (5)$$

A large number of fatigue tests have been performed using strain as a controlling parameter. If the stress amplitude-plastic strain range response of the material in the cyclically stabilized state can be described by

$$\sigma = A\,\Delta\varepsilon_p^{n'} \quad (6)$$

where $\Delta\varepsilon_p$ is the plastic strain range, n' is the cyclic strain hardening coefficient, and A is a constant, then Equation (5) can be modified as

$$f_c = \frac{A' \Delta\varepsilon_p^{7n'} \Omega^4 C_o \delta D_b^o \exp(-Q_b/RT)}{16\pi^2 K^4 T^4 \gamma^3 F_v} \quad (7)$$

It is apparent that the critical frequency for intrinsic creep damage to occur is a function of material characteristics, applied stress (or strain) and temperature. It has been found that the predicted values using Equation (5) or (7) agree with experimental results very well (Tang, 1983). For example, for strain-controlled fatigue tests, Coffin (1980) reported that for AISI 304 stainless steel at $\Delta\varepsilon_p = 0.01$ and T = 923 K, the critical frequency was lower than 1.6×10^{-4} Hz, and for the iron-nickel based alloy A286 at $\Delta\varepsilon_p = 0.01$ and T = 866°K, the critical frequency was 1.6×10^{-4} Hz. The predicted values using the newly advanced

model are 1.5×10^{-5} Hz and 5×10^{-5} Hz respectively. The agreement is very good when the vagaries of diffusion data are considered. Having developed the quantitative model, we are now able to more accurately interpret experimental results and provide answers to several previously unresolved questions.

Creep Effect or Environmental Degradation

Balanced cycle. Table 1 lists the predicted critical frequencies for several engineering materials at their half absolute melting points and at a stress amplitude σ_o = 100 MPa. It can be seen that for stainless steels and nickel base superalloys the values of the critical frequencies at this stress level are 1×10^{-7} and 5×10^{-6} Hz respectively. Due to the strong dependence of the critical frequency on the stress amplitude, the critical frequency is about 10^{-5} Hz for strain-controlled fatigue tests at $\Delta\varepsilon_t$ = 1%, i.e. one cycle per day. Only in rare cases have balanced cycle fatigue tests been performed at such a low frequency. It is apparent that for most laboratory tests the reported reduction in fatigue life at elevated temperatures was mainly due to environmental degradation, rather than the creep effect.

Hold time. For fatigue tests with a hold period at the peak tensile stress point the critical radius for a cavity to grow is considerably reduced. The expression is given by

$$r_c = \frac{2\gamma}{\sigma_{mean}} \quad (8)$$

where σ_{mean} is the time mean of the stress over one cycle, and γ is the surface energy (the details of the derivation of Equation 8 will be published in due course). Consequently, Equation 5 should be modified to take this into account. The minimum hold period required for the intrinsic creep damage to occur is decided by the incubation time required for a cavity embryo to reach the critical radius given by Equation 8. This problem is essentially a problem of cavity nucleation in creep. For cavity nucleation in creep, Raj (1978) theoretically predicted that the cavity nucleation rate was strongly dependent on the local stress. This strong stress dependence suggested an introduction of a threshold stress, below which the nucleation rate is negligible. The existence of the threshold stress was experimentally observed by Sadananda and Shahinian (1977), and Fleck et al (1975). The magnitude of this threshold varies from material to material. For example, the threshold stress was about 10 MPa for OFHC copper (Fleck, Taplin and Beevers, 1975), and about 170 MPa for Inconel 718 (Sadananda and Shahinian, 1977). Based on the above discussion, it is apparent that the cause of the reduction in life in fatigue with a hold time may either be environmental degradation or a true creep effect. If the hold time is relatively short and the stress level at which the hold was introduced is low, the cause of the reduction in life is mainly due to environmental effect. Under such circumstances, introducing a compressive dwell period would have no healing effect (Thorpe and Smith, 1981). On the other hand, if the hold time is relatively long, and the stress level at which the hold time was introduced is high, the cause of the reduction in life is mainly due to creep damage. In this case, a compressive dwell period would have a large healing effect, as reported by a number of researchers (Cheng and co-workers, 1973; Majumdar and Maiya, 1978).

TABLE 1 Predicted Critical Frequency for Selected Engineering Materials (at $T = 0.5\ T_m$, $\sigma_o = 100$ MPa)

Material	f_c Hz	Melting Point T_m °K	Surface Energy γ J/m	Atomic Volume Ω m^3 x 10^{-29}	Diffusion Coefficient δD_b^o m^3/s	Q_b KJ/mole
Aluminum	4.2×10^{-3}	933	1.1	1.66	5×10^{-14}	84
OFHC Copper	2.3×10^{-4}	1356	1.5	1.18	5×10^{-15}	104
α - Iron	8.5×10^{-6}	1810	1.8	1.18	1.1×10^{-12}	174
Nickel	2.5×10^{-5}	1726	1.9	1.09	3.5×10^{-15}	115
Ni - 20Cr	3.2×10^{-6}	1673	1.6	1.1	2.8×10^{-15}	115
Inconel X-750	3.2×10^{-6}	1673	1.6	1.1	2.8×10^{-15}	115
Inconel 718	3.2×10^{-6}	1673	1.6	1.1	2.8×10^{-15}	115
A286	1×10^{-7}	1680	1.7	1.21	2.0×10^{-13}	167
AISI 304	1×10^{-7}	1680	1.7	1.21	2.0×10^{-13}	167
AISI 316	1×10^{-7}	1680	1.7	1.21	2.0×10^{-13}	167
1%Cr - Mo-V	5×10^{-7}	1753	1.7	1.18	1.1×10^{-12}	174
α - Titanium	8×10^{-4}	1933	1.7	1.76	3.6×10^{-16}	97
Niobium	3×10^{-7}	2741	2.1	1.80	5.0×10^{-14}	263

Data are mainly taken from Deformation-Mechanism Maps for Metals and Alloys by H.J. Frost and M.F. Ashby (1983).

Environment-induced creep damage

There is ample experimental evidence indicating that an air environment encourages intergranular fatigue cavitation. The accelerating onset of the intergranular cavitation in the presence of the air environment at elevated temperature is termed environment-induced creep damage in this paper. This subject is a rather complicated topic, and our understanding of the mechanisms at work is still in an early stage of development. A few possible mechanisms are suggested below:

1. Oxygen absorption on the cavity surface reduces the surface energy thereby reducing the critical radius of a cavity for growth. Since the maximum reduction in surface energy due to oxygen absorption is about 50% (Seah, 1975), the critical frequency in Equation 5 would correspondingly be increased by 8 times.

2. Oxygen may react with reactive species in the grain boundaries and deposit oxides at them. If the Pilling-Bedworth ratio of the oxides is greater than 1, the oxides would wedge grain boundaries apart, therefore effectively increasing the normal traction on the grain boundaries. This will again increase f_c.

3. Oxide deposition on grain boundaries would reduce their cohesion and introduce some grain boundary porosity. This would increase grain boundary diffusivity, and further increase f_c.

4. Finally, the air environment may introduce a new mechanism for fatigue cavity nucleation. It has been suggested by a number of researchers (Bricknell and Woodford, 1981; Dyson, 1982; Nich and Nix, 1981; Taplin, Topper and Tang, 1983) that if the reaction products of oxygen with reactive species in grain boundaries are gases, they would precipitate out in grain boundaries as gas bubbles. These will become cavity nuclei. It was reported that pre-oxidation of a Udimet 700 specimen at 1800 F followed by fatigue testing at 1400 F produced many surface intergranular cracks, whereas testing of specimens without prior oxidation produces a single intergranular crack (Paskiet, Boone and Sullivan, 1972). Similar observation was made on the AMAX MZC alloy (Tang, 1983). For the former, oxygen reacting with grain boundary carbides forming CO_2 bubbles was probably the responsible mechanism (Bricknell and Woodford, 1981; Dyson, 1982). For the latter, oxygen absorbed by the grain boundaries and reacting with trace hydrogen in the lab atmosphere to form water vapour bubbles was suggested as the responsible mechanism. It can be mentioned here that a large internal pressure within the bubbles was introduced during the gas forming reaction. Raj theoretically predicted that an internal pressure as high as 1000 MPa may occur during the carbide oxidation reaction (Raj, 1982). Spherical grain boundary bubbles as small as 50 Å in diameter were observed in the fatigue tested specimens which were prior oxidized (Tang and Taplin, 1983). This was a clear indication of the large internal pressure which stabilized these small bubbles.

The Effect of the Cycle-Dependent Stress Response of the Materials in Strain Controlled Fatigue on the Fracture Mode

Engineering materials commonly reveal rather complicated stress response in strain-controlled fatigue. For example, AISI 304, 316 stainless steels revealed an initial cyclic hardening period which was followed by a continuous softening process after conventional processing (Berling and Slot, 1969). This phenomenon is expected to influence the onset of intergranular cavitation in fatigue at a given frequency. According to Equation 5, the critical frequency for intergranular cavitation is strongly dependent on stress. The ratio of the stress amplitude at the hardening peak to the initial stress was 1.4 for AISI 316 in fatigue at $\Delta\varepsilon_t = 2\%$ and 650°C (Berling and Slot, 1969). This would reduce the critical frequency f_c by one order of magnitude. Therefore, for fatigue testing at a given frequency and a given temperature intergranular cavitation may be delayed until the hardening peak is reached. Thus, the fracture process would be a transgranular initiation followed by an intergranular propagation in this case. Since intergranular propagation of fatigue cracks is an extremely rapid process, the transgranular initiation would dominate the fatigue life. Since the frequency has only a minor effect on transgranular crack initiation, one would expect a saturation of the frequency effect on fatigue life. This kind of saturation effect was observed by Conway et al (1973) in fatigue testing of smooth specimens of AISI 304 stainless steel. Based on the same argument, one can conclude that such a saturation effect will not be observed when notched specimens are tested. This was proved by Coffin and his colleagues (1973, 1974).

For fatigue tests with hold time imposed in tensile strain, the reduction in life with increasing hold time also revealed a saturation effect (Conway, Berling and Stentz, 1973; Rezgui, Petrequin and Mottot, 1981). In addition to the explanation mentioned earlier, the hold time saturation effect was believed to be related to the saturation in stress relaxation during the hold period as well. As an example, for fatigue tests of AISI 316 S.S. at 650°C and at a total strain range of 1%, 7 hours tension holding leads to a total relaxation amount 28%, 25% of which has been consumed during the first 2 hours. The stress relaxation seems to have a strong effect on cavity nucleation and growth.

A Qualitative Understanding of the Fracture Modes

At elevated temperatures a variety of combinations of modes of fatigue crack initiation and propagation is possible. Table 2 outlines the effect on the fracture modes of the frequency and the materials properties at a fixed temperature, stress (or strain) range and environment. With decreasing frequency one expects more intergranular fracture. Pre-oxidation of the material at high temperature would considerably weaken the grain boundary cohesion, thereby encouraging intergranular crack initiation.

Both intergranular and transgranular fracture may be accelerated by an air environment. However the intrinsic creep damage is an extremely rapid process. When the cycle time exceeds a certain limit and the intrinsic creep damage occurs, the intergranular cavity growth would overwhelm the environmental degradation. Under such a circumstance, a vacuum test will be nearly as damaging as the corresponding air test. This may explain Coffin's observation (1980).

FATIGUE MECHANISM MAPS

The concept of maps as a means of describing the change in mechanical properties of materials with temperature is largely due to Ashby (1972). Since the first group of deformation mechanism maps was introduced in 1972, the mapping concept has been extended to cover the mechanisms of fracture, sintering, formability, and fatigue. A review of this subject can be found in Taplin, Pandey and Tang's article (1983).

Early development of fatigue maps (Collins, 1979) ignored the fact that fatigue is a two stage process, i.e., crack initiation and crack propagation, therefore the development was largely unsuccessful. Recently, fatigue cavity nucleation maps and fatigue crack propagation maps have been introduced by Tang and others (1983). Fig. 1 is a fatigue cavity nucleation map for AISI 304 stainless steel. It clearly displays that how the cyclic frequency and applied stress affect the nucleation of cavities

TABLE 2 Summary of Fracture Modes

	Damage Mode: Initiation	Damage Mode: Propagation	frequency	material's condition
1.	T*	T	$\gg f_c$	
2.	T	I	f_c	cyclic hardening under strain control
3.	I*	T	$> f_c$	surface preoxidized
4.	I	I	$< f_c$	

T* transgranular
I* intergranular
f_c critical frequency given by Equation 5

in fatigue at elevated temperature. The effect of the air environment on the cavity nucleation can also be seen clearly. The map is divided into six regions. At high stress levels (region A) transgranular void formation is dominant, the specimen will fail within a few cycles. At low stresses (below the endurance limit) no significant damage takes place (region B). In atmospheres other than vacuum, this level is reduced at lower frequencies. Intergranular cavity nucleation takes place at intermediate stress levels. The line which separates the region of cavity nucleation at stress-concentration sites (region C) from that where transgranular surface crack dominates (region D) is constructed by assuming that the surviving cavities are the same size as those in creep. In the centre of the map (region E), the dominant damage mechanism will be environment-assisted cavitation and/or transgranular surface crack formation. In general, environment-assisted cavitation will facilitate nucleation at stress relaxed grain boundaries (region F), thereby lowering the dividing line between regions E and F. Intergranular fracture will then shift to higher frequencies. It is apparent that transgranular surface crack formation will be more dominant at the higher frequencies and lower stress (i.e., bottom-right of region E). The position of the boundary separating the two is totally dependent on the specific environment under consideration. Some cases of intergranular initiation of cavities reported in the literature are depicted in Fig. 1. It can be seen that an air environment assisted intergranular cavitation to occur in regions E and C.

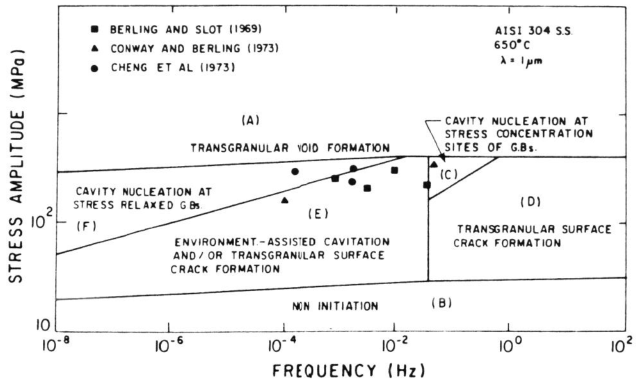

Fig. 1 Fatigue cavity nucleation map for AISI 304 stainless steel at 650°C. All data points refer to intergranular fracture.

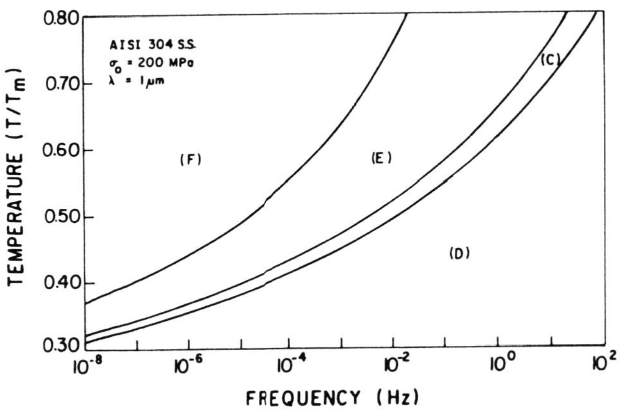

Fig. 2 Temperature-frequency version of the fatigue cavity nucleation map for AISI 304 stainless steel at a stress amplitude of 200 MPa, Regions C, D, E and F shown are the same as those in Fig. 1.

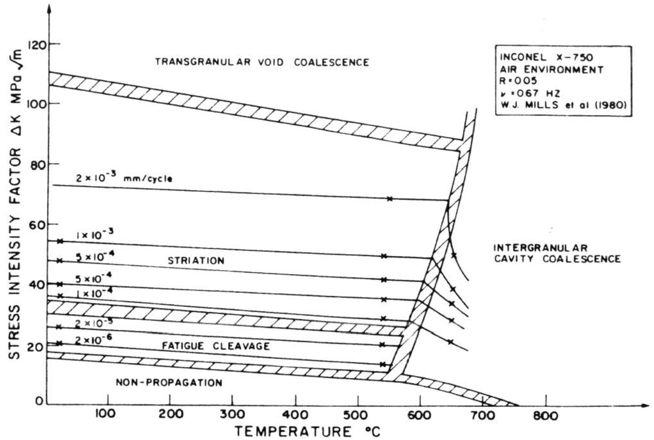

Fig. 3 Fatigue crack propagation mechanism map for Inconel X-750, at a frequency of 0.67 Hz.

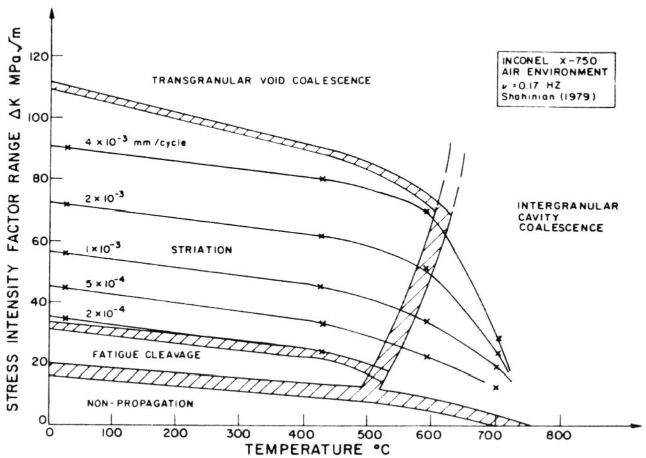

Fig. 4 Fatigue crack propagation mechanism map for Inconel X-750, at a frequency of 0.17 Hz.

Fig. 2 is another version of the map produced by plotting temperature versus frequency for a given stress (in this case 200 MPa). A similar map for Udimet 700, based on experimental data rather than theoretical prediction, was constructed by Runkle and Pelloux (1978). Since their tests were performed in a rather narrow range of frequency, the boundary lines which separated different regions were assumed to be straight lines. Nevertheless, the general trend of these boundary lines was the same as in Fig. 2.

Fig. 3 and 4 are experimental fatigue crack propagation mechanism maps for Inconel alloy X-750 at two frequencies. Once again the map is divided into six regions, with one mechanism being dominant in each domain. It can be seen that when the fracture path is transgranular, the crack propagation rate is marginally affected by the temperature. However, the crack propagation rate is considerably increased when intergranular cavitation becomes the dominant mechanism for damage accumulation. By comparing Fig. 3 with 4, one can find that at the lower frequency of 0.17 Hz the temperature dependence of the crack propagation rate increases. This is apparently due to increasing environmental attacks. Also, one finds that the lower frequency advances the onset of intergranular propagation to a lower temperature at a given stress intensity factor range. An interesting comparison of the FCP behaviour can be made between the Inconel alloy X-750 (Fig. 4) and Inconel alloy 718 (Fig. 5). The Inconel alloy 718 was more vulnerable to intergranular cavitation. This is due to the high strength of Inconel alloy 718 in comparison with Inconel alloy X-750. Figures 6-8 illustrate our recent developments

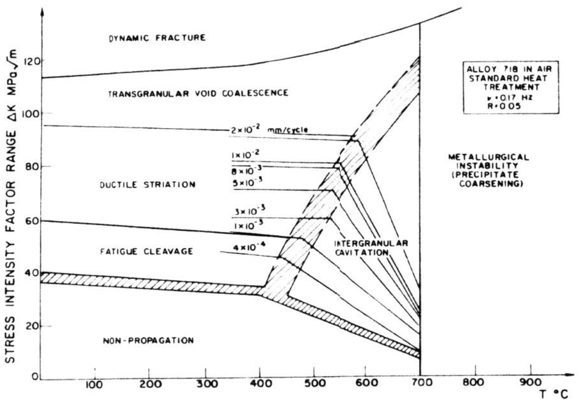

Fig. 5 Fatigue crack propagation mechanism map for Inconel 718.

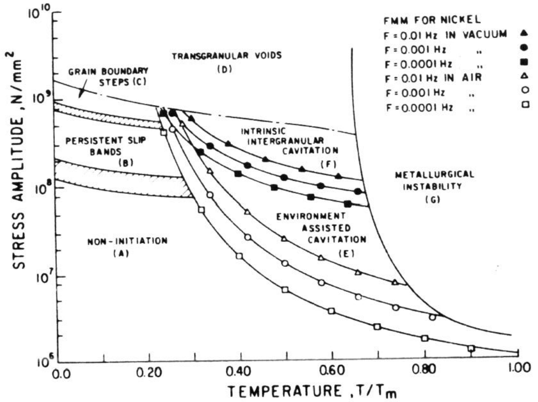

Fig 6 Recent Fatigue Mechanism Map for Nickel

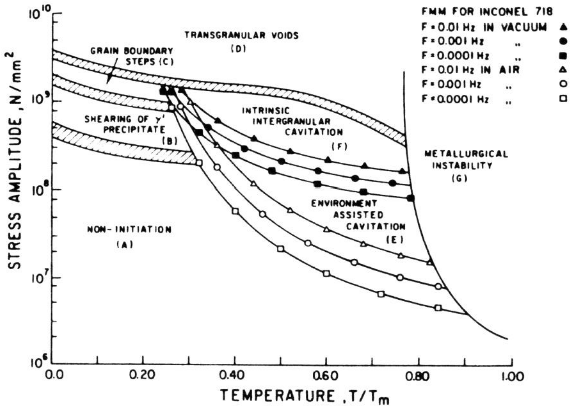

Fig 7 Recent Fatigue Mechanism Map for Inconel 718

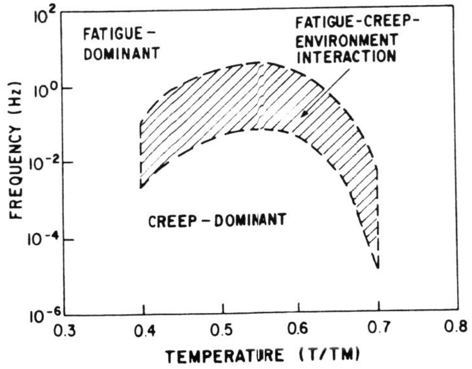

Fig 8 Schematic Indicating Frequency-Temperature Regime where strong Fatigue-Creep-Environment Interaction Takes Place

These maps, at their early stage of development, are by no means perfect. However, they do provide valuable information of the initiation and accumulation of fatigue damage. They have been proved to be very useful teaching tools. With further improvement, they will find useful applications for life prediction and alloy design.

SUMMARY

A quantitative model has been introduced in this paper which can be used to find the critical cyclic frequency below which intrinsic creep damage would occur. A better understanding of the relative importance of the environmental degradation versus the creep effect in determining the fatigue properties in high temperature fatigue has been achieved through the development of this model. It has been shown that for most laboratory tests the frequency dependence of the fatigue life and the detrimental effect of hold time are due to environmental degradation rather than creep effect. Fatigue mechanism maps clearly display the dominant mechanisms for fatigue damage initiation and accumulation. This is a promising area for future research.

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