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Fatigue Crack Propagation Behaviour of a TiAl-based Alloy

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Abstract: In the present study the fatigue crack propagation resistance of a nearly fully lamellar quaternary TiAl based is investigated at room-temperature. The fatigue crack growth rates material are shown to be highly sensitive on the applied stress intensity factor amplitude. A special attention is also paid to the influence of extrinsic factors. Thus crack-closure is proved to strongly influence the propagation. As regards environmental effects, ambient air induces fatigue crack growth rates two orders of magnitude higher than those measured in vacuum. The role of moisture in this deleterious effect is discussed.

1. INTRODUCTION

TiAl-based intermetallic compounds are currently extensively studied due to their potential in high temperature structural applications. Indeed, they exhibit attractive properties: low density, good oxidation resistance, excellent modulus retention, reasonable fracture toughness and good elevated temperature strength[1,2]. However, their poor room-temperature ductility and fracture toughness, and their sensitivity to environment restrict any industrial application. In addition, for some specific applications like rotating machines, the damage tolerance assessment might constitute a key issue for material development as for structural design. With this respect, it is necessary to investigate the fatigue crack growth behaviour over a wide range of temperatures including temperatures below the ductile-brittle transition. However, the fatigue crack propagation (FCP) behaviour of such brittle materials is still ill-known. In particular, a critical point is to determine whether some of the mechanisms which have been identified in the FCP behaviour of ductile alloys still applies in this case.

In FCP studies the fatigue crack growth rates (FCGRs) noticed da/dN are commonly determined with respect to the stress intensity factor amplitude ΔK . Nevertheless several factors have been shown to affect the FCGRs. These factors might be classified as intrinsic parameters (microstructure, elastic constants,

plastic behaviour, etc.) and extrinsic parameters (load ratio R defined as: $R = \frac{K_{min}}{K_{max}}$, environment,

temperature). As regards the latter class and more particularly the influence of the R ratio, experimental results bring out that, for a given ΔK value, the lower the R value, the slower the FCGR. Actually it has been proved that this influence is related to the crack closure phenomenon. Indeed, for low R values, a premature re-contact of the crack lips is observed at low loading levels[3]. Therefore the lower part of the loading cycle does no longer contribute to the fatigue damage at the crack tip. That means that the effective stress intensity factor amplitude ΔK_{eff} at the crack tip is lower than the applied nominal value ΔK . The use of ΔK_{eff} has been proved to successfully account for the observed R influence but almost is the only way to determine the intrinsic FCP behaviour of a given material. However, in this aim, one must also get free from the influence of the environment. Indeed, FCGR measured in ambient air are generally much higher than those obtained in an inert atmosphere like vacuum[4]. The observed FCP enhancement is assumed to result from a complex interaction between the cyclically strained material at the crack tip and moisture present in the test environment. This interaction has recently been analysed according to two distinct processes[5,6]:

- the adsorption of water vapour molecules onto freshly created surfaces which does not alter the basic FCP mechanism,
- a FCP regime assisted by the hydrogen released by the dissociative surface reaction of the adsorbed molecules and subsequently swept by the mobile dislocations.

The applicability of these concepts to the FCP behaviour of intermetallics at room temperature is however under question.

Hence, the present study has a double purpose:

- i) to investigate the FCP resistance of a quaternary TiAl-based alloy at room temperature,
- ii) to get a first insight into the fatigue mechanism, the role of crack closure and the environmental contribution to the damage process in such a brittle material.

2. TEST MATERIAL AND EXPERIMENTAL PROCEDURE

The material is a quaternary alloy Ti-48Al-2Mn-2Nb provided by the IRC in Birmingham. It is a part of an ingot produced in a large-scale plasma furnace and cooled relatively slowly in air. Details are available elsewhere[7,8]. The microstructure consists of coarse alternating γ and α_2 plates grains. The plates are perpendicular to the grain longitudinal axis. The size of these grains is approximately 300-400 μm . Some sparse equiaxed gamma-grains can be found between the lamellar colonies. The colonies are elongated in the longitudinal direction near the centre part of the ingot while they exhibit a radial elongation near the edge.

Fatigue crack growth experiments were carried out using CT type specimens ($W=22\text{mm}$, $B=5\text{mm}$) taken with different orientations within the ingot. The conventional notch was prolonged by short slit ($\approx 0.05\text{ mm}$) performed by electrodischarge machine. Two types of specimen with notch planes globally perpendicular to the lamellae planes were considered. In the first orientation, designed as notch "A"[7,8] or as "crack arrester" by Chan [9], the alternate plates of γ and α_2 are stacked at high angles with respect to the notch plane and the propagation direction. The second orientation, noticed notch "B"[7,8] or "crack divider"[9] is characterised by a stacking mode still almost perpendicular to the crack plane but nearly parallel to the crack direction.

The fatigue crack growth tests were conducted at room temperature in tension-tension loading with a load ratio $R=0,1$ on a servohydraulic machine at 35 Hz. Since this procedure might result in a premature catastrophic failure of the specimen, it requires a highly time-consuming care. However it allows to avoid changes in the R values, and especially a change from a compression/compression loading to a tension/tension, which might induce transient effect on the closure behaviour.

The crack length was optically determined on the polished side of the specimens with a travelling microscope. Crack closure measurements were performed by the unloading compliance method using a strain gauge bonded on the back face of the specimens at test frequency. The U parameter ($U=\Delta K_{\text{eff}}/\Delta K$) is computed for each data point on the basis these compliance plots [7].

3. RESULTS AND DISCUSSION

3.1. Fatigue crack propagation

The results obtained in air and in vacuum are presented on figures 1 and 2 respectively. The crack growth rates have been determined over nearly five decades, namely from 10^{-10} to 10^{-5} m/cycle. The salient feature of these curves consists in the high Paris law exponent exhibited. In other words, crack growth rates are extremely sensitive to the applied ΔK value. The apparent m exponents, although sometimes hard to estimate, are much higher than those encountered for ductile alloys but are closer to those encountered in ceramics. Whatever the environmental conditions and the lamellae orientations, the crack propagation process was intermittent as noticed by Davidson and Campbell in a similar titanium aluminide[10]. While the crack remained dormant, features similar to slip bands appeared and developed, even far ahead of the crack tip. These bands were parallel to the lamellae (figure 3). Numerous crack branching, bridging and secondary cracking were also observed. This spasmodic growth process and this diffuse damage in the crack wake might explain the scatter on crack growth rates.

The examination of fracture surfaces in SEM does not bring out any significant difference between air and vacuum. The main distinction comes from the different orientation of the lamellae. For the notch "A" samples, the crack propagation was essentially translamellar, characterised by a "fibre-like" aspect as seen in figure 4. However, the crack sometimes followed an interlamellar boundary, at high angle with respect to the notch plane and then, propagated again in a translamellar manner. Interlamellar decohesion is observed, mostly at high ΔK values. Figure 5 presents an evidence of interlamellar secondary cracking. It can be seen that microcracks formed along a lamellar interface, even far below the main crack surface and linked to the next interface by translamellar microcracking. For the notch "B" specimens, the "fibre-like" features are not the main features presented on the surface but they alternate with interlamellar features. The interlamellar microcracking process is closely related to the slip-bands mentioned above. It is noteworthy that during fracture testing of a lamellar TiAl alloy, Chan and Kim [11] have observed similar

slip lines. According to these authors, interlamellar slip is linked to the easy slip system on the (111) plane, parallel to the interface. Subsequent interface debonding or microcracking is linked to the respective crystallographic orientations at the plates interface. Using a micromechanical model, Chan [12] has shown that in this case the maximum stress is redistributed far ahead of the crack tip, inducing an enhanced damage process in this region.

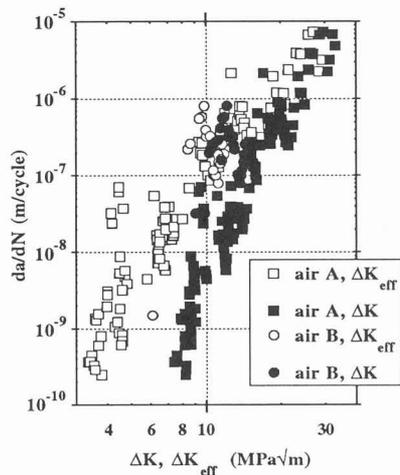


figure 1: da/dN vs. ΔK and ΔK_{eff} in air

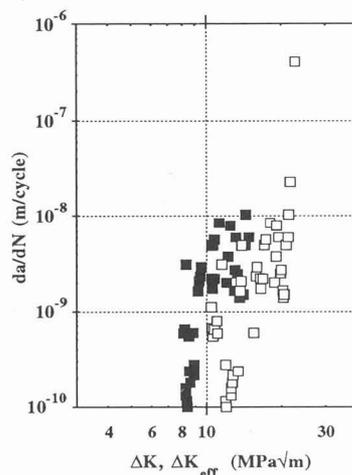


figure 2: da/dN vs. ΔK and ΔK_{eff} in vacuum (notch B)

These observations combined all together suggest that the fatigue damage is not confined to a small process zone at the crack tip. In particular, the observed diffuse damage process consisting in microcracking, crack branching, crack deflection, and crack bridging is expected to induce a significant crack-tip shielding, improving the crack growth resistance. However, shielding might also depend on the notch orientation. For example, in notch "B" orientation, lamellae boundary debonding is assumed to be less operative than in the notch "A" orientation. This is not clear when considering the effective curves in air and in vacuum, for the two orientations. It is rather suggested that the basic fatigue mechanism is fundamentally the same and is not deeply influenced by toughening mechanisms.

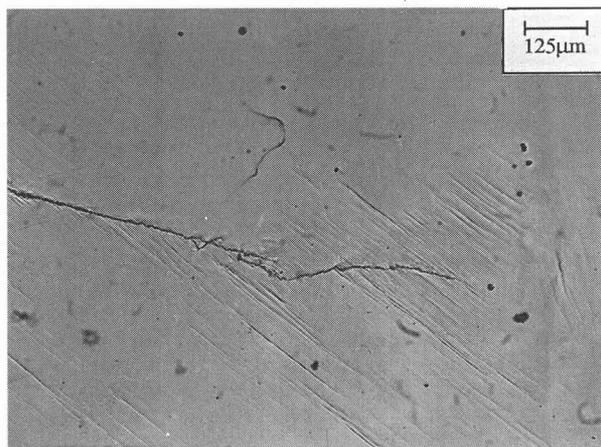


figure 3: Slip-band like features at the crack tip (notch "B", vacuum).

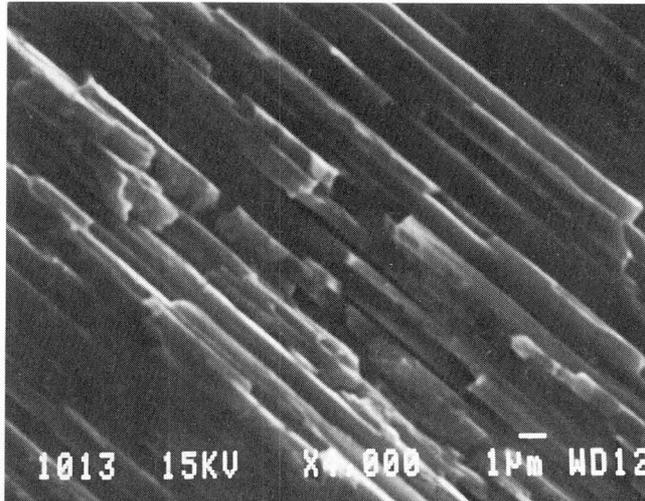


Figure 4: Fibre-like features (notch "A", air, $da/dN=10^{-7}$ m/cycle)

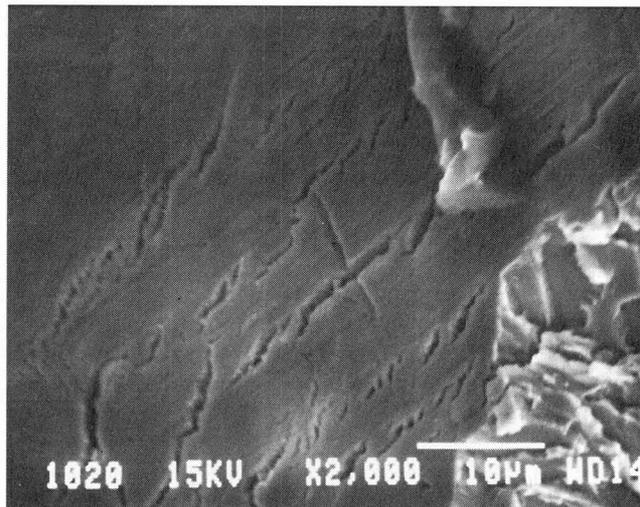


Figure 5: Interlamellar and translamellar microcracking (notch "A", air, $da/dN=10^{-9}$ m/cycle)

3.2. Crack closure

The comparison of nominal (i. e. da/dN vs. ΔK) and effective (i. e. da/dN vs. ΔK_{eff}) curves brings out a significant contribution of closure in the slow growth rate regime for both environments (figures 1&2). The validity of the crack closure measurements method has been demonstrated by previous testing carried out on another TiAl-based alloy with a nearly fully γ microstructure. FCGRs were determined in air for different R values. Actually, the minimum load was higher than the opening load, calculated on the basis of the closure measurements performed at $R=0.1$. These closure-free data provide an effective curve in perfect agreement with the effective curve obtained for $R=0.1$ after crack-closure corrections[13]. In these materials crack-closure is thought to result from surface mismatch and subsequent wedging[14].

The U ratios in air and in vacuum are presented in figures 6 and 7 respectively. In air closure progressively developed as the crack grew, inducing a FCGR decrease. The load amplitude was then periodically increased in order to complete the test. Conversely, the closure effects tends to disappear at higher growth rates. In vacuum, the U ratio decreased rapidly after initiation to stabilise at a value of $U=0,67$, even for the high growth rates. Data represented in figure 4 for testing in vacuum have been

collected for two different notch "B" specimens. Cracks have been initiated for two different ΔK values: $\Delta K=8.9\text{MPa}\sqrt{\text{m}}$ for the first one and $\Delta K=14.2\text{MPa}\sqrt{\text{m}}$ for the second one. The general behaviour is the same, but for the second sample, the minimum U value is reached at a much higher value of ΔK . As a consequence, if the two nominal curves (da/dN vs. ΔK) do not perfectly match the agreement between the two effective curves (da/dN vs. ΔK_{eff}) is fairly nice (figure 8). When the initial ΔK value is high enough, it enables to overcome the development of closure and prevent any slowing down of the propagation. These observations highlight the need to clearly define the precracking procedure which can influence the FCP. From a similar point of view, it is worth noticing that the differences exhibited on the nominal curves for the two different lamellae orientation tend to disappear when considering effective curves. (cf. figure 9). This suggests that the basic fatigue mechanisms is not strongly dependent on the lamellar orientation.

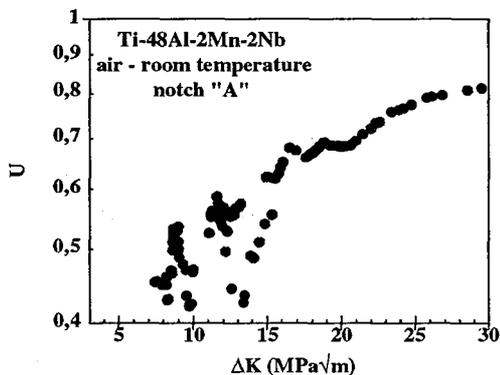


figure 6: Stress intensity factor range vs. U ratio for notch A orientation in air

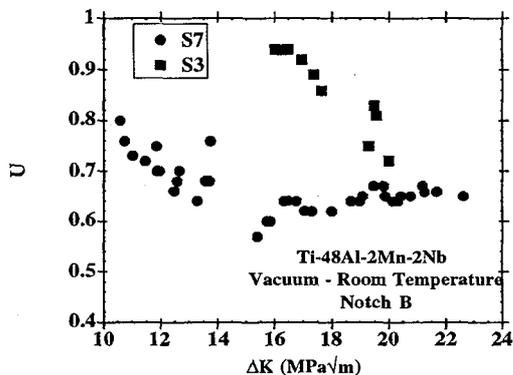


figure 7: Influence of precracking on the closure evolution in vacuum

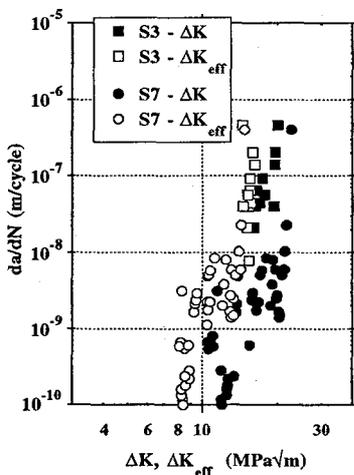


figure 8: Stress intensity factor range and effective stress intensity factor range vs. crack growth rate for two notch B specimens in vacuum

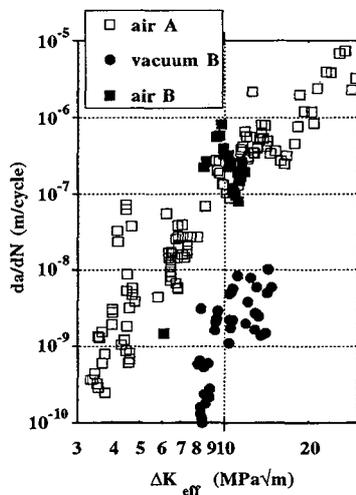


figure 9: Effective stress intensity factor range vs. crack growth rate in air and in vacuum for both orientation.

3.3. Environmental effects

The FCP behaviours exhibited in air and in vacuum are compared on figure 6, taking into account the crack-closure effects. The crack growth in air is enhanced by about two orders of magnitude for both orientations. The enhancement is of the same order in the case of a nearly fully γ microstructure[13]. As discussed in section 1 the slow FCP regime of conventional engineering alloys is known to be strongly influenced by a moist environment. Besides many intermetallics are prone to environmental

embrittlement[15-19]. This embrittlement is characterised by a loss of ductility without concomitant change on yield stress. In the case of aluminides it is suggested that this phenomenon results from a kind of hydrogen embrittlement. The sequential process would be as follows: surface reaction between water vapour and aluminium[15], hydrogen release and subsequent embrittlement.

However in the case of titanium aluminides titanium might be reactive as well. In addition the precise nature of the embrittling process is not clear. Nevertheless all these remarks suggest that FCP in titanium aluminides might be significantly enhanced by a moist environment. Balsone et al.[20] have noticed an influence of frequency on FCGR in presence of traces of moisture in the case of a Ti₃Al. Similarly Aswath and Suresh[21] have noticed an increase in FCGR when the water vapour pressure is increased. These results suggest that the FCP enhancement might be controlled by the exposure (pressure x time). It can be legitimately assumed that the same processes still applied for TiAl alloys. However on-going work will aim to assess this point.

4. CONCLUSION

The conclusions drawn out from the present investigation might be summarised as follows:

1. The fatigue crack growth is intermittent and results from a diffuse damage process encompassing the crack wake and a large region ahead of the crack tip.
2. Crack closure plays a significant role in fatigue crack propagation in air as in vacuum, especially at slow growth rates. In addition, the closure evolution seems closely related to experimental procedure, in particular crack initiation and precracking.
3. Fatigue cracks propagate much faster in air than in vacuum. This enhancement is assumed to be mainly due to a crack tip reaction with moisture present in the test environment.

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References

- [1]. Y.W. Kim, *Mater Sci Eng A-Struct Mater* Vol. **193** (1995), 519-533.
- [2]. S. C. Huang and J. C. Chesnutt, *Intermetallic Compounds: Vol.2, Practice* (edited by J. H. Westbrook and R. L. Fleisher, John Wiley & Sons Ltd, Vol. 1994) 73-90.
- [3]. W. Elber, presented at the Damage Tolerance in Aircraft Structures, ASTM STP 486, Toronto, Ontario, Canada Eds, American Society for Testing and Materials, Philadelphia, Pa., 1971) 230-242.
- [4]. J. de Fouquet, J. Petit, and G. Hénaff, *Handbook of Fatigue Crack Propagation in Metallic Structures* (edited by Andrea Carpinteri, Elsevier Science B. V., Vol. 2, 1994) 1159-1203.
- [5]. G. Hénaff, K. Marchal, and J. Petit, *Acta Metall et Mater* Vol. **43 8** (1995), 2931-2942.
- [6]. G. Hénaff and J. Petit, presented at the Corrosion-Deformation Interactions International Conference, Fontainebleau, France (T. Magnin et al. Eds, les éditions de physique, 1992) 599-617.
- [7]. G. Hénaff, B. Bittar, C. Mabru, J. Petit, and P. Bowen, presented at the Seventh International Conference on Mechanical Behaviour of Materials ICM 7, The Hague, The Netherlands Eds, 1995).
- [8]. P. Bowen, R.A. Chave, and A.W. James, *Mater Sci Eng A-Struct Mater* Vol. **193** (1995), 443-456.
- [9]. K. S. Chan, *Jom-J Min Metall Mater Soc* Vol. (1992), 30-38.
- [10]. D. L. Davidson and J. B. Campbell, *Metall. Trans.* Vol. **24A July** (1993), 1555-1574.
- [11]. K.S. Chan and Y.W. Kim, *Metall Mater Trans A* Vol. **25 6** (1994), 1217-1228.
- [12]. K. S. Chan, M. Y. He, and J. W. Hutchinson, *Mater Sci Eng* Vol. **A167** (1993), 57-64.
- [13]. G. Hénaff, S. A. Cohen, C. Mabru, and J. Petit, presented at the Micromechanics of Advanced Materials, TMS fall meeting, Cleveland, Ohio Eds, 1995).
- [14]. R. Gnanamoorthy, Y. Mutoh, N. Masahashi, Y. Mizuhara, and M. Matsuo, *J Mater Sci* Vol. **29 19** (1994), 5199-5206.
- [15]. C. T. Liu and M. Takeyama, *Scripta Metall. Mater.* Vol. **24 8** (1990), 1583-1586.
- [16]. C.T. Liu, C.L. Fu, E.P. George, and G.S. Painter, *Isij Int*
Isij International Vol. **31 10** (1991), 1192-1200.
- [17]. C.T. Liu and W.C. Oliver, *Scr Metall Mater*
Scripta Metallurgica et Materialia Vol. **25 8** (1991), 1933-1937.
- [18]. N.S. Stoloff and C.T. Liu, *Intermetallics* Vol. **2 2** (1994), 75-87.
- [19]. E.P. George, C.T. Liu, H. Lin, and D.P. Pope, *Mater Sci Eng A-Struct Mater* Vol. **193** (1995), 277-288.
- [20]. S. J. Balsone, D. C. Maxwell, M. Khobaib, and T. Nicholas, presented at the Fatigue'90, Honolulu, Hawaii (K. Kitagawa and T. Tanaka Eds, MCEP, Birmingham, 1990) 1173-1178.
- [21]. P. B. Aswath and S. Suresh, *Metall. Trans.* Vol. **22A April** (1991), 817-828.