

INFLUENCE OF MOIST ATMOSPHERES ON FATIGUE CRACK PROPAGATION.

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INTRODUCTION

During the last decades, fatigue crack propagation (FCP) has been widely investigated, specially in the near-threshold area, where environment, microstructure and closure have been shown to play a major role (Davidson and Suresh, 1984 ; Petit *et al.*, 1994). Such pronounced interest for the near-threshold crack propagation behaviour is two fold. From an engineering point of view there has been a need of reliable and reproducible data in order to assess the damage tolerance of components which have to withstand a high number of low-amplitude loading cycles. On the fundamental side, the basic mechanisms governing FCP in this region were ill-known in sharp contrast with the so-called Paris regime formally extensively investigated.

Following the initial work of Dahlberg (1965), Hartman (1965) and Bradshaw and Wheeler (1966) and (1969), the deleterious effect of ambient air as compared to an inert environment like high vacuum, has been clearly related to the presence of moisture in the surrounding environment for most of the metallic materials fatigued at room temperature (Lynch, 1978 ; Simmons *et al.*, 1978 ; Wei, 1979 ; Petit, 1983 ; Pao *et al.*, 1988 ; Lynch, 1988 ; Gudladt and Petit, 1991 ; Henaff *et al.*, 1995 ; Petit and Mendez, 1996 ; Enochs and Devereux, 1975 ; Piasiek and Gangloff, 1991). At higher temperatures the respective role of water vapour and oxygen is more disputed (Petit *et al.*, 1992).

The main difficulty encountered to understand the role of water vapour resides in the complex interactions of an active environment with other parameters which influence the propagation, including intrinsic parameters as alloy composition and microstructure or extrinsic parameters as loading conditions, specimen geometry, crack depth, crack closure and temperature. This paper proposes a brief survey of the influence of gaseous moist environments on fatigue crack propagation at mid and low rates, on the basis of a framework describing the intrinsic fatigue crack propagation which is essential to uncouple the respective influence of microstructure, closure, environment, and to analyse their interactions.

INTRINSIC FATIGUE CRACK PROPAGATION.

The characterisation of the intrinsic behaviour of fatigue cracks means the elimination of any environmental and closure effect. High vacuum ($\sim 10^{-4}$ Pa) for tests run at frequencies of about 20 to 50 Hz, can be considered as an inert environment in most cases. Closure correction can be made using the compliance method as initially suggested by Elber (1971) and Kikukawa *et al.* (1982).

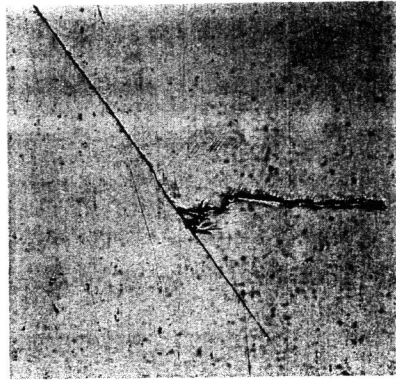


Figure 1. Stage I to stage II transition in high-purity Al - 4.5% Zn - 1.25 wt % Mg single crystal in overaged condition and oriented for single slip. High vacuum, 35 Hz, R = 0.1. The transition was obtained at a ΔK_{eff} range of $4 \text{ MPa}\sqrt{\text{m}}$.

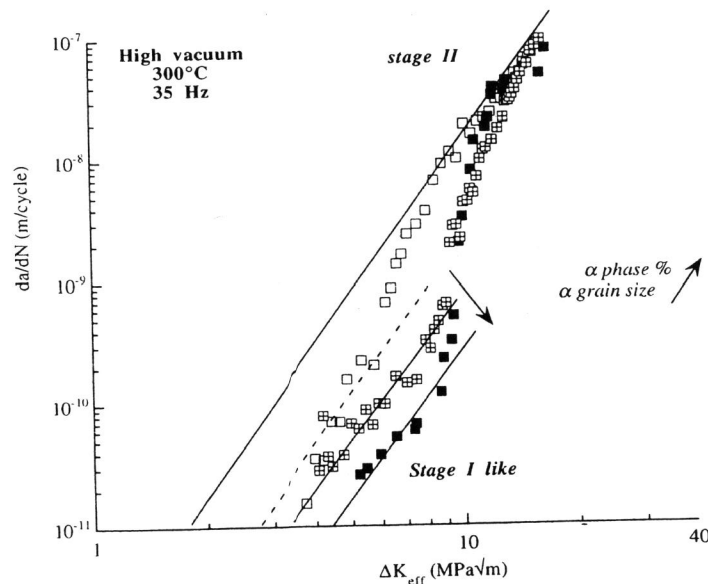


Figure 2. Influence of microstructure on the intrinsic near-threshold fatigue crack propagation of Ti-6Al-4V alloy. High vacuum, 35 Hz, R = 0.1.
 □ 40 % bimodal (grain size $\varnothing \sim 20 \mu\text{m}$), ■ globular ($\varnothing \sim 8 \mu\text{m}$) + elongated α grain ($\varnothing \sim 50 \mu\text{m}$), ⊠ 75 % α_p globular ($\varnothing \sim 8 \mu\text{m}$)

On the basis of numerous experimental data obtained on various Aluminium alloys (including technical alloys and high purity single crystals), steels and Titanium alloys, the intrinsic FCP has been analysed in accordance to three basic crack propagation regimes (Petit *et al.*, 1994) :

– The fastest intrinsic stage I, has been identified for long cracks on single crystals of high purity Al-Zn-Mg alloy figure 1. Typically, the crack develops within a $\{111\}$ planes pre-oriented for single slip (Petit, 1992 ; Petit *et al.*, 1992). This regime is also observed on various materials in the early growth of microstructural short cracks (de Los Rios, 1985).

– The slowest regime, or intrinsic stage I-like propagation corresponds to a crystallographic crack growth observed near the threshold in polycrystals or in the early stage of growth of naturally initiated microcracks when ageing conditions or low stacking fault energy generate heterogeneous deformation along single slip systems, within individual grains (Starke *et al.*, 1984 ; Selines, 1971). Lamellar microstructures can also facilitates the stage I like regime as observed in Ti alloys (Lesterlin *et al.*, 1996) and as illustrated in figure 2. Crack branching, crack deviation (Suresh, 1985) and barrier effect of grain boundaries (de Los Rios *et al.*, 1985) are assumed to lower the stress intensity at the crack tip of the main crack. This regime and the associated retardation effect are highly sensitive to the microstructure.

– The intermediate intrinsic stage II is commonly observed on polycrystals and single crystals when crack propagation proceeds at macroscopic scale along planes normal to the loading direction. Such propagation is favoured by microstructures which promote homogenous deformation and wavy slip as large or non coherent precipitates or small grains size.

The stage II regime is in accordance with a propagation law derived by Petit *et al.* (1991) from the models initially proposed by Rice (1965) and Weertman (1966) :

$$\frac{da}{dN} = \frac{A}{D_0^*} \left(\frac{\Delta K_{eff}}{E} \right)^4 \quad (1)$$

where A is a dimensionless parameter, E the Young modulus and D_0^* the critical cumulated displacement leading to rupture over a crack increment ahead of the crack tip.

Intrinsic data for well identified stage II propagation are plotted in figure 3 in a da/dN vs ΔK_{eff} diagram for a wide selection of Al alloys, and in figure 4 in a da/dN vs $\Delta K_{eff}/E$ diagram for a selection of steels and some data for a TA6V alloy in comparison to the mean curve for Al alloys. The diagrams constitute an excellent validation of the above relation and confirm that the LEMF concept is very well adapted to describe the intrinsic growth of a stage II crack which clearly appears to be nearly independent on the alloy composition, the microstructure (when it does not introduce a change in the deformation mechanism), the grain size, and hence the yield stress. The predominant factor is the Young modulus of the matrix, and the slight differences existing between the three base metals can be interpreted as some limited change in D_0^* according to the alloy ductility (Petit and Henaff, 1991).

Finally most of the differences observed in the intrinsic regime of long cracks are related to the transition from stage II to stage I-like and to the enhancement of the retardation effect of microstructure in the near threshold region. A good illustration is given in figure 2. For the TA6V alloy, the stage II regime governing the propagation in the mid rate range appears to be independent on the microstructure while the crystallographic stage I-like regime is highly sensitive to the grain size and to the α volumic fraction.

In conclusion of this part of the paper the intrinsic behaviour of a fatigue crack can be in any case described on the basis of the framework of the three main regimes. For example, the intrinsic behaviour of naturally initiated microcracks can be analysed on such basis (Petit *et al.*, 1996). A microcrack initiated at the surface of a specimen of a 7075 type alloy in T651 peak aged condition, grows in the stage I regime in the first grain. Such propagation is favoured by GP and S' shearable precipitates which promote the localisation of the deformation within PSB's (Petit, 1992). When the crack has crossed several grains, the stage I-like propagation regime is encountered giving highly retarded rates.

– at growth rates lower than $(da/dN)_{cr}$, the Hydrogen assisted crack growth mechanism becomes operative. Hydrogen being provided by adsorbed water vapour when some critical conditions are fulfilled. Both mechanisms will be detailed in the following.

Adsorption assisted propagation

Results recently obtained under very low water vapour partial pressure (10^{-3} Pa) and under closure-free conditions in the case of a high-strength low-alloy steel (Henaff *et al.*, 1995) clearly demonstrated that this residual moisture content is sufficient to induce a significant increase in crack growth rates especially when the load frequency is decreased (fig. 8). In addition these results provide insight about the kinetics of this adsorption process. The curve obtained at low frequency (0.2 Hz) exhibits a linear relationship on the explored rate range with a Paris law exponent $m = 4$ like the intrinsic stage II regime (Petit and Henaff, 1991) as shown on figure 8. This regime is characterised by a fourfold increase in crack growth rates as observed in air in the upper range. At intermediate frequencies the behaviour first follows this regime and then, above a critical value of da/dN tends to recover the intrinsic behaviour.

A model has been proposed to provide a comprehensive picture of the processes involved in the behaviour (Henaff *et al.*, 1995). The approach developed here is actually extremely close to the problem formerly addressed by Orowan (1944) and the general framework is based on the concept of Langmuir isotherms (1918).

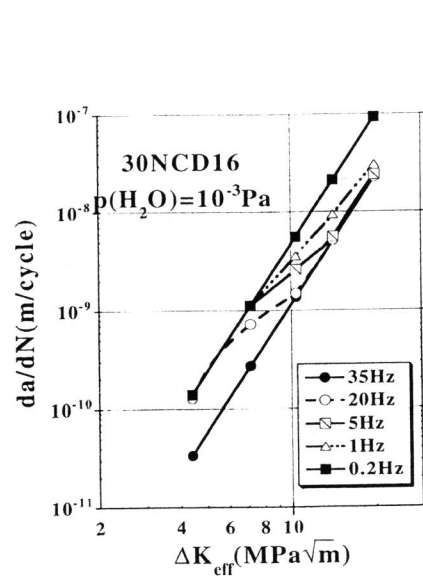


Figure 8 - Influence of test frequency on the propagation of a fatigue crack in high vacuum on a high strength steel $R = 0.7$

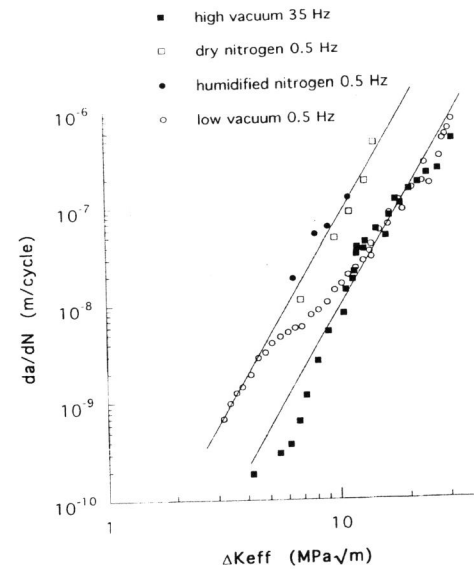


Figure 9 - Influence of test frequency and water vapour partial pressure on the fatigue crack propagation behaviour in a TA6V alloy at 300°C. Characterization of an adsorption assisted regime induced by low pressures of water vapour.

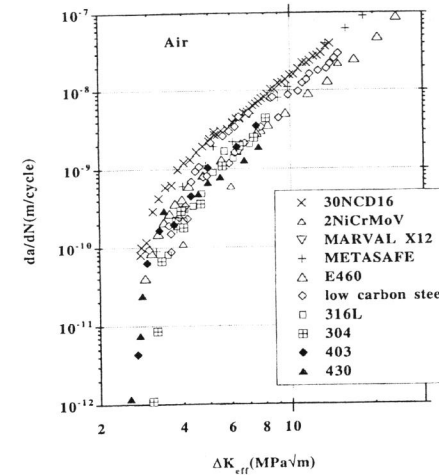


Figure 5- Effective crack propagation of steels tested in ambient air at frequencies of 20 to 50 Hz.

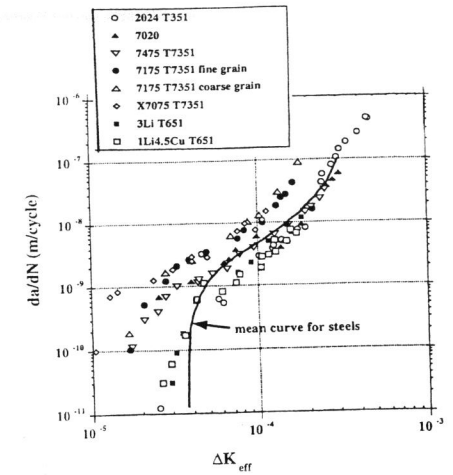


Figure 6- Effective crack propagation of Al alloys tested in ambient air at frequencies of 20 to 50 Hz.

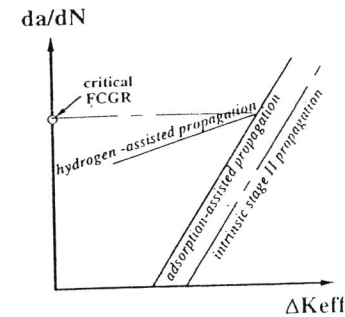


Figure 7 - Schematic illustration of environmentally assisted stage II fatigue crack growth mechanisms.

On the basis of experimental data obtained on Aluminium alloys and steels in air, high vacuum and purified nitrogen containing traces of water vapour, a comprehensive model has been established by J. Petit et al. including these two different mechanisms for environmentally assisted crack growth (Petit, 1983 ; Petit *et al.*, 1994 ; Henaff *et al.*, 1995) as schematically illustrated in Figure 7 :

– at growth rates higher than a critical rate $(da/dN)_{cr}$, which depends upon several factors as surrounding partial pressure of water vapour, load ratio, test frequency, chemical composition and microstructure, the crack growth mechanism is assisted by water vapour adsorption but it is still controlled by plasticity as in vacuum.

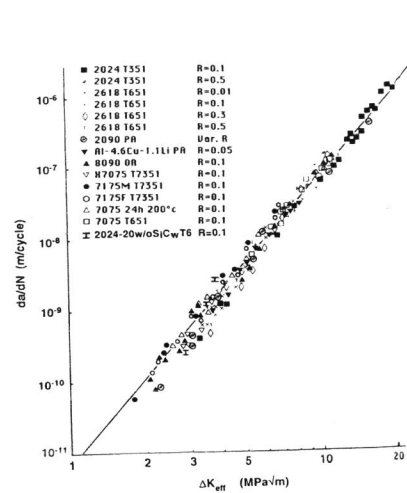


Figure 3 - Intrinsic stage II propagation for a range of Aluminium based alloys. Tests performed in high vacuum at frequencies of 20 to 50 Hz.

For larger crack extend and increased ΔK range, the propagation switches to the intermediate stage II regime. So, when the relation of the crack propagation with respect to the microstructure is well established, the LEFM concept, i.e. the ΔK concept, can be applied as well for short cracks, as for long cracks after correction for closure.

ENVIRONMENTALLY-ASSISTED FATIGUE CRACK PROPAGATION.

3.1. Identification of the governing mechanisms.

Following the rationalization of the intrinsic stage II propagation as presented above, some similar rationalization of FCP in air could be expected after correction for crack closure and temperature effects ($\Delta K_{eff}/E$). Figures 5 and 6 present a compilation of stage II propagation data obtained in ambient air for almost the same alloys as in vacuum. Obviously any rationalization does not exist in air. The sensitivity to air environment is shown strongly dependent as well on base metals, addition elements, and microstructures (see 7075 alloy in three different conditions) as on R ratio and growth rate. However a common critical rate range at about 10^{-8} m/cycle can be pointed out for all materials. This critical step is associated to stress intensity factor ranges at which the plastic zone size at the crack tip is of the same order as grain or sub-grain diameters. In addition there is a general agreement to consider that, for growth rates lower than this critical range, crack propagation results from a step-by-step advance mechanism instead of a cycle-by-cycle progression as generally observed in the Paris regime in air (Davidson and Lankford, 1981).

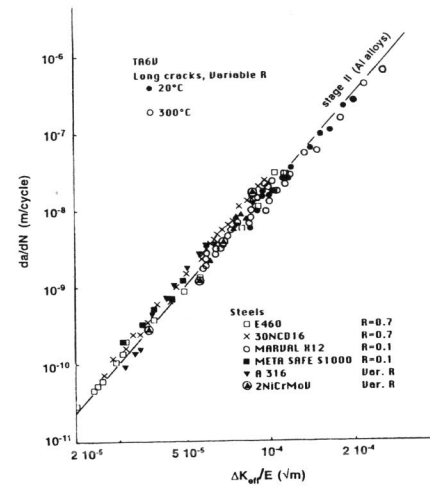


Figure 4 - Intrinsic stage II propagation for a selection of steels and some data on TA6V alloys. The mean curves for Al alloys is reported from figure 3.

The basic propagation mechanism might be described by a relationship similar to Eq. [1] :

$$\frac{da}{dN} = \frac{A}{D^*} \left(\frac{\Delta K_{eff}}{E} \right)^4 \quad (2)$$

The critical displacement D^* is thus closely related to the surface coverage rate θ by the following equation :

$$D^* = \frac{D_0' D_1'}{(D_1' + \theta (D_0' - D_1'))} \quad (3)$$

The computations show a nice agreement with experimental data (Henaff *et al.*, 1995). This model has also been proved to succeed in explaining results obtained under gaseous atmospheres by Liaw *et al.* (1982). It also reveals successful in accounting for the adsorption-assisted propagation at elevated temperature in the case of titanium alloy (figure 9) throughout a broad spectrum of frequency and water vapour partial pressure (Lesterlin, 1996). However in this case it was assumed that equilibrium does no longer correspond with a saturating coverage rate, that means that θ is lower than 1 due to a promoted influence of desorption as temperature raises. A compendium of data from ancillary testings and from literature proves that the adsorption-assisted regime might be encountered for a wide variety of materials and environmental conditions (J. Petit *et al.*, 1996).

Hydrogen-assisted fatigue crack propagation

Wei and co-workers (Wei, 1979 ; Pao *et al.*, 1988) have proposed a sequential process for environmentally-assisted fatigue crack propagation in an hydrogenous atmosphere. The sequential events would be :

- transport of the active species to the crack tip
- physical adsorption
- chemical adsorption and dissociation
- hydrogen penetration
- diffusion and embrittling reaction.

The same process is considered to describe environmentally assisted fatigue crack propagation. Hydrogen production is assumed to result from the dissociative chemical adsorption of molecules after physical adsorption on surfaces freshly created at the crack tip. Hydrogen then diffuses into the process zone where the embrittling reaction takes place at some specific but ill-identified sites. However most of the experimental results support the existence of such a regime in moist atmosphere only near threshold conditions and for growth rates lower than a critical rate as initially proposed by Achter (1968). In the Paris regime and at room temperature, in most cases, the adsorption assisted propagation is generally observed (Petit *et al.*, 1994 ; Henaff *et al.*, 1995). Indeed the Hydrogen assisted regime becomes operative only when several conditions favouring high Hydrogen concentration into the process zone at the crack tip are fulfilled :

- conditions of access to the crack tip for active species which lead to sufficient partial pressure of water vapour to create an instantaneous adsorbed monolayer : surrounding pressure, frequency, growth rate, R ratio ; in such conditions, the mechanism is reaction-controlled.
- sufficiently low stress intensity factor to reach a regime with a stationary crack and localized plastic deformation in a limited number of slip systems within a single grain at the crack front ;
- a long time enough to allow Hydrogen to diffuse by dislocation dragging so as to attain a critical Hydrogen concentration for metal embrittlement.

Such conditions are encountered in ambient air or in humidified inert gas for growth rates lower than the a critical rate $(da/dN)_{cr}$ (Fig. 10) which can be derived from the expression proposed by Bowles (1978):

$$\left(\frac{da}{dN}\right)_{crit} \text{ (mm/cycle)} = 3.6 \times 10^{-3} \frac{P}{f\sqrt{MT}} \quad (3)$$

However this expression does not take into account the crack impedance effect highlighted by Bradshaw (1967).

CONSTITUTIVE FATIGUE CRACK PROPAGATION LAW.

Up to now no fatigue crack growth law taking into account the different processes exposed above is available. Theoretical models such as Weertman's model do not account for environmental effects and therefore they are merely valid for fatigue crack propagation in inert atmospheres. Similarly fatigue crack growth laws considering the strained material at the crack tip as a low-cycle fatigue microsample (Lanteigne and Bailon, 1981); Roven and Nes, 1991) indirectly integrates environmental effects mainly through the use of a Manson-Coffin law.

The occurrence of hydrogen-assisted crack growth is associated to a typical change in the slope of the propagation curves which becomes close to 2 to 1 at low rates, and the transition from one regime (adsorption assisted) to the other (Hydrogen-assisted) often corresponds to a more or less well marked plateau range (Fig. 5-7):

The present authors have proposed the following relationship based on a superposition principle:

$$\frac{da}{dN} = \frac{A}{D_1^*} \left(\frac{\Delta K_{eff}}{\mu}\right)^4 + \frac{B}{\sigma\mu} (\Delta K_{eff}^2 - \Delta K_{eff,th}^2) \quad (4)$$

where $\Delta K_{eff,th}$ denotes the threshold obtained under closure-free conditions, B is a dimensionless constant and σ a strength parameter.

The first term accounts for the adsorption-saturating regime previously described. The second one was subsequently added in an attempt to describe the Hydrogen-assisted propagation regime. The ΔK_{eff}^2 dependence might be viewed as a coarse description of the dislocation dragging via the CTOD. $\Delta K_{eff,th}$ would thus denote a threshold value of this sweep-in mechanism to enable the attainment of a critical hydrogen concentration at the crack tip. However, up to now, one needs to experimentally determine this value. However such a formulation for Hydrogen-assisted propagation is still highly empirical. The problems raise from the lack of a sound understanding of what happens ahead of the crack tip. Some critical issues are require for a detailed knowledge to achieve this goal. Finally the temperature dependence of these phenomena obviously constitutes a prime issue to investigate.

INFLUENCE OF HIGH TEMPERATURE

It is generally admitted that the effect of air environment at elevated temperature results from oxidation. The role of oxides in fatigue crack propagation has been, for example, recently reviewed by J.E. King and P.J. Cotteril (1990). The role of oxide built up and subsequent oxide wedging of crack in the near threshold area has been widely demonstrated (Suresh *et al.*, 1981; Liaw *et al.*, 1982). But the relation between oxide formation observed on the crack surfaces of postmortem specimens and the crack growth rates in the mid-range, is mostly inferred, and the active role of oxygen at the very crack tip during the cycling process is not really demonstrated. Moreover recent experiments on Titanium alloys have clearly demonstrated a predominant detrimental effect of water vapour even at very low partial pressure and at temperatures ranging up to 500°C (Lesterlin *et al.*, 1996; Sarrazin-Baudoux *et al.*, 1996; Petit *et al.*, 1996).

Similar experiments were performed at 500°C. The crack propagation behaviour was determined in air, high vacuum, low vacuum (10^{-3} Pa) and in humidified Argon on a Ti 6246 alloy with a fine Widmanstätten microstructure (Lesterlin, 1996). In figure 10, the growth rates in moist

argon are shown to be faster than in air with same partial pressure of water vapour (1.3 KPa). Indeed, the presence of oxygen appears to be more inhibiting than active. But oxide built-up explains for both environments the enhanced contribution of closure in the near threshold area, the thickness of the oxide film on the cracked surfaces being evaluated by mean of RBS technique at about 100 nm which is of the same order as the $\Delta CTOD$ near the threshold. Lowering the frequency down to 0.1 Hz induces a sharp acceleration as well in air as in moist argon with still faster growth rates in the latter environment (Figure 11). These experiments conducted in the mid-rate range and in conditions without closure or with closure correction, support a great enhancement of the effect of water vapour at low frequency.

As a creep-fatigue process could be operative, cyclic loadings with long hold time were applied. These experiments clearly demonstrate the existence of a saturation of the influence of environment since increased holding times have no additional effect (Petit *et al.*, 1996; Sarrazin-Baudoux *et al.*, 1996). These observations substantiate the existence of a corrosion fatigue mechanism due to the presence of water vapour and are in contradiction with a time-dependent creep fatigue process. Careful observations and analysis of surface oxidation by mean of the RBS technique (Lesterlin, 1996) support the absence of correlation between surface oxidation and crack growth rates.

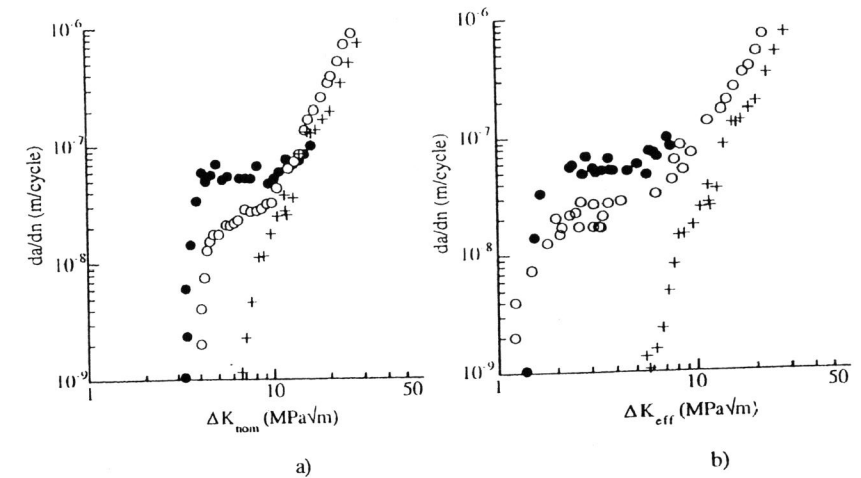


Figure 10. Fatigue crack propagation behaviour of Ti6246 at 500°C in different environmental conditions: a) nominal propagation curves ($R = 0.1$, 35 Hz). b) Effective data after closure correction (or from tests at constant K_{max} without closure)

Very recent tests were conducted in low vacuum (10^{-2} Pa) (Lesterlin, 1996). Mass spectrometer analysis showed a predominant amount of water vapour (80 %) and a very small residual pressure of oxygen ($< 10^{-4}$ Pa). Crack growth rates determined at 10^{-2} Hz in comparison to 35 Hz are shown to induced acceleration of more than one order of magnitude at a ΔK range of $6\text{MPa}\sqrt{\text{m}}$ leading to rates like as in air (figure 11). These last experiments demonstrate the major role of water vapour. As plotted in figure 11, the experimental data provided by different testing

conditions and the literature support the existence of a quite well defined corrosion assisted regime described by a relation based on the crack tip opening displacement as :

$$\frac{da}{dN} = \frac{1}{2\pi} \left(\frac{\Delta K^2}{\sigma_y E} \right) \quad (5)$$

and experimental data obtained in air at 593°C on a Ti 1100 alloy by reported from Ghonem and co-authors (Ghonem and Foerch, 1991 ; Foerch et al., 1992) between the above presented results.

A good agreement is obtained supporting the existence of a similar FCP corrosion assisted mechanism in the range 500°C - 600°C. However these authors evoke mainly a creep-fatigue controlling mechanism. The exploration of the range 500°C to 600°C would give more precise informations on the respective role of water vapour and oxygen. Similar water vapour assisted crack propagation has to be expected on all metallic alloys including steels and Ni-based alloys in some specific temperature ranges.

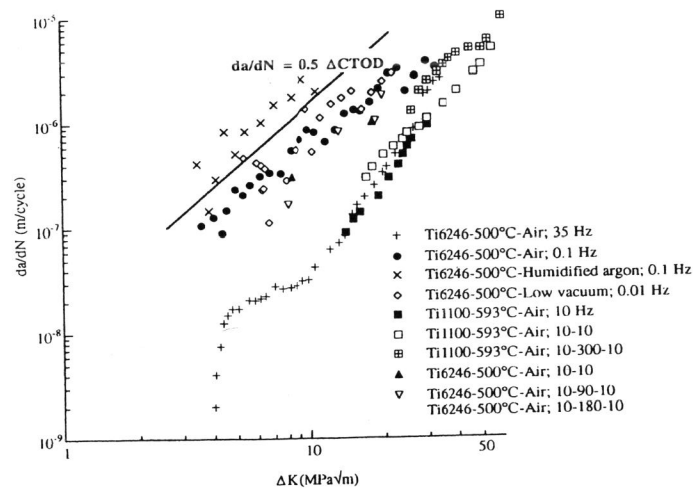


Figure 11. Comparison of FCP at 500°C in Ti 6246 and at 593°C on Ti 1100 (20, 21).

CONCLUSIONS

On the basis of numerous experimental data obtained in high vacuum and with closure correction, on Al alloys with various ageing conditions, steels and Titanium alloys, three intrinsic crack growth regime have been identified :

- i) The faster stage I, identified on Al based single crystals, is also observed in the early growth of short cracks.
- ii) The intermediate stage II is commonly observed when the crack advance proceeds along planes normal to the load axis.
- iii) The slower regime called stage I-like corresponds to a crystallographic crack growth when the deformation is localised within single slip system in each individual grains. This regime prevails near threshold conditions.

The behaviour in moist environment of these materials has been described by superimposing two distinct processes :

- i) Adsorption of water vapour molecules which promotes the growth process without altering the basis intrinsic mechanism of damage accumulation.
- ii) Hydrogen-assisted propagation as initially described by Wei and co-authors, generally operative at rates below a critical range.

At 500 °C in Ti 6246 alloy, a predominant effect of water vapour has been demonstrated while the role of oxygen appears limited to oxidation of cracked surfaces which can hamper the detrimental action of water vapour and favors near-threshold closure. A similar behaviour can be expected on Ti alloys and most of the metallic alloys.

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