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CORROSION FATIGUE OF AIRCRAFT MATERIALS, (U)  
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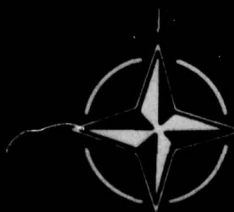
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## Corrosion Fatigue of Aircraft Materials

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Four papers presented at the 44th Meeting of the Structures and Materials Panel of AGARD,  
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## PREFACE

Environmental effects on the fatigue of aircraft materials have been neglected in the past. Damage behaviour of aircraft structures was analysed by fatigue and fracture mechanics without paying much attention to the environment.

A few years ago, some laboratories of NATO countries started to investigate corrosion fatigue. A proposal was made to the Structures and Materials Panel for new activities in corrosion fatigue with the intention of presenting the state of the art in this field of interdisciplinary R & D of corrosion and fatigue engineers.

The SMP approved the presentation of four pilot papers on the corrosion fatigue of high strength aluminium, titanium and steel alloys at the 44th Panel Meeting in April 1977. The four papers are of interest to materials and structural engineers and give detailed information on experimental results from four laboratories together with recommendations of areas for future research.

The discussion of the pilot papers by members of an ad hoc Group led to the constitution of a Sub-Committee on Corrosion Fatigue. In addition, a cooperative testing programme on the corrosion fatigue of coated joints was initiated and will be discussed at a future Specialists' Meeting of the Panel.

Wolfgang BUNK  
Chairman, Materials Group

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## CORROSION FATIGUE OF ALUMINIUM AND TITANIUM ALLOYS AND STEELS

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## SUMMARY

Literature on the effects of various gaseous and aqueous environments on the fatigue life and fatigue crack growth behaviour of aluminium and titanium alloys and steels is reviewed. The degree of reduction in life, and enhancement in crack growth rate, is presented, and the efficiency of the protective schemes: surface treatments, applied potentials and control of solution chemistry, are detailed. It is shown that whereas all three protection schemes can be effective to varying degrees in fatigue life situations, the use of applied potentials and solution chemistry control must be approached with caution in crack growth situations. The importance of surface reactions is brought out in the review, and some areas for future work are suggested. It is concluded that the degree of control over corrosion fatigue is at present limited, and awaits future multidisciplinary approaches to elucidate mechanisms in more detail.

## INTRODUCTION

Corrosion fatigue is said to occur when the conjoint action of corrosion acting during fatigue produces considerably more damage than either acting alone. The materials most likely to be at risk are the high strength aerospace alloys - aluminium and titanium alloys and steels - and in this review we shall consider some of the significant results for these important structural materials, illustrated wherever possible by work carried out in the UK. Within the space available this cannot be an exhaustive review, rather it attempts to outline the ways in which fatigue properties are degraded by aggressive environments, and some of the methods of alleviating the problem.

Points of clarification are made first, concerning the aggressiveness of various environments, alloy characterization and the fatigue process itself, and this is followed by a presentation of results on corrosion fatigue life (and crack initiation) and corrosion fatigue crack growth for the three alloy systems. Constant amplitude testing only is considered here; the paper by Wanhill in this report covers variable amplitude testing, at least for aluminium and titanium alloys.

## AGGRESSIVENESS OF ENVIRONMENTS

The aggressiveness of any environment must always be relative, and much discussion has taken place on the type of environment which should be used as a datum(1). It is generally agreed that vacuum (1.3 m-1.3  $\mu$ Pa) or a dry inert gas such as argon or nitrogen are the most suitable, but there is disagreement on which is the better:

- the use of vacuum has been criticized since rewelding may occur and therefore cause crack retardation(2). This is particularly important in crack growth, and evidence has recently been presented to show that rewelding can occur(3) (and retard growth) albeit after part cycling in compression (zero mean loading;  $R = -1 = \frac{\text{min stress}}{\text{max stress}}$ ).
- the use of dry argon or nitrogen has been criticized since it is difficult to be certain that the gas is dry(4) - the water vapour level should be below 10 ppm(1). The presence of other impurity gases may also be important(4).

Bearing these points in mind, and remembering that a reference environment must neither enhance nor retard fatigue behaviour, ultra high purity dry argon or nitrogen ( $H_2O < 10$  ppm) would thus seem to be the preferred reference. (Consideration could perhaps also be given to an inert oil, which while excluding the environment, does not retard crack growth by any wedging action(5).)

If we now consider the relative aggressiveness of environments, there is little disagreement on ranking - these are usually in the order: vacuum or dry gas; laboratory air; wet air; distilled water; sodium chloride solution. These will be the environments considered here, although special cases, such as hydrogen gas, will be mentioned briefly.

## ALLOY CHARACTERIZATION

Alloys can be characterized by the parameters: composition and microstructure, and the effects of these variables on corrosion fatigue behaviour are detailed in subsequent sections. For complete alloy characterization, however, a third structural parameter, namely texture, should also be considered. While it could be argued that texture is not of primary significance in cubic materials because of the extensive nature of their slip modes (although in the author's view such an attitude is fallible), the importance of texture in  $\alpha$  based hcp titanium alloys cannot be stressed too strongly(6). (There is very little data available on the fatigue of metastable  $\beta$ -bcc titanium alloys, and these are not considered here.). As an example, Fig 1 shows the type of property variations due to crystallographic anisotropy in Ti-6Al-4V, for the same microstructure. The crystal plane most sensitive to the environment is (0002), and while only a few investigations have been reported in recent years on the effects of texture on corrosion fatigue, it

has been known for some time that in stress corrosion tests a very significant decrease in  $K_{ISCC}$  can occur when fracturing on the basal plane (Fig 2). Note that  $K_{IC}$  is almost the same for all three orientations and  $K_{IC} \sim K_{ISCC}$  for fracture normal to (0002).

#### THE FATIGUE PROCESS

The effect of corrosion on the fatigue process can be studied in a number of ways - either by measuring fatigue life (as in much of the earlier work), or by considering the various stages in the fatigue process(8):

- crack initiation
- stage I growth
- stage II growth.

Results based on total life are difficult to interpret in terms of these individual stages, and in recent years, therefore, there has been a move to confine studies to one particular aspect of fatigue.

Until very recently the only information available on environmental effects on stage I growth was the work of Duquette and Gell(9) on a nickel alloy single crystal (MAR M200), where oxygen adsorption at the crack tip at low stresses promoted cleavage and reduced life compared to that in vacuum. However, data is now available on aluminium alloy single crystals(10-12), where moisture was observed to produce similar crack acceleration to that observed in the nickel alloy.

Stage II growth in different environments, on the other hand, has been widely studied. This has been assisted by two factors:

i the use of fractography, which has shown that fatigue crack growth is by striation formation, thus providing a one for one record of load applications (each striation is produced by a single load cycle, but the converse is not always true). The effect of aggressive environments is to change the normally ductile striations to brittle ones(8), thus producing enhanced growth (Fig 3). The formation of these brittle striations has been discussed at some length, and it is possible that a cleavage component contributes to such striation formation in steels (on 001) and titanium alloys (on 0002) because dislocation mobility can be exceeded by the crack velocity(13). But in the case of aluminium alloys, fracture (probably on 001) must, it would seem, be assisted by some lowering of the surface energy(13).

In addition, if there is a change in fracture mode (eg hydrogen induced embrittlement) additional evidence, of intergranular(14) or quasi-cleavage(14,15) failure, can be found in steels; and in some of these cases it may be difficult, due to the absence of striations, to differentiate between corrosion fatigue and stress corrosion(14);

and

ii the application of fracture mechanics, and in particular the analysis of crack growth behaviour in terms of the Paris(16) equation:  $da/dN = C(\Delta K)^m$ , where  $da/dN$  is the growth rate,  $\Delta K$  the stress intensity factor range =  $K_{max} - K_{min}$ , and  $C$  and  $m$  are constants. This has enabled quantification of crack growth behaviour, thus making data comparison much easier.

A complication which arises during stage II crack growth in corrosive environments is that the growth rate at a particular stress intensity must always be viewed relative to the threshold stress intensity for stress corrosion cracking ( $K_{ISCC}$ ). This contribution of stress corrosion to corrosion fatigue can arise either as the cyclic frequency is decreased, or as the mean stress is increased, particularly if the alternating stress amplitude is small. In fact, it is convenient to consider fatigue crack growth in three broad categories(17):

a. pure fatigue in an inert environment where growth should be insensitive to frequency (except for intrinsic effects) and waveform, and a distinct threshold should be observed (Fig 4).

b. true corrosion fatigue, where growth is moderately dependent on frequency and waveform, but  $K_{max} < K_{ISCC}$  (Type A, Fig 4).

c. stress corrosion assisted corrosion fatigue, which is very dependent on frequency and waveform, and  $K_{max} > K_{ISCC}$ . In some cases this enhancement can occur abruptly at  $K_{ISCC}$ , as in type B, Fig 4, or in other cases after some enhanced growth below  $K_{ISCC}$ , as in type C, Fig 4. In this region it has been suggested that growth can be approximated by a superposition of pure fatigue and stress corrosion cracking(18), ie

$$\left(\frac{da}{dN}\right)_{total} = \left(\frac{da}{dN}\right)_{pure} + \left(\frac{da}{dN}\right)_{scc}$$

and evidence both for (19-22) and against(23-25) this equation can be found in the literature. A process competition model(25) has been proposed recently as an alternative to the superposition model(18), but available data is, at present, too meagre to assess its overall applicability.

At best such modelling can only approximate corrosion fatigue behaviour, in view of synergistic effects arising out of complex stress, waveform and frequency interactions. Information demonstrating this complex corrosion fatigue-stress corrosion interaction has been provided very recently by Parkins and Greenwell(26).



They concluded that any superposition will be most nearly valid when the process is dominated by physical factors such as hydrogen ingress or rapid film formation. These were likely to be the extremes of electrochemical activity, and for regions between these two extremes, where marked electrochemical activity is still possible, the value of  $K_{Isc}$  may be lowered under cyclic conditions. Thus for the 0.16C steel considered in their work  $K_{Isc}$  was measured to be  $\sim 22 \text{ MPa}\sqrt{\text{m}}$ , the stress intensity for stress corrosion cracking at a fatigue frequency of 19 Hz was also about  $22 \text{ MPa}\sqrt{\text{m}}$ ; however at a frequency of 0.1 Hz this value for stress corrosion cracking had dropped to  $\sim 12 \text{ MPa}\sqrt{\text{m}}$ . Similar changes under fatigue cycling have also been noted in titanium alloys(22), but in the opposite sense to that found by Parkins and Greenwell(26). The variance of  $K_{Isc}$  has also been noted by Atkinson and Lindley(27).

#### FATIGUE LIFE

##### Aluminium Alloys

In gaseous environments such as air, water vapour has been shown to be the species damaging to the fatigue life of aluminium and its alloys(28,29). The data in Table 1, taken from the work of Broom and Nicholson(29), shows the magnitude of the life reduction, which these authors ascribed to a hydrogen embrittlement mechanism.

The effect of the more aggressive environment of NaCl solution was demonstrated by Stubbington(30) and Stubbington and Forsyth(31). The sustainable fatigue life of Al-Zn-Mg alloys at  $10^7$  cycles could be approximately halved compared to that in laboratory air (Fig 5). A comparison of a high purity Al-7.5Zn-2.5Mg alloy and a commercial version (DTD 683) revealed little difference between the fatigue properties in 3% NaCl solution, although the DTD 683 alloy showed superior properties in air and a somewhat higher tensile strength(31). Heat treatment to alter the microstructure did little to change the corrosion fatigue strength of either alloy(30,31). Stage I fracture was found to be very sensitive to the chloride environment, particularly at low stresses. In addition, for stage II growth, brittle striations were observed on (001) planes on transcrystalline facets(30,32). These observations were explained in terms of electrochemical dissolution, the Kramer effect or film removal for enhanced stage I cracking, and ion adsorption lowering the surface energy during accelerated stage II cracking. An Al-4Zn-5Mg alloy was also studied in later work, and similar susceptibility to NaCl solution was recorded(33). However, in this case failure was intercrystalline; an effect attributed to large amounts of anodic grain boundary precipitation.

The above tests were all carried out at a zero mean load ( $R = -1$ ), but from work by Corsetti and Duquette(34) it would not appear that increasing the mean stress fundamentally alters fatigue behaviour, ie the relative times spent in crack initiation should remain the same both for air and sodium chloride environments.

##### Titanium Alloys

Contrary to the situation for aluminium alloys, little in the way of systematic studies of fatigue life of titanium alloys have been carried out in corrosive environments, although there are reports for tests carried out in laboratory air(35). The importance of texture has already been stressed(6), and if the results of Larson and Zarkades(36) for 'c' and 'a' oriented specimens of Ti-6Al-4V tested in laboratory air are taken as an indication (Fig 6), then testing in, for example, 3.5% NaCl solution would produce a considerable divergence of these already significantly different results. The 'c' curve (stressing normal to (0002)) would be expected to be depressed much more than the 'a' curve (stressing parallel to (0002)). The only corrosion fatigue data generated to date on titanium alloys has been that of Smith and Hughes on Ti-6Al-4V(37), who confirmed earlier work of Wells and Sullivan(38) on the same alloy. In both cases the alloy was found to be immune to NaCl solution, in spite of considerable electrochemical activity(38). This non-susceptibility was attributed to the plane stress nature of surface deformation and the shear nature of stage I cracking, since in plane strain conditions some susceptibility was indicated(38). Other work, however, has suggested some degradation of the fatigue strength of Ti-6Al-4V in chloride solutions(39,40), and such discrepancies could well be due to differences in texture. Waterhouse and co-workers(40,41) have also shown that, irrespective of the tensile or fatigue strength of a number of titanium alloys measured in air, the fretting fatigue strength in argon, air and sodium chloride solution is always in the range 77-150 MPa. This relative insensitivity to environment under fretting conditions was tentatively ascribed to the high reactivity of titanium when the protective oxide film is damaged.

##### Steels

The response of a 0.18C steel to gaseous and aqueous environments has been reported by Duquette and Uhlig (Figs 7 and 8). The alloy was susceptible to moisture in air and argon, but not in oxygen. Note, however, that the fatigue limit is affected only in aqueous environments (Fig 8). In the case of moist oxygen, preferential adsorption of water vapour improved somewhat the fatigue performance compared with that in dry oxygen; this is in contrast to the effect of argon which is only weakly adsorbed (Fig 7). Similar effects have been recorded for a 4140 steel(43), and by increasing the strength of this alloy, increasing sensitivity to moisture was demonstrated (Table 2). The importance of dissolved oxygen in aqueous solutions (Fig 8) will be considered later.

Detailed crack initiation studies on a number of steels have been carried out by Rollins and co-workers(44,45), and they were able to show that initiation in aqueous environments is enhanced by environmentally strain controlled activated dissolution at the tips of corrosion pits and persistent slip bands (PSB). Transient current bursts were found to be associated with the tension and compression cycles (Fig 9), which were believed to emanate from surface slip steps. (It may be noted that similar cyclic variations, in potential, have also been recorded for Ti-6Al-4V(38), and an aluminium alloy and steel(46).) One criticism which might be levelled at this work is the very high strains involved ( $\pm 0.05$ , Fig 9), but more recent work(47) has extended the technique to smaller strains, and hence longer lives, and also to a

number of other alloy systems. Similar effects were recorded, and the authors now feel confident that this technique can be used as a method for predicting whether a material is susceptible to a particular environment. The basis of the technique is to measure the rate of change in transient current magnitude ( $\Delta I = I_{\max} - I_{\min}$ ) in relation to the change in cycles,  $N$ . Thus if the value of  $d\Delta I/dN$  is negative the alloy is not susceptible, whereas when the value is positive, the opposite is true. Results of this survey are given in Table 3.

This observation of current transients is also interesting in view of a minimum corrosion rate which seems to be associated with corrosion fatigue behaviour. For example, Duquette and Uhlig(48) found that an anodic current of  $2 \text{ mA/cm}^2$  had to be applied to their steel before there was any significant reduction in fatigue life. This seemed to be independent of pH(48). Similar behaviour has since been found for a number of materials(43,49), and Smith and Hughes(37) suggest that the immunity shown by Ti-6Al-4V may be due to the corrosion current in this alloy being below the minimum required to affect life. It may also indicate that corrosion resistance is more important than strength in these situations.

#### Protective Measures

**Surface Treatments.** A number of investigations have reported on the efficacy of coatings on aluminium alloys. Broom and Nicholson(29) found that a vinyl rubber coating gave good protection, recording a life equivalent to that found in vacuum. Similar results were reported by Jenkins(50). The application of paints to DTD 683 alloy(51), tested subsequently in NaCl solution, also improved life to that observed in air, but anodic coatings were either ineffective or harmful. Cladding on aluminium alloy L73 degraded the fatigue properties of holed test pieces both in air and sodium chloride solution(52), but the effect was much smaller in the chloride solution. A vinyl coating on an unclad alloy (L71) gave properties almost equal to those in air. Metal sprayed coatings have also been examined, the effectiveness being found to decrease as  $R$  was increased(53,54). This applied to both notched and unnotched situations.

Surface treatment such as peening can also be used on a number of materials, but these run into the problem of relaxation with time; hence they may be ineffective in the long term. Nitriding (of steels) may be a useful alternative in these cases. Anodising of Ti-6Al-4V has been evaluated, and found to have little effect on life compared with a ground finish(55). Plating is also a widely used surface treatment, and for steels it has been suggested that, based on an electrochemical argument, the protection afforded can vary from poor for magnesium to good for cadmium, with zinc being intermediate(43).

**Applied Potentials.** The application of a cathodic current during fatigue testing of Al-4Zn-5Mg has been shown to prolong life indefinitely, and conversely an anodic current of equal magnitude decreased life considerably (Fig 10). Interestingly, such an effect on the Al-Zn-Mg alloy DTD 5054 was found to be marked only over a relatively narrow current range, and furthermore, decreased with increasing  $R$ . (Fig 11). This latter point was attributed to the ability of the applied cathodic currents to suppress cracking in stage I to a much greater extent than in stage II. Protection applied in terms of cathodic potentials has also been shown for 5052 aluminium(58), and for an Al-Zn-Mg alloy(59). In the latter case, however, potentials more negative than  $-1.4\text{V}$  (SCE - standard calomel electrode) produced lower fatigue properties, the degradation being explained in terms of hydrogen embrittlement.

As far as titanium alloys are concerned, the immunity to NaCl solution shown by Ti-6Al-4V(37) has also been found at potentials  $+1.4\text{V}$  (SCE) to the free corrosion potential. Commercial purity titanium showed similar immunity(55). However, changes in properties, both increases and decreases, were recorded at potentials more extreme than these (up to  $\pm 4\text{V}$  SCE)(60).

The beneficial effect of cathodic protection of steels was first demonstrated many years ago(61), and similar effects have been repeated on a number of occasions subsequently(42,43,48). Where the yield strength can be varied, it can be shown that whereas hydrogen evolution does not seem to affect fatigue strength at low strength levels(43), the highest strength condition in 4140 steel is susceptible to hydrogen embrittlement - showing a reduction in fatigue life at the more negative potentials; note that some protection is afforded at less negative potentials(43). In lower strength steel cathodic protection is equally effective in pH2 chloride solutions(48). It should be remembered that any beneficial effect in steel may well be due to the high frequency of cycling and the  $R = -1$  loading employed in these fatigue life tests(62), since as will be shown later, cathodic polarization is clearly detrimental during crack growth in steels, irrespective of the strength level.

**Solution Chemistry Control.** The effect of oxygen level in both distilled water and sodium chloride solution has already been presented (Fig 8). Removal of oxygen restored the fatigue properties of steel to values higher than those found in air, and the mechanism of this improvement lies in the removal of the oxygen reduction process at cathodic areas, thus limiting the amount of dissolution in anodic areas. Similar experiments do not appear to have been carried out for either aluminium or titanium alloys.

Controlling the pH of aqueous solutions can be a very effective means of changing fatigue life. The more acidic the solution the lower are the fatigue properties, and conversely the more alkaline the solution, the higher is the fatigue strength(48,60,63). For marked improvements, however, pH values greater than 12 are required, and Fig 12 shows such an improvement for a high C steel, indicating that properties equal to those in air occur at pH 12(63). This protection in alkaline solutions is believed to arise due to the presence of a passive surface layer. Similar but smaller improvements have been shown for commercial purity titanium and stainless steel on increasing the pH from 2 to 6(60).

#### General Comments

The important point to note from the work reviewed in this section is that although drastic reductions in fatigue life are encountered in aggressive environments, a high degree of control can be exercised over the crack initiation process, and hence produce fatigue lives equal to, or better than, those found in air. More importantly, one can go from a fail to a safe life situation, ie restore the fatigue limit. These

improvements are achieved often purely by the exclusion of the environment (eg many coatings), but in some cases chemical reactions are modified eg oxygen removal, reduced corrosion rate, oxide film alteration or modification to the nature and amount of adsorbed species. The effect is to prevent or moderate the environmental attack at PSBs or particle/matrix interfaces, where greater slip steps intensify slip and reduce lives.

While the simplest and most economical protective against this attack will probably continue to be a coating, there is still room for further work to study surface reactions, since all of the studies described here demonstrate clearly that these are the key to environmental effects during crack initiation. Thus in future it might be possible to stipulate a specific surface structure (which could be achieved by, for example, surface alloying(64)) as a requirement for environmental immunity. Before such a stage is reached, however, it is required to know, for instance:

- 1 Whether environment affects only the formation of PSBs, or also the development of cracks from these PSBs. Recent work has shown that there is a difference in near surface dislocation configuration(65) caused by the environment, which was claimed to be sufficiently large to influence PSB formation. Further work is required to see if such differences are maintained at later stages in fatigue. A factor which must also be taken into account is PSB spacing, since a direct relationship has been shown to exist between this spacing and dissolution rate(66).
- 2 The physical significance of the minimum corrosion rate. If this is used to calculate an overall dissolution rate, the values are much too small to indicate general corrosion(67); hence dissolution must be occurring on a heterogeneous scale in order to produce failure in the short times observed. Thus the overall density and extent of the dissolution sites must be ascertained in order to arrive at a dissolution rate which can be interpreted in terms of the volume of metal removed.
- 3 The kinetics of these local reactions. Here the straining and scratching electrode tests would seem to be useful tools(68,69).
- 4 The rate of initiation in terms of the frequency and depth of cracking. Sectioning specimens and carrying out statistical analyses will be necessary in this area. These tests may also indicate the extent to which corrosion influences the relative fractions of life spent in initiation and growth of cracks. Short cracks are likely to be influenced by metallurgical structure to a greater degree than larger cracks, and moreover, the stage I growth mode may be much more prevalent for these short cracks. Restricting growth to this mode, say by controlling shear stress, would therefore be of benefit since greater electrochemical control might also be possible.

#### FATIGUE CRACK GROWTH

Considerable effort has been expended in recent years on studying the effects of stressing variables on fatigue crack growth in corrosive environments. And since these have an important bearing on the interpretation of the data for each alloy system, the main conclusions from these investigations are first summarized here, before considering effects such as composition, different environments etc.

#### Stressing Variables

The effect of frequency changes on fatigue behaviour has been established for a considerable time, but only in recent years has the significance of load profile been revealed. The advent of modern servo-hydraulic testing machines has allowed the incorporation of function generators, thus extending the normally available sine wave to other waveforms. The result is that additional waveforms such as: triangular, square, positive and negative sawtooth have been studied, and a number of investigations have now shown that the rise time during the loading cycle is a very critical factor(2,21,22,25,27,70-72), ie the slower the rise time the greater is the environmental enhancement. This is analogous to a lowering of the frequency, indicating that waveform and frequency are very interdependent and cannot be considered in isolation.

For aluminium alloys both Selines and Pelloux(2) and Proctor and Moss(21) have shown the extent to which waveform influences growth, and their results are in broad agreement. Fig 13 shows the data of Proctor and Moss, where frequency effects are also included. These results fell into two groups: For  $\Delta K < \Delta K_{Isc}$  the significant observation was the marked effect of frequency only for the triangular waveform, which could be explained completely on the basis of enhanced anodic dissolution. (The square waveform had no effect because the rise times were the same for both frequencies.) For  $\Delta K > \Delta K_{Isc}$  the lower frequency produced the faster growth rate for both waveforms. The increment in growth for the square waveform could be explained completely on an additive effect of stress corrosion and fatigue, as proposed by Wei and Landes(18), but for the triangular waveform anodic dissolution also had to be included.

In the case of titanium alloys waveform and frequency changes have also been performed in the same series of tests (Fig 14). Here, however, an inversion in frequency dependent-growth occurred at  $\Delta K \sim \Delta K_{scc}$ , growth  $< \Delta K_{scc}$  showing a direct relationship with frequency. (Note again the little effect of frequency for the square waveform below  $\Delta K_{scc}$ .) This change in frequency dependent growth was explained on the basis of repassivation being sufficiently rapid to prevent transgranular cleavage for  $\Delta K < \Delta K_{scc}$ , leading in these cases either to minimised dissolution or slip dispersal.

The behaviour of steels below  $K_{Isc}$  is summarized by the work of Barsom(71,73), where again, positive rise times only led to enhancement (Fig 15). Normal inverse frequency effects were also observed (Fig 16). Similar results have been reported for other steels, both for stressing below(27,74) and above(19,20,25,72)  $K_{Isc}$ .

A third parameter which also needs to be considered in this area is that of R ratio, since increasing the mean stress usually increases growth rate, at a constant  $\Delta K$ , because  $K_{max}$  is increased ( $\Delta K = K_{max}(1 - R)$ ). Such a means of increasing growth rate must be of importance in corrosive environments since it could move the crack growth regime from below to above  $K_{Isc}$ . When this occurs much flatter curves (and often a plateau) are observed(25,72,74,75). As an example of this the data in Fig 17 shows the shape of crack growth curves for an R ratio of 0.8 for 4340 steel. Some idea of the plateaus for aluminium and titanium alloys can be gauged from Figs 13 and 14.

#### Aluminium Alloys

The damaging species in gaseous environments, when measuring fatigue crack growth in these alloys, has also been isolated as water vapour. For instance, crack growth rates in vacuum for DTD 683 were found to be a factor of ten slower than those in laboratory air(76,77) (Fig 18). These authors concluded, in agreement with Broom and Nicholson(29), that hydrogen embrittlement was the most likely explanation of their results. Testing outside the UK has found substantial agreement with these findings(eg 78-80); moreover, Wei has shown that crack growth is thermally activated with the activation energy being very sensitive to  $\Delta K$ (78).

The effect of alloy composition on corrosion fatigue crack growth is shown by the data reviewed by Hahn and Simon(81), and conveniently illustrated by the results of Feeney et al(82) for  $\Delta K < K_{Isc}$  (Table 4). The much smaller change in environment and  $\Delta K$  for 2024-T3 is very clearly apparent. In contrast there is little difference between the results for the two 7000 series alloys, with 7178-T6 being slightly worse. The high sensitivity of the 7000 series alloys is also shown by the work of Pelloux(83) and Stoltz and Pelloux(84), and over- or under-ageing 7075 does little to reduce the growth rate. Of significance with overageing is the change from planar to wavy slip, which probably modifies deformation at the crack tip and reduces the growth rate(83,84).

#### Titanium Alloys

A study of growth rates in the threshold region (low  $\Delta K$  values) in Ti-6Al-4V has been made by Irving and Beevers(4). Much higher threshold values were observed in vacuum (Fig 19) for all three microstructures studied, the growth rates in vacuum always being at least 2-3 times slower than in air. At  $\Delta K < 12 \text{ MPa}\sqrt{\text{m}}$  structure sensitive growth occurred, but for higher  $\Delta K$  values the growth rates were approximately the same in vacuum and air for all three microstructures. This latter point is disputed by the data of Paton et al(85) obtained in laboratory air. Irving and Beevers claimed that the changeover from structure sensitive to non-sensitive growth occurs when the plastic zone size exceeds the dimensions of a microstructural parameter.

The effects of other gaseous environments on crack growth show similar ranking to those for aluminium alloys, and this has been shown by James(86), Bucci and Paris(70) and Wei and Ritter(87). It should be noted, however, that in gaseous hydrogen very rapid growth occurs in titanium alloys, in marked contrast to the absence of any effect for aluminium alloys(88). In these cases fracture is on (10 $\bar{1}$ 0) and not (0002) planes(89).

If we now consider aqueous environments, a number of workers have shown how distilled water and sodium chloride solution increasingly degrade crack growth resistance(22,90-93). The importance of texture can be estimated from the data of Speidel et al(91) and Wanhill(93), but it must be stressed that a systematic study of textural effects on corrosion fatigue is still awaited. Evaluation of titanium alloys produced to specified textures would seem to be an urgent necessity (this would apply equally to crack initiation and crack growth), particularly since some of these can now be produced commercially(94,95); and improved fatigue properties are claimed(95). To demonstrate the high degree of corrosive attack on (0002) planes, the work of Stubbington and Pearson(96) on strongly textured Ti-6Al-4V, possessing a single crystal type of texture, is presented here. These workers applied a hold time at maximum load, thus approximating the stressing mode to one of a trapezoid waveform of low frequency. Stressing to cause failure normal to (0002) showed no effect of this dwell cycling in laboratory air (Fig 20), but stressing to cause fracture on (0002) showed an increase in growth rate (Fig 21). This effect could not be removed by testing in vacuum. The very significant effect of adding sodium chloride solution to test pieces of this orientation is shown in Fig 22. Here there was a marked increase in growth rate merely by decreasing the frequency from 25 to 0.3 Hz, and failure then occurred in only 11 cycles of dwell. Strain assisted hydrogen embrittlement at or just ahead of the crack tip was thought to be the mechanism of premature failure.

#### Steels

Wei and co-workers have considered a number of high strength steels and shown how their fatigue crack growth behaviour varies in gaseous environments(15,97-99). For a 0.45C steel(97) marked sensitivity to the presence of water vapour was found, but a 18 Ni(250) maraging steel proved non-sensitive(15). If the strength of this steel was increased to the 18(300) grade, however, the alloy was susceptible(98). Testing in hydrogen gas showed very rapid crack growth, but in this case, if water vapour was added, preferential adsorption of water at the crack tip reduced growth rates(15). Though the effect of hydrogen gas is drastic, it has been shown that hydrogen sulphide is, in fact, more aggressive(23,25,72), and from work on these and other aggressive environments Austen and Walker(72) concluded that for below  $K_{Isc}$  behaviour the degree of aggressiveness could be represented by changes in the values of  $m$  and  $C$  in the Paris equation (Table 5). Thus, for example,  $m$  varied from 2.33 for growth in vacuum to 3.11 for growth in  $\text{H}_2\text{S}$  gas, at similar  $C$  values. A detailed discussion of mechanisms is also presented in this paper.

As in the case of aluminium and titanium alloys, fatigue behaviour of steels is also marked by the transition to stress corrosion fatigue at high  $K$  values, and some idea of these differences in above and below  $K_{Isc}$  behaviour for 4340 steel can be gained by comparing Figs 17 and 23. Below  $K_{Isc}$

increased yield strength has a twofold effect: it produces greater environmental enhancement, and also results in a lowering of the transition to above  $K_{Isc}$  behaviour (Fig 23). Unfortunately similar work to that on aluminium and titanium alloys, where the transition could be observed in the same test, does not appear to have been carried out. This would be possible in steel (eg 4340) by choosing an appropriate tempering temperature and starting stress intensity level. At the moment, however, the magnitude of the transition must be judged by comparing separate tests, either on the same heat treated alloy (by eg varying R(25,72)), or on different alloys (eg HY 80 and 4340(19)).

Purity of steel also seems to be important since Evans et al(101) have shown that a high purity version of an En 24 steel proved less sensitive to a laboratory air environment (compared with vacuum) and exhibited much less grain boundary cracking. It is not known whether such effects would be carried over to more aggressive environments.

#### Protective Measures

**Applied Potentials.** The first demonstration of cathodic protection applied to aluminium alloys would seem to have been that of Forsyth and Sampson(56). In their experiments a crack was allowed to initiate in Al-4Zn-5Mg, and alternating periods of cathodic and anodic currents of equal magnitude were applied subsequently. No growth was detected under cathodic conditions (Fig 24), whereas applied anodic currents produced considerable crack enhancement. On examination of the fracture surfaces some crack growth could in fact be detected under cathodic protection, but much more widely spaced striations were found for anodic attack (Fig 25). Although this mode of failure was intergranular, similar effects of anodic and cathodic potentials, if somewhat moderated, have been observed in transgranular cracking in 7075-T6(83,84,102) and 7079-T651(91). This difference in magnitude may be due to the stress corrosion cracking nature of crack advance in Al-4Zn-5Mg, since Speidel et al(91) have shown that for 7079 much greater suppression of crack advance could be achieved for stress corrosion cracking compared with corrosion fatigue. An interesting observation related to this work is that at potentials more positive and negative than -1.4V(SCE) in an Al-Zn-Mg alloy, striations were much wider (and brittle) than the closely spaced ductile ones produced at -1.4V(59).

Protection at negative potentials has also been shown for Al-7Mg (Fig 26) where in this case growth rates close to those in argon were achieved. In this work hydrogen embrittlement was categorically eliminated as the cause of environmental enhancement, which is in direct conflict with the conclusions of Endo et al(59).

Contrary to the situation on fatigue life of titanium alloys, a number of investigations have shown that applied potentials can affect the growth rates of these alloys in halide solutions(22,90,91). Speidel et al(91) tested a Ti-6Al-4V alloy with (0002) planes parallel to the crack plane in 5M KI solution and found that -1.5V(SCE) gave rates identical to those measured in distilled water, both above and below  $K_{Isc}$  (Fig 27). Care must be taken in the use of this data since for corrosion fatigue in 0.6M KCl solution complex behaviour was recorded - at low  $\Delta K$  values ( $< K_{Isc}$ ) a changeover in the effectiveness of applied potentials was observed, ie the more positive potentials resulted in the lower growth rates(91). These latter results are in agreement with other data which showed that for true corrosion fatigue anodic potentials(22,90) reduced growth rates and cathodic potentials(91) had the reverse effect. For a Ti-6Al-2Nb-1Ta-0.8Mo alloy, however, no effect of cathodic polarization was observed(103). In view of these conflicting results it would seem prudent to exercise caution in the use of applied potentials on titanium alloys, at least until further work clarifies the position. The importance of microstructural condition on the corrosion fatigue of Ti-6Al-2Sn-4Zr-6Mo under applied potentials has been shown by Williams et al(104).

Unlike the situation for aluminium or titanium alloys, cathodic polarization increases growth rates in steels and a comparison of all three alloy systems has recently been completed by Crooker et al(103), from which Fig 28 is taken. The trends shown in this figure are in agreement with other work on various strength steels(14,19,74). Negative potentials degrade the resistance of steels (and presumably of aluminium and titanium alloys) by producing hydrogen which, by ingress at the crack tip, results in embrittlement even at low strength levels(74,105). It is important to note that fatigue life would not be expected to be affected at these strength levels(42,48).

A point worthy of note regarding these types of experiments on applied potentials concerns the position of the Luggin probe of the reference electrode. This should be placed as close as possible to the crack, thus eliminating any (iR) drop which has been shown to produce an elongation of the potential axis(102).

**Solution Chemistry.** Control of solution pH, as in fatigue life, can be very effective in modifying fatigue crack growth rates. The pH at the tip of a fatigue crack is ~3(90) (for a nominally neutral bulk solution), thus making the overall solution more acidic would not be expected to increase growth rates markedly. This has been borne out by work on Ti-6Al-6V-2Sn in pH 1 sodium chloride solution (Fig 29); although above  $\Delta K_{sc}$  a greater effect of pH was observed. By making the chloride solution alkaline Barsom demonstrated for a 12Ni-5Cr-3Mo steel that growth could be reduced to rates lower than those observed in air (Fig 30). Buffering of the sodium chloride solution to pH 8.3 doubled the  $K_{Isc}$  value for 4340 steel but did not alter growth rates(19).

The addition of inhibitors is another means of controlling solution chemistry, and for 7075 aluminium alloy Stoltz and Pelloux(84,106) showed that by adding a solution of sodium nitrate to sodium chloride solution it was possible to reduce growth rates to levels observed for dry air and argon (Fig 31). The mechanism of improvement was thought to be the replacement of chloride ions by nitrate ions at the crack tip, thus reducing markedly the anodic dissolution rate. The important fractographic observation was the transition from brittle to ductile striations in the presence of an inhibitor. Care would seem to be necessary in the application of inhibitors to titanium alloys since  $Na_2SO_4$  additions to distilled water

were found to be effective, at least for  $\Delta K < 30 \text{ MPa}\sqrt{m}$ , whereas nitrate ions decreased rates below  $\Delta K_{\text{sc}}$  but raised  $\Delta K_{\text{sc}}$  and accelerated growth above the transition(22); this behaviour is similar to the effect recorded by changes in anodic polarization(91). Inhibitors may also be ineffective in acidic solutions.

#### General Comments

In many ways consideration of crack growth is much simpler than crack initiation since damage is restricted to a small known region which is relatively easy to observe. The solution chemistry, however, is more complicated because the solution in the crack is usually more acidic than in the bulk(90); although the pumping action of the cyclic loading, particularly at high frequency, may act as a means of raising the pH, even for  $\Delta K > K_{\text{Isc}}$ (90). In practical terms the best that can probably be achieved, since one cannot stop growth completely, is to aim to reduce growth rates to those observed in dry argon or vacuum.

Waveform experiments have shown that the important phenomenon during crack growth is the dynamic interaction between dislocations and the environment. Moreover, similar indications are apparent from the reverse polarity tests, inhibitor additions and pH control. These would all suggest that enhancement of growth rates is a short range effect, and if such embrittlement is local (within  $\lambda$  of the crack tip?) then Auger electron spectroscopy might be a useful tool in determining whether any ions preferentially collect in the crack tip region. It would also be desirable to ascertain whether degradation occurs as a result of a critical straining rate at the crack tip, such as that which has been recorded in stress corrosion cracking(26). It is also important to establish whether, alternatively, these effects arise due to differences in crack closure. Other methods of crack tip analysis would also be useful, particularly if testing is carried out under controlled conditions of potential and solution chemistry. It would also be useful to carry out simple programmed loading, provided that constant amplitude type testing is sufficiently understood. Allied to all of this work should be extensive fractography.

#### CONCLUDING REMARKS

It is clear from this review that under certain conditions the detrimental effect of aggressive environments can be moderated or even removed for all three of the alloy systems considered here. This is most likely to occur in situations where protectives prove effective and no cracks develop in the life of the component, and/or in closed circuit conditions where the solution chemistry can be controlled or cathodic protection applied. Closed circuit conditions exist in eg boiler tubes, and here the situation could also be assisted by the solution being deaerated. The more pressing needs arise in open circuit conditions where control of solution chemistry is not practicable. One is then forced to resort to protective schemes, possibly using coatings of sophisticated compositions which might include inhibitors which are preferentially leached out; however, the degree of control in these instances cannot be very high. Problems might also arise in in-service inspection. With cathodic protection sacrificial anodes may be attached but there is dispute in all three systems as to the truly beneficial effects to be realized; and moreover, for crack growth in steels there is no evidence of any benefit. One is forced to conclude that extreme caution must be exercised in the use of such electrochemical protection.

Changes in all stages of design could also assist in alleviating the problem: eg quenched in stresses could be minimised by careful heat treatment; triaxial stressing (possibly inducing hydrogen migration to the crack tip thus causing failure) could be avoided by designing wherever practicable in thinner materials. (It may be noted here that little has been done on the effect of thickness in corrosion fatigue crack growth, except for aluminium alloys(1).) There would also be advantages in using torsional loading whenever possible, since this would introduce shear (mode III), as opposed to the usual tensile (mode I) stress. This would also result in non-opening cracks, from which the environment might be excluded.

As far as composition and microstructure of the alloys themselves are concerned, there is again little evidence of marked changes in fatigue behaviour. For example, although there are improvements in  $K_{\text{Isc}}$  in the newer (purer) aluminium alloys (eg 7050), there do not seem to be any significant changes in corrosion fatigue behaviour(107). The best that can be achieved at the present time would seem to be either:

- a. choose the composition within an alloy series eg 2024 instead of 7075; maraging instead of low alloy steels;
- b. choose a heat treatment schedule for a particular alloy eg T73 instead of T6 for 7000 series aluminium alloys; a higher tempering temperature in 4340 type steels, but avoiding temper embrittlement;
- c. change from one alloy system to another (more resistant) one, eg titanium alloys for fatigue life situations;
- d. choose an alloy most amenable to a particular protective scheme, eg one showing immunity, or near immunity, for a particular inhibited coating.

More sophisticated choices than these are not usually possible because the specific mechanisms of environmental attack are not sufficiently well understood. Probably the most widely accepted mechanism is that of hydrogen embrittlement in steels(108), but there is considerable dispute, for instance, regarding its relevance to aluminium alloys (eg 59,68). A discussion of this and the other mechanisms of corrosion fatigue can be found in the review of Duquette(67) and the Storrs Corrosion Fatigue Conference(17).

Hopefully future work will lead to some improvements in this situation; for instance, by specifying mechanisms in more detail, ultimately possibly providing a general prediction of environmental effects

that can transcend the treatment of each material and environment as a separate study. A big stumbling block to this advancement is the multidisciplinary nature of the phenomenon of corrosion fatigue, and improved communication between workers in the metallurgical, stressing and electrochemical areas would therefore seem essential for any rapid progress.

Table 1

EFFECT OF GASEOUS ENVIRONMENTS ON THE FATIGUE LIFE OF ALUMINIUM ALLOYS(29)

Test Environment	High Purity Al-4Cu	(Al-Cu-Mg) L65	(Al-Zn-Mg) DTD 683
	(Cycle to Failure $\times 10^{-5}$ )		
Vacuum (cold trap)	8.9	-	-
Nitrogen (cold trap)	6.3	5.4	7.3
Oxygen (cold trap)	8.4	3.5	2.6
Hydrogen (cold trap)	3.2	2.7	1.4
Air	0.98	0.66	1.2
Water Vapour	0.59	-	-
Stressing ( $\pm$ ) (MPa)	185	317	301

Table 2

FATIGUE STRENGTH OF 4140 STEEL IN AIR AND 3% NaCl SOLUTION, AS A FUNCTION OF YIELD STRENGTH(43)

Hardness (Rc)	Yield Stress (MPa)	$10^6$ Cycle fatigue strength		Decrease in fatigue strength %
		Air	NaCl solution	
20	662	400	310	22
37	1131	634	983	24
44	1407	758	462	39
52	1669	1069	620	42

Table 3  
CORROSION-FATIGUE TEST AND CYCLIC STRAIN DISSOLUTION DATA FOR VARIOUS SYSTEMS(47)

Material	Corrosion-fatigue test data		Cyclic strain dissolution data I peak Values, First cycle (mA)	Sign of $\frac{d\Delta I}{dN}$	Cor fat end limit air fat end limit $\left(\frac{\sigma_c}{\sigma_a}\right)$		
	Air fatigue limit ( $\sigma_a$ ) (MPa)	Corrosion-fatigue endurance ( $\sigma_c$ ) (MPa)				Test conditions, material, test solution and potential	
0.1% steel susceptible	250	55 (freely corroding in 3%NaCl of neutral pH)	0.2% C steel in 3.5%NaCl solution pH 8.2 (at the natural potential)	0.9 Tensile Compressive	1.85 } +ve 1.15 }	0.22	
	138	69 (freely corroding in 3%NaCl of neutral pH)	Al-Zn-Mg in 3%NaCl solution pH 6.5 (at the natural potential)	70 $\mu$ A Tensile Compressive	111 $\mu$ A } +ve 79 $\mu$ A }	0.5	
0.1% steel 0.1% steel 0.4% steel Titanium non-susceptible	275	275 (freely corroding in 3%NaCl pH 12)	0.2% C steel in 3.5%NaCl solution pH 12 (at the natural potential)	1.25 Tensile	0.4 } -ve	1.0	
	185	194 (in 10%NH <sub>4</sub> NO <sub>3</sub> sol at 0.6 V (SCE))	0.2% C steel in 10%NH <sub>4</sub> NO <sub>3</sub> (at 0.6 V (SCE))	1.0 Compressive Tensile	0.2 } -ve 0.4 }	1.05	
	270	270 (freely corroding in 3%NaCl solution + 0.05 M Na <sub>2</sub> CrO <sub>4</sub> )	0.2% C steel in 3.5% NaCl solution + 0.05 M Na <sub>2</sub> CrO <sub>4</sub> pH 8.2 (at the natural potential)	0.5 Compressive Tensile Compressive	0.25 } -ve 0.4 } 0.3	1.0	
	466	466 (freely corroding in 3%NaCl solution of neutral pH)	Titanium in 3.5% NaCl solution pH 8.2 (at the natural potential)	2.0 Tensile Compressive	0.2 } -ve 0.1 }	1.0	
	265	265 (freely corroding in 3%NaCl solution of neutral pH)	Monel in 3.5%NaCl solution pH 8.2 (at the natural potential)	0.5 Tensile Compressive	0.2 } -ve 0.12 }	1.0	



Table 4

COMPARISON OF FATIGUE CRACK GROWTH RATES OF ALUMINIUM ALLOYS IN VARIOUS ENVIRONMENTS(82)

$\Delta K$ (MPa $\sqrt{m}$ )	da/dN ( $\mu$ in/cycle)			
	Dry Air	Wet Air	Distilled Water	3 $\frac{1}{2}$ % NaCl solution
2024-T3				
3.8	0.1	0.1	0.3	2.2
5.5	1.8	1.8	2.8	5.2
9.9	19	19	20	21
7075-T6				
3.8	0.42	1.3	2	3.2
5.5	3.8	7.4	13	18
9.9	27	50	80	>100
7178-T6				
3.8	0.8	2	4.4	2.9
5.5	3.6	12	19	14
9.9	4.4	>100	>100	>100

Table 5

INFLUENCE OF AGGRESSIVE ENVIRONMENTS ON THE PARAMETERS  $m$  AND  $C$ , FROM THE PARIS EQUATION(16), FOR A 835M30 STEEL(72)

Environment	$m$	$C$ ( $\times 10^8$ ) (mm/cycle)
Vacuum	2.33	3.07
Argon	2.42	7.34
Air	2.28	11.2
3.5% NaCl solution	3.06	1.44
Oils	3.14	0.987
H <sub>2</sub>	3.37	0.912
H <sub>2</sub> S	3.11	3.17

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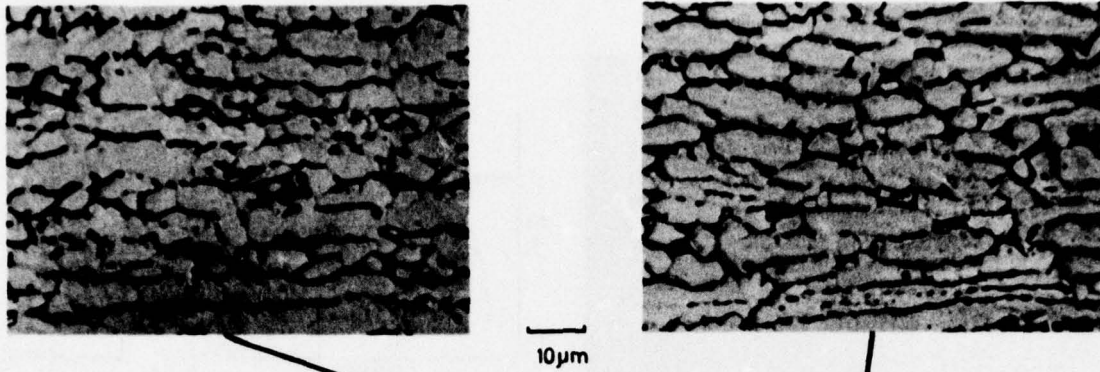
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The author would like to thank C A Stubbington and P J E Forsyth for their comments on the manuscript.

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Mechanical Property	Stress Axis	
	<1120> SL Orientation	[0001] TL Orientation
0.2% Proof Stress (MPa)	859	1052
E (GPa)	110	131
R of A (%)	23	33
$n$ in $\frac{da}{dn} = C(\Delta K)^m$	2.4	3.1
$K_{Ic}$ (MPa $\sqrt{m}$ )	80	46
$C_v$ (J)	47.5	19

Fig 1 Variation in mechanical properties in strongly textured Ti-6Al-4V (different crystallographic orientation; same microstructure)(6).

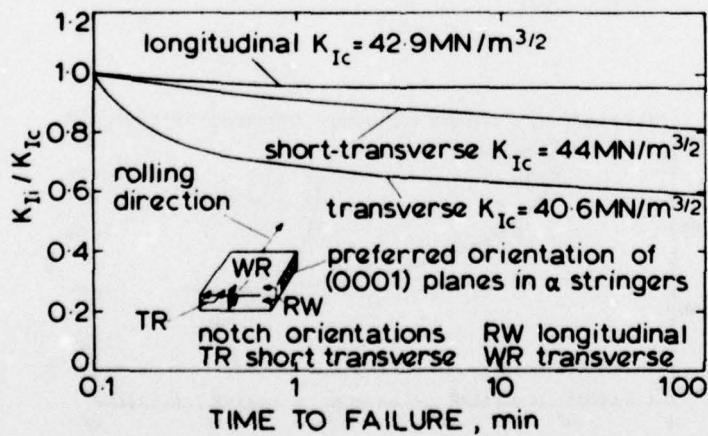


Fig 2 Stress corrosion susceptibility as a function of specimen orientation in Ti-8Al-1Mo-1V annealed plate(7).

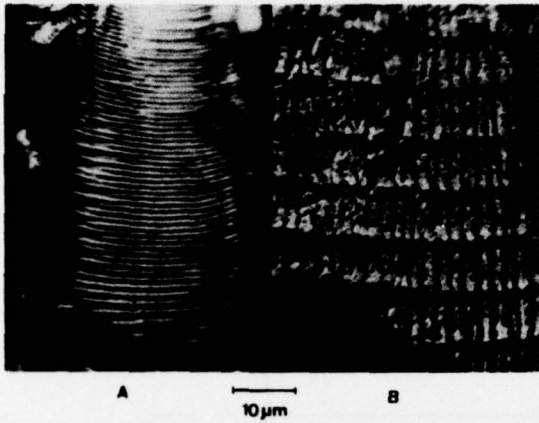


Fig 3 Fatigue striations produced in (A) air (ductile) and (B) NaCl solution (brittle). (Test Pieces are Al-7.5Zn-2.5Mg)

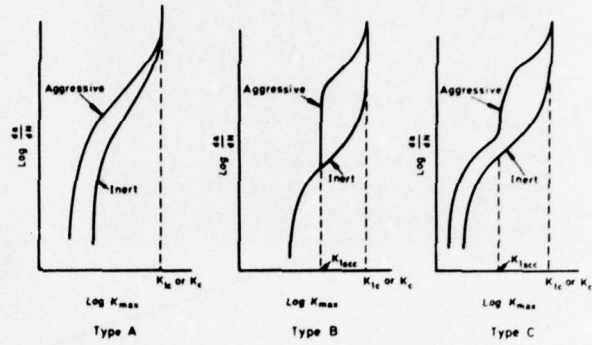


Fig 4 Types of fatigue crack growth behaviour(17).

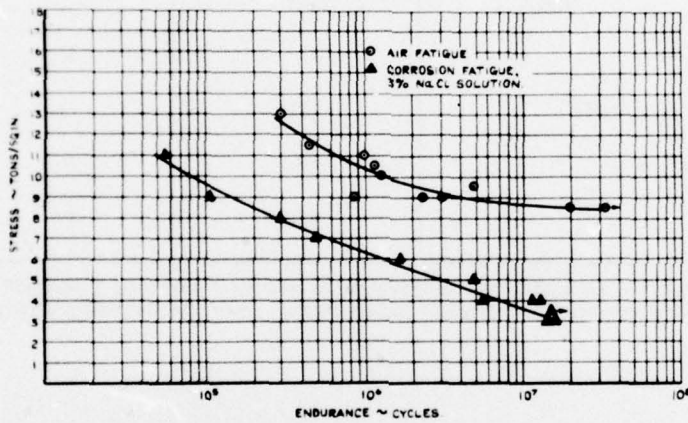


Fig 5 Stress-cycles curves for air and corrosion fatigue of Al-7.5Zn-2.5Mg solution treated 450°C, cold water quenched and aged 0.1 days at 150°C (rotating bending tests)(30).

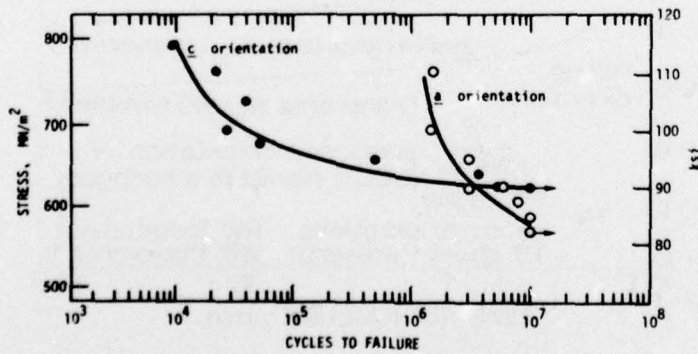


Fig 6 Stress-cycles curves for 'c' and 'a' oriented test pieces in strongly textured Ti-4Al-4V, tested in air, (R = 0.1; frequency (f) = 30 Hz(36).



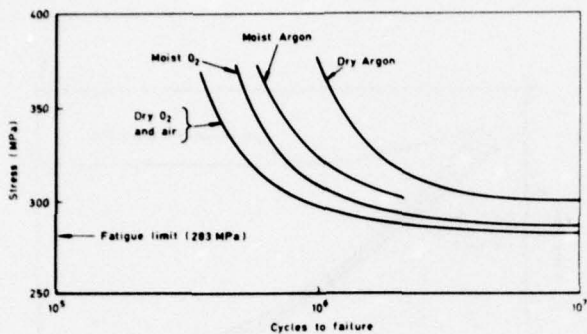


Fig 7 Stress-cycles curves for fatigue tests on 1015 steel in air, oxygen and argon (reverse bending tests;  $f = 30$  Hz)(42).

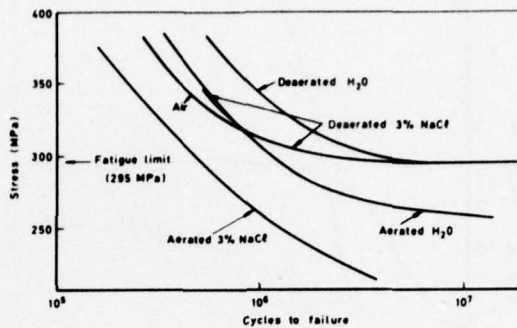


Fig 8 Stress-cycles curves for fatigue tests on 1015 steel in air, distilled water and 3% NaCl solution (reverse bending tests;  $f = 30$  Hz)(42).

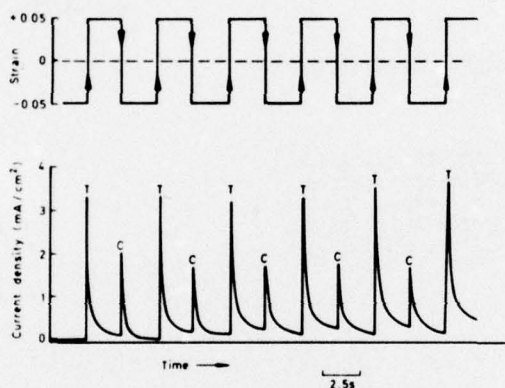


Fig 9 Dissolution transients in 18-8 stainless steel strain cycled in 3.7M  $H_2SO_4$  at 1.042 V (NHE-normal hydrogen electrode)(45).

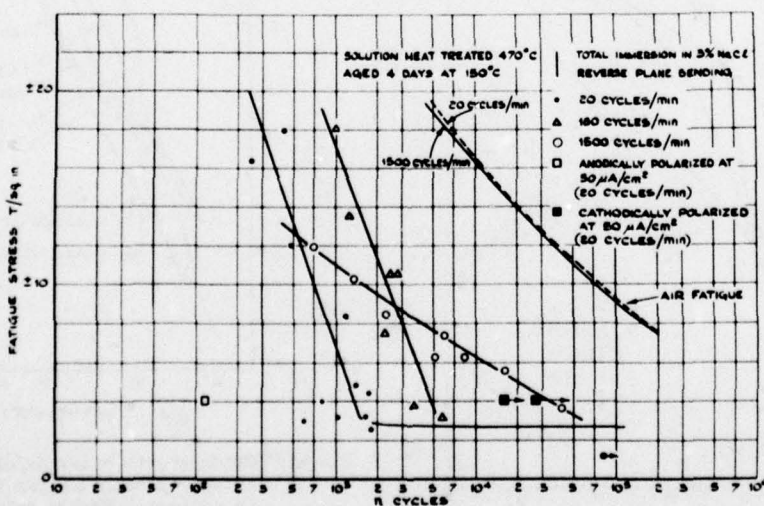


Fig 10 Stress-cycles curves for Al-4Zn-5Mg tested in air and 3% NaCl solution (reverse plane bending tests)(56).

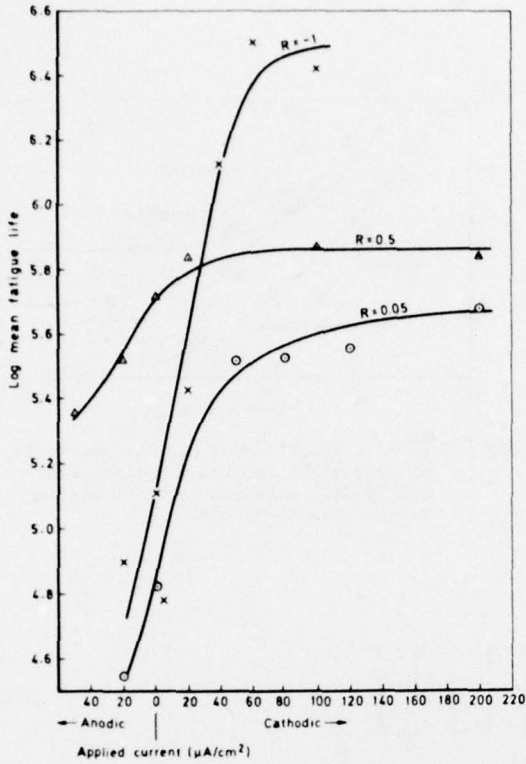


Fig 11 Mean fatigue life for DTD 5054 aluminium alloy under anodic and cathodic applied currents at different R ratios. ( $R = -1$  for  $+154 \text{ MPa}$ ;  $R = 0.05$  for  $136 \pm 124 \text{ MPa}$  and  $R = 0.5$  for  $255 \pm 85 \text{ MPa}$  stressing;  $f = 50 \text{ Hz}$ )(57).

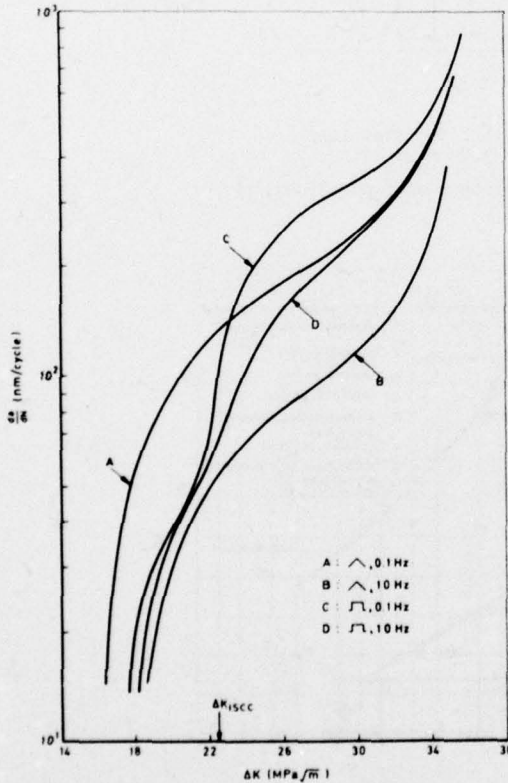


Fig 13 Crack growth rates  $da/dN$  versus stress intensity range  $\Delta K$  for Al-Cu-Mg alloy (HE 15) - effect of waveform and frequency. ( $R = 0.04$ )(21).

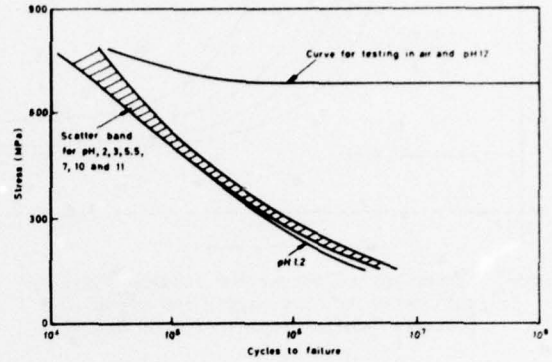


Fig 12 Stress-cycle curves showing effect of pH of chloride solution on corrosion fatigue strength of 1% C steel (rotating bending tests)(63).

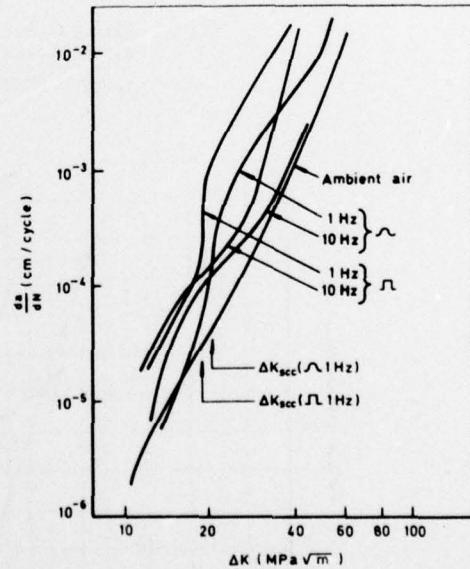


Fig 14 Crack growth rates  $da/dN$  versus stress intensity range  $\Delta K$  for Ti-6Al-6V-2Sn in aqueous 0.6M KBr solution - effect of waveform and frequency. ( $R = 0.1$ )(22).

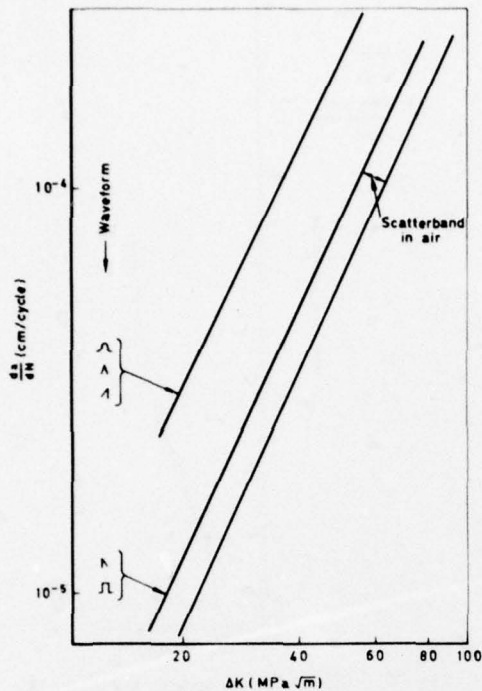


Fig 15 Crack growth rates  $da/dN$  versus stress intensity range  $\Delta K$  for 12Ni-5Cr-3Mo steel in 3% NaCl solution - effect of waveform. ( $R = 0.25$ ;  $f = 0.1$  Hz)(71).

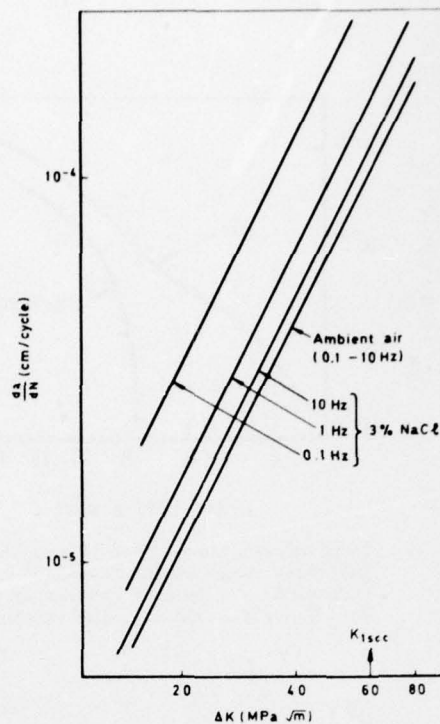


Fig 16 Crack growth rates  $da/dN$  versus stress intensity range  $\Delta K$  for 12Ni-5Cr-3Mo steel in 3% NaCl solution - effect of frequency. ( $R = 0.25$ )(73).

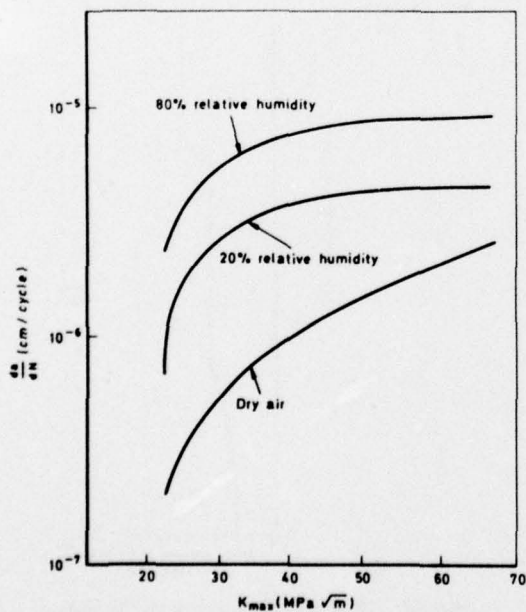


Fig 17 Crack growth rates  $da/dN$  versus maximum stress intensity  $K_{max}$  for 4340 steel - effect of water vapour content. ( $R = 0.8$ )(75).

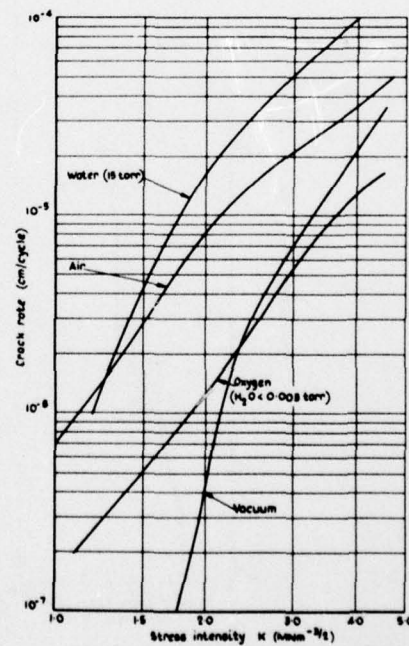


Fig 18 Crack growth rates  $da/dN$  versus stress intensity range  $\Delta K$  for DTD 683 aluminium alloy - effects of gaseous environments. ( $R = 0.1$ ;  $f = 100$  Hz; sine waveform)(77).

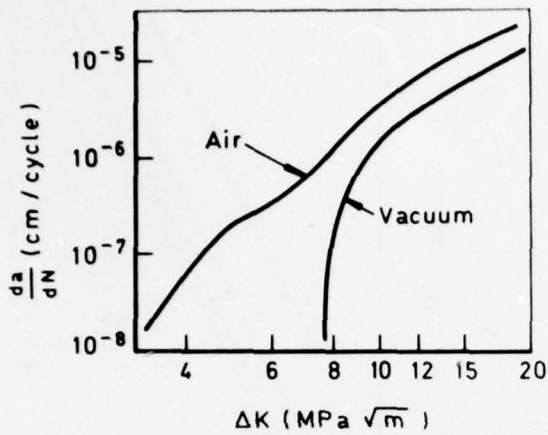


Fig 19 Crack growth rates  $da/dN$  versus stress intensity range  $\Delta K$  for Ti-6Al-4V (annealed) - effect of gaseous environment. ( $R = 0.35$ ;  $f = 100$  Hz; sine waveform)(4).

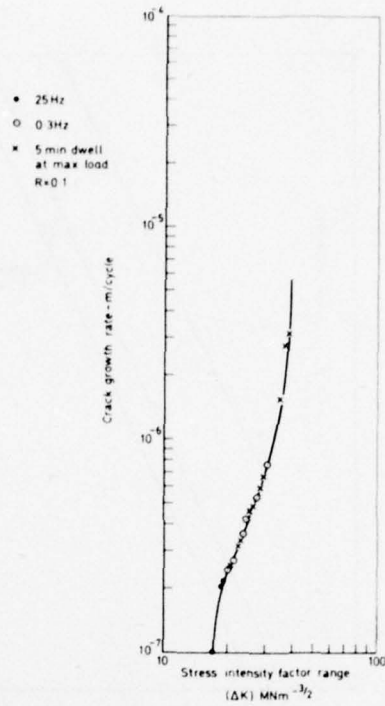


Fig 20 Crack growth rates  $da/dN$  versus stress intensity range  $\Delta K$  for strongly textured annealed Ti-6Al-4V - air environment; SL orientation (fracturing normal to (0002) planes)(96).

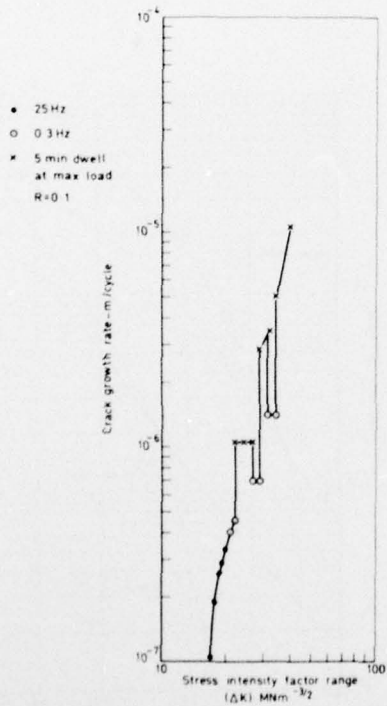


Fig 21 Crack growth rates  $da/dN$  versus stress intensity range  $\Delta K$  for strongly textured duplex annealed Ti-6Al-4V - air environment, TL orientation (fracturing on (0002) planes)(96).

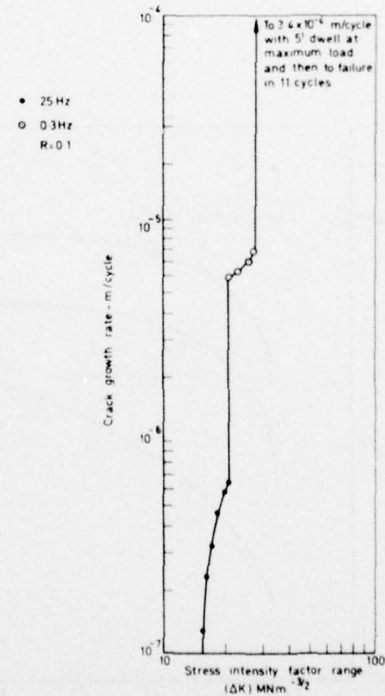


Fig 22 Crack growth rates  $da/dN$  versus stress intensity range  $\Delta K$  for strongly textured duplex annealed Ti-6Al-4V - 3.5% NaCl solution, TL orientation (fracturing on (0002) planes)(96).

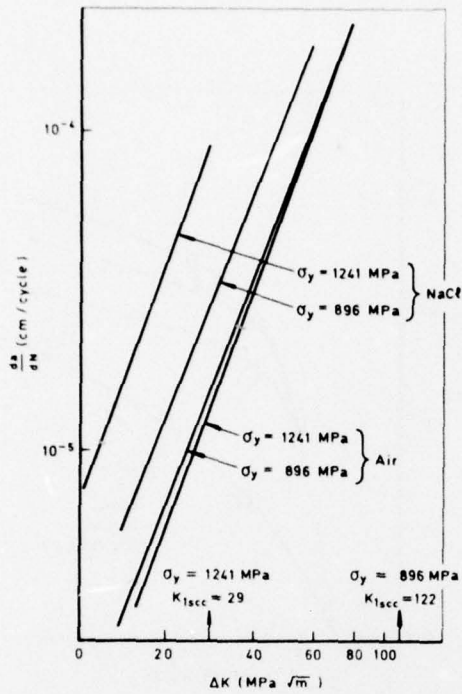


Fig 23 Crack growth rates  $da/dN$  versus stress intensity range  $\Delta K$  for 4340 steel tempered to 896 and 1241 MPa yield strengths. Tested in air and 3% NaCl solution. ( $R = 0.1$  and  $0.2$  respectively;  $f = 0.1$  Hz)(100).

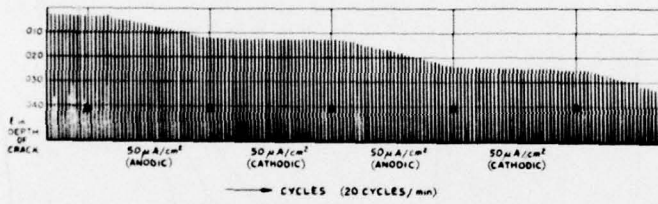


Fig 24 Changes in crack depth with impressed current in Al-4Zn-5Mg(56). (33 cycles per reversal).

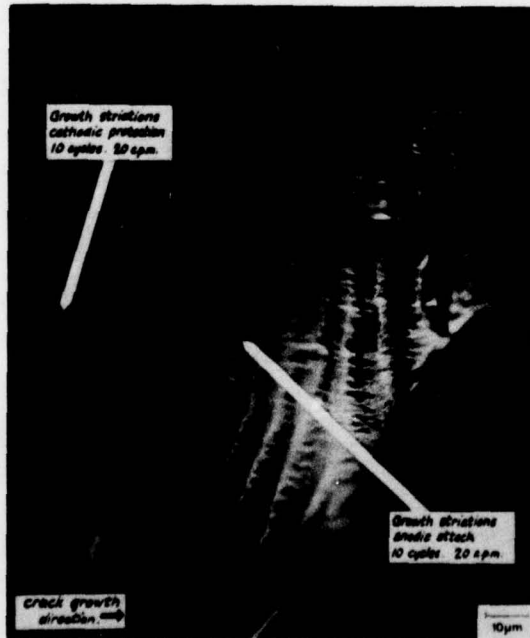


Fig 25 Changes in striation spacing with impressed current in Al-4Zn-5Mg(56). (10 cycles per reversal).

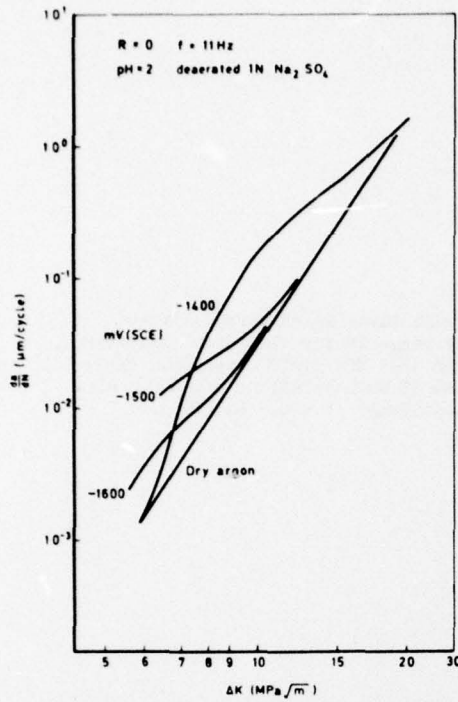


Fig 26 Crack growth rates  $da/dN$  versus stress intensity range  $\Delta K$  for Al-7Mg - effect of cathodic protection(68).

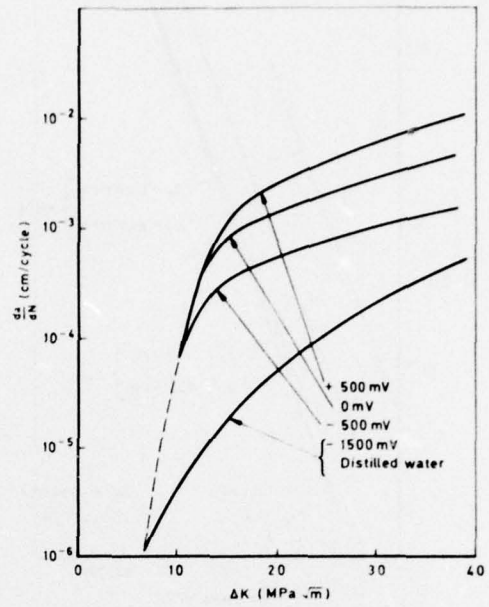


Fig 27 Crack growth rates  $da/dN$  versus stress intensity range  $\Delta K$  for Ti-6Al-4V (fracturing on (0002) planes) in aqueous 5M KI solution - effect of cathodic protection vs SCE. ( $R = 0$ ;  $f = 4$  Hz)(91).

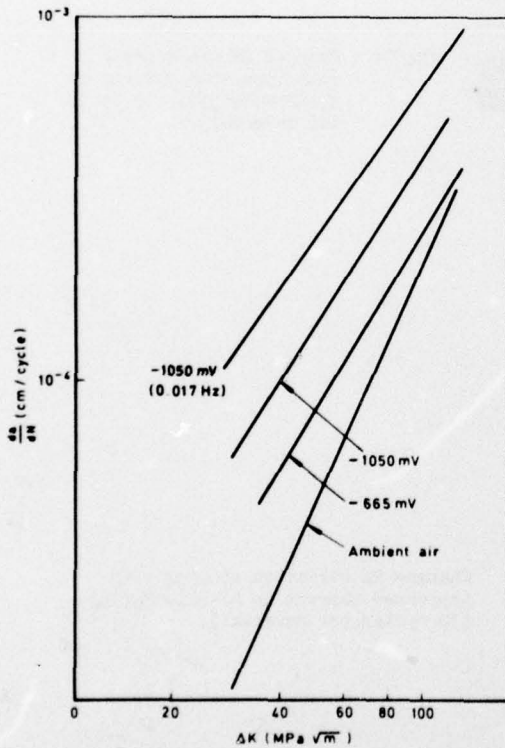


Fig 28 Crack growth rates  $da/dN$  versus stress intensity range  $\Delta K$  for HY 130 steel in sea water - effect of cathodic potentials vs Ag/AgCl. ( $R = 0$ ;  $f(\text{air}) = 0.5$  Hz;  $f(\text{sea water}) = 0.17$  Hz except where shown; triangular waveform)(103).

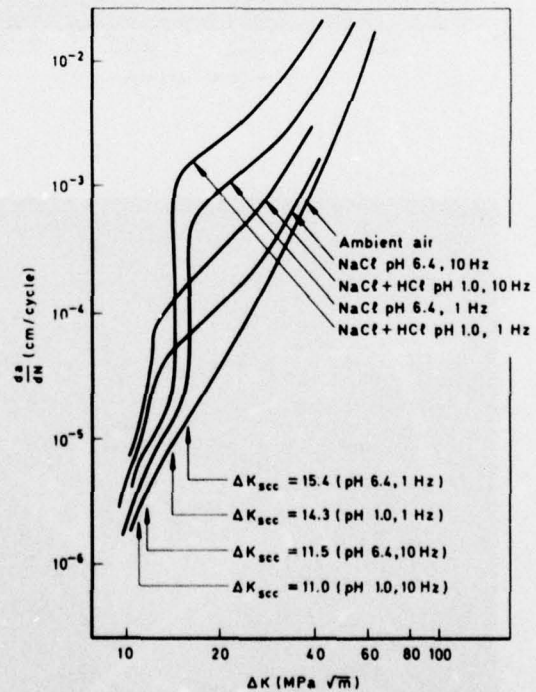


Fig 29 Crack growth rates  $da/dN$  versus stress intensity range  $\Delta K$  for Ti-6Al-6V-2Sn in aqueous 0.6M NaCl solution - effect of pH at different frequencies. ( $R = 0.1$ ; sine waveforms)(22).

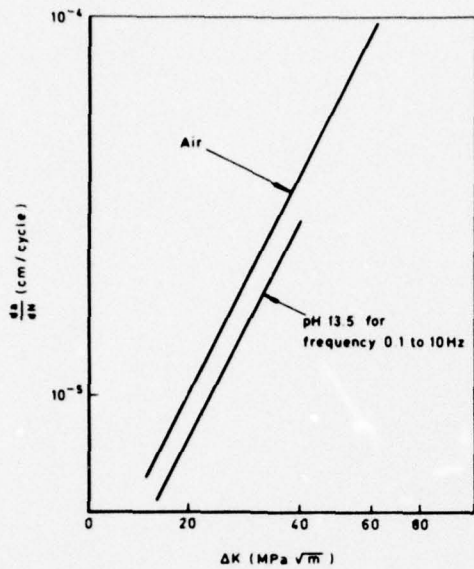


Fig 30 Crack growth rates  $da/dN$  versus stress intensity range  $\Delta K$  for 12Ni-5Cr-3Mo steel in 3% NaCl solution + NaOH-pH 13.5. ( $R = 0.25$ )(73).

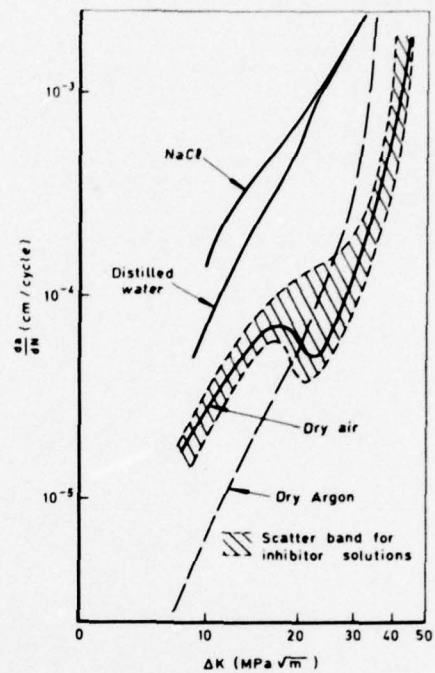


Fig 31 Crack growth rates  $da/dN$  versus stress intensity range  $\Delta K$  for 7075-T6 aluminium alloy - effect of inhibitor additions. ( $R = 0.033$ ;  $f = 1$  Hz (dry argon  $f = 10$  Hz))(106).

ENVIRONMENTAL EFFECTS ON FATIGUE OF ALUMINIUM  
AND TITANIUM ALLOYS

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SUMMARY

Environmental effects on the fatigue strength and crack propagation resistance of aluminium and titanium alloys are reviewed. Particular attention is paid to the influence of air and aqueous environments on aerospace structural materials. Recommendations are made for future research activity, notably with regard to material selection and service performance.

INTRODUCTION

A survey is made of environmental effects on the fatigue of aluminium and titanium alloys, especially the influence of air and aqueous solutions on aerospace structural alloys. The paper consists of the following sections:

- ALUMINIUM ALLOY FATIGUE STRENGTH
  - Vacuum and gas pressure effects
  - Atmospheric humidity effects
  - Prior exposure, fatigue in air
  - Prior exposure, fatigue in aqueous environments; alternating periods of exposure and fatigue
  - Aqueous environment effects
- ALUMINIUM ALLOY FATIGUE CRACK PROPAGATION
  - Vacuum and gaseous effects
  - Atmospheric humidity effects
  - Aqueous environment effects
- TITANIUM ALLOY FATIGUE STRENGTH
- TITANIUM ALLOY FATIGUE CRACK PROPAGATION
  - Vacuum and gaseous effects
  - Atmospheric humidity effects
  - Aqueous environment effects
- DISCUSSION
- RECOMMENDATIONS FOR FUTURE RESEARCH
- REFERENCES

ALUMINIUM ALLOY FATIGUE STRENGTH

Vacuum and gas pressure effects

Many investigations have demonstrated that the unnotched fatigue life of aluminium alloys can be shorter in air than in vacuo. Some results show life differences for up to  $10^7$ - $10^8$  cycles [1-4], e.g. figures 1a, 1b. Other data indicate little or no effect at long lives [1, 5-7], figures 1c, 1d.

Cracks initiate at the same time [3,4,8] or earlier [7,9] in air than in vacuo. Even in the latter instance, however, it appears that the major environmental effect is on crack propagation. For example, figure 1e gives life ratios for  $\sim 15 \mu\text{m}$  crack initiation ( $N_i$ ) and failure ( $N_F$ ) of 1100 Al in air and vacuo; lives in vacuo were about twice those in air, figure 1c. For vacuum fatigue  $N_i/N_F$  is generally much less, and this is most likely due to slower crack propagation [7].

Reduction in air pressure increases fatigue life either according to a sigmoidal curve [2-4, 10], or somewhat more gradually [8,11]. Examples are given in figures 1f and 1g. Sometimes the sigmoidal trend is slight [4,10], e.g. figure 1h. Possible reasons for this are the dependence on frequency [2,3] and stress or strain amplitude [3]. The most significant characteristic of pressure effects is that large increases in fatigue life usually do not occur until the pressure is lowered beyond 10 Pa.

The influence of different gaseous environments follows a consistent trend. Fatigue lives in dry oxygen and nitrogen are essentially equivalent to those in high vacuum [4,9,12]. The same gases containing traces of water vapour are much more detrimental [9], and water vapour by itself is more detrimental than air or air plus water vapour [4,8]. The behaviour of an aluminium alloy in various gases at different pressures is depicted schematically in figure 1i. The detrimental influence of water vapour (alone and in laboratory air) is evident.

Atmospheric humidity effects

The significance of water vapour is reflected in results of fatigue tests in air of varying humidity [8,13-18] and some tests comparing clean and organically coated specimens [9,13,14,16]. Significant differences in fatigue strength up to long lives have been reported for peak aged alloys, figures 2a-2e, but for superpure Al [8] and 2024-T3 Alclad [16] there is little effect at  $10^6$  cycles, e.g. figure 2f.

Fatigue lives have usually been related to relative humidity, but absolute humidity may be the more relevant parameter, cf. identical data in figures 2g and 2h; and figure 2i. Note in figure 2i that coating with vaseline resulted in negligible variation in fatigue life with atmospheric humidity, but that the endurance values corresponded to an intermediate level of humidity. Also, the apparent lack of effect of humidity shown in figure 2j is interesting: it was found that higher humidity increased crack propagation rates but decreased crack initiation density and hence the likelihood of coalescence to form a macrocrack



in such a way that there was no significant dependence of fatigue life on relative humidity [18].

#### Prior exposure, fatigue in air

Tests in alternately moist and dry air [14,15] indicated no effect of humidity until fairly late in the life. This result is consistent with a dominating influence on crack propagation. However, in more aggressive environments the situation may change, as is most clearly borne out by exposure before fatigue testing, figure 3. Surface pitting accelerating crack initiation appears to be responsible for the fatigue life reductions [19-21]. Cladding is highly beneficial [22], but anodizing, at least in the long term, is only slightly so, figure 3h.

Prior exposure results in very large reductions in fatigue life of smooth specimens of high strength alloys, figures 3a-3c, the difference increasing with longer lives, figure 3c. Lesser, but still large reductions occur for notched elements, figures 3d-3f, and lower strength materials, figure 3g. The greatest effect usually is achieved early in the exposure period. There may be exceptions, e.g. the behaviour of bolt-filled countersunk holes in figure 3f.

The data in figure 3e are of particular interest because a realistic stress concentration and load history were employed for several aircraft structural materials. The fatigue performance of all the alloys was markedly degraded by prior alternate immersion in salt water. The controlling parameter is probably overall corrosion resistance (only fair in all cases) and hence surface pitting, as mentioned previously, since for 2024 and 7075 ageing to T851 and T73 conditions, which have greatly increased resistance to exfoliation and stress corrosion, did not improve performance with respect to control life.

#### Prior exposure, fatigue in aqueous environments; alternating periods of exposure and fatigue

Effects of exposure followed by fatigue in aqueous environments and of alternate periods of exposure and fatigue are shown in figure 4. With respect to air an aqueous environment nearly always shortened the total life of pre-exposed specimens. A (slight) exception is 2024-T851 in figure 4a. There is also a tendency for both crack initiation and propagation lives to be shortened. The severity of fatigue in an aqueous environment is emphasized by the data for "Alodine" coated specimens, which with presumably superior surface conditions after exposure failed earlier than uncoated specimens tested in air.

A few data show that alternate periods of exposure and cycling are less severe than exposure before fatigue, figure 4c. Differences would probably be smaller for longer exposures. Figure 4d compares clad and bare 2024-T3 and 7075-T6 fatigued indoors or exposed to seacoast air and cycled daily. The atmospheric effects reduced the average lives of bare specimens by a factor of 3, by about 1.5 for clad 7075-T6, and had negligible influence on clad 2024-T3. The overall performance of 2024-T3 was slightly superior to that of 7075-T6.

#### Aqueous environment effects

By far the majority of environmental fatigue tests on aluminium alloys have involved more or less continuous access to an aqueous environment during the cycling of smooth specimens, usually under rotating or reversed bending. For salt water or spray, the most commonly used environments, the high cycle fatigue strength is typically 0.4-0.6 of that in air, there being a slight tendency to lower values in higher strength materials, figure 5a. Reductions in fatigue life appear to be independent of material strength above 250 MPa, figure 5b, and tend to be very large. The wide gap between the trend lines for  $10^6$  and  $5 \times 10^6$  cycles is probably a consequence of much increased exposure time allowing more severe corrosion and hence facilitating crack nucleation. Nevertheless, simultaneous exposure and cycling seem to be synergistically detrimental. Figure 5c compares the life reductions with those for prior exposure of two high strength (>230 MPa) alloys. Despite long prior exposure the alloys tested in air were less affected than high strength materials subjected to simultaneous exposure and fatigue for periods ~1-7 hours only.

Limited data for saline environments indicate that fatigue strength and life reductions are not much affected by mean stress variation, figure 5d, or by changing from rotating bending to axial loading [33]. For example, the direct stress,  $R = 0$ , high cycle unnotched fatigue strength of various product forms of 7050 alloy in salt fog is 0.45 - 0.5 of that in air [37]. Of more importance are geometrical effects, namely specimen shape and stress concentrations. Fatigue life reductions are much larger in sheet than in round specimens, figure 5e, presumably because of the greater amount of material exposed to the environment. Notched fatigue tests,  $K_t = 1.9$  and 3, on round specimens in salt water or spray gave high cycle fatigue strengths about 0.7 of those in air [25,34], e.g. figure 5f. However, for notched sheet, figure 5g, this ratio is only 0.5.

The concentration of salt in the environment does not appear to be critical, since the data in figures 5a-5g were obtained for a very wide range of salinity. However, the importance of there being some salt present is made clear by tests in water, giving high cycle unnotched fatigue strengths 0.6-0.8 of those in air [28,32,39].

Anodizing and cladding lessen the reductions in fatigue strength and life in aqueous solutions [31, 32,39-43], e.g. figures 5h and 5i, but both treatments often lower the fatigue properties in air [32, 42-46]. Painting is free of this disadvantage [32,47], e.g. figure 5h, and so are porous anodic layers impregnated with polar organic compounds [48].

Last, figures 5j-5l present environmental fatigue data for realistic stress concentrations and load histories. In air the spectrum loaded materials exhibited mean crack initiation and propagation lives equivalent to between  $5 \times 10^5$  and  $4 \times 10^6$  cycles. Salt water reduced the lives by 10-65%. Although important, these reductions are somewhat less than those for notched constant amplitude tests, figures 5f, 5g, and considerably less than those for smooth specimens, figure 5b. Figure 5l gives results of load history simulation for specimens representing landing gear cylinders. Exposure to salt fog caused the peak aged alloy cylinders to fail early by stress corrosion [50]. 7075-T73 and X7080-T7 cylinders failed by fatigue originating from the unexposed bores: life reductions compared to continuous cycling in air were apparently due to the hold periods, as evidenced by a 7075-T73 cylinder test in air with hold times.

## ALUMINIUM ALLOY FATIGUE CRACK PROPAGATION

Vacuum and gaseous effects

At low stress intensities fatigue crack propagation rates in vacuo may be up to 10 times slower than in normal air, figures 6a-6d. For  $R \geq 0$  the differences become negligible at  $\Delta K$  values above 15-20  $\text{MPa}\sqrt{\text{m}}$ , but reversed stressing allows a large difference to be maintained to higher  $\Delta K$ , figure 6d.

Crack rates in dry gases are similar to those in vacuo, figures 6e-6f. Humidity has a dominating influence, amply demonstrated by tests in dry and wet gases [51,58,59] and in water vapour alone, figures 6e,6f. Exerted in normal air this influence is greater on 7000 series alloys than on 2000 series alloys, figure 6g.

In vacuo and very dry gases crack rates are usually independent of cycle frequency [51,55,60], e.g. DTD 5070A and 7075-T651 in figures 6a and 6c, but an intrinsic frequency effect was reported for DTD 683, figure 6a. Test temperature has a significant influence on crack rates at lower  $\Delta K$ , figure 6h. The apparent activation energy depends strongly on  $\Delta K$ , ranging from 2 kJ/mole for  $\Delta K = 15.5 \text{ MPa}\sqrt{\text{m}}$  to about 8 kJ/mole for  $\Delta K = 10 \text{ MPa}\sqrt{\text{m}}$  [54].

Atmospheric humidity effects

Tests in air indicate that for 7000 series alloys there are usually higher crack rates and greater sensitivity to humidity than in 2000 series alloys [49,51,60-67], e.g. figures 7a-7f. For high frequencies the crack rates in nominally dry air at low  $\Delta K$  may be an order of magnitude slower than in wet air, figures 7a, 7b, the differences becoming small at  $\Delta K$  values of 20-30  $\text{MPa}\sqrt{\text{m}}$ , depending on  $R$ . However, the behaviour is frequency dependent. Figures 7g and 7h show that reducing the frequency increased crack rates more in dry air than in wet air at low  $\Delta K$ , but did the opposite at high  $\Delta K$ . A similar result was found for other  $R$  values [61]. Equally important at low  $\Delta K$  is the amount of water vapour in the environment [1,51,58,59,61], e.g. figure 7i, which illustrates that the greatest effect of increasing humidity is achieved at lower water vapour contents.

The effects of frequency and humidity are interrelated. The influence of water vapour content on crack rates at low  $\Delta K$  and two frequencies is depicted in figures 7j and 7k. The crack rate data often exhibit a sigmoidal trend [1,51,57,60], with a critical pressure above which the rates are increased only slightly. The critical pressure increases with increasing frequency and crack rates [51,60] and decreases for higher  $R$  [57]. It varies also for different alloys: in figure 7j the critical pressure for 7075-T6 is about 10 Pa, and that for 2024-T3 is ~500 Pa. The overall behaviour is such that there should be relatively little variation in crack growth rates at low frequencies and/or in natural air ( $\geq 5\%$  R.H.). That this is so is evident from constant amplitude tests at 0.0167, 1 and 100 Hz in air with water vapour partial pressures of 0.67 - 1.5 kPa, figure 6a; from random and programmed load tests at 0.33 Hz indoors (40-60% R.H.) and outdoors (70-90% R.H.) [63]; and from flight simulation tests at 0.1, 1 and 10 Hz in air of 40-60% R.H. [68].

Constant amplitude tests in normal and nominally dry air demonstrate that crack rates frequently tend to be higher in thicker sections [37,38,69-73]. A similar trend is observed for flight simulation, figure 7l. For constant amplitude loading the thickness effect is basically related to that of the environment [64,74]. In a reactive environment crack rates are higher for flat fracture than for slant fracture, and increasing thickness postpones the flat-to-slant transition to higher  $\Delta K$  [64,74]. This transition is also postponed by more aggressive environments [73,75]: increasingly reactive environments maintain crack rate differences with respect to a mild environment to higher  $\Delta K$ , e.g. figure 7i.

For flight simulation additional factors are involved in the thickness effect. Crack growth delays due to peak loads are reduced by increasing thickness because plastic zone sizes and hence residual compressive stresses and crack closure are less [76-78]. Also, lower fracture toughness in thick sections may result in static crack extension at peak loads, e.g. the large crack rate alternations for 10 mm 7075-T6 in figures 7f, 7l.

Aqueous environment effects

Most tests with aqueous environments have used salt water. In this environment average crack rates are 2-4 times higher than in air for  $\Delta K$  values between 4 and 30  $\text{MPa}\sqrt{\text{m}}$ , figure 8a: at very low and very high  $\Delta K$  there is no difference [79]. Different  $R$  values result in air and salt water crack rate curves being shifted by similar amounts, figure 8b.

7000 series alloys are more sensitive than 2000 series alloys to the change from air to aqueous environments [49,64,75,78,82-84] and, as in air, generally have higher crack rates, e.g. compare figures 8c and 8d; and figures 8e-8g. Figure 8e also illustrates that despite various processing histories the ageing treatments tend to determine crack propagation resistance in an aqueous environment. For 7000 series alloys overageing usually increases crack propagation resistance [56, 85-87]: the data in figure 8d for  $\Delta K > 9 \text{ MPa}\sqrt{\text{m}}$  are an exception. For 2000 series alloys the crack propagation resistances of naturally aged materials (2024-T3 in figure 8c, UHP 2000-T36 and T4 in figure 8e) are greater than those of artificially aged ones. In addition to overageing, higher copper contents are beneficial for 7000 series alloys [85,88].

Frequency effects are usually small in aqueous environments [55,56,65,79,89], especially if an alloy is immune to stress corrosion, e.g. 2219-T87 in figure 8h. However, the same figure shows that an alloy highly susceptible to stress corrosion (7079-T651) and loaded in the short transverse direction exhibits an extreme frequency dependence at frequencies low enough to allow stress corrosion.

Tests on 7075-T651 in distilled water demonstrated that crack rates are very dependent on temperature at low  $\Delta K$ , much less so at  $\Delta K \sim 9-10 \text{ MPa}\sqrt{\text{m}}$  [54]. The apparent activation energy is considerably lower than for dry gaseous environments at equivalent  $\Delta K$ , and ranges from 3.8 kJ/mole for  $\Delta K = 10 \text{ MPa}\sqrt{\text{m}}$  to 15.5 kJ/mole for  $\Delta K = 6.6 \text{ MPa}\sqrt{\text{m}}$  [54]. A strong temperature dependence has also been found for 7075-T6 tested in salt water [73].

The effect of specimen thickness on crack growth rates in aqueous environments appears to be slight [37,38,74,90] or negligible [37,38,78] for constant amplitude loading. On the other hand, there are significant thickness effects for flight simulation, figure 8i. Basically this is because in thicker sections peak loads cause shorter delays in crack growth and sometimes static crack extension, as mentioned previously.

An unusual effect of aqueous environments is elimination of crack rate differences between longitudinal specimens taken from various thick section mill products. Figures 8j and 8k show that in dry air the crack propagation behaviour of forgings and extrusions of 7050 alloy was very different from that of plate, but in salt water or fog the differences virtually disappeared. The slower rates for forgings and extrusions in dry air are due to crack reorientation leading to growth in the longitudinal direction [37]. Presumably this change was suppressed by the saline environments. Although these environments had an equalising effect on crack rates in longitudinal specimens, specimens of different orientations maintained differences in crack propagation behaviour, e.g. figure 8l.

#### TITANIUM ALLOY FATIGUE STRENGTH

The fatigue strength of titanium alloys is only slightly altered, if at all, by variations in atmospheric humidity [91], outdoor exposure [92], and changing from air to aqueous environments [34,49,93,94], figure 9. This environmental insensitivity is attributable to fairly thick [95], highly protective and fast healing [96] oxide films. Figure 9d demonstrates the importance of oxide films by including test data for a methanol-HCl environment, in which a passive film is absent [97].

#### TITANIUM ALLOY FATIGUE CRACK PROPAGATION

##### Vacuum and gaseous effects

Crack propagation behaviour of titanium alloys in vacuo and in gaseous environments is very diversified owing to the marked influences of heat treatment, texture, and a transition from faceted, structure sensitive fracture at low growth rates to structure insensitive crack growth with the material acting more like a continuum [98-107]. Combinations of these influences with that of the environment are responsible for three types of crack propagation behaviour:

- (1) Threshold  $\Delta K$  values. Figures 10a and 10b present data for weakly textured ( $\alpha\beta$ ) heat treated Ti-6Al-4V. There are well-defined threshold  $\Delta K$  values for fatigue in vacuo but not in air. The same material  $\beta$  annealed (randomized texture) gave an indication of a threshold  $\Delta K$  of  $5.5 \text{ MPa}\sqrt{\text{m}}$  in air at  $R = 0.35$  [99].

The origin of threshold  $\Delta K$  values is a transition from highly irregular and unsustained faceted fracture to structure insensitive crack growth [98]. This transition seems to correspond with the cyclic plastic zone size exceeding the grain size [98]. Thus a larger grain size and higher yield strength should increase the threshold  $\Delta K$ , but ideally there should be no effect of  $R$ . Figure 10a shows that material in the highest strength condition (annealed, quenched and aged) had the lowest threshold  $\Delta K$ : apparently, microstructural refinement by heat treatment prevailed over the strength increase [98]. In figure 10b the threshold  $\Delta K$  is seen to be slightly dependent on  $R$ .

- (2) Abrupt increases in crack growth rates. For annealed Ti-6Al-4V and Ti-8Al-1Mo-1V plate a strong {0002} orientation in the macroscopic crack plane resulted in sustained faceted fracture (cleavage on or near {0002}, {106}) in vacuo even at very low growth rates, figure 10c. However, there was an abrupt increase in growth rates at the transition from cleavage to structure insensitive propagation.

Sudden crack rate increases at the transition to structure insensitive propagation have also been found for Ti-6Al-4V and Ti-5Al-2.5 Sn L-T sheet specimens in argon, figures 10d, 10e. The Ti-6Al-4V had a strong transverse texture ({0002} in the T-L crack plane) and the faceted fracture was not cleavage [105] but possibly cracking along slip planes. The Ti-5Al-2.5 Sn texture was moderate with {0002} tending to be in the plane of the sheet: nevertheless, faceted fracture was predominantly cleavage [107].

Probably in the same category is the sudden increase in crack growth rates for Ti-5Al-2.5 Sn in nitrogen, figure 10f. Here only a macroscopic change in fracture surface roughness was reported [111]. A similar but less pronounced transition was observed for commercial purity titanium in nitrogen, figure 10g.

- (3) No abrupt alterations in crack rates. This is the most common type of behaviour accompanying transition from faceted fracture to structure insensitive cracking, and is favoured in more reactive environments, e.g. annealed, quenched and aged Ti-6Al-4V in air, figure 10a; Ti-6Al-4V and Ti-8Al-1Mo-1V in air, figure 10c; Ti-6Al-4V and Ti-5Al-2.5 Sn L-T specimens in air, figures 10d, 10e; Ti-5Al-2.5 Sn in air and oxygen, figure 10f; and commercial purity Ti in air, water vapour and oxygen, figure 10g. Note that crack propagation curve inflexions are not entirely absent for annealed Ti-6Al-4V in air, figure 10a, and Ti-5Al-2.5 Sn in water vapour, figure 10f.

Also, there is no crack propagation curve inflexion for Ti-6Al-4V T-L specimens in argon, figure 10d: here cleavage gave way gradually to structure insensitive crack growth. Possibly the residual moisture content of the argon facilitated a gradual transition, which contrasts with that of similarly textured and oriented specimens of Ti-6Al-4V and Ti-8Al-1Mo-1V in vacuo, figure 10c.

For relatively inert environments a weak texture may preclude a sudden increase in crack rates at the transition to structure insensitive cracking. This possibility is suggested by the behaviour of nearly isotropic (i.e. weakly textured) Ti-6Al-4V in very carefully dried argon, figure 10h.

Another kind of behaviour has been somewhat arbitrarily defined: namely, a threshold  $\Delta K$  below the transition from faceted to structure insensitive fracture and at which crack growth rates become insignificant [100]. On this basis threshold  $\Delta K$  values are exhibited by Ti-6Al-4V and Ti-8Al-1Mo-1V in air, figure 10c, although the results in vacuo indicate that true thresholds do not exist.

As is to be expected from the foregoing considerations different alloys tested under identical conditions show significant differences in environmental sensitivity and in crack growth rates in a particular environment, figure 10e and of figures 10f, 10g. Comparison of data for air and argon in figure 10e indicates that differences between alloys are less in air.

Crack propagation rates are always slower in vacuo, figures 10f-10h. Titanium alloys are very sensitive to the residual moisture content of nominally dry gases [112] and even to the degree of vacuum at very low pressures, figure 10h. Thus, although figures 10f and 10g depict different crack rates in oxygen and nitrogen at low  $\Delta K$ , these differences may be mainly due to variations in residual moisture content.

From tests in argon and air it appears that crack rates in mild or nominally inert gaseous environments are virtually independent of cycle frequency [100, 112-114], e.g. figure 10i. Also, test temperature has been found to have negligible influence on Ti-6Al-4V crack rates in argon, figure 10j, and in vacuo at 212 °K and room temperature [112]. However, crack rates in hydrogen are markedly frequency [115, 116] and temperature [117, 118] dependent, figures 10k, 10l.

The frequency dependence in hydrogen is related to microstructure, heat treatment and the susceptibility to static crack growth. Figure 10k shows that at 0.5 Hz hydrogen enhanced crack growth, as compared to fatigue in vacuo, occurred only for the fine acicular microstructure. At 0.005 Hz hydrogen enhanced crack growth occurred for the coarse acicular microstructure. In both cases accelerated crack propagation appears to depend greatly on allowing sufficient time for static fracture. Also, for mill annealed Ti-6Al-4V crack rates were similar in hydrogen, oxygen and argon at 5 Hz, figure 10h; but at a fairly equivalent temperature (297 °K) and frequency (8.33 Hz) crack rates were much higher in hydrogen than in helium for ( $\alpha+\beta$ ) solution treated and aged Ti-6Al-4V, figure 10l.

The temperature dependence in hydrogen also varies with microstructure. At 297 °K maximum enhancement of crack growth in Ti-6Al-4V by hydrogen was greater for weld metal than for ( $\alpha+\beta$ ) solution treated and aged material, but the opposite at 200 and 255 °K [118].

#### Atmospheric humidity effects

The importance of humidity is evident from the results for normal air and water vapour in figures 10f and 10g and from tests in air of different moisture contents, figures 11a, 11b. However, for Ti-6Al-4V crack growth delays due to peak loads have been found to be little affected by changing the environment from dry argon to air of 30 - 60 % relative humidity [122].

Ti-6Al-4V in dry [72, 104] and normal [98, 99, 123-126] air and Ti-6Al-6V-2Sn in normal air [127, 128] consistently exhibit the lowest crack growth rates in the  $\beta$  processed [127] or  $\beta$  heat treated [72, 98, 99, 104, 123-126, 128] conditions. In comparison to the mill annealed condition recrystallization annealing is generally beneficial for Ti-6Al-4V [104, 124, 126], but only slightly [129] or not at all [126, 130] beneficial for Ti-6Al-6V-2Sn. Figure 11c compares Ti-6Al-4V plate materials in the same three heat treatment conditions in dry and normal air. In each environment there were large differences in crack rates. Further, the relative significance of atmospheric humidity is to some extent indicated by the fact that owing to differing R values and possible material and specimen differences (e.g. texture and orientation) the crack rates for each heat treatment condition were higher in dry air.

Tests in normal air at different temperatures might be expected to alter the influence of water vapour. However, temperature effects on crack propagation are either slight until higher  $\Delta K$  [131-133] or negligible [131], e.g. figure 11d. The lack of temperature dependence at lower  $\Delta K$  is possibly caused by compensatory changes in the effects of deformation and water vapour [112]. At higher  $\Delta K$  the generally slower crack growth with increased temperature probably results from greater plasticity, which is also most likely responsible for the increase with temperature of crack growth delays due to peak loads [134].

#### Aqueous environment effects

Many constant amplitude data [72, 82, 103, 105, 108, 110, 113, 114, 127, 135-146] show that in the absence of cyclic stress corrosion the crack rates in salt water are 1-4 times faster than in air over the frequency range 0.1-50 Hz, e.g. figure 12a, which also illustrates the marked influence of stress corrosion on alloy ranking.  $\beta$  heat treated Ti-6Al-4V is one of the most resistant materials, with aqueous environment crack rates only 1-2 times higher than in dry air [72, 147, 148], figure 12b; note that the effect of R is not simply to shift crack rate curves for each environment by similar amounts.

Frequency effects in the absence of stress corrosion were found to be small [114, 138] or insignificant [140, 149] for Ti-6Al-4V [149], Ti-6Al-6V-2Sn [114] and Ti-8Al-1Mo-1V [138, 140, 149] tested in distilled and salt water. However, a strong frequency dependence in these environments has been observed for IMI 230 (Ti-2.5 Cu), figure 12c. The effect of temperature has not been thoroughly investigated; for Ti-6Al-4V crack rates were essentially the same in distilled water at 293 and 358 °K [112].

Flight simulation loading results in greater differences between crack rates in air and aqueous environments [49, 121, 147]. Figure 12d demonstrates that crack rates for Ti-6Al-4V were up to 30 times faster in salt water than in air, and the overall difference was such that the influences of specimen orientation and texture were relatively small. This large environmental effect was most likely caused by the basically faster crack propagation in salt water reducing crack growth delays due to severe flights and also reducing the number of severe flights encountered.

Cyclic stress corrosion in constant amplitude tests [105, 110, 114, 135-138, 140, 141, 143, 144, 149] leads to crack rates typically 5-20 times faster than in air. Much higher accelerations can occur from the combination of a high R and low frequency [138]. Frequency variation strongly affects crack rates [114, 140, 149], e.g. figures 12e, 12f, and also the cyclic stress corrosion threshold stress intensity factor range,  $\Delta K_{Isc}$ , which increases with decreasing frequency owing to the competition at the crack tip between repassivation and cyclic deformation exposing fresh metal surface to the environment [114].

In  $\alpha$  and  $\alpha-\beta$  titanium alloys aqueous stress corrosion is characterised by cleavage at 13-17 ° from (0002) [150-153]. For Ti-6Al-4V undergoing cyclic stress corrosion in salt water this anisotropic fracture mode resulted in a pronounced effect of texture on crack propagation, figure 12g. In transverse specimens fracture was almost entirely cleavage and crack rates were up to 50 times higher than in air; but for longitudinal specimens, whose fracture surfaces consisted of ductile fatigue striations linking occasional cleavage facets, crack rates were no more than 8 times faster than in air.

Finally, stress corrosion has been implicated in the frequency dependence of crack growth delays in Ti-6Al-4V [122, 134]. Figure 12h summarises the results. The top left diagram shows that for cycling at 5 Hz delays were shorter in salt water than in air, but that reducing the high load frequency in salt

water increased subsequent delays. At top right are shown delays following manually applied high loads, which included hold times of 0.5-30s at maximum load [122]: the behaviour in salt water was anomalous in that for increasing  $K_{I\max}$  the delays decreased initially and then increased to a constant value, instead of monotonically decreasing. The lower diagram depicts the influence of hold time on subsequent delays in salt water, and demonstrates a much larger influence at the higher stress intensity levels. All these effects probably have their origin in stress corrosion, which was observed during high load cycles [134]. It was suggested that increased delay for longer hold times was caused by stress corrosion extending the crack through a fatigue-damaged zone into material more resistant to subsequent fatigue fracture [134]. It is also possible that increased delays resulted from stress corrosion crack branching, since fatigue crack growth in Ti-6Al-4V is slower if secondary cracks are present [104,125].

#### DISCUSSION

Aerospace programmes go through three main stages [154]: design, prototype construction and test flights, and service use. In each stage the significance of fatigue and environmental effects varies. Table 1 broadly surveys programme aspects involving fatigue and likely or certain to include environmental effects.

TABLE 1  
SURVEY OF AIRCRAFT FATIGUE ASPECTS LIKELY OR CERTAIN TO INCLUDE ENVIRONMENTAL INFLUENCES

DESIGN STAGE	<ul style="list-style-type: none"> <li>• materials selection</li> <li>• structural concepts</li> <li>• corrosion protection systems</li> <li>• estimation of fatigue lives and crack propagation lives</li> <li>• exploratory fatigue tests for design studies and corroboration of life estimates</li> </ul>
PROTOTYPE CONSTRUCTION AND TEST FLIGHTS	<ul style="list-style-type: none"> <li>• component or full-scale test</li> <li>• materials change</li> <li>• structural modifications</li> <li>• inspection procedures for service</li> </ul>
AIRCRAFT IN SERVICE	<ul style="list-style-type: none"> <li>• corrections to predicted fatigue properties</li> <li>• relation of service cracking to predicted damage</li> <li>• materials change and/or replacement</li> <li>• structural modifications</li> <li>• corrosion protection systems modifications</li> <li>• revised inspection and maintenance procedures</li> </ul>

- (1) Design stage. Environmental influences on fatigue strength are not usually considered in the design stage since corrosion protection systems are applied to actual structures. However, these protection systems may rupture, especially during crack growth in the underlying metal. Thus in fatigue crack propagation analysis and testing attention should be paid to environmental effects.

There are several guidelines for 2000 and 7000 series aluminium and  $\alpha$ - $\beta$  titanium alloy selection for environmental fatigue crack propagation resistance. For aluminium alloys:

- 2000 series alloys are generally more resistant than 7000 series alloys
- for the 2000 series naturally aged materials are more resistant than artificially aged materials
- for the 7000 series overageing usually increases resistance
- higher Cu contents are beneficial to 7000 series alloy resistance

For  $\alpha$ - $\beta$  titanium alloys [155]:

- $\beta$  processed and/or  $\beta$  heat treated materials tend to be the most resistant
- texture control to result in {0002} parallel to the principal stress axis should be beneficial for ( $\alpha$ + $\beta$ ) processed and heat treated alloys
- recrystallization annealing can be beneficial with respect to mill annealing
- lower oxygen and hydrogen contents can improve resistance

Material selection may also be influenced by the structural concept. For example, compared to equivalent monolithic structures honeycomb sandwich and laminated constructions provide increased crack propagation resistance by load shedding to uncracked face sheets and laminations and by decrease in gauge [156-161]. This structurally increased resistance may permit use of materials intrinsically somewhat inferior in fatigue crack propagation resistance but advantageous in other respects. Another example is given by composite-reinforced structures. In the presence of aqueous media aluminium corrodes severely when exposed (as could occur during fatigue cracking) to carbon-epoxy composite [162,163]. This problem would be mitigated if it were feasible to use titanium instead of aluminium [162,163] or to use other composites, e.g. boron-epoxy [164] and Kevlar-epoxy [165].

- (2) Prototype construction and test flights stage. In this stage environmental simulation during some phase of full-scale testing might be considered. Correlations between fatigue defect lives for full-scale tests and service [166-171] have revealed that tests, which are generally done in laboratory air, gave average lives 3-5 times longer [172]. Also, many service defects had no test equivalents [169]. The main reason for these disparities appears to be that test loads did not represent service experience, but differences between test and service environments accounted for a significant minority of cases [169]. However, environmental simulation during fatigue life testing is impractical because testing time is much shorter than the anticipated service life and it is a sine qua non for corrosion protection systems that they maintain their effectiveness for at least a major part of the service life, i.e. the test structure would experience very little or no corrosion.

Because corrosion protection systems are likely or certain to rupture during crack growth in the underlying metal, environmental simulation could well be employed during full-scale fatigue crack propagation testing in order to obtain more realistic assessments of inspection procedures and intervals and possibly to disclose the need for material changes and/or structural modifications.

- (3) Aircraft in service. Service experience amply demonstrates that despite all precautions corrosion will occur and that in-service control of corrosion is necessary. Corrosion has been and is continuing to be a severe problem for some older types of aircraft, especially since there is a tendency to extend service lives because of financial constraints [173].

Corrosion is clearly detrimental to the fatigue properties of high strength aluminium alloys, and the consequences of in-service corrosion are that life estimations must be reassessed and alterations made to inspection and maintenance procedures. Material changes and/or structural modifications may also be necessary.

There are few available data for enabling life reassessments owing to corrosion representative of service use. Such data are probably obtained mainly on an ad hoc basis, but it would be worthwhile to conduct investigations of more general applicability.

Several parameters are involved in environmental effects, namely the environment per se; ambient pressures and temperatures; and loading types and frequencies, table 2.

TABLE 2  
PARAMETERS INVOLVED IN ENVIRONMENTAL EFFECTS

ENVIRONMENT	<ul style="list-style-type: none"> <li>• air of different humidities</li> <li>• periodic or continuous wetting with aqueous solutions</li> <li>• jet fuel, hydraulic fluid</li> </ul>
PRESSURE	<ul style="list-style-type: none"> <li>• ambient</li> <li>• periodic reductions representative of flight conditions</li> </ul>
TEMPERATURE	<ul style="list-style-type: none"> <li>• ambient</li> <li>• cycling representative of flight conditions</li> </ul>
LOADING TYPE	<ul style="list-style-type: none"> <li>• constant amplitude</li> <li>• programme</li> <li>• random</li> <li>• flight simulation</li> </ul>
LOADING FREQUENCY	<ul style="list-style-type: none"> <li>• "real time"</li> <li>• accelerated</li> </ul>

Variation of atmospheric humidity within natural limits and at the fairly low frequencies (< 20 Hz) typical of most types of aircraft loads [174] appears to be relatively unimportant for aluminium alloy fatigue crack propagation. Titanium alloys might be more sensitive in this respect. Choice of aqueous environments is moot: salt water is probably too aggressive, especially if continuously applied; on the other hand an alternative such as rainwater is variable in pH and e.g. chloride content. For bilge areas sump tank water has been recommended [175].

Air pressure reductions do not appear to be necessary. Even at an altitude of 30 km the atmospheric pressure is ~ 1 kPa [176], whereas for aluminium alloys the pressure has to be  $\leq$  500 Pa before significant decreases in fatigue crack propagation rates occur.

Air temperature rapidly decreases to 218 °K at an altitude of 10 km [176], such that flights above about 3 km experience below freezing conditions [172]. The temperature remains at 218 °K until 20 km height, followed by a gradual increase to 227 °K at 30 km [176]. At ground level the ambient temperature varies from below freezing to ~ 323 °K [172]. For aluminium and titanium alloys temperature cycling within the foregoing regime will affect mainly the environmental contribution to fatigue crack growth. Higher temperatures increase reaction rates, and testing continuously at laboratory temperatures is, on the whole, likely to be conservative except if aerodynamic heating is significant (e.g. CONCORDE).

For composite-reinforced structures actual or simulated temperature cycling may be necessary at some stage in testing in order to assess the influence of stresses due to thermal expansion incompatibilities. A problem with simulated temperature cycling (applying additional external forces) is that the stresses in metal and composite are of the same sign. However, if crack propagation in the metal is of prime concern this problem may be unimportant.

The type of loading can greatly influence environmental fatigue test results. For example, the effect of sheet thickness on aluminium alloy crack growth rates in aqueous environments was slight or negligible for constant amplitude loading [37,38,74,78,90] but large for flight simulation [78]. Also for Ti-6Al-4V the differences between crack growth rates in air and aqueous environments were greater in flight simulation than for constant amplitude loading.

Aircraft loading frequencies vary widely. Order of magnitude values are listed in table 3.

TABLE 3  
CYCLE FREQUENCIES OF AIRCRAFT FATIGUE LOADS [174]

LOAD CYCLES	FREQUENCIES (Hz)
Ground-Air-Ground	0.00003 - 0.001
Cabin pressurization	0.00003 - 0.0005
Manoeuvres	0.005 - 0.2
Gusts	0.1 - 10
Taxiing	0.5 - 20
Buffeting	10 - 100
Acoustic	100 - 1000

Ground-air-ground and cabin pressurization cycles are far too long to be used in tests, as are some manoeuvre and gust load cycles. However, in the absence of stress corrosion the rates of fatigue crack propagation for aluminium and titanium alloys in air and aqueous environments are usually little affected by frequency variations at frequencies below about 20 Hz. Also, acceleration of low frequency cycles for aluminium alloy environmental fatigue strength tests seems feasible in view of the apparently synergistic effect of simultaneous exposure and cycling.

#### RECOMMENDATIONS FOR FUTURE RESEARCH

From the previous review and discussion it is suggested that environmental influences on fatigue be considered for

- design stage fatigue crack propagation testing and analysis
- full-scale fatigue crack propagation testing
- generally applicable data for fatigue life reductions due to corrosion in service

In the design stage research topics for environmental fatigue crack propagation include the evaluation of

- candidate materials, e.g. titanium as compared to aluminium alloys, 2000 series as compared to 7000 series aluminium alloys
- candidate structures, e.g. monolithic versus laminated or sandwich panels, mechanical fastening versus adhesive bonding, all-metal versus metal/composite
- possible influences of inhibitors in primers and of aluminium alloy cladding

Investigation of fatigue life reductions owing to corrosion is especially relevant for high strength aluminium alloys and should include

- comparison of different corrosion protection and/or fastener systems [21]
- comparison of different geometries, e.g. flat and round specimens [37,38] and specimens with open or bolt-filled holes [20]
- comparison of specimens corroded in the laboratory with those exposed outdoors [20]

Testing variables of prime concern are

- the environment
- loading type
- loading frequency

- (1) The environment. As mentioned before, the choice of environment is debatable. At present, in addition to tests in laboratory air it appears advisable to use readily reproducible neutral aqueous solutions except for very specific service simulations.
- (2) Loading type. Both constant amplitude and flight simulation loading should be employed. Although constant amplitude loading is highly artificial and irrelevant for many aircraft components, the data serve as a basic reference and are a prerequisite for making and improving cumulative damage computations.

Flight simulation is a necessity for full-scale testing [154], but also in other stages, since results are obtained which are not even qualitatively predicted by constant amplitude tests. For comparative testing of lower wing skin materials and structures there are two well-defined reference spectra available. These are the gust spectrum TWIST (Transport Wing Standard) developed by the Laboratorium für Betriebsfestigkeit and the NLR [177]; and the manoeuvre spectrum FALSTAFF (Fighter Aircraft Loading Standard For Fatigue evaluation) [178,179].

- (3) Loading frequency. In the absence of stress corrosion the acceleration of low frequency load cycles to minimise testing time seems reasonable, although more data are needed to corroborate this.

At the NLR the current approach for gust and manoeuvre spectrum loading is to test all-metal specimens at 10 - 15 Hz in dry or normal air and 5 Hz in more aggressive environments. Metal/composite specimens are tested at 5 Hz irrespective of environment in order to avoid possible heating effects in the composite.

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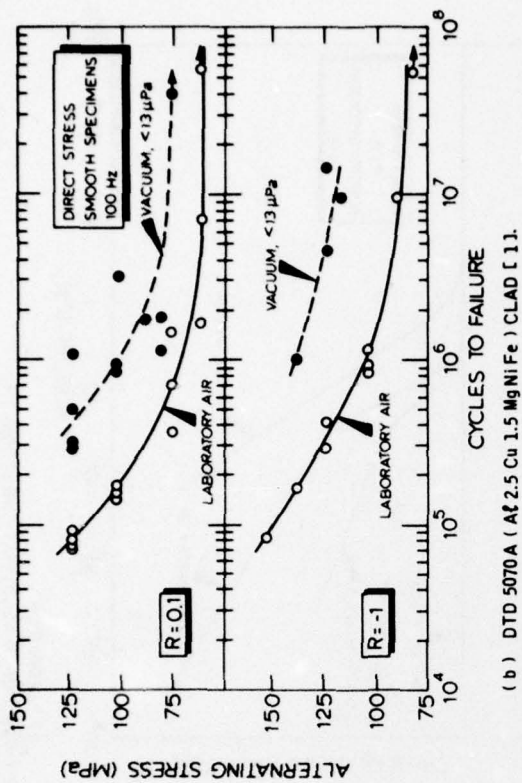
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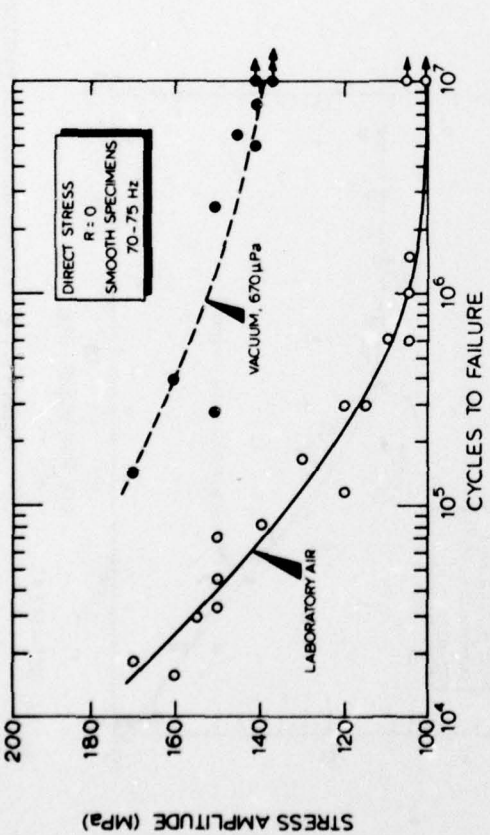
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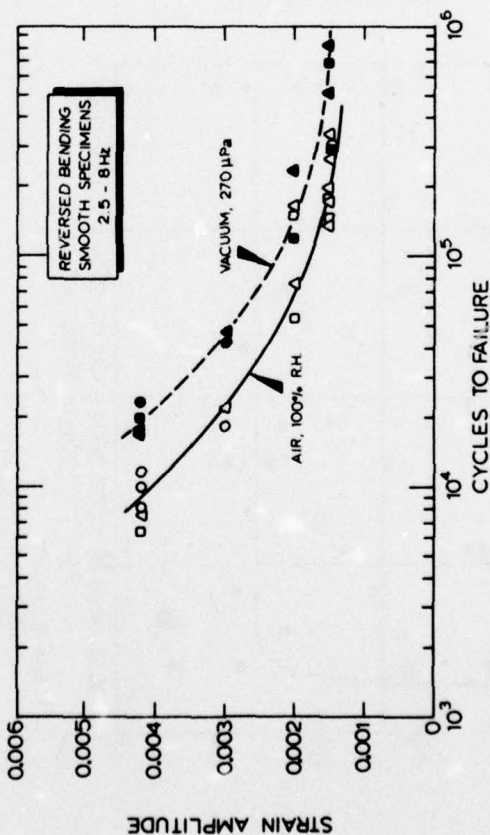
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178. G.M. van Dijk and J.B. de Jonge, Paper 3.6, *Proceedings of the Eighth ICAF Symposium*, ed. J. Branger and P. Berger, Lausanne, June 1975.
179. *Description of a Fighter Aircraft Loading STandard For Fatigue evaluation*, Combined report of the NLR, LBF, IABG and F+W, March 1976.



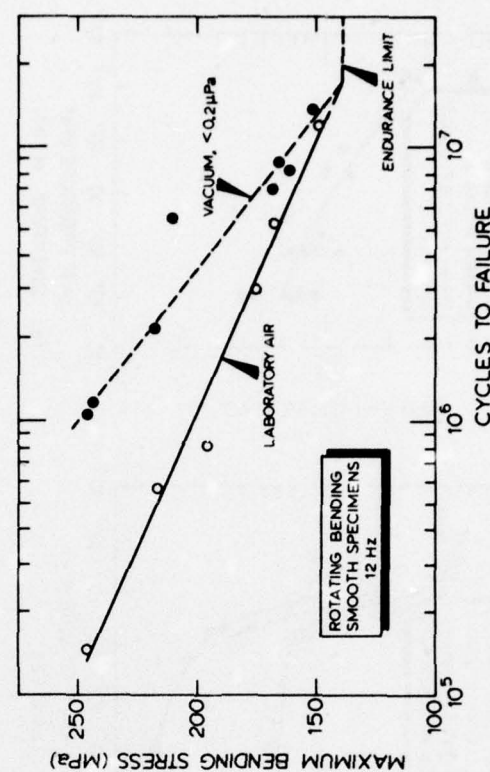
(a) AlMg3 [4].



(b) DTD 5070A (Al 2.5 Cu 1.5 Mg Ni Fe) CLAD [1].

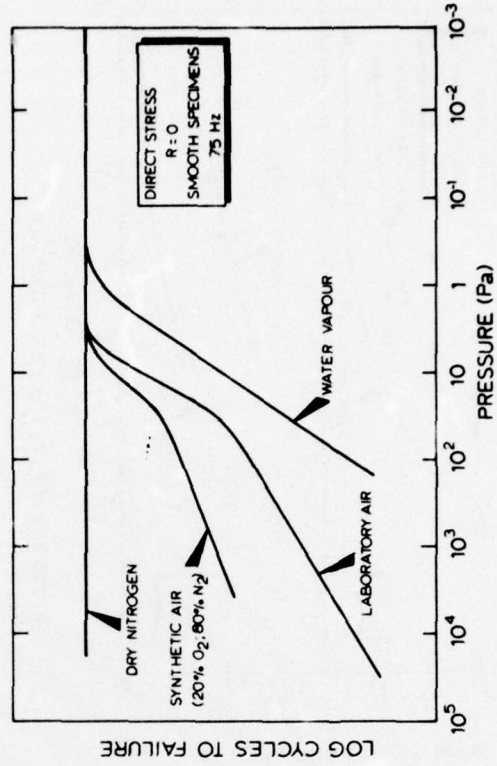
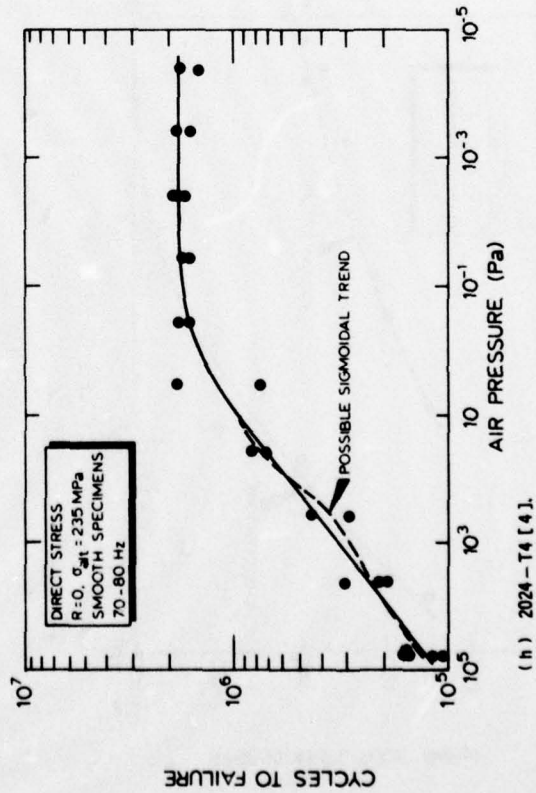
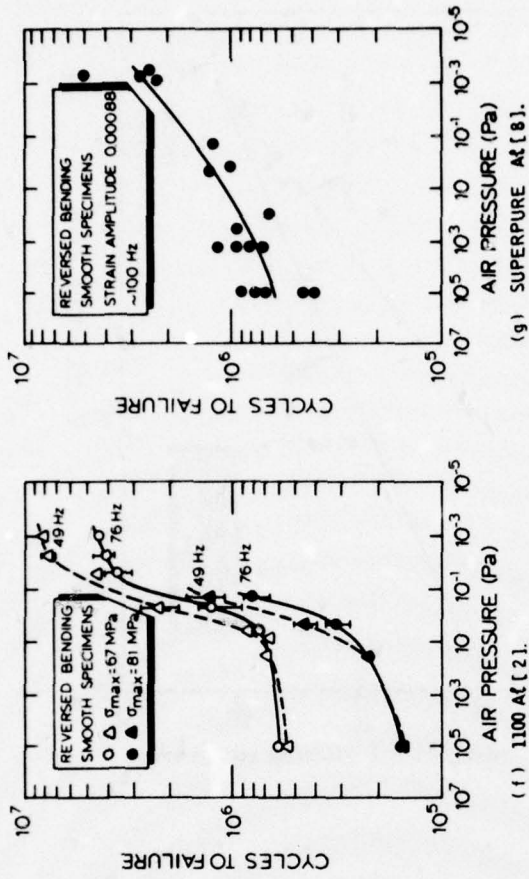
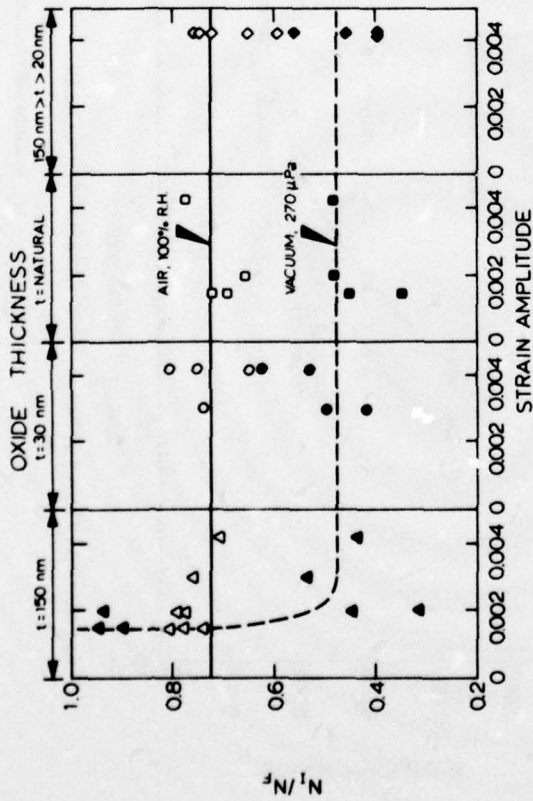


(c) 1100 Al [7]. FOR EXPLANATION OF SYMBOLS SEE FIGURE 1e.

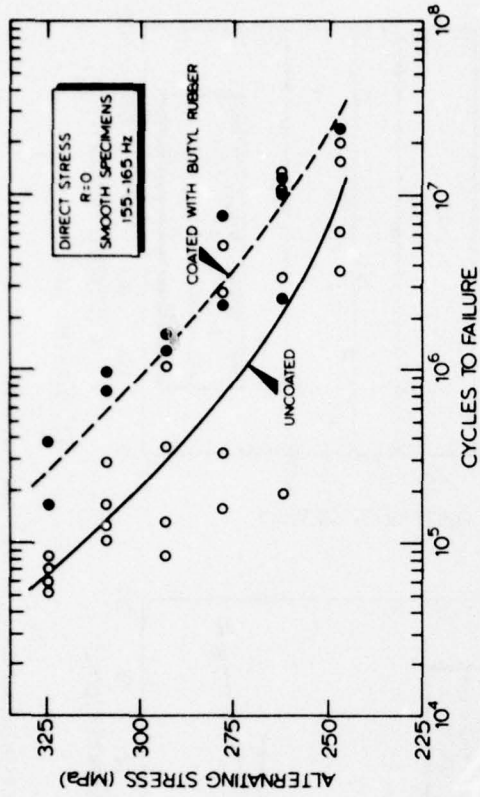


(d) 2024-T3 [6].

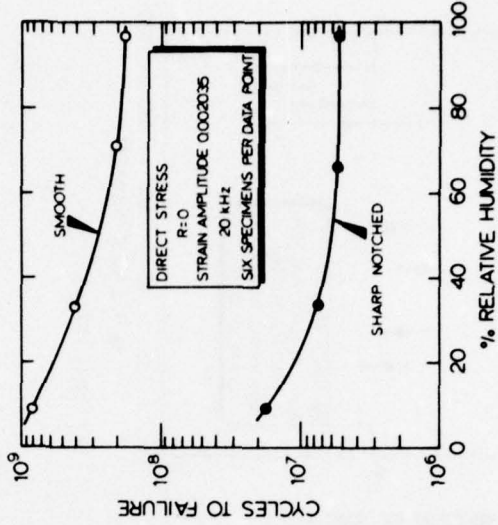
FIGURES 1a - 1d: VACUUM AND GASEOUS PRESSURE EFFECTS ON ALUMINIUM ALLOY FATIGUE STRENGTH.



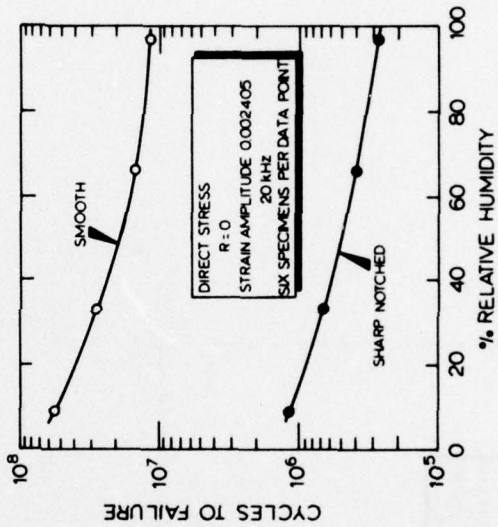
FIGURES 1e - 1i: VACUUM AND GASEOUS PRESSURE EFFECTS ON ALUMINIUM ALLOY FATIGUE STRENGTH.



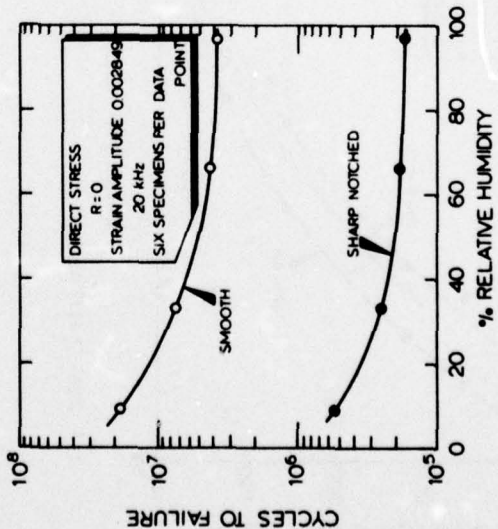
(a) 6061 - T6 [14, 15].



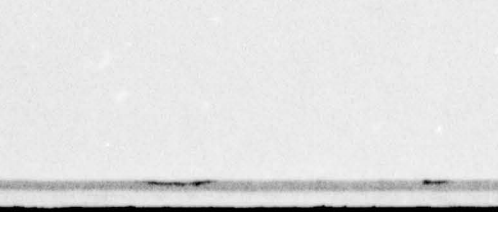
(b) 2024 - T6 [9].



(c) 2024 - T6 [17].

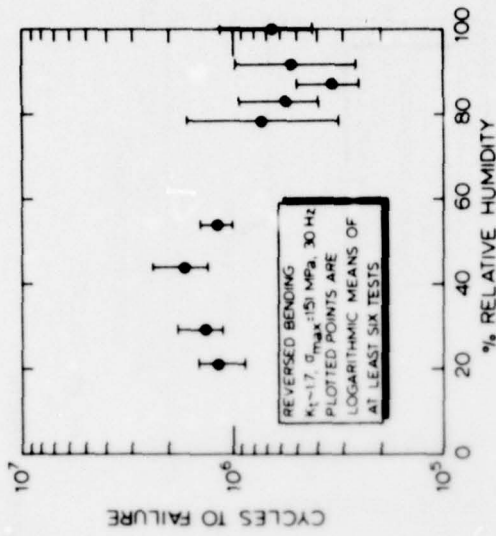


(d) 2024 - T6 [17].

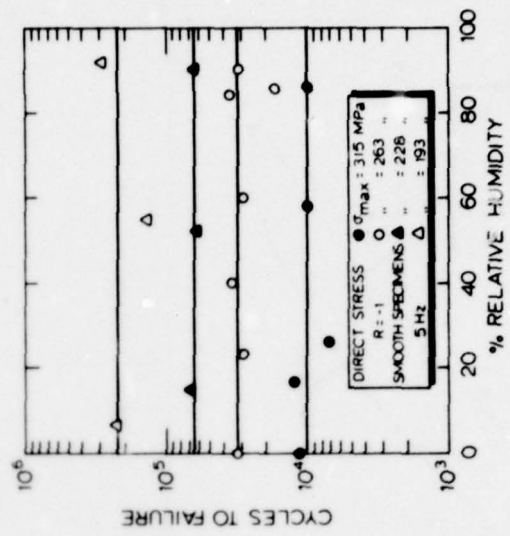


(e) 2024 - T6 [17].

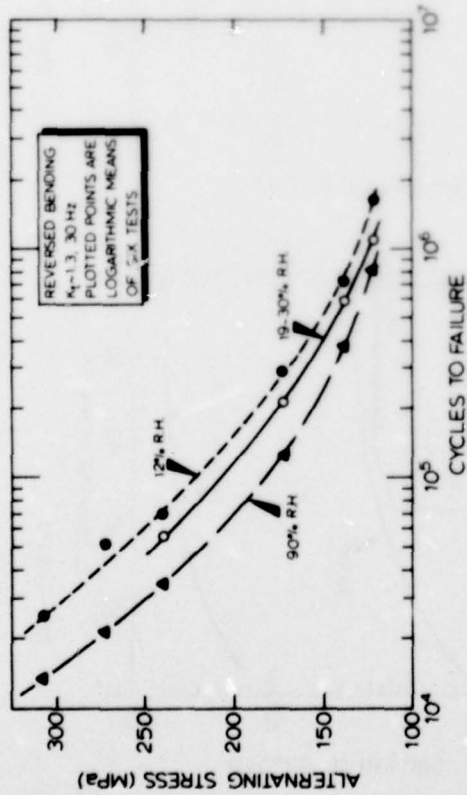
FIGURES 2a - 2e : ATMOSPHERIC HUMIDITY EFFECTS ON ALUMINIUM ALLOY FATIGUE STRENGTH.



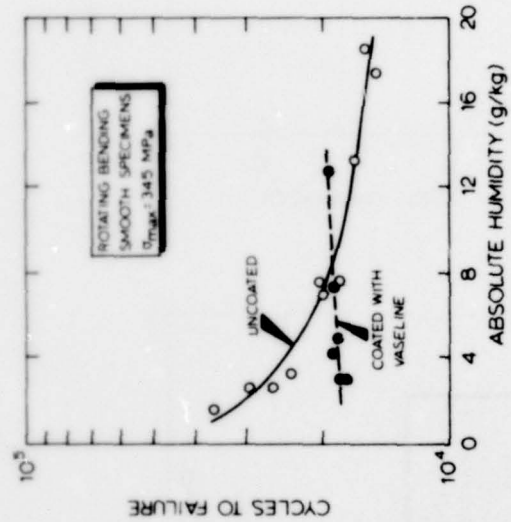
(g) 2024-T3 [16].



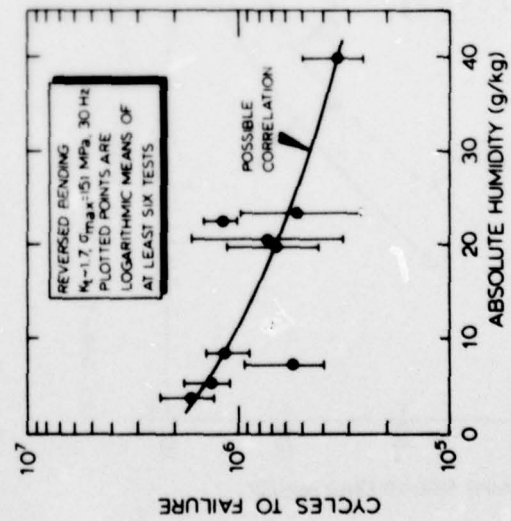
(j) 2219-T851 [18].



(f) 2024-T3 ALCLAD [16].

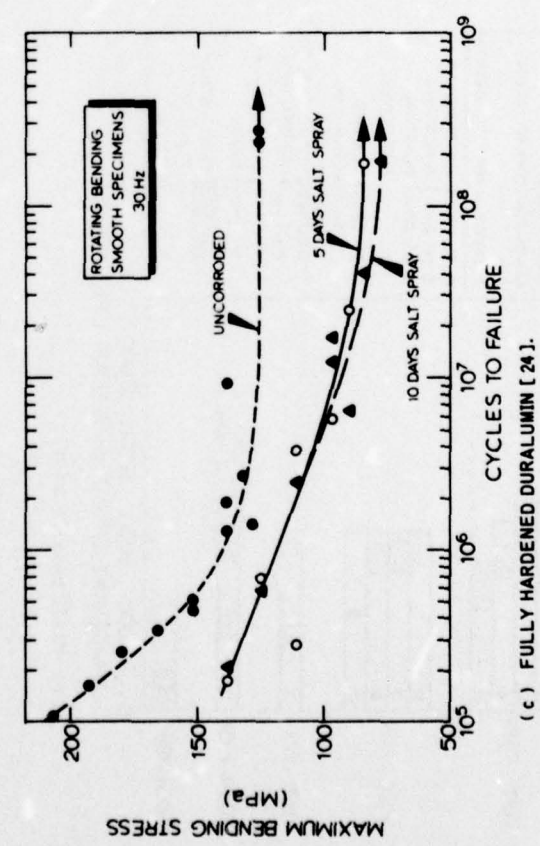
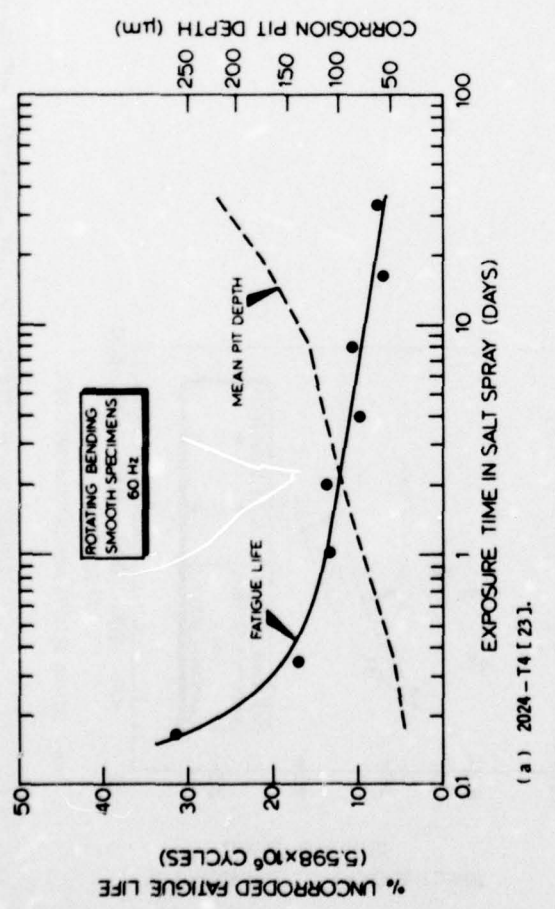
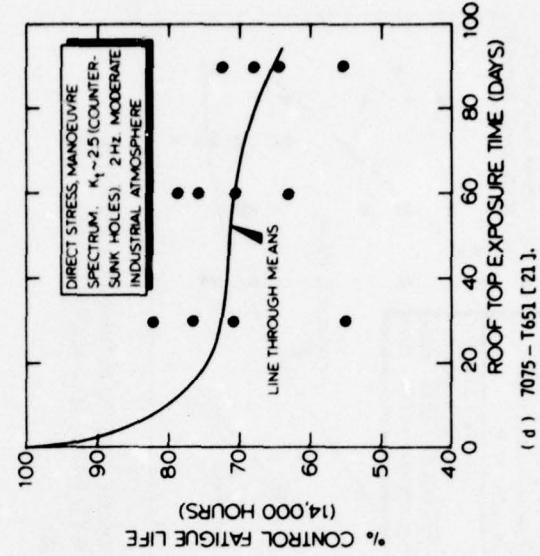
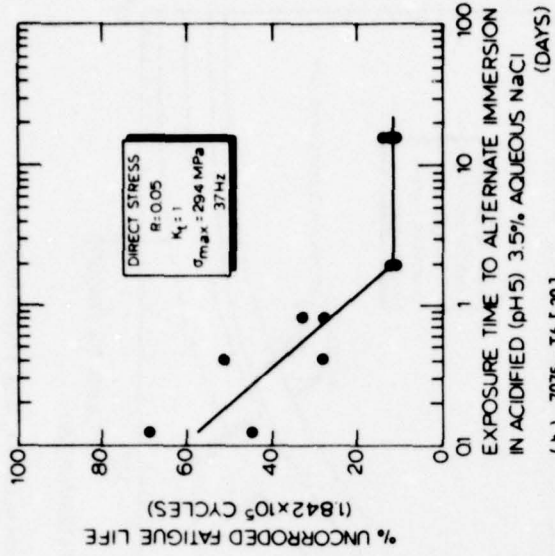


(i) 7075-T6 [13].



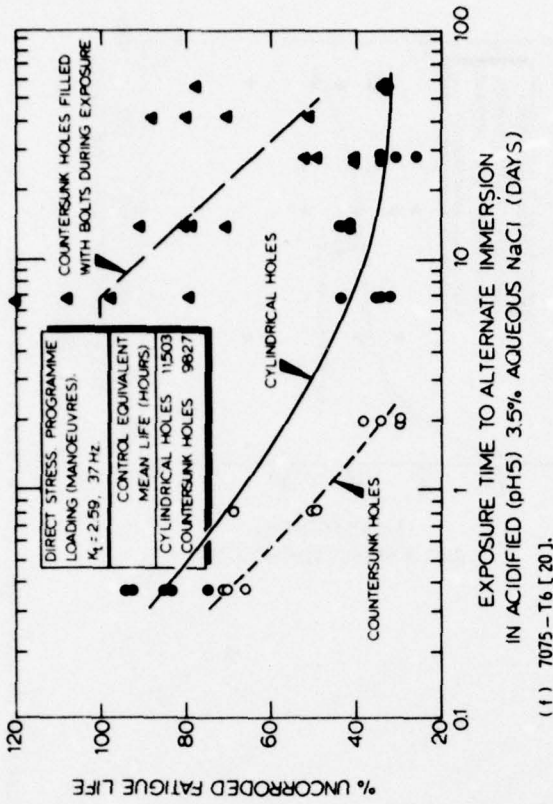
(h) 2024-T3 [16].

FIGURES 2f - 2j: ATMOSPHERIC HUMIDITY EFFECTS ON ALUMINIUM ALLOY FATIGUE STRENGTH.

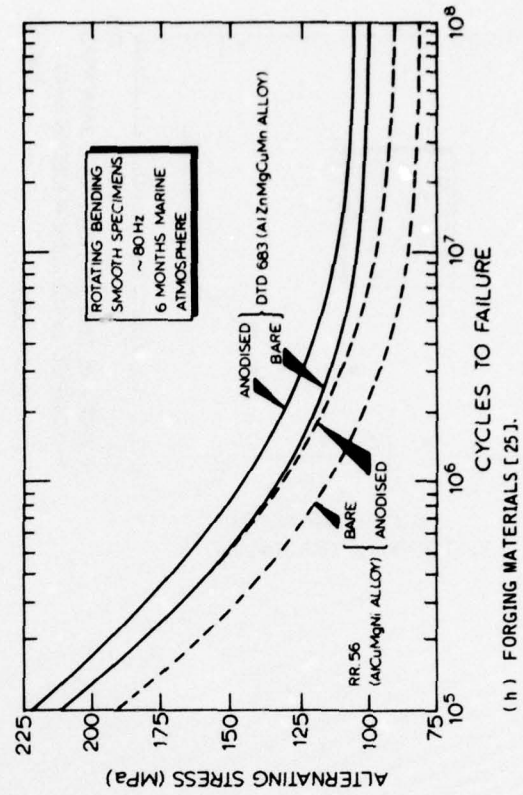


FIGURES 3a - 3d: EFFECTS OF PRIOR ATMOSPHERIC AND AQUEOUS CORROSION ON ALUMINIUM ALLOY FATIGUE STRENGTH IN AIR.

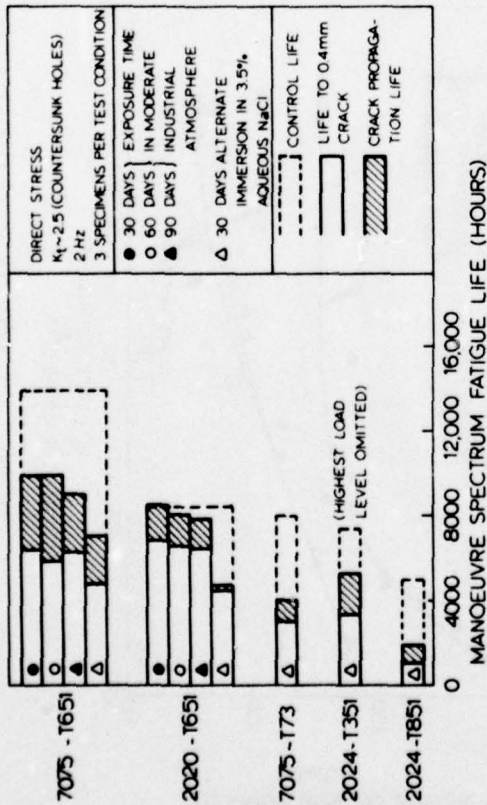




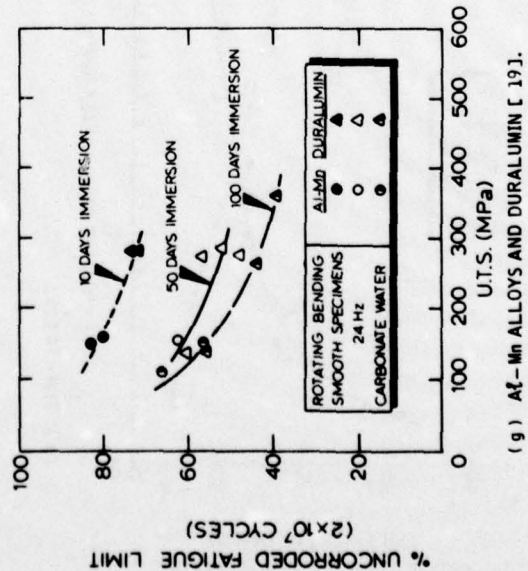
(f) 7075-T6 [20].



(h) FORGING MATERIALS [25].

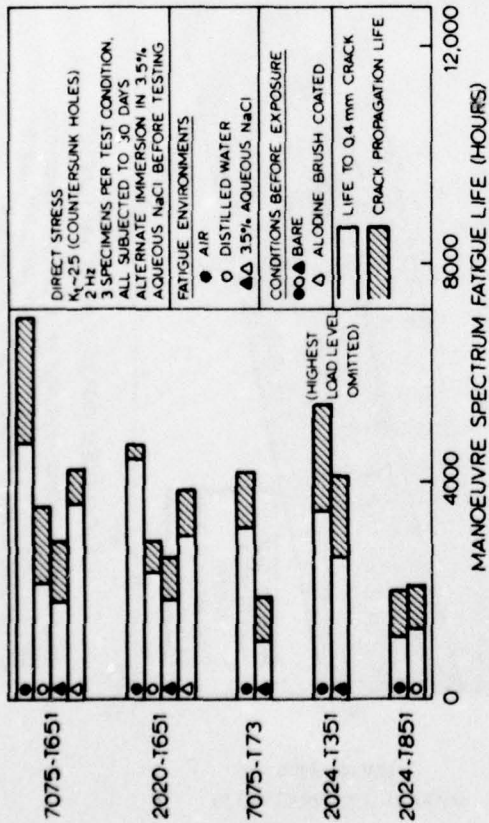


(e) PLATE MATERIALS [21].

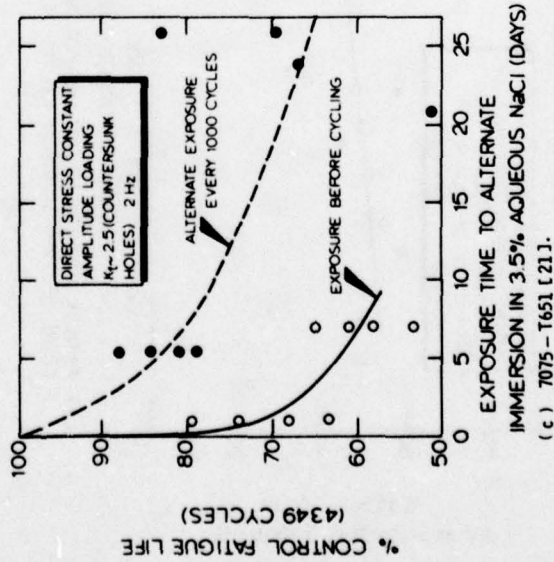


(g) Al-Mn ALLOYS AND DURALUMIN [19].

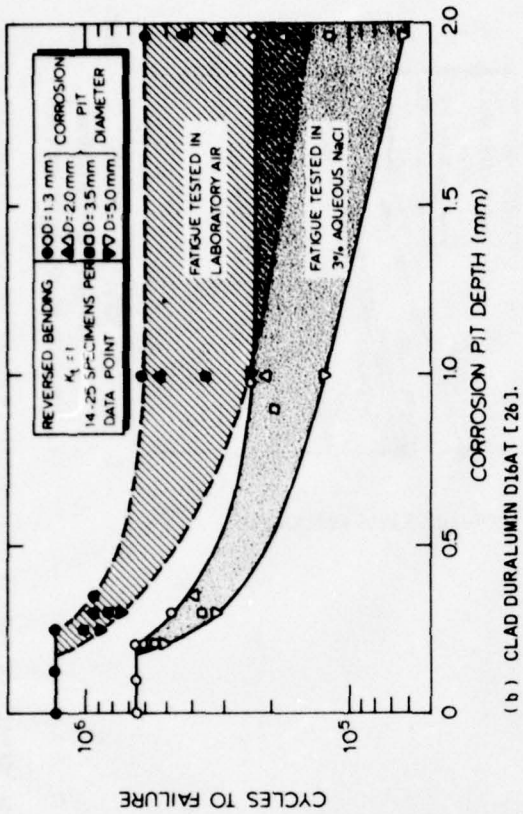
FIGURES 3e - 3h: EFFECTS OF PRIOR ATMOSPHERIC AND AQUEOUS CORROSION ON ALUMINIUM ALLOY FATIGUE STRENGTH IN AIR.



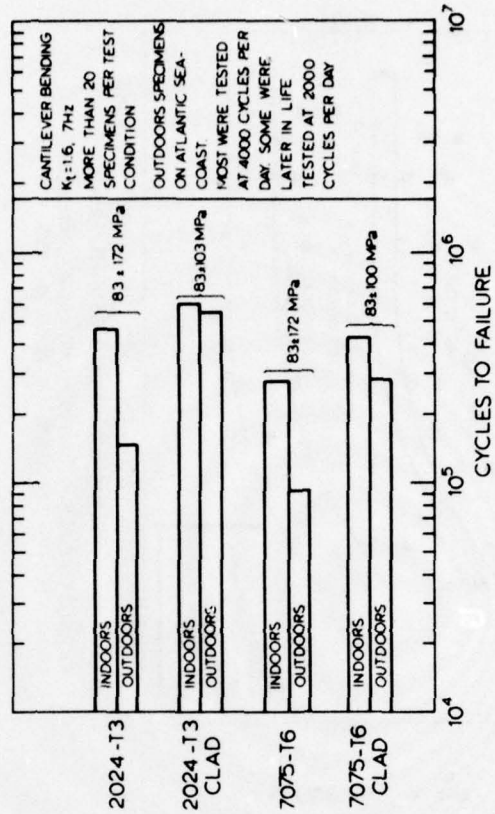
(a) PLATE MATERIALS [ 21 ].



(c) 7075 - T651 [ 21 ].

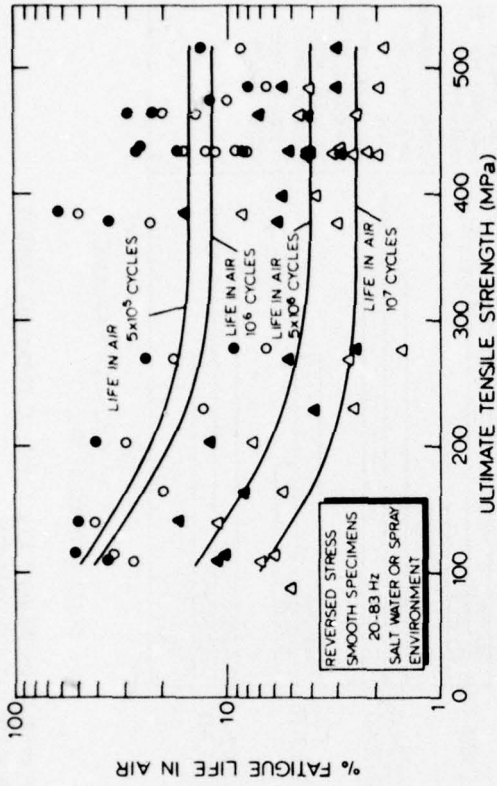


(b) CLAD DURALUMIN DIGAT [ 26 ].

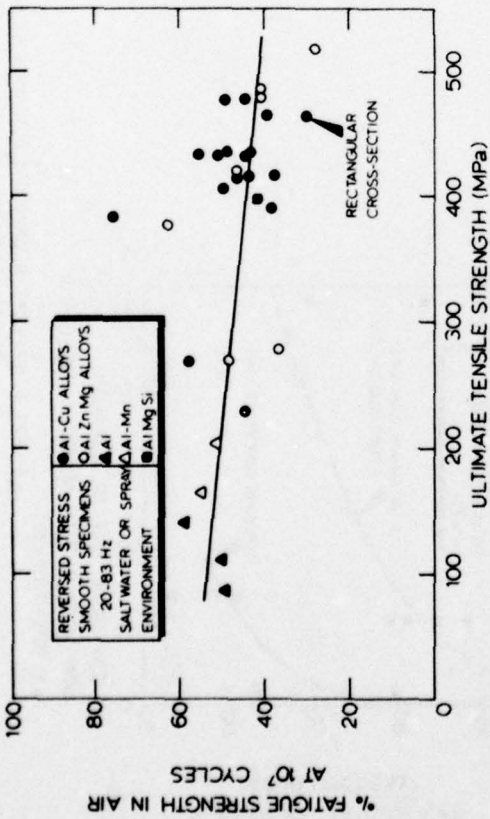


(d) SHEET MATERIALS [ 27 ].

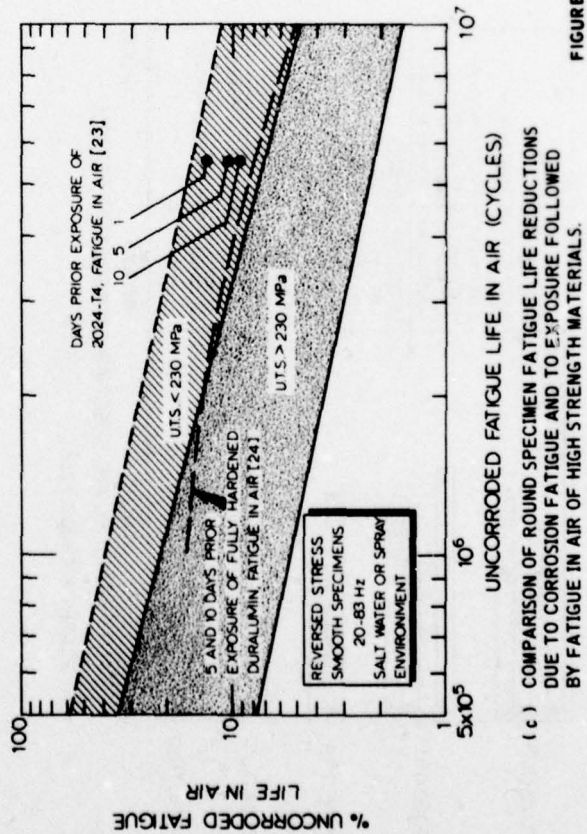
FIGURE 4 : EFFECTS OF EXPOSURE FOLLOWED BY FATIGUE IN AQUEOUS ENVIRONMENTS ( a, b ) AND ALTERNATING PERIODS OF EXPOSURE AND FATIGUE ( c, d ) ON ALUMINIUM ALLOY FATIGUE STRENGTH.



(a) FATIGUE STRENGTH REDUCTIONS IN ROUND AND RECTANGULAR CROSS-SECTION SPECIMENS [ 25, 28 - 35 ].

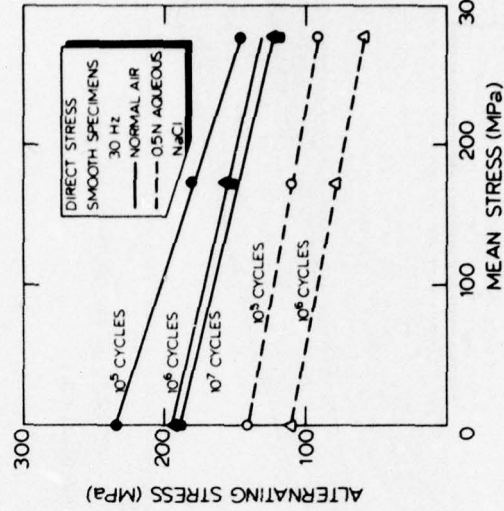


(b) FATIGUE LIFE REDUCTIONS IN ROUND AND RECTANGULAR CROSS-SECTION SPECIMENS [ 25, 28, 30, 32 - 35 ].

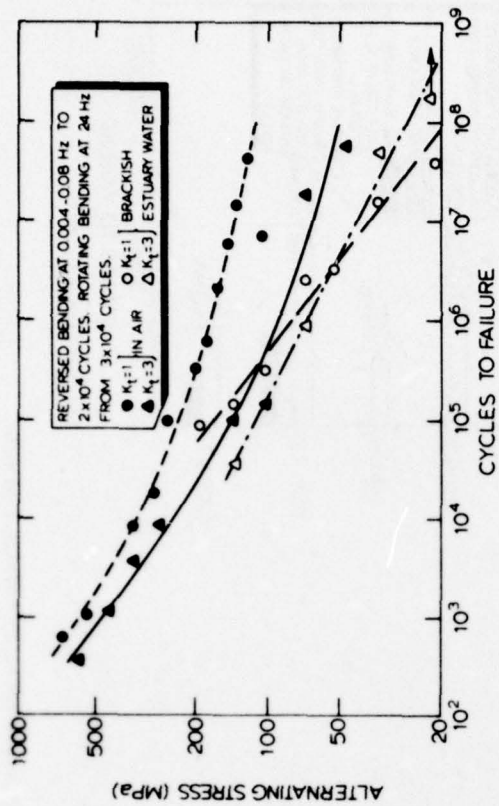


(c) 7075-T6 ROUND SPECIMENS [ 36 ].

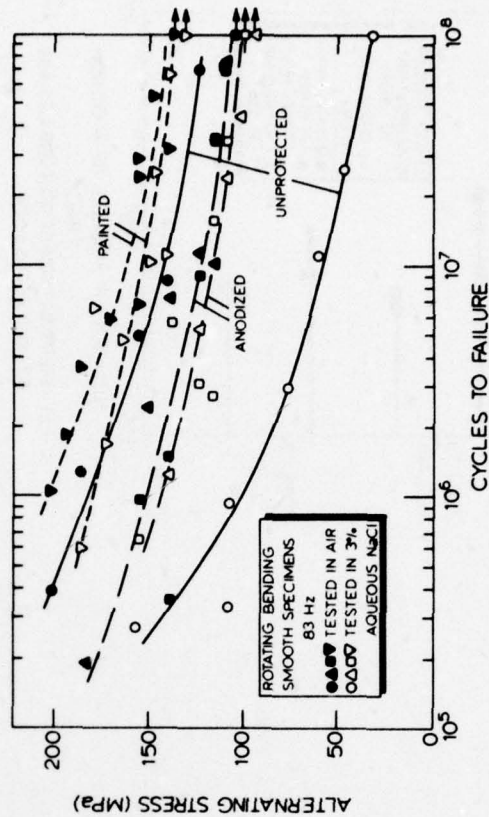
FIGURES 5a - 5d: EFFECTS OF AQUEOUS ENVIRONMENTS ON ALUMINIUM ALLOY FATIGUE STRENGTH.



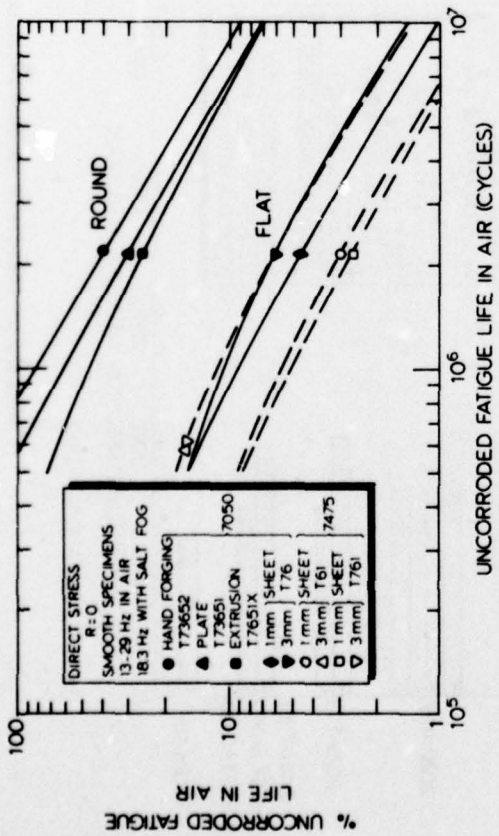
(d) 7075-T6 ROUND SPECIMENS [ 36 ].



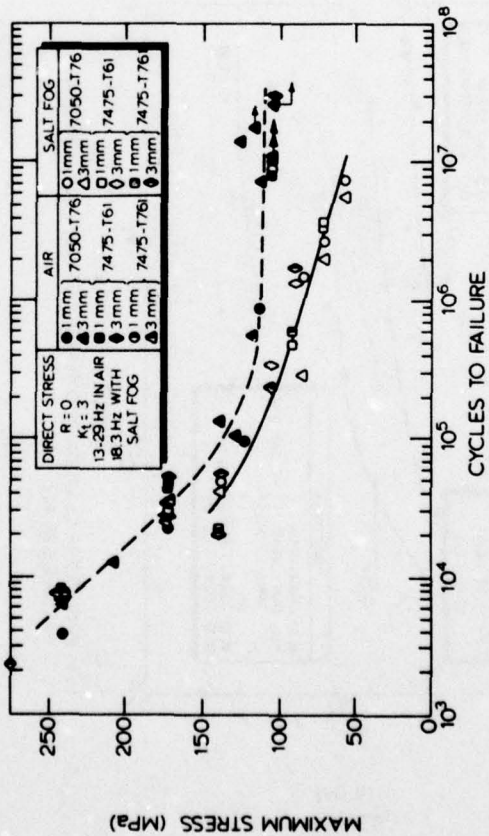
(f) 7079-T6 PLATE [34].



(h) H10 (Al-Mg-Si) [32].

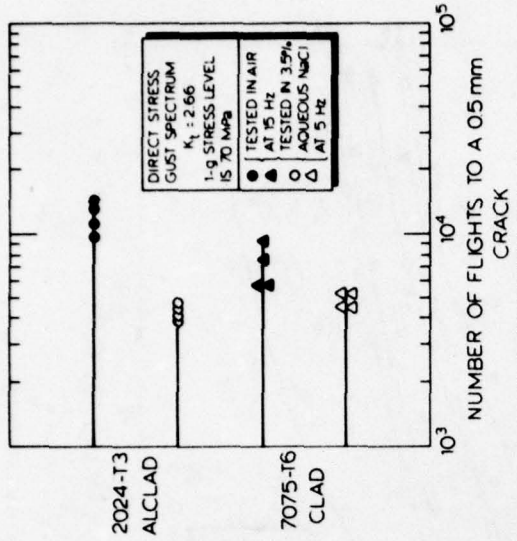


(e) DEPENDENCE OF LIFE REDUCTION DUE TO CORROSION FATIGUE ON SPECIMEN GEOMETRY [37, 38].

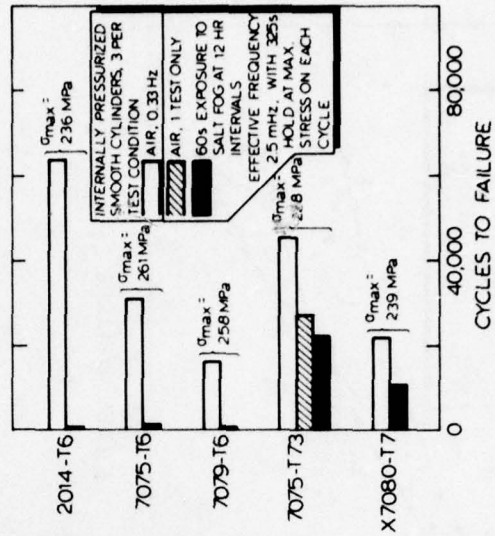


(g) 7475 AND 7050 SHEET [37, 38].

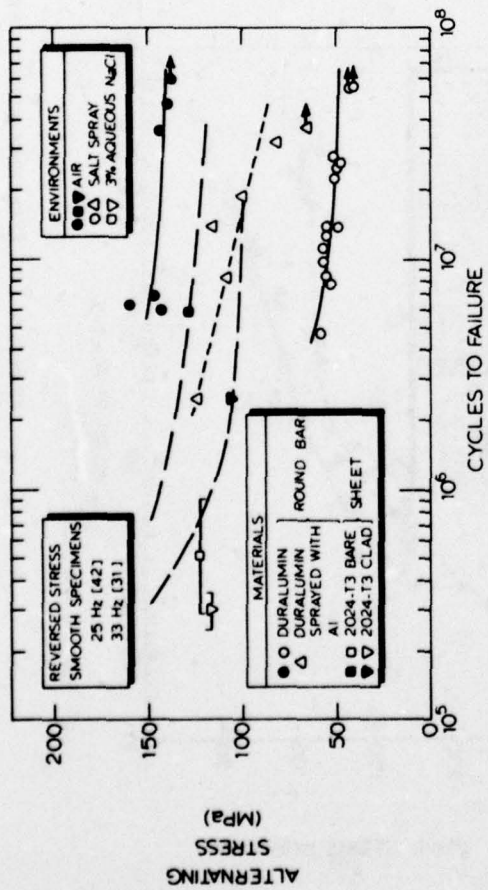
FIGURES 5a - 5h: EFFECTS OF AQUEOUS ENVIRONMENTS ON ALUMINIUM ALLOY FATIGUE STRENGTH.



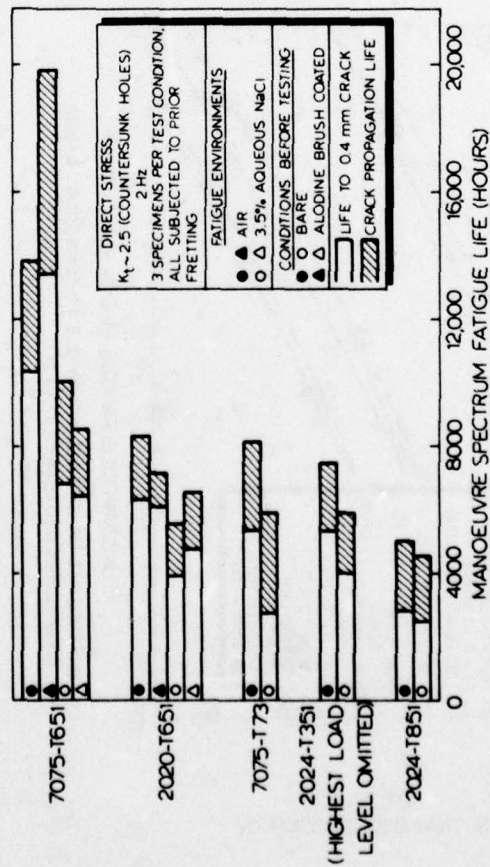
(j) SHEET MATERIALS WITH DRILLED AND REAMED HOLES [49].



(k) DIE FORGINGS [50].

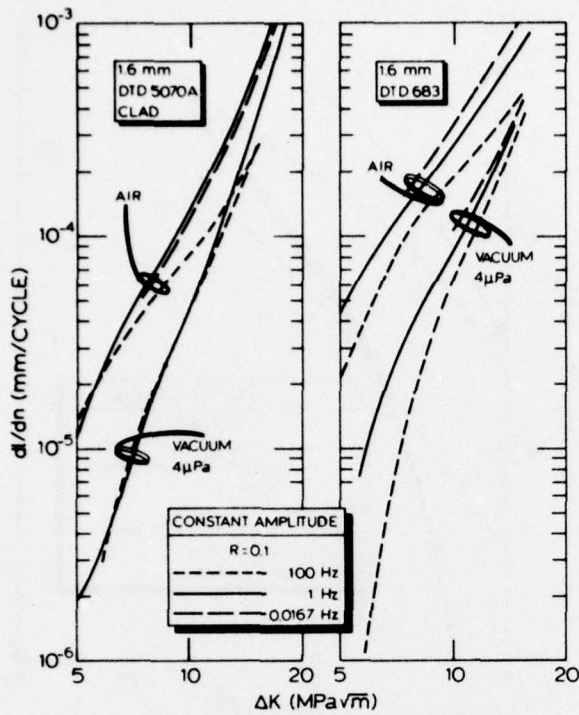


(l) EFFECTS OF CLADDING AND AL SPRAY COATING ON CORROSION FATIGUE [25, 42].

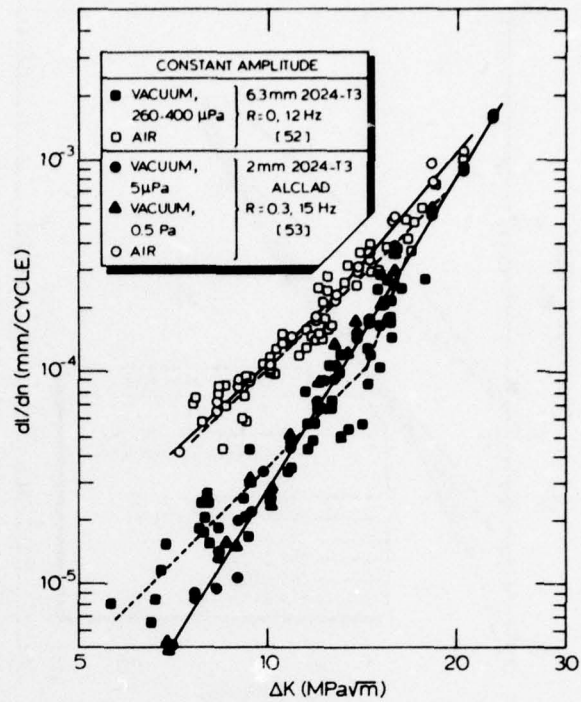


(m) PLATE MATERIALS [21].

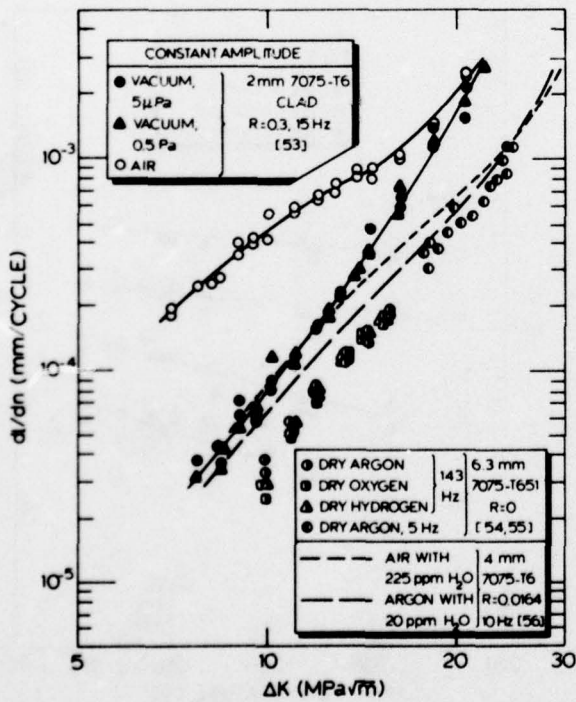
FIGURES 5i - 5k: EFFECTS OF AQUEOUS ENVIRONMENTS ON ALUMINIUM ALLOY FATIGUE STRENGTH.



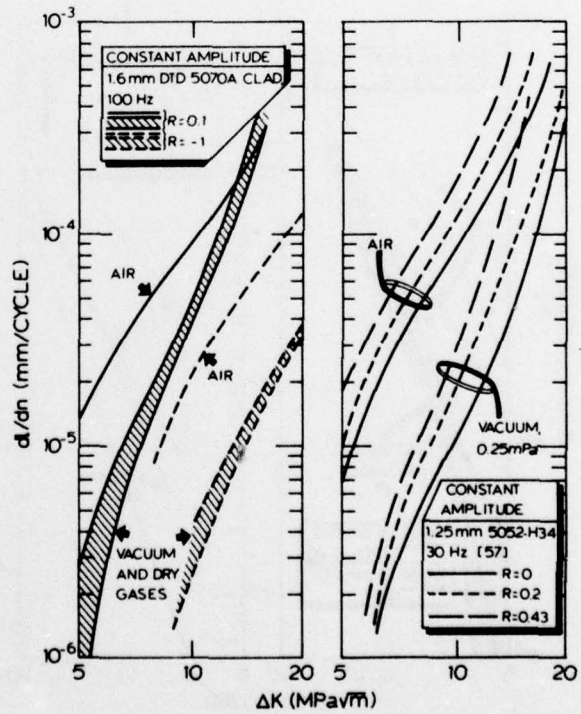
(a) DTD 5070A (Al 2.5 Cu 1.5 Mg Ni Fe) CLAD AND DTD 683 (Al 6 Zn 3 Mg 1.5 Cu Mn) SHEET [51].



(b) 2024-T3 BARE AND ALCLAD SHEET AND PLATE [52, 53].

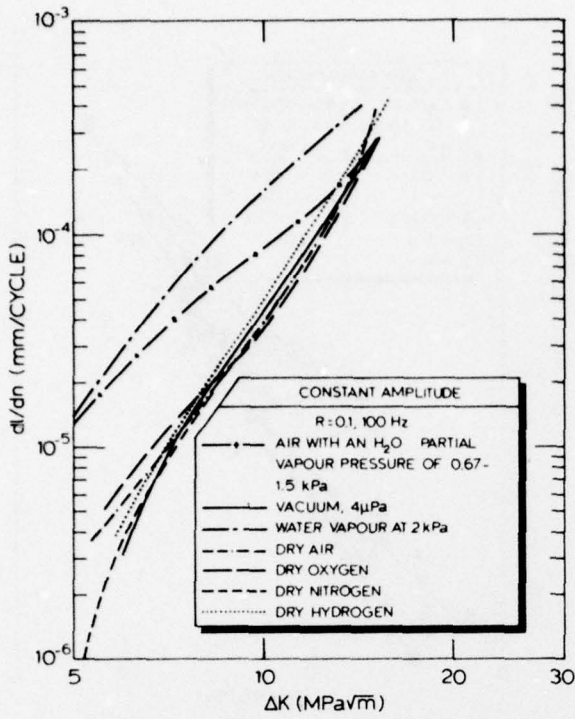


(c) 7075-T6 BARE AND CLAD SHEET AND 7075-T651 PLATE [54 - 56].

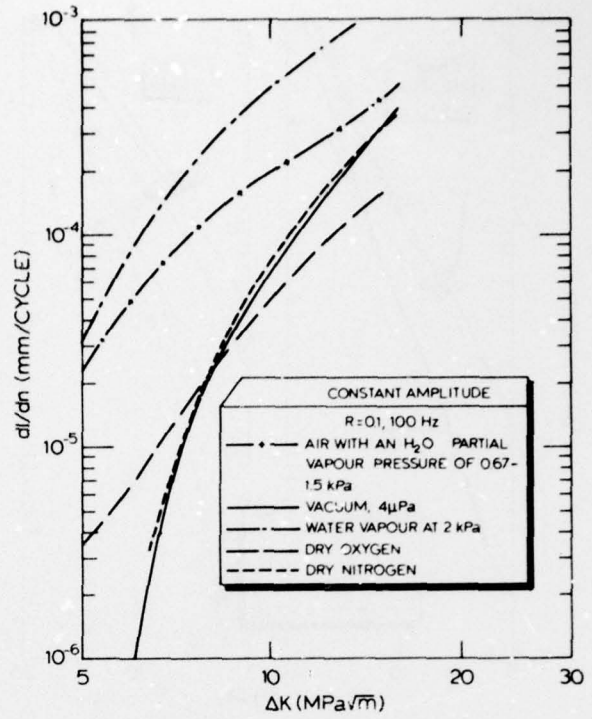


(d) EFFECT OF MEAN STRESS ON SHEET MATERIALS [1,57].

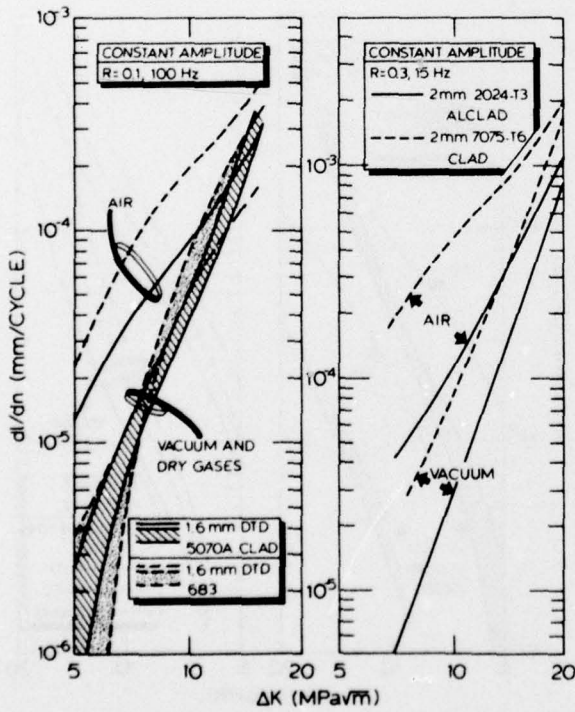
FIGURES 6a - 6d: VACUUM AND GASEOUS EFFECTS ON ALUMINIUM ALLOY FATIGUE CRACK PROPAGATION.



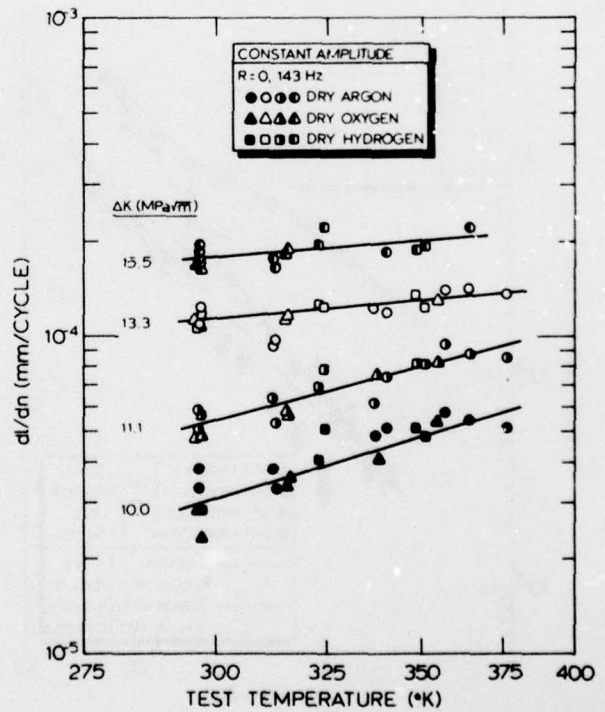
(e) 1.6mm DTD 5070A (Al 2.5 Cu 1.5 Mg Ni Fe) CLAD SHEET [1,51].



(f) 1.6mm DTD 683 (Al 6 Zn 3 Mg 1.5 Cu Mn) SHEET [1,51].

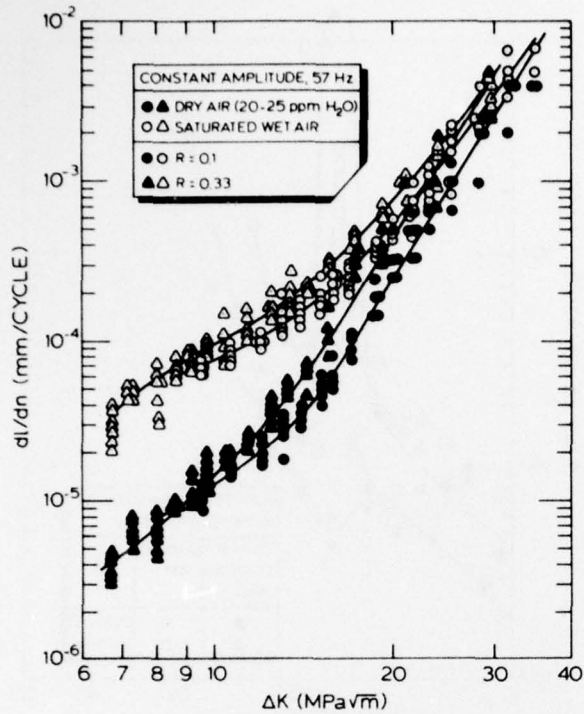


(g) 2000 AND 7000 SERIES SHEET [1,51, 53].

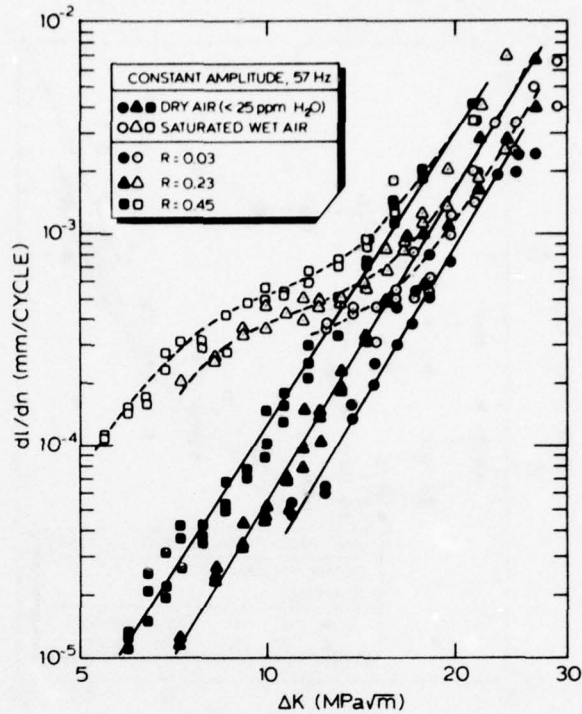


(h) EFFECT OF TEMPERATURE FOR 6.3mm 7075 - T651 PLATE [54].

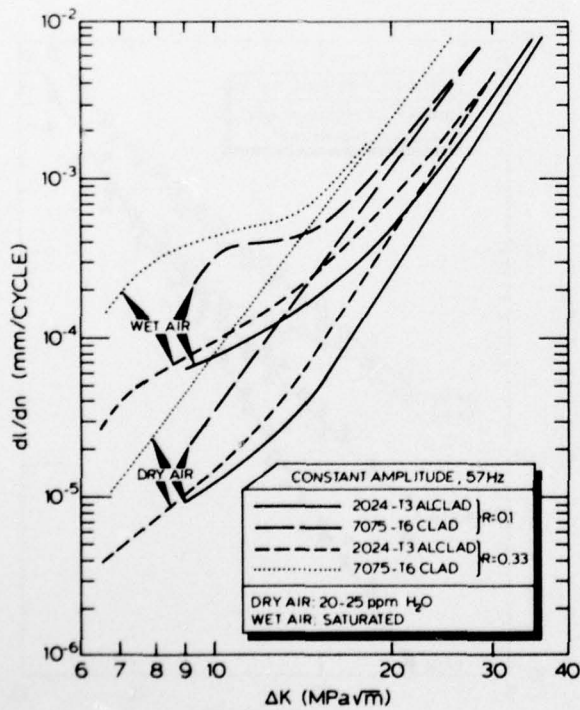
FIGURES 6e - 6h: VACUUM AND GASEOUS EFFECTS ON ALUMINIUM ALLOY FATIGUE CRACK PROPAGATION.



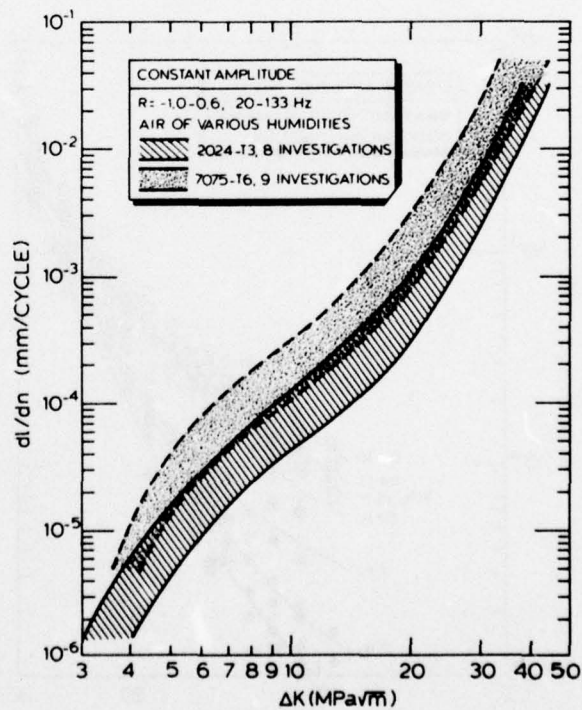
(a) 1mm 2024 - T3 ALCLAD SHEET [61].



(b) 1mm 7075 - T6 CLAD SHEET [61].



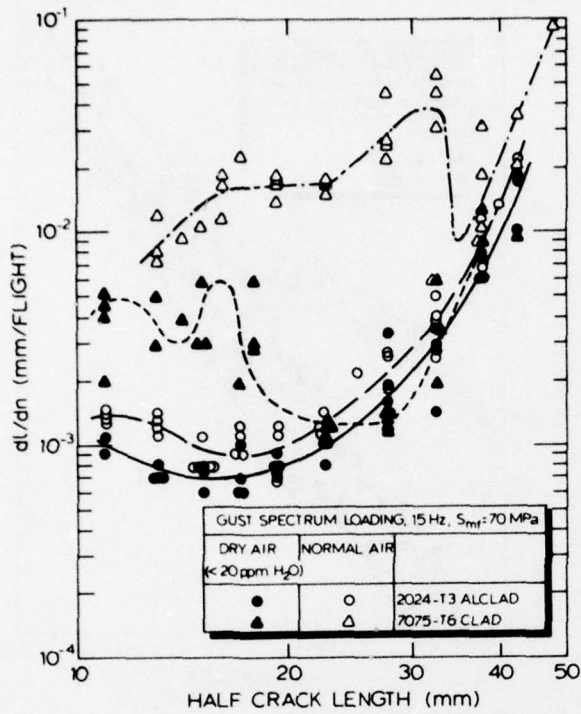
(c) 1mm 2024 - T3 AND 7075 - T6 SHEET [61].



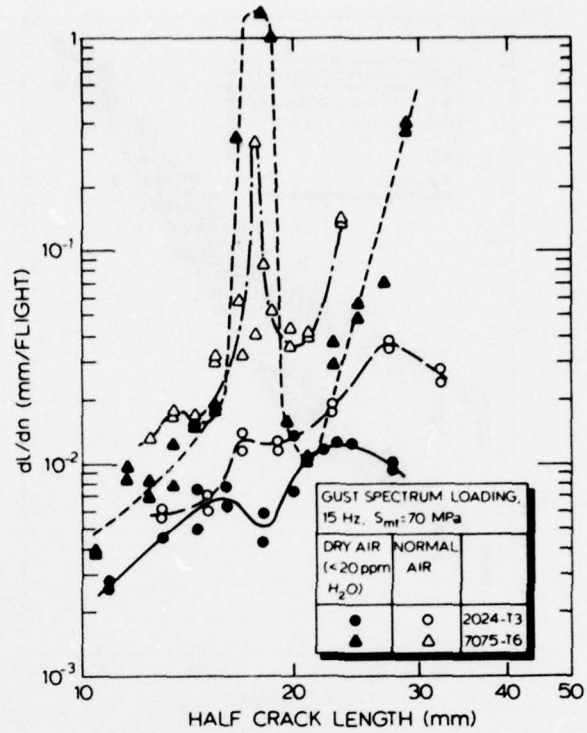
(d) 2024 - T3 AND 7075 - T6 SHEET AND PLATE [66].

FIGURES 7a - 7d: ATMOSPHERIC HUMIDITY EFFECTS ON ALUMINIUM ALLOY FATIGUE CRACK PROPAGATION.

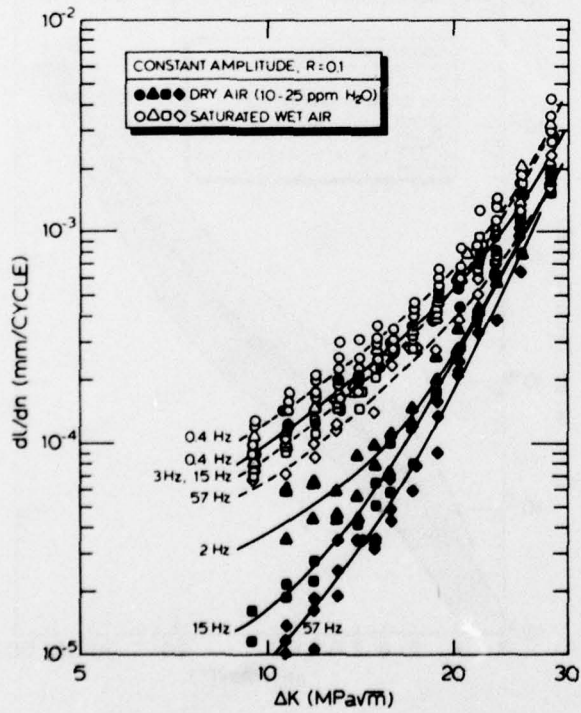




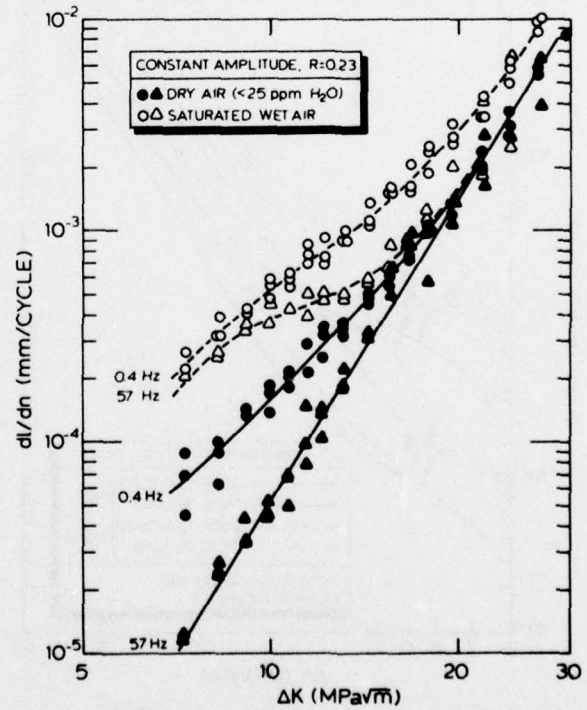
(e) 2mm 2024 - T3 AND 7075 - T6 SHEET [49]: CRACK RATE ALTERNATIONS ARE DUE TO SEVERE FLIGHTS.



(f) 10mm 2024 - T3 AND 7075 - T6 SHEET [49]: CRACK RATE ALTERNATIONS ARE DUE TO SEVERE FLIGHTS.

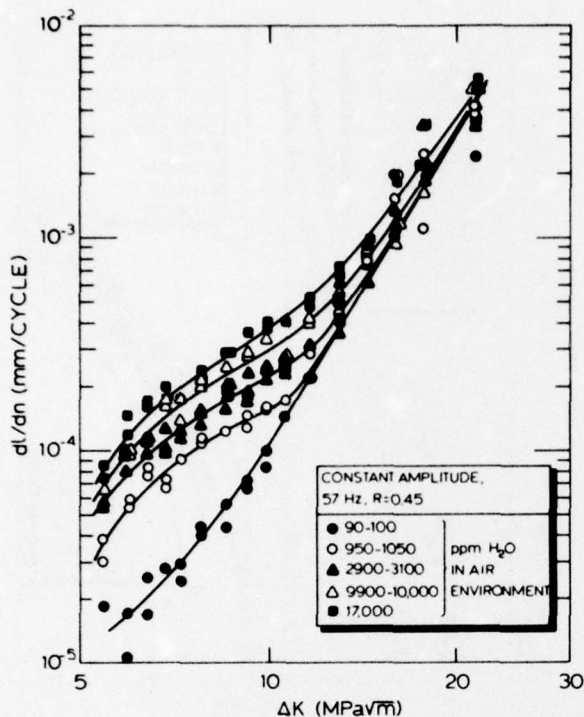


(g) FREQUENCY EFFECTS ON 1mm 2024 - T3 ALCLAD SHEET [61].

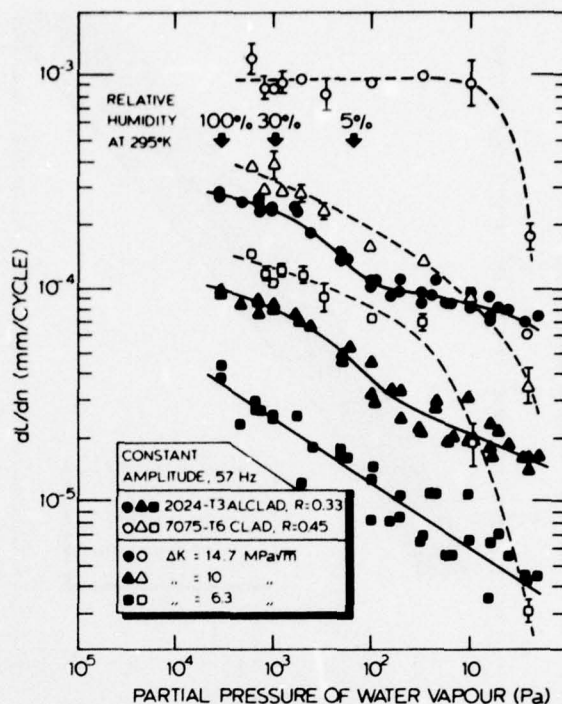


(h) FREQUENCY EFFECTS ON 1mm 7075 - T6 CLAD SHEET [61].

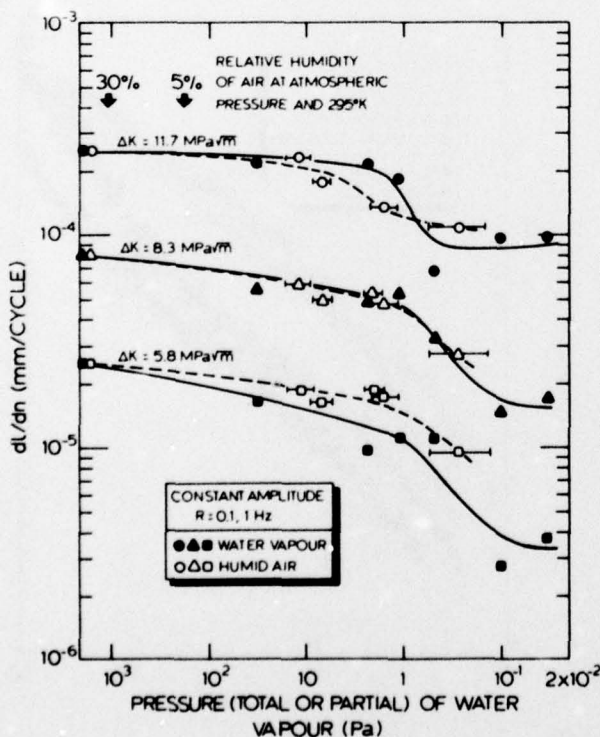
FIGURES 7e - 7h: ATMOSPHERIC HUMIDITY EFFECTS ON ALUMINIUM ALLOY FATIGUE CRACK PROPAGATION.



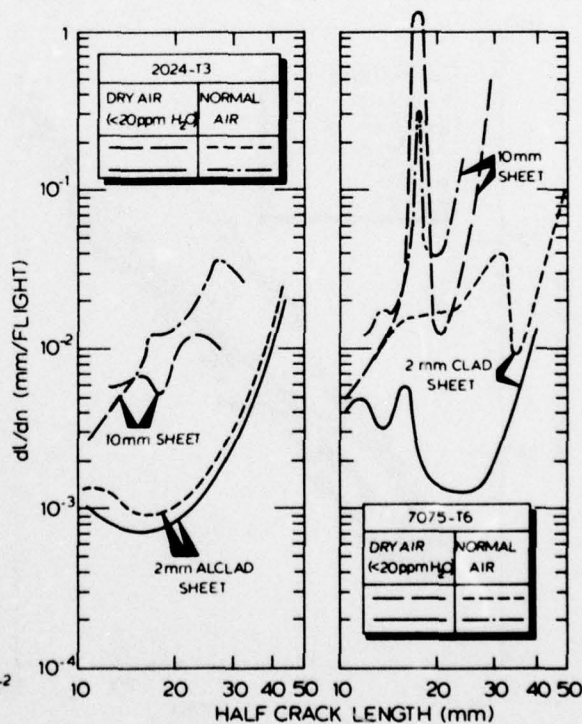
(i) EFFECT OF WATER VAPOUR CONTENT OF THE AIR ON 1mm 7075 - T6 CLAD SHEET [ 61].



(j) EFFECT OF WATER VAPOUR CONTENT AT LOW ΔK FOR 1mm 2024 - T3 AND 7075 - T6 SHEET AT 57 Hz [ 58, 61].

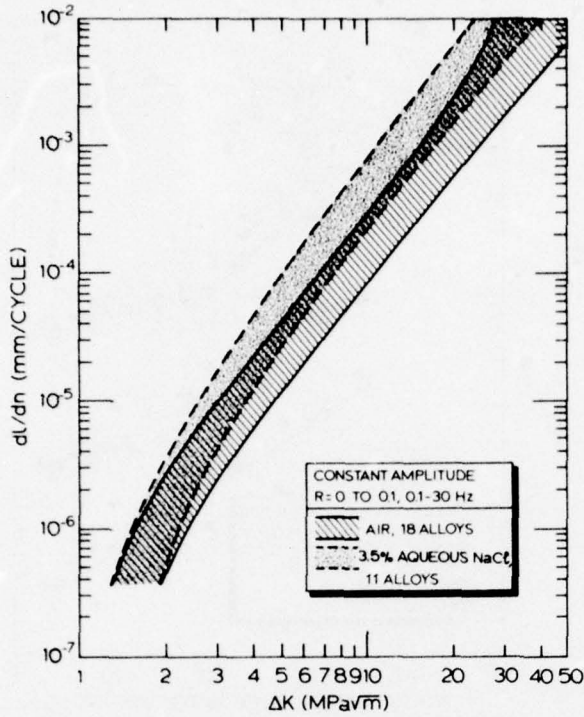


(k) EFFECT OF WATER VAPOUR CONTENT AT LOW ΔK FOR 1.6mm DTD 5070A CLAD SHEET AT 1 Hz [ 51, 60].

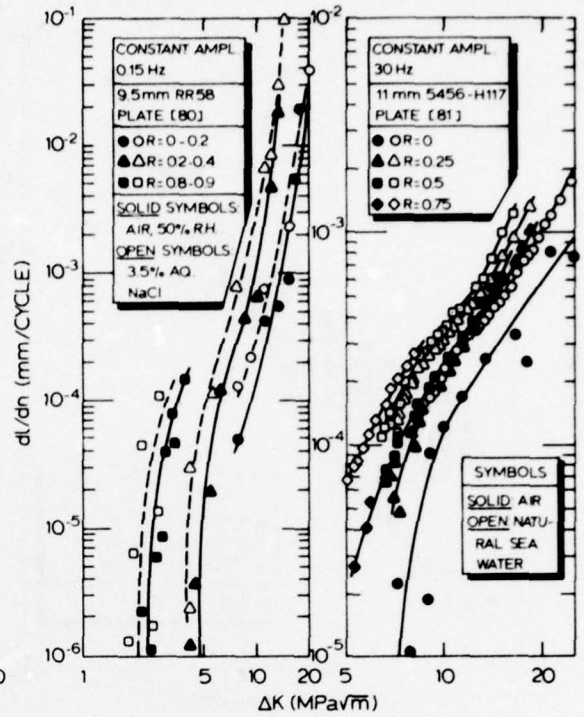


(l) 2 AND 10mm 2024 - T3 AND 7075 - T6 SHEET [ 49].

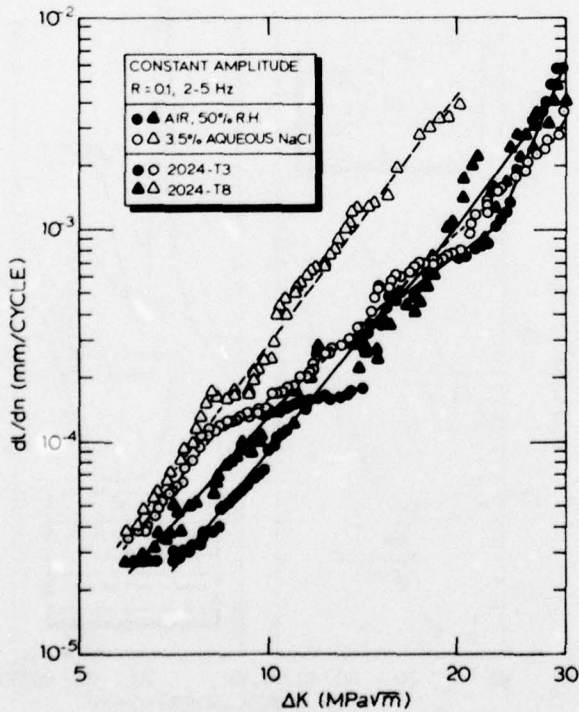
FIGURES 7i - 7l: ATMOSPHERIC HUMIDITY EFFECTS ON ALUMINIUM ALLOY FATIGUE CRACK PROPAGATION.



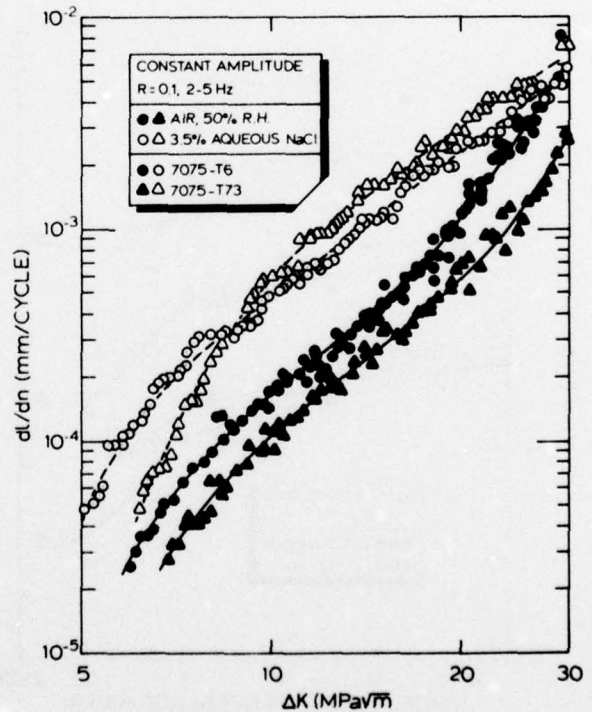
(a) 2000, 5000, 6000 AND 7000 SERIES ALLOYS [79].



(b) EFFECT OF MEAN STRESS [80, 81].

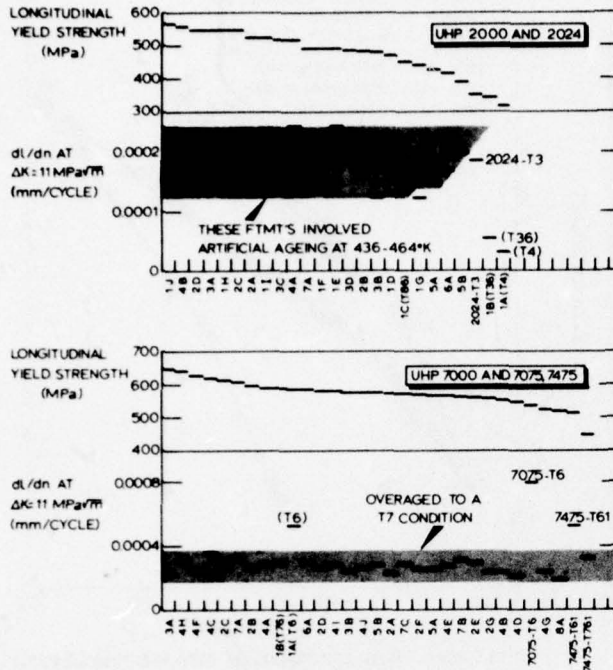


(c) 1.6mm 2024 SHEET [84].

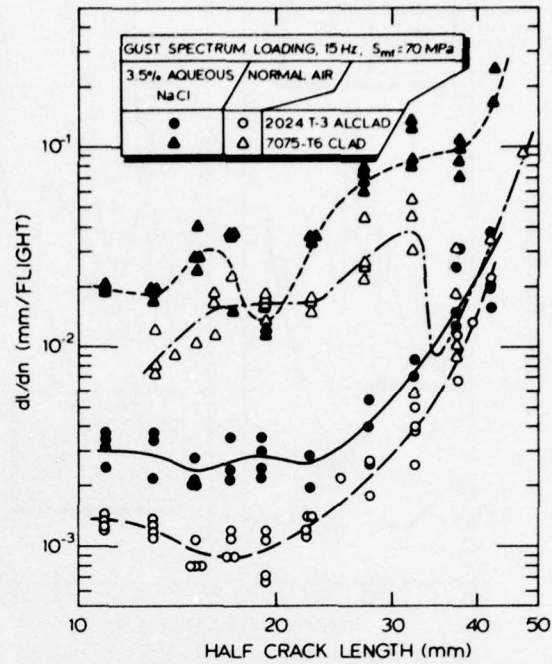


(d) 1.6mm 7075 SHEET [84].

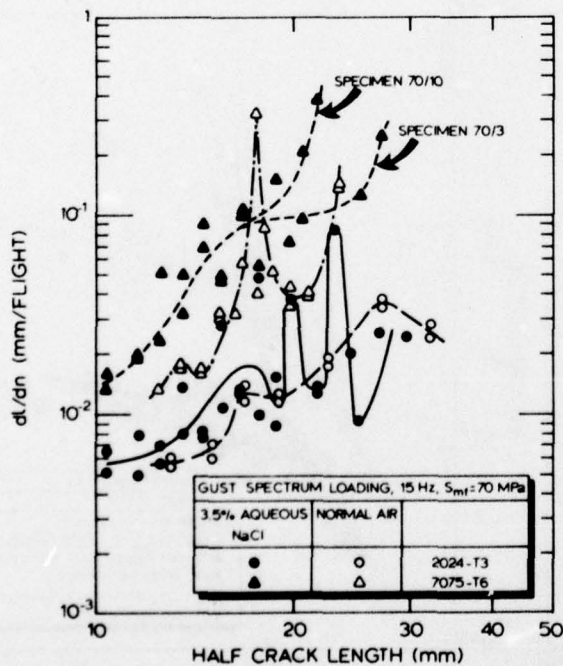
FIGURES 8a - 8d: EFFECTS OF AQUEOUS ENVIRONMENTS ON ALUMINIUM ALLOY FATIGUE CRACK PROPAGATION.



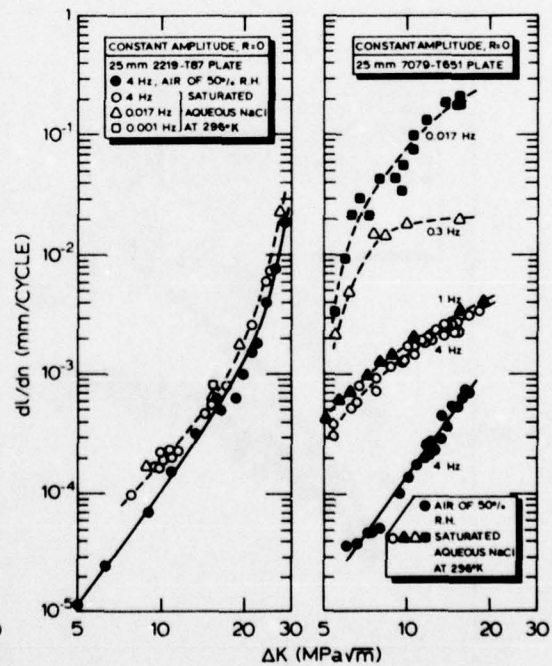
(e) 2.5 - 5 mm 2000 AND 7000 SERIES ALLOY SHEET CONVENTIONALLY PROCESSED OR SUBJECTED TO FINAL THERMOMECHANICAL TREATMENTS (FTMT) AND TESTED IN DISTILLED WATER [85].



(f) 2 mm 2024 - T3 AND 7075 - T6 SHEET [49, 78]: CRACK RATE ALTERNATIONS ARE DUE TO SEVERE FLIGHTS.

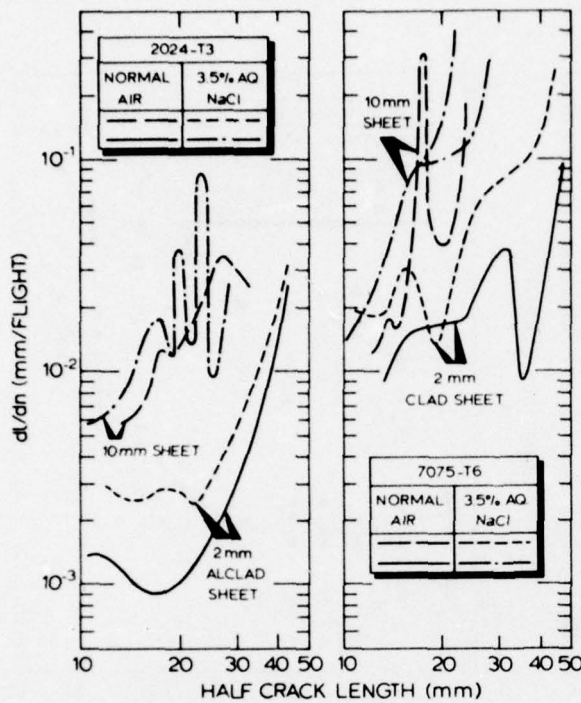


(g) 10 mm 2024 - T3 AND 7075 - T6 SHEET [49, 78, 83]: CRACK RATE ALTERNATIONS ARE DUE TO SEVERE FLIGHTS.

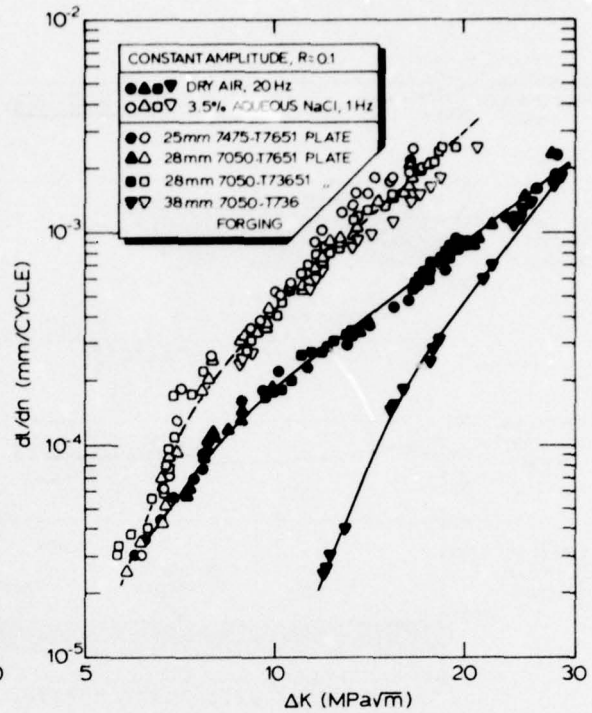


(h) FREQUENCY EFFECTS FOR SHORT TRANSVERSE CRACK GROWTH IN PLATE MATERIALS [79].

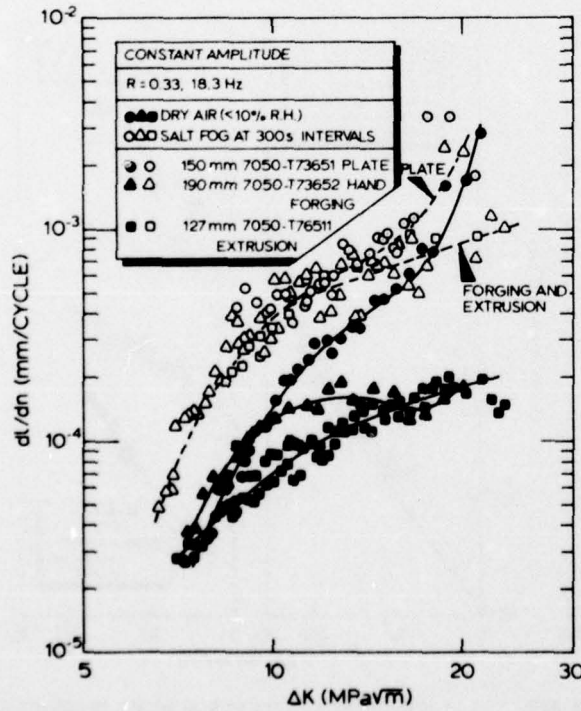
FIGURES 8e - 8h: EFFECTS OF AQUEOUS ENVIRONMENTS ON ALUMINIUM ALLOY FATIGUE CRACK PROPAGATION.



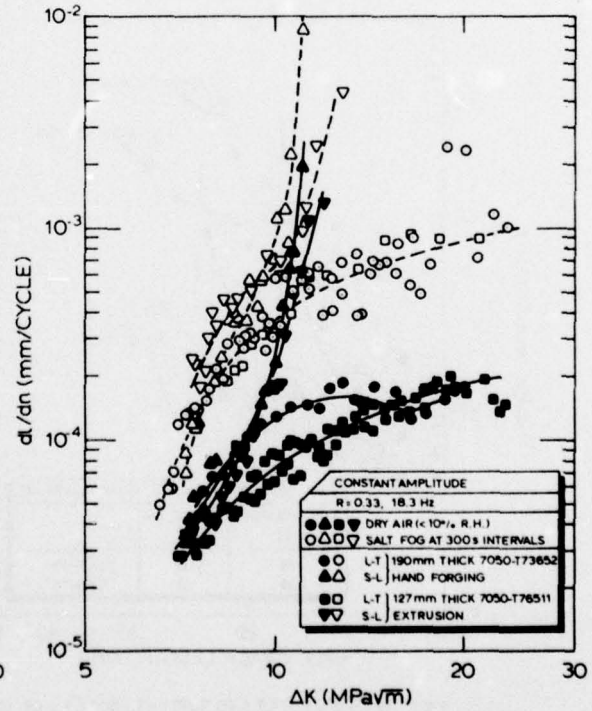
(i) 2 AND 10mm 2024 - T3 AND 7075 - T6 SHEET [49, 78, 83]



(j) 25mm THICK SPECIMENS OF 7475 AND 7050 LOADED IN THE LONGITUDINAL DIRECTION [72].



(k) 25mm THICK SPECIMENS OF 7050 LOADED IN THE LONGITUDINAL DIRECTION [37].



(l) 25mm THICK SPECIMENS OF 7050 LOADED IN THE LONGITUDINAL AND SHORT TRANSVERSE DIRECTIONS [37].

FIGURES 8i - 8l: EFFECTS OF AQUEOUS ENVIRONMENTS ON ALUMINIUM ALLOY FATIGUE CRACK PROPAGATION.

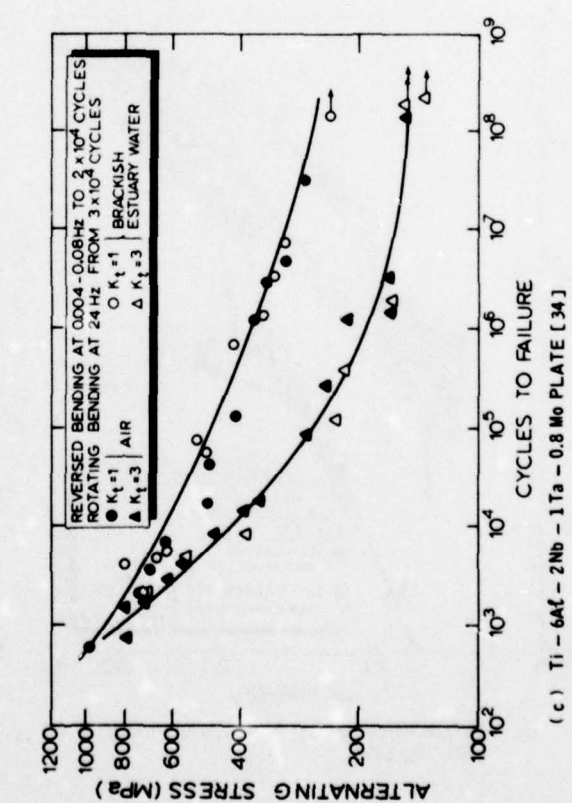
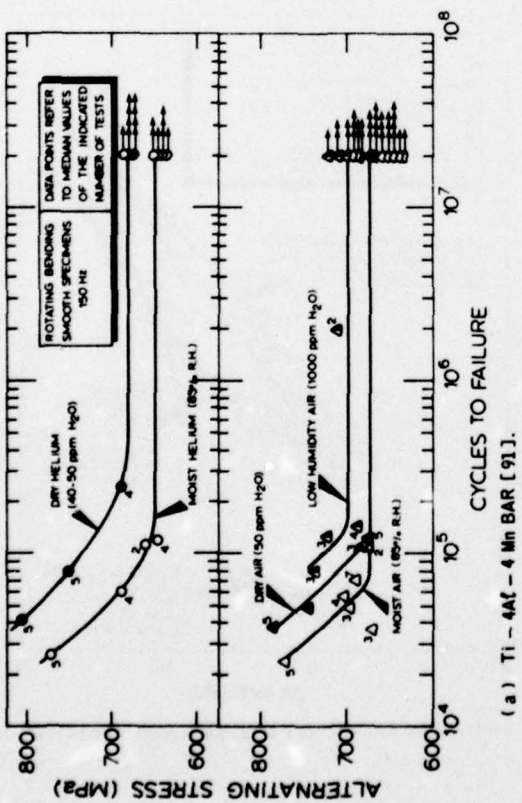
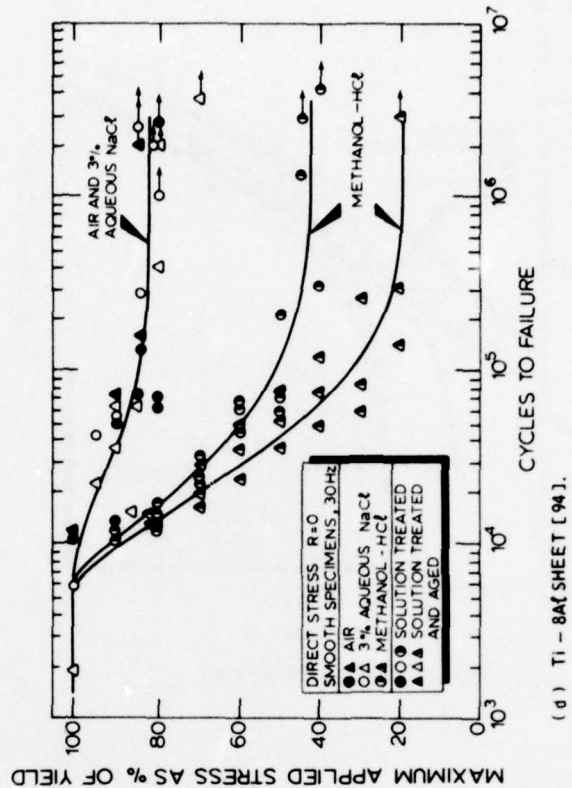
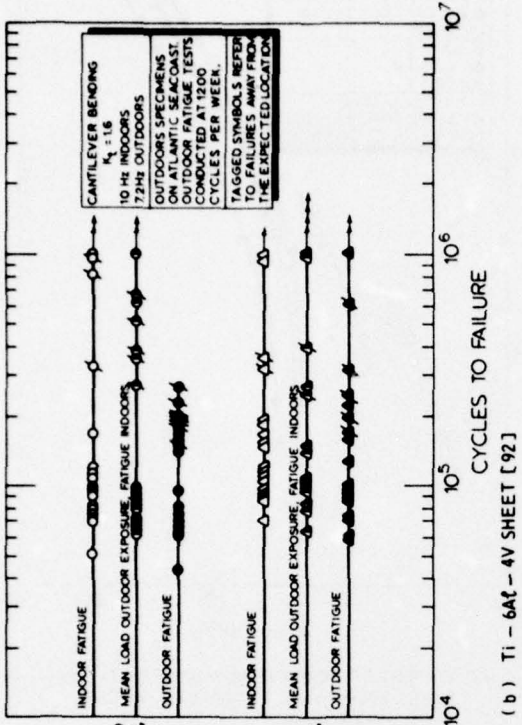
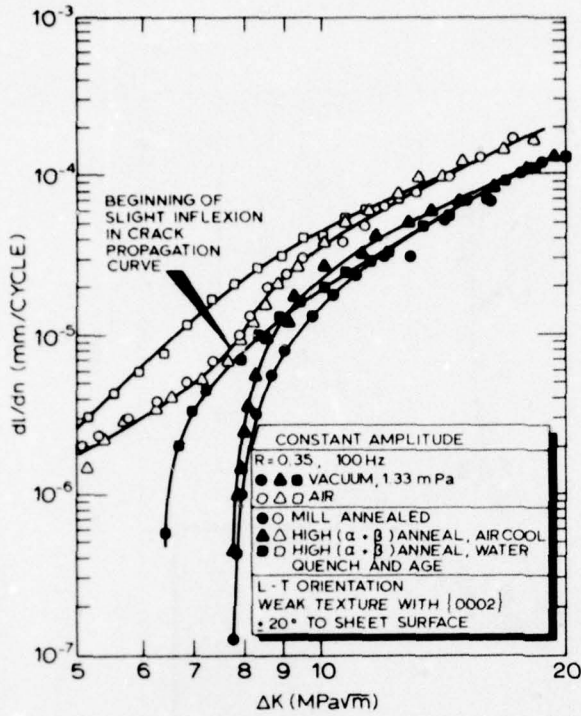
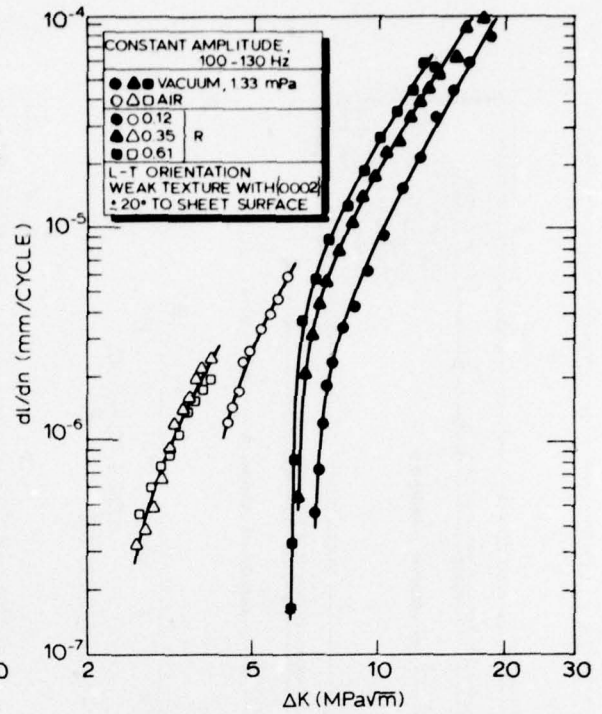


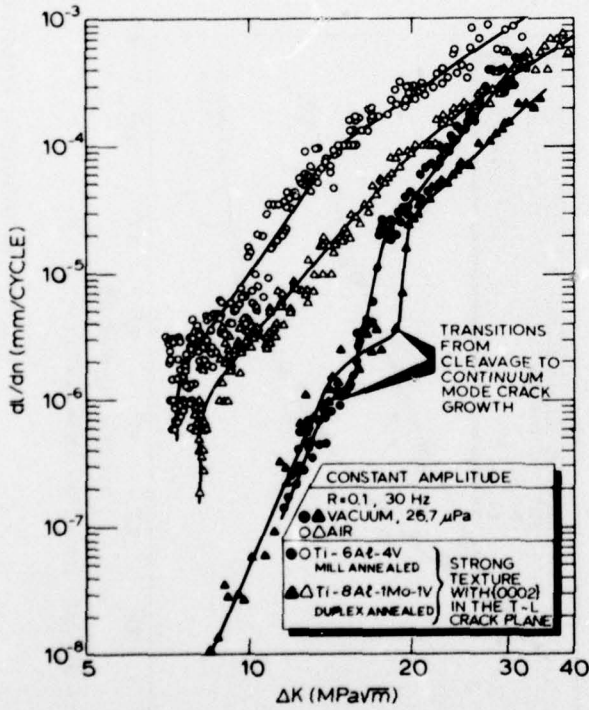
FIGURE 9: ENVIRONMENTAL EFFECTS ON TITANIUM ALLOY FATIGUE STRENGTH.



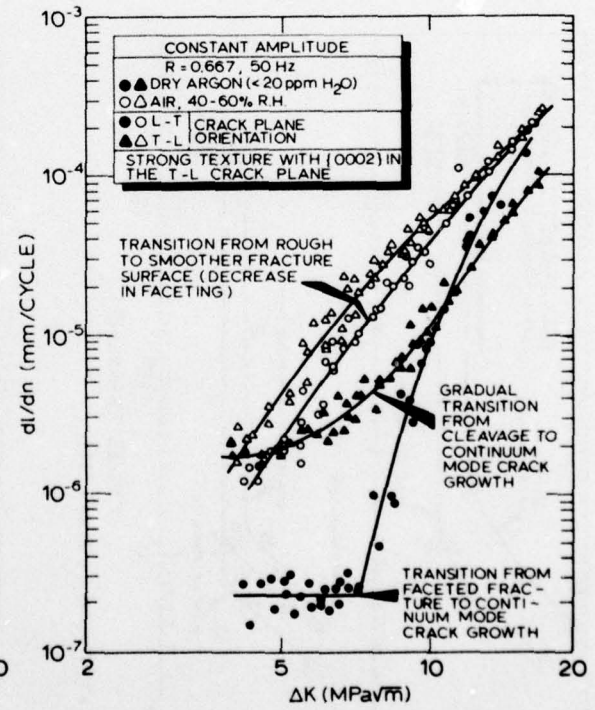
(a) 10mm Ti - 6Al - 4V SHEET [98].



(b) EFFECT OF MEAN STRESS FOR 10mm Ti - 6Al - 4V SHEET HIGH (α + β) ANNEALED, WATER QUENCHED AND AGED [98, 99].

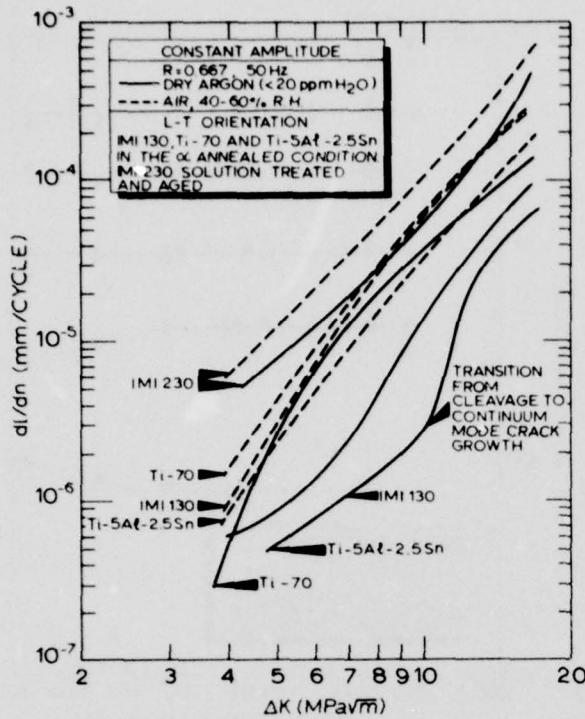


(c) 12.7mm THICK SPECIMENS OF 16mm PLATE [106, 108, 109].

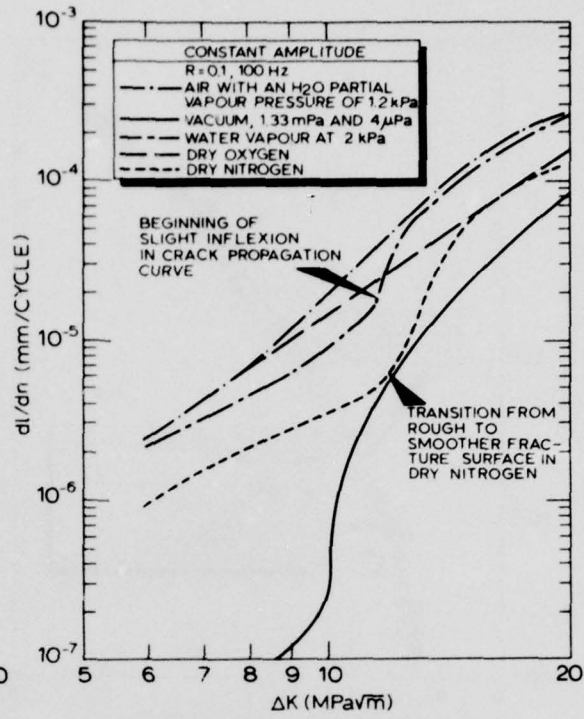


(d) 2.2mm Ti - 6Al - 4V MILL ANNEALED SHEET [105].

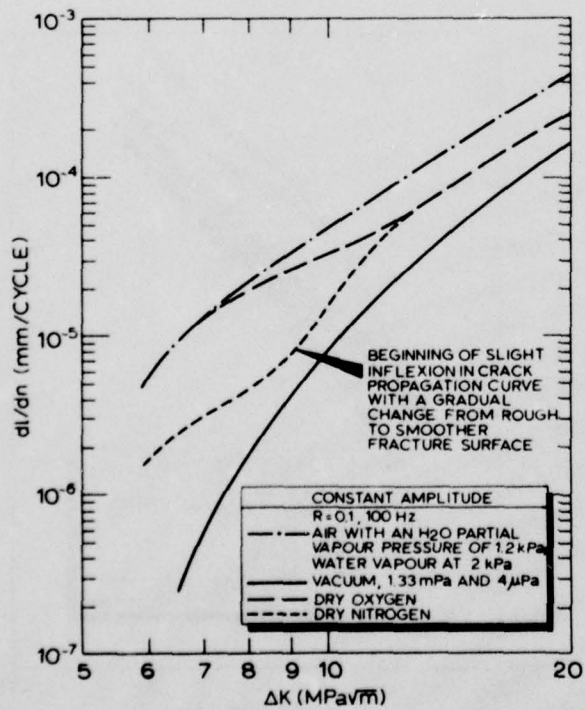
FIGURES 10a - 10d: VACUUM AND GASEOUS EFFECTS ON TITANIUM ALLOY FATIGUE CRACK PROPAGATION.



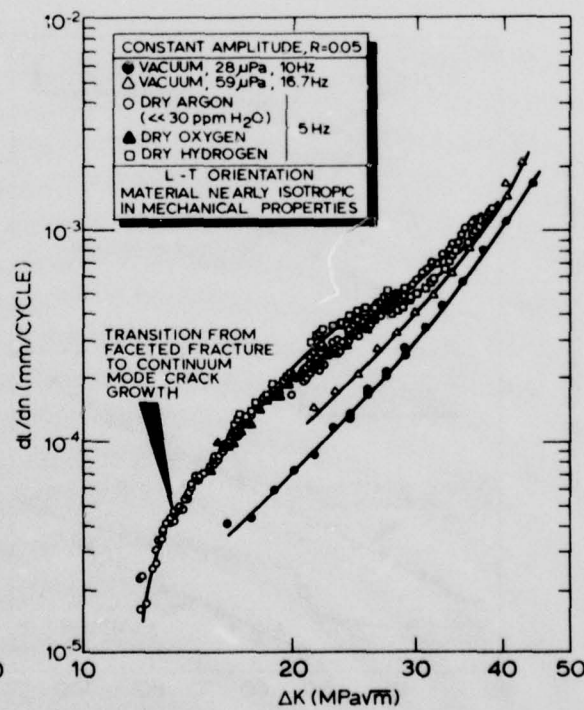
(e) MEDIUM STRENGTH 2mm SHEET MATERIALS [107, 110].



(f) 1.6mm Ti-5Al-2.5Sn α ANNEALED SHEET [111].



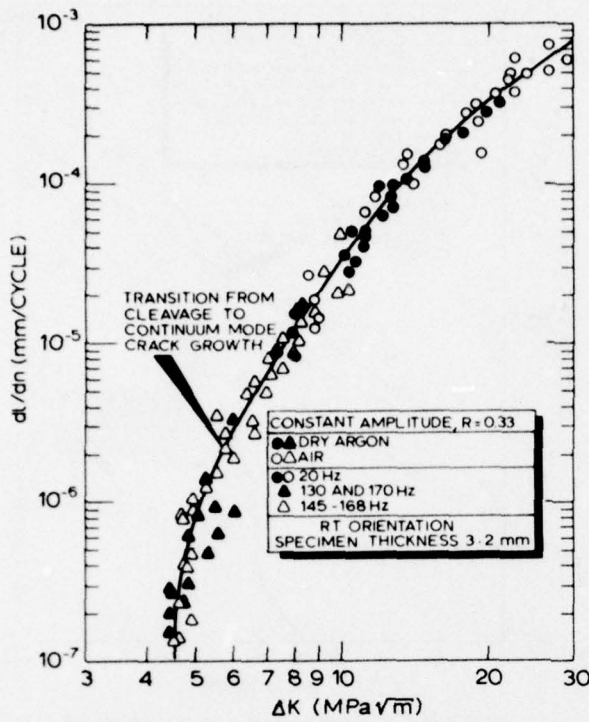
(g) 1.6mm IMI 130 (C.P.Ti WITH LOW OXYGEN CONTENT) α ANNEALED SHEET [111]



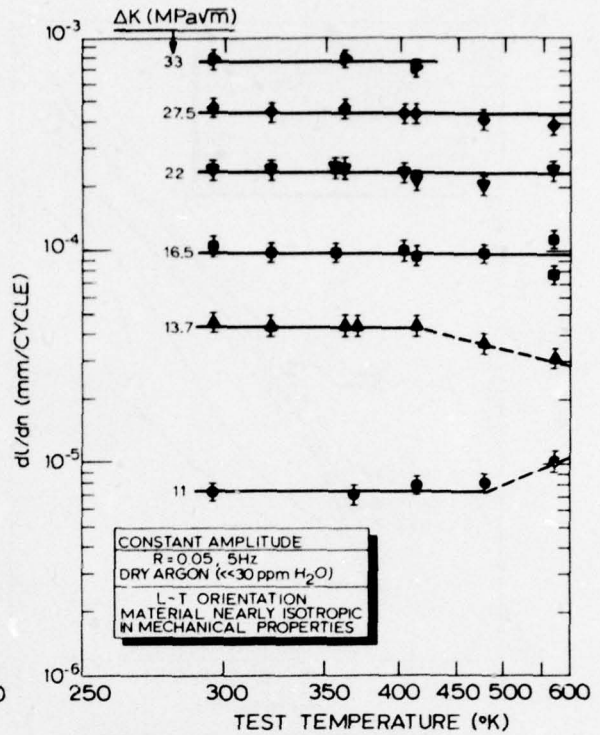
(h) 6.35mm Ti-6Al-4V MILL ANNEALED SHEET [112].

FIGURES 10e - 10h: VACUUM AND GASEOUS EFFECTS ON TITANIUM ALLOY FATIGUE CRACK PROPAGATION.

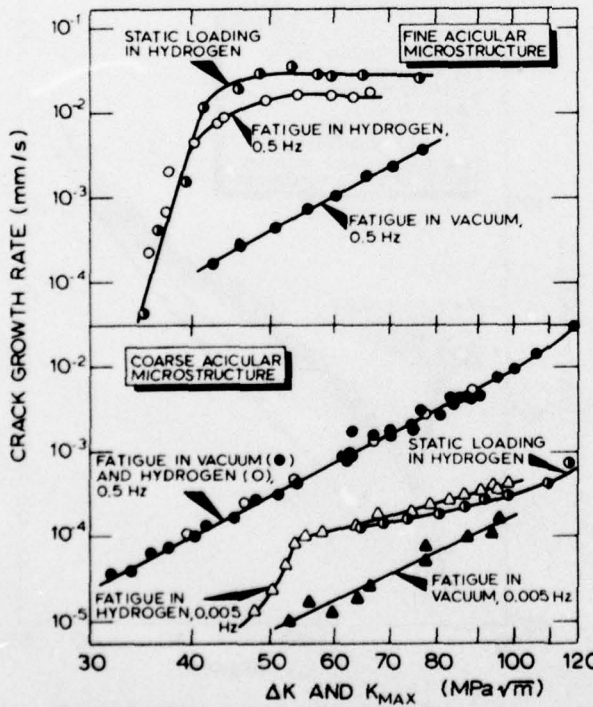




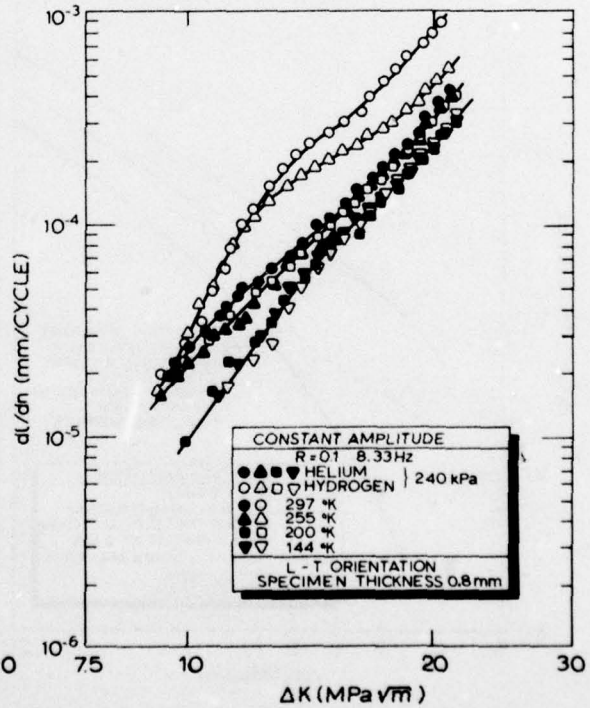
(i) FREQUENCY EFFECTS FOR A Ti - 6Al - 4V ( $\alpha + \beta$ ) SOLUTION TREATED AND AGED COMPRESSOR DISC FORGING [ 100].



(j) EFFECT OF TEMPERATURE FOR 6.35 mm Ti - 6Al - 4V MILL ANNEALED SHEET [ 112].

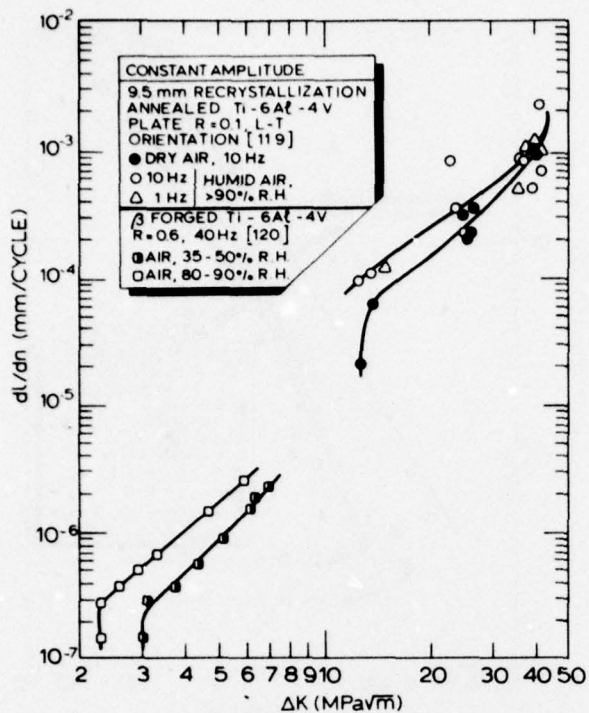


(k) 12.7 mm Ti - 5Al - 2.5 Sn  $\beta$  ANNEALED PLATE TESTED IN VACUO (< 0.67 Pa) AND HYDROGEN (91 kPa) AT R=0 UNDER SQUARE WAVE LOADING [ 115, 116].

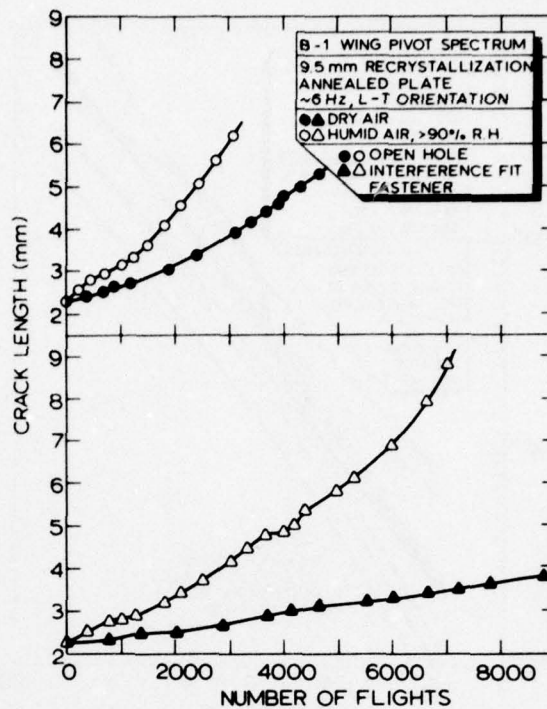


(l) ELI Ti - 6Al - 4V ( $\alpha + \beta$ ) SOLUTION TREATED AND AGED FORGINGS [ 118].

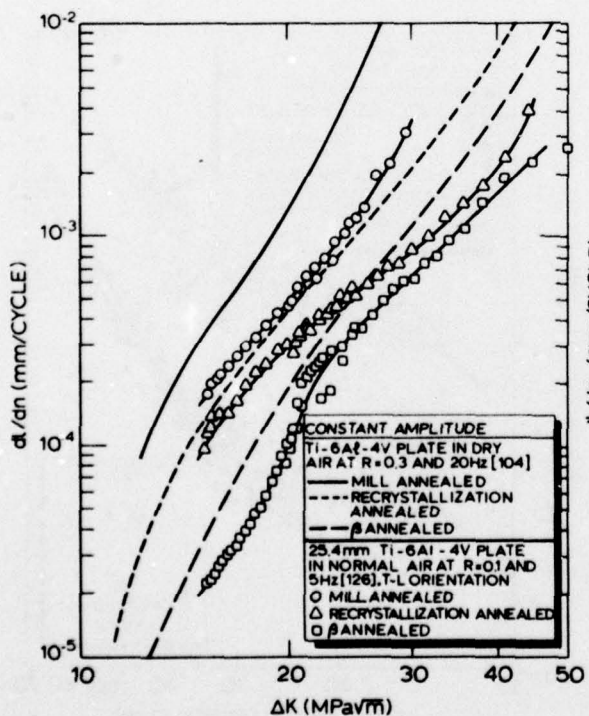
FIGURES 10i - 10l: VACUUM AND GASEOUS EFFECTS ON TITANIUM ALLOY FATIGUE CRACK PROPAGATION.



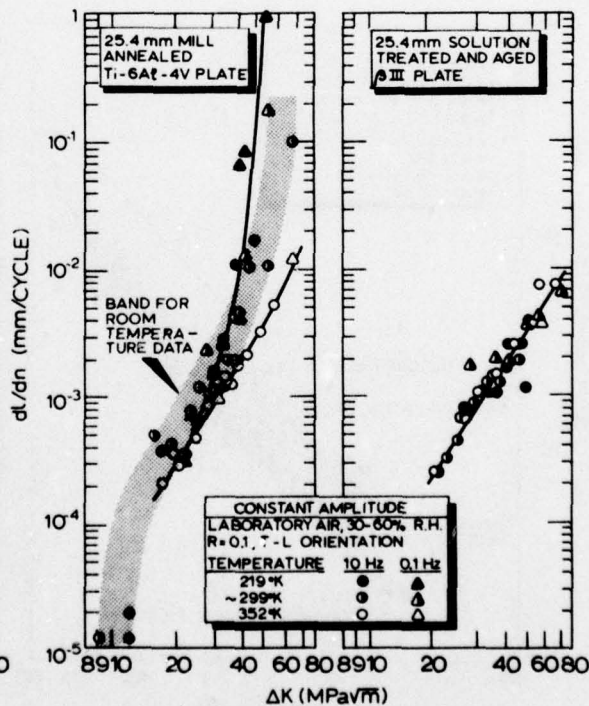
(a) Ti-6Al-4V PLATE AND FORGINGS [119, 120].



(b) Ti-6Al-4V RECRYSTALLIZATION ANNEALED PLATE [121].

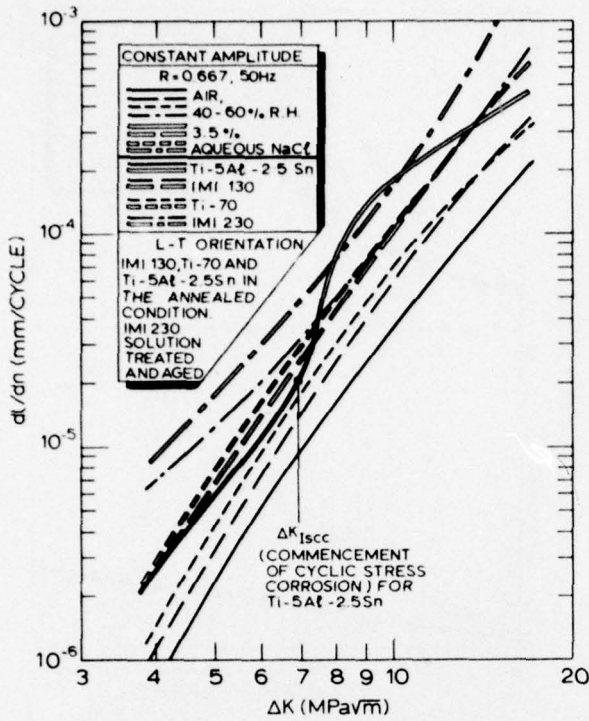


(c) Ti-6Al-4V PLATE [104, 126].

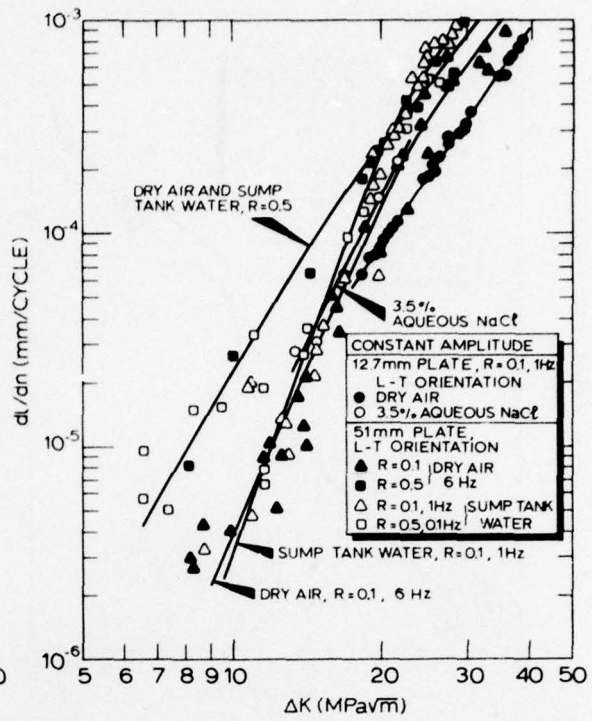


(d) EFFECT OF TEMPERATURE FOR Ti-6Al-4V AND β III PLATE [131].

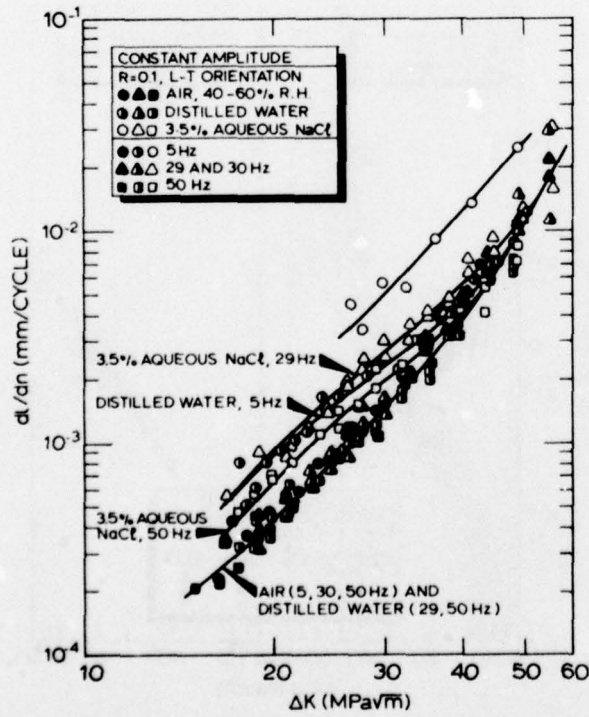
FIGURE 11: ATMOSPHERIC HUMIDITY EFFECTS ON TITANIUM ALLOY FATIGUE CRACK PROPAGATION.



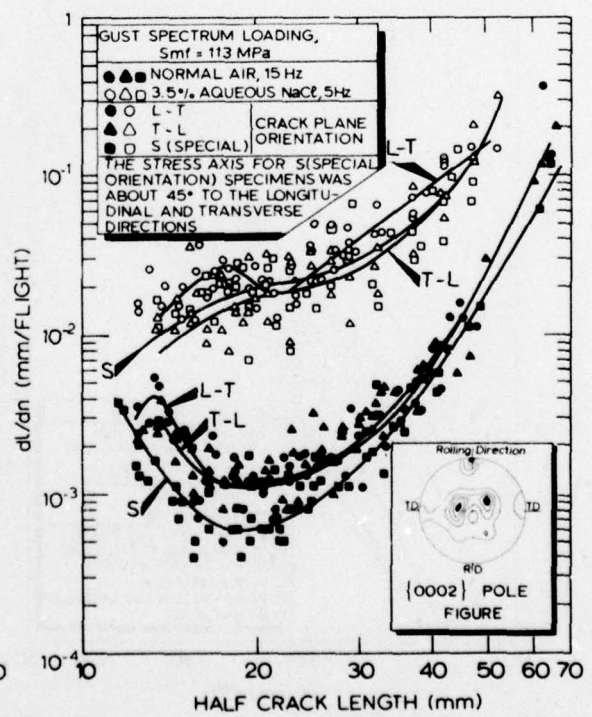
(a) MEDIUM STRENGTH 2mm SHEET MATERIALS [ 110].



(b) Ti-6Al-4V MILL ANNEALED PLATE [ 72, 147]

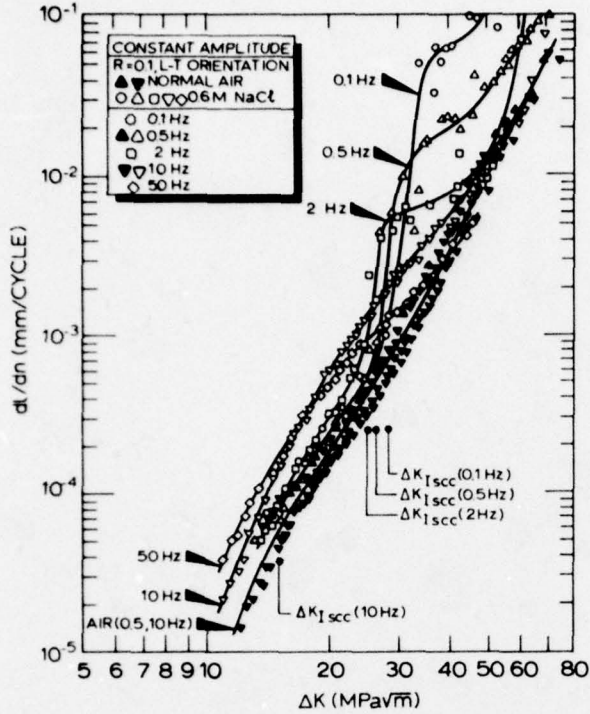


(c) FREQUENCY EFFECTS FOR 2mm IMI 230 SOLUTION TREATED AND AGED SHEET [ 113].

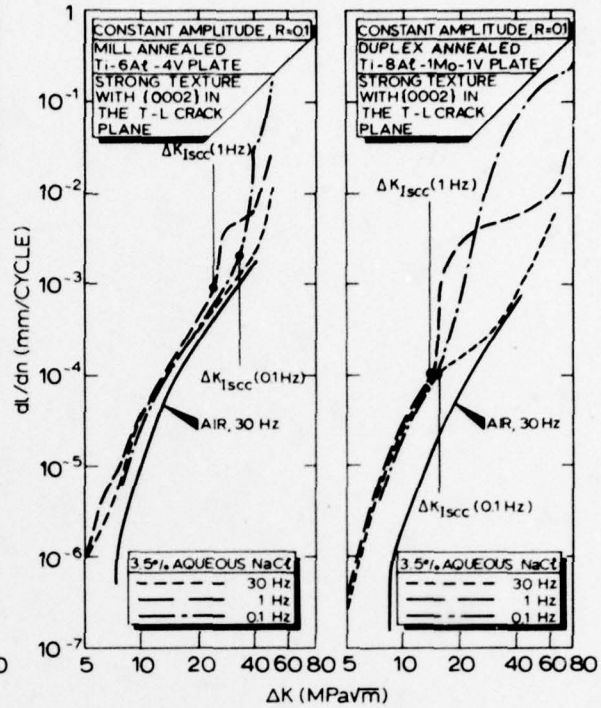


(d) 0.5mm Ti-6Al-4V MILL ANNEALED SHEET [ 49].

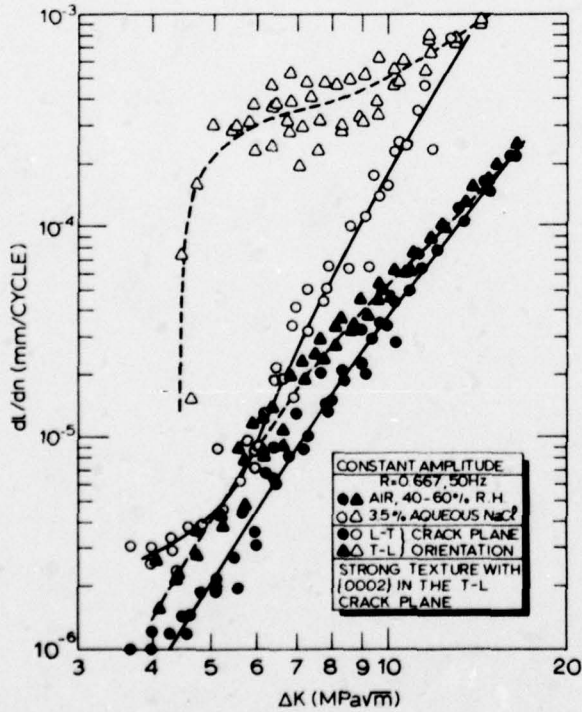
FIGURES 12a - 12d: EFFECTS OF AQUEOUS ENVIRONMENTS ON TITANIUM ALLOY FATIGUE CRACK PROPAGATION.



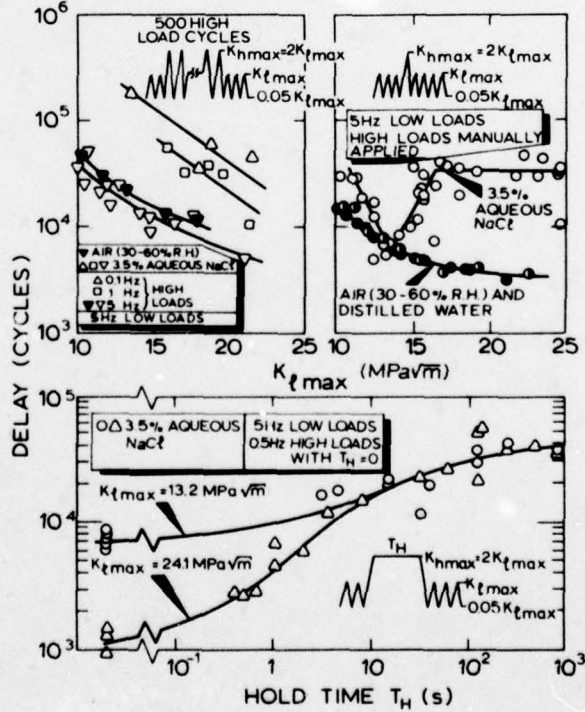
(e) FREQUENCY EFFECTS FOR 3mm Ti - 6Al - 6V - 2Sn (α + β) DUPLEX ANNEALED SHEET [ 114 ].



(f) FREQUENCY EFFECTS FOR 12.7mm THICK SPECIMENS OF 16mm PLATE [ 149 ].



(g) 2.2mm Ti - 6Al - 4V MILL ANNEALED SHEET [ 105 ].



(h) EFFECTS OF ENVIRONMENT AND HOLD TIME ON CRACK GROWTH DELAYS IN 5.1mm Ti - 6Al - 4V MILL ANNEALED PLATE, T - L ORIENTATION [ 122, 134 ].

FIGURES 12a - 12h: EFFECTS OF AQUEOUS ENVIRONMENTS ON TITANIUM ALLOY FATIGUE CRACK PROPAGATION.

## CORROSION FATIGUE OF TITANIUM ALLOYS

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## SUMMARY

Titanium alloys are susceptible to stress corrosion cracking and corrosion fatigue in aqueous halide solutions especially in a precracked state. The fatigue limit of smooth specimen can be up to 15 % lower than in air. But also no effect and sometimes a higher fatigue strength was observed.

For all titanium alloys an effect of aqueous environments on crack propagation rate can be expected. At high  $\Delta K$  crack propagation rate  $da/dN$  increases with decreasing frequency, at medium  $\Delta K$  the  $da/dN$ - $\Delta K$ -curves of different frequencies intersect, at low  $\Delta K$  crack propagation rate is independent of frequency. The superposition model cannot explain the onset of cyclic stress corrosion cracking at  $\Delta K_{ISCC} < (1-R)K_{ISCC}$ .

Fractographic and electrochemical investigations can give valuable informations on the corrosion fatigue process.

Crack propagation under varying amplitudes in a corrosive environment is very complex and needs further investigations on delay effects and the transition effects after change of frequency or wave form.

## 1. INTRODUCTION

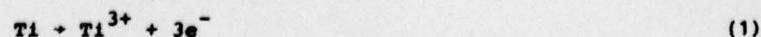
Titanium is a very reactive material, but is protected by an oxide layer from the environment. In addition, after a destruction of the surface layer in air and in aqueous environments there is a rapid repassivation. Therefore titanium and its alloys are known as corrosion resistant materials. Also under constant or alternating stresses the effect of most environments is small, if smooth specimens are tested. Since the fundamental observation of Brown [1] in 1966, however, it has been known that there can be a dangerous effect of the environment if stress raisers are existent. Thereafter many investigations have shown the stress corrosion susceptibility of precracked titanium alloys. From these results it is evident that the environment can also affect the fatigue behavior, especially the fatigue crack propagation. There is a wide range of applications where the behavior of titanium alloys in corrosive environments is of interest, for instance for aircraft and vessel components or for surgical implants.

This report summarizes some results on the effect of environment, especially of salt solutions, on the fatigue behavior of titanium alloys.

## 2. THE ELECTROCHEMICAL BEHAVIOR OF A FRESHLY EXPOSED TITANIUM SURFACE IN A SALT SOLUTION

The mechanisms of stress corrosion cracking and corrosion fatigue in titanium alloys are not fully understood. Some information exist on the electrochemical reactions of a freshly generated surface with the environment. The most important observation is the rapid passivation of a freshly generated surface. This can be shown by measuring the time dependence of the corrosion current of a specimen broken in aqueous halide solutions [2 - 8] or by ellipsometry [9]. Buhl and Raetzer-Scheibe [5 - 8] observed after the formation of fresh surface a constant current and afterwards an exponential decrease. The time during the constant current was in the order of  $10^{-1}$  to  $10^{-2}$  sec, dependent on the anion concentration. Ambrose and Kruger [9] measured for Ti-8Al-1Mo-1V in 5,8 % NaCl-solution 38 m sec for the formation of an oxide monolayer. The current density in the constant current phase is not exactly known. The results vary between  $10 \text{ mA/cm}^2$  [7] and more than  $10 \text{ A/cm}^2$  [4].

The main reaction in the constant current range is the dissolution of titanium:



The repassivation reaction can be



forming an oxide layer, or



forming a salt layer. Beck [4] favours the formation of a salt layer at the tip of a growing crack.

Also hydrogen is formed:



At the crack tip these reactions can occur. Under constant load conditions the pH of the solution at the crack tip decreases because of reaction (2) [10]. In an alternating load test there is an exchange of the solution during each cycle and therefore such an increase is not expected.

According to the reactions (1) and (4) two different mechanisms for stress corrosion cracking and corrosion fatigue are proposed:

- a) Anodic dissolution according to Eq. (1) along crystallographic planes. A prerequisite for such a mechanism is a high current density. Beck [4] and Buhl [8] favour this mechanism.
- b) Hydrogen assisted crack growth. Hydrogen can diffuse into the material and embrittle the material by the formation of titanium hydride or accelerate crack growth by the formation of hydrogen voids. The hydrogen mechanism is favoured by Nelson [11] and Vassel [12].

### 3. CORROSION FATIGUE AND STRESS CORROSION CRACKING

There are connections between stress corrosion cracking and corrosion fatigue. The fundamental processes for both loading conditions - constant load and alternating load - are identical. In both cases first the oxide layer has to be destroyed by plastic deformation and after the destruction the same reactions are occurring in principle in both cases. Because of the rapid repassivation process, restoring a protective layer, in both cases a strong effect of the environment occurs especially if there is a concentration of plastic deformation, i.e. at the tip of a crack. For a smooth part the corrosion process after plastic deformation can only occur for a very small time.

There is one fundamental difference between constant and alternating load. In an alternating load test there is a continuous production of metallic surface by cyclic plastic deformation. Therefore a continuous reaction can occur between the material and the environment. In a constant load test plastic deformation and production of metallic surface occurs especially during the loading. After the constant load is reached there is much less plasticity and therefore the corrosion process ceases. Only at sufficiently high loads in the presence of stress raisers there is a continuous production of metallic surface in a corrosive environment. For this reason a reaction between the material and the environment can occur in fatigue at much lower loads than in stress corrosion cracking.

### 4. S-N-CURVES

#### 4.1 Smooth specimens

Only a limited number of investigations on the effect of environment on the fatigue life and fatigue strength of smooth specimens are published. In Table 1 the ratio of fatigue strength in a corrosive environment  $\sigma_{fc}$  to the fatigue strength in air is shown. Nearly all environmental tests were performed in NaCl-solution.

In pure titanium, Waterhouse and Dutta [13] and Smith [14] found a reduction of the fatigue strength in NaCl-solution of 15 %. For the same material Hughes et al. [15] observed also a lower (11 %) fatigue strength in NaCl-solution at  $10^7$  cycles. At  $10^8$  cycles, however, fatigue strength in NaCl was higher than in air due to the further decrease of the S-N-curve in air between  $10^7$  and  $10^8$  cycles.

In the binary alloys Ti-2,5Cu and Ti-8Al the same fatigue strengths were found in NaCl-solution and in air [13, 17].

For Ti-6Al-4V, Waterhouse and Dutta [13] observed in 1 % NaCl a 13 % lower fatigue strength than in air. Smith and Hughes [18] found no difference and Hughes et al. [15] measured a higher fatigue strength in NaCl-solution at  $10^6$  cycles.

The reason for the higher fatigue strength in salt solution, observed in some investigations, is possibly due to the cooling effect of the liquid environment.

The only investigation available for other environments than salt solution is the work of Wanhill [17] in methanol-HCl. For Ti-8Al a considerable reduction in the fatigue strength compared with air or NaCl-solution was found.

#### 4.2 Notched specimens

Hughes et al. [15] determined the fatigue strength of notched specimens of some titanium alloys. Cylindrical specimens with a radial notch with  $K_t = 4$  were tested. The results are shown in Table 2. In salt water a fatigue strength reduction up to 24 % was observed.

### 5. CRACK INITIATION

There are no quantitative comparative investigations in air and salt solution concerning crack initiation. An indirect conclusion can be drawn from the results on the fatigue strength. In titanium and titanium alloys no cracks occur below the fatigue limit. If a crack is formed by repeated cycling, then this crack propagates until failure. From the lower fatigue strength found in some investigations in NaCl-solution it can therefore be concluded that fatigue cracks can be initiated at a lower amplitude than in air.

Wells and Sullivan [20] found the same mode of crack initiation in NaCl-solution and air for Ti-6Al-4V. Also Wanhill [17] observed for Ti-8Al a similar crack initiation process by reversed slip in NaCl-solution and in air. In methanol-HCl environment, where the fatigue strength is considerable lower, a mixture of intergranular and transgranular fracture was observed.

### 6. STRESS CORROSION CRACKING OF PRECRACKED SPECIMENS

Because of the connection between stress corrosion crack propagation and fatigue crack propagation the fundamental observations of stress corrosion cracking of precracked specimens are summarized. Tests with precracked titanium alloy specimens usually are performed in cantilever beam loading. In this test the stress intensity factor increases with crack extension. Using a number of specimens which are loaded to different load levels a critical  $K_{ISCC}$  is determined, below which no fracture of the specimens occurs. The value  $K_{ISCC}$  or the ratio  $K_{ISCC}/K_{Ic}$  characterizes the stress corrosion susceptibility of a given combination material/environment.

The results of tests in salt solutions and of comparative tests in air or vacuum can be summarized as follows:

1. Stable crack growth below  $K_{Ic}$  can occur not only in salt solutions but also in air or in vacuum [21 - 27].
2. The threshold in air  $K_{Ith}$  can be below or above the threshold  $K_{ISCC}$  in salt solution [25, 27], see Fig. 1 and 2.
3. Crack growth rate in salt solution is much higher than in air. In air there is a continuous increase of  $da/dt$  with  $K_I$ , whereas in salt solution a plateau region exists [27 - 29].
4. There exists a region  $K_{Ith}^* < K_{Ii} < K_{ISCC}^*$ , in which stress corrosion cracks are initiated. After some crack extension, however, the crack stops [27, 30 - 33]. For an alloy with high stress corrosion susceptibility such as Ti-8Al-1Mo-1V stress corrosion cracks are initiated if  $K_{Ii} > K_{ISCC}^* = K_{Imax}$ , where  $K_{Imax}$  is the maximum stress intensity factor during precracking. For the alloy Ti-6Al-4V, stress corrosion cracks are initiated for  $K_{Ii} > K_{ISCC}^* = K_{Ith}$  [50].

## 7. FATIGUE CRACK PROPAGATION

### 7.1 Stage 1 and Stage 2

It is often useful to subdivide fatigue crack propagation into two stages:

stage 1: crack propagation along slip bands favourably oriented to the stress axis.

Stage 1 occurs if cracks are initiated in slip bands. After crack propagation in one or several grains the crack path changes to stage 2.

stage 2: crack propagation macroscopically perpendicular to the stress axis. The microscopic crack path can follow crystallographic planes.

Crack propagation in stage 1 is difficult to investigate because of the small crack length and the low crack propagation rate. Investigation of stage 2 usually is done with precracked specimens by measuring crack propagation rate  $da/dN$  as a function of  $\Delta K_I$ .

### 7.2 General form of the $da/dN$ - $\Delta K_I$ -curve

In Fig. 3 the general form of a log  $da/dN$ -log  $\Delta K$ -curve in an inert environment and in an aqueous environment for a titanium alloy is shown. In both environments the curve has a lower limit  $\Delta K_{Ith}$  and an upper limit  $\Delta K_{Imax}$ . At least at high frequencies  $\Delta K_{Imax}$  is independent of the environment, whereas  $\Delta K_{Ith}$  can be lower in a corrosive environment than in an inert environment. In the corrosive environment at a critical  $\Delta K$ -value, called  $\Delta K_{ISCC}$ , a more or less sudden increase of the crack propagation rate can occur. Some authors distinguish between cyclic stress corrosion cracking for  $\Delta K_I > \Delta K_{ISCC}$  and true stress corrosion fatigue for  $K_I < \Delta K_{ISCC}$  [28, 34].

### 7.3 The superposition model

For the region of cyclic stress corrosion cracking Wei and Landes [35] developed a simple model for the calculation of the crack propagation rate. In this model it is assumed that the crack growth during one cycle ( $da/dN$ )<sub>cf</sub> is the sum of two components. One is the crack growth in the inert environment ( $da/dN$ )<sub>inert</sub> and the other the crack growth due to stress corrosion cracking ( $da/dN$ )<sub>SCC</sub>. The component of stress corrosion cracking is calculated under the assumption that there exists an unique relation between stress intensity factor and crack growth rate. Then the crack growth during one cycle can be calculated by integration of the  $da/dt$ - $K$ -curve. The integration is done along the  $K$ - $t$ -curve for  $K_I > K_{ISCC}$ . From this model three conclusions can be drawn:

- A deviation of the corrosion fatigue curve from the curve in an inert environment occurs if the maximum stress intensity factor of the fatigue cycle exceeds  $K_{ISCC}$ , i.e. for  $\Delta K_I > (1-R)K_{ISCC}$ .
- For  $\Delta K_I > (1-R)K_{ISCC}$  the component of cyclic stress corrosion cracking is proportional to  $1/f$ , where  $f$  is the frequency.
- Comparing different wave forms crack growth rate should increase with increasing portion of load above  $K_{ISCC}$ . For instance a square wave form should have a higher crack growth rate than a sinusoidal wave form.

### 7.4 Susceptible alloys

Crack propagation was especially investigated in NaCl-solution or in seawater. In some investigations the effect of other salt solutions was investigated. Nearly all investigations revealed higher crack growth rate in salt solutions than in air, at least at high  $\Delta K_I$ -values. An effect of salt solutions was found for the following alloys:

Ti in 3,5 % NaCl [36]  
 Ti-2,5 Cu in 3,5 % NaCl [36 - 38]  
 Ti-6Al-4V in 3,5 % NaCl [33, 39 - 44], 0,6M KCl [28], 5M KJ [28]  
 Ti-6Al-6V-2Sn in 3,5 % NaCl [40 - 42], 1 % NaCl [19], 0,6M KBr [42]  
 Ti-8Al-1Mo-1V in 3,5 % NaCl [30, 33, 34, 40, 45]  
 Ti-5Al-2,5Sn in 3,5 % NaCl [36]  
 Ti-7Al-2,5Mo in 3,5 % NaCl [39]  
 Ti-4Al-4Mo-2Sn-0,5Si in 3,5 % NaCl [46]  
 Ti-6Al-2Sn-4Zr-6Mo in 3,5 % NaCl [47]  
 Ti-7Al-2Nb-1Ta in 3,5 % NaCl [39, 48, 49]

From this list, including  $\alpha$  and  $\alpha/\beta$ -alloys, it can be concluded that for all titanium alloys an effect of salt solution on crack propagation rate can be expected.



### 7.5 Effect of composition and microstructure

Because of the limited investigations on crack propagation in salt solutions in titanium alloys it is not easy to find general rules for the effect of alloy composition and metallurgical structure on the corrosion fatigue susceptibility. It can be supposed that all materials with high stress corrosion susceptibility are also highly susceptible to corrosion fatigue. Wanhill [50] recently summarized the metallurgical variables influencing stress corrosion susceptibility with the following trends:

- stress corrosion susceptibility is increased by increasing content of aluminum, oxygen and hydrogen
- the  $\alpha$ -phase is the susceptible phase; precipitation of  $Ti_3Al$  increases the susceptibility;  $\alpha'$  and  $\alpha''$  martensites are almost immune, but tempered  $\alpha'$  and  $\alpha''$  are susceptible
- decreasing grain size and volume fraction of  $\alpha$ -phase decreases the susceptibility
- texture is an important parameter.

In Fig. 4 and 5 two examples are shown of the effect of microstructure on corrosion fatigue susceptibility. In Fig. 4 results of Wanhill [42] are plotted as  $lg da/dN - lg \Delta K_I$ -curves for two materials. One is called Ti-6Al-4V, the other IM1 318, which is also a 6Al-4V-alloy. For IM1 318 crack propagation rate was measured in the two orientations T - L and L - T. From Fig. 4 it can be seen that in air crack propagation rate is very similar. In NaCl-solution, however, there is a great difference showing that the microstructure including the texture has a much stronger influence in NaCl than in air.

Fig. 5 shows results of Williams et al. [46] for the alloy Ti-6Al-2Sn-4Zr-6Mo in two different heat treatments. In one condition (15 min 1000 °C/vacuum cooled, 24 h 600 °C) with a Widmanstätten microstructure no effect of the environment was found. In the other condition (step cooled from 650 °C) with equiaxed ( $\alpha + \alpha_2$ ) +  $\beta$  in NaCl-solution a higher crack growth rate was observed.

### 7.6 Effect of frequency

For all time-dependent processes in environmental fatigue an effect of frequency can be expected. Fatigue crack propagation in titanium alloys in corrosive environments at different frequencies was measured by Bucci [30], Meyn [34], Dawson and Pelloux [41, 42] and by Döker and Munz [33].

In Fig. 6 - 8 some results are given. From these and other investigations the  $da/dN - \Delta K$ -curve can be subdivided into different regions.

1. At very high crack growth rates there is no effect of the environment. The curve for air and corrosive environment merge above a critical  $\Delta K$ -value. This critical value increases with decreasing frequency.
2. At lower  $\Delta K$ -values there is a region of increasing crack growth rate with decreasing frequency in qualitative agreement with the superposition model of Wei and Landes.
3.  $\Delta K_{ISCC}$  - the  $\Delta K_I$ -value at the rapid increase of crack growth rate - increases with decreasing frequency [33, 41]. Therefore a region exists with an intersection of the  $da/dN - \Delta K_I$ -curves of different frequencies. In contrast to the prediction of the superposition model there is  $\Delta K_{ISCC} < (1-R)K_{ISCC}$ . For the higher frequencies no  $\Delta K_{ISCC}$  can be evaluated from the  $da/dN - \Delta K$ -curves.
4.  $\Delta K_{Ith} < \Delta K_I < \Delta K_{ISCC}$   
Meyn [34] and Döker and Munz [33] observed a frequency-independent region at low  $\Delta K_I$ . Bucci [30] and Dawson and Pelloux [41, 42] observed for  $\Delta K_I < \Delta K_{ISCC}$  a lower crack growth rate for the lower frequency.

There are two observations not predicted by the superposition model of Wei and Landes: 1.  $\Delta K_{ISCC} < (1-R)K_{ISCC}$  2. The decreasing crack growth rate with decreasing frequency in the region 3. In section 6 it was shown that in a stress corrosion test stress corrosion cracks can occur also below  $K_{ISCC}$ . Therefore it is not astonishing that also for  $\Delta K < (1-R)K_{ISCC}$  cyclic stress corrosion cracking can occur. The explanation for the frequency dependence of  $\Delta K_{ISCC}$  and the intersection of the  $da/dN - \Delta K$ -curves in region 3 is more difficult. Dawson and Pelloux [41], Döker and Munz [33] and Döker [51] tried to give explanations referring to the repassivation time, the active region behind the crack tip and the amount of hydrogen developed.

### 7.7 Effect of wave form

Crack growth rate in a corrosive environment can be effected by the wave form. Important parameters characterizing the wave form are the rise time until maximum load and the time spent above  $K_{ISCC}$ . In the range of cyclic stress corrosion cracking the time above  $K_{ISCC}$  is the most important parameter. Below  $\Delta K_{ISCC}$  the rise time can be important.

Dawson and Pelloux [41] compared sine and square wave form for Ti-6Al-6V-2Sn in 0,6M KBr at 1 and 10 Hz. Above and below  $\Delta K_{ISCC}$  a higher crack growth rate was observed for the square wave form. Above  $\Delta K_{ISCC}$  this observation is in qualitative agreement with the superposition model. Below  $\Delta K_{ISCC}$  the wave form with the shorter rise time (square) has a higher crack growth rate. This is in agreement with the frequency effect at low  $\Delta K$ , observed by the same authors.

### 7.8 Fractographic results

Fractographic investigations with the scanning electron microscope can give some information on the corrosion fatigue process. Döcker [51] has examined the fracture surfaces of Ti-6Al-4V and Ti-8Al-1Mo-1V in different environments at crack growth rates between  $2 \times 10^{-4} \mu\text{m}/\text{cycle}$  to  $10^2 \mu\text{m}/\text{cycle}$ . The appearance of the fracture surface depends on environment, crack growth rate and frequency. Characteristic features, observed in 3,5 % NaCl-solution, were:

- a) cleavage facets at low crack growth rates (Fig. 9a)
- b) striations at medium crack growth rates
- c) dimples at high crack growth rates, mostly in combination with striations (Fig. 9c)
- d) cleavage facets, as observed in stress corrosion cracking (Fig. 9b); similar to a) but more serrated.

The range of crack growth rate for d) depends on the frequency. The higher the frequency, the lower the  $da/dN$ -range of cyclic stress corrosion cracking. Therefore for high frequencies (30 Hz), in a test with increasing  $\Delta K$  there is a continuous transition from the cleavage facets at low crack growth rates to the stress corrosion cleavage facets. For low frequencies a region with striations can be seen between the two cleavage regions.

From the similarity between cleavage at low crack growth rate, cyclic stress corrosion cleavage and stress corrosion cleavage and the effect of frequency some difficulties are involved in the characterization of service failures.

### 7.9 Electrochemical measurements

During a cyclic test in a salt solution the electrochemical potential changes in phase with the load [51]. The amplitude of this fluctuation, the wave form, and the mean potential depends on frequency and crack growth rate. In Fig. 10 the mean potential is plotted against crack growth rate for two frequencies for the alloy Ti-6Al-4V measured against a Ag/AgCl-electrode. The electrochemical potential decreases from about 200 mV at low crack growth rates to -400 mV for 1 Hz and -600 mV for 30 Hz. When the test is stopped, the potential increases. When the test is started again the potential gradually decreases to the value before the interruption. It takes some time until the steady state potential is reached.

The change of the potential during one cycle gives some information on the active time of the material. As an example Fig. 11 shows the potential change for a square wave with three amplitudes, all in the range of cyclic stress corrosion cracking. For the lowest  $\Delta K_I$  (curve a) at first the potential decreases at maximum load, showing that the crack is propagating. Then the potential increases because of a decrease in the crack propagation rate. At minimum load there is a more rapid increase of the potential due to a repassivation. At higher  $\Delta K_I$  (curve b) after some time at maximum load, the potential is constant, suggesting a constant crack growth rate. At still higher  $\Delta K$  after a sudden decrease of the potential, an increase and afterwards a considerable decrease occurs, showing that the crack growth rate at the constant maximum load is not constant.

Such measurements are useful for investigations concerning the superposition model, the effect of frequency and of the damage in random load fatigue.

### 7.10 Crack propagation under varying amplitudes

There is a lack of investigations on the effect of environment on the crack growth rate under random load conditions for titanium alloys. In an investigation by Lockheed Aircraft Corporation [52] it was found that under flight simulation loads Ti-6Al-4V is highly sensitive to corrosive environment. To obtain a crack extension of 5 mm in salt water 120 flights, in moist air 2440 flights and in dry air 4400 flights were required.

Because of the interacting effect (effect of previous load cycles on crack extension of the following load cycle) a prediction of crack growth rate under random load conditions is difficult. It is much more difficult if the effect of environment is included. From the limited results for titanium alloys only some trends can be shown:

- a) The delay in crack growth rate after one or several overload cycles can depend on the environment. For mildly aggressive environments such as moist air or distilled water Shih and Wei [44] found the same delay as in dehumidified argon. For 3,5 % NaCl-solution, however, a larger delay occurred than in air.
- b) The frequency of random load testing in an aggressive environment can be very important. In titanium alloys at high amplitudes the frequency effect is most important (see Fig. 6 - 8). Therefore an increase of crack growth rate with decreasing frequency should be expected. Some additional effects, however, complicate the behavior:

Shih and Wei [43] found a delay after 500 high load cycles which increased with decreasing frequency of the 500 high load cycles (fig. 12). This effect counteracts the increase of fatigue crack growth with decreasing frequency at high  $\Delta K$ .

After a change in frequency there are transition effects in the crack growth rate [53]. Similar effects are observed in experiments on the effects of hold time at maximum load on the fatigue crack propagation [54]. In these tests the wave form was changed from triangular with a frequency of 0,33 Hz to trapezoidal with the same load increase and decrease as for the triangular load cycle and a hold time of 5 min at maximum load. Immediately after the change of the wave form to trapezoidal the crack growth rate was higher by a factor of more than ten. During further cycling the crack growth rate decreased considerably.

There are different effects which can contribute to the complex crack growth behavior in random loading or after a change in frequency or wave form: change in electrochemical potential, build-up of corrosion products, change in crack closure load, crack tip blunting. To separate these and additional effects further investigations are necessary.

## 8. PROPOSALS FOR FURTHER RESEARCH

It is possible to regard corrosion fatigue from two points of view. It is possible to investigate the fundamental processes of corrosion fatigue, the micromechanism, the reactions between the environment and the material, especially at the tip of a crack. The other side is a more practical one. The question is: is it possible to predict the fatigue life of a component under random loading in a special environment from tests under simple loading conditions. Is it at least possible to predict the ranking of different materials or of different microstructures of one material under random loading from constant amplitude tests in a corrosive environment. After reviewing the literature on titanium alloy it can be concluded that both sides of corrosion fatigue - the fundamental one and the practical one - are not well illuminated. Further investigations are necessary which should include the following:

1. Crack initiation in notched specimens
2. Transition effects after change of frequency and wave form
3. Effect of hold times
4. Correlation between  $da/dN-\Delta K_I$ -curves and crack propagation in random load fatigue, beginning with change in amplitude or mean load and overload effects
5. Further investigation of the correlation between stress corrosion cracking and  $da/dN-\Delta K_I$ -curves
6. General rules of the effect of microstructure
7. Electrochemical investigations in crack propagation experiments.

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Table 1 Ratio of fatigue strength in a corrosive environment  $\sigma_{fc}$  to fatigue strength in air  $\sigma_f$

material	environment	$\sigma_{fc}/\sigma_f$	reference	remarks
titanium	1 % NaCl	0,85	Waterhouse, Dutta [13]	10 <sup>8</sup> cycles 10 <sup>7</sup> cycles
	3,5 % NaCl	0,85	Smith [14]	
	3,5 % NaCl	1,09	Hughes et al. [15]	
	3,5 % NaCl	0,89	Hughes et al. [15]	
	synth. seawater	1,04	Cotton, Downing [16]	
Ti-2,5Cu	1 % NaCl	1	Waterhouse, Dutta [13]	
Ti-8Al	3 % NaCl	1	Wanhill [17]	solution treated aged
	methanol-HCl	0,50	Wanhill [17]	
	methanol-HCl	0,22	Wanhill [17]	
Ti-6Al-4V	1 % NaCl	0,87	Waterhouse, Dutta [13]	
	3,5 % NaCl	1	Smith, Hughes [18]	
	3,5 % NaCl	1,15	Hughes et al. [15]	
Ti-6Al-6V-2Sn	3,5 % NaCl	1	Levy et al. [19]	
Ti-4Al-4Mo-2Sn	3,5 % NaCl	1,17	Hughes et al. [15]	
Ti-4Al-4Mo-4Sn	3,5 % NaCl	1	Hughes et al. [15]	
Ti-8Al-1Mo-1V	3,5 % NaCl	1	Levy et al. [19]	
Ti-22Sn-4Mo-2,25Al	3,5 % NaCl	0,86	Hughes et al. [15]	

Table 2 Ratio of fatigue strength in 3,5 % NaCl-solution  $\sigma_{fc}$  to fatigue strength in air  $\sigma_f$  for notched specimens,  $K_t = 4 \text{ } 15$

material	$\sigma_{fc}/\sigma_f$
titanium	0,76
Ti-6Al-4V	0,95
Ti-4Al-4Mo-2Sn	0,76
Ti-4Al-4Mo-4Sn	1,0
Ti-22Sn-4Mo-2,25Al	0,79

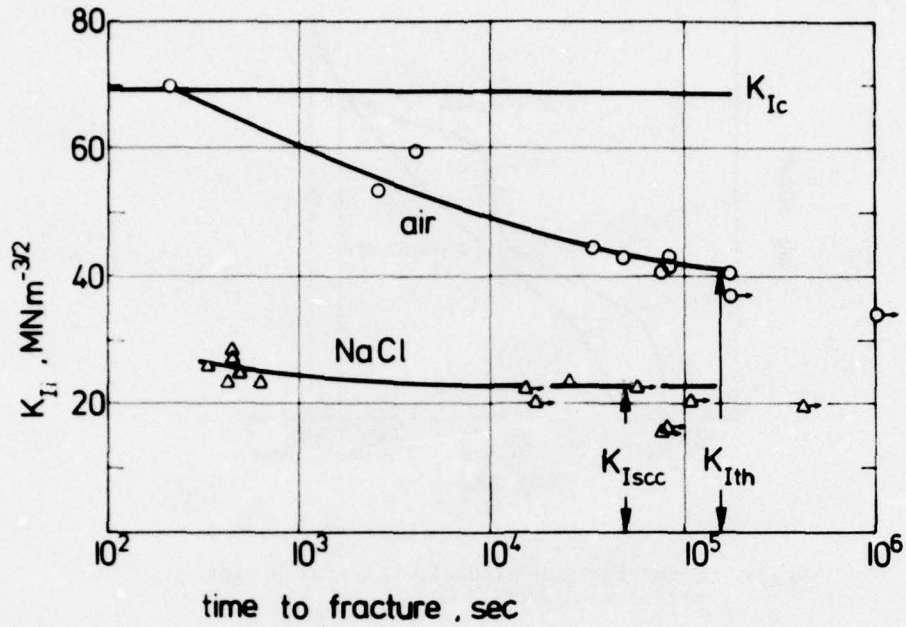


Fig. 1 Initial stress intensity factor  $K_{Ii}$  versus time to fracture in cantilever beam tests, Ti-8Al-1Mo-1V in air and 3,5 % NaCl-solution (Döker, Munz [27])

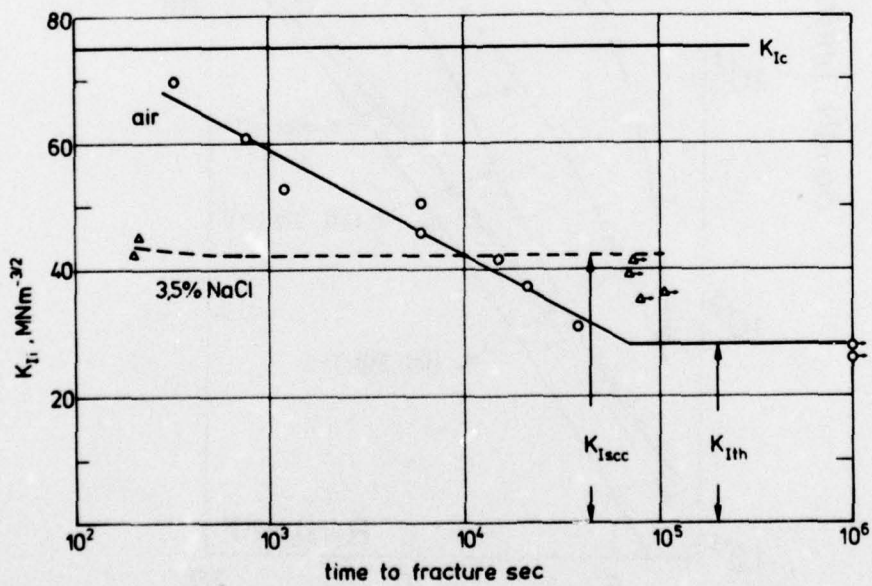


Fig. 2 Initial stress intensity factor  $K_{Ii}$  versus time to fracture in cantilever beam tests, Ti-6Al-4V in air and 3,5 % NaCl solution (Döker, Munz [27]).

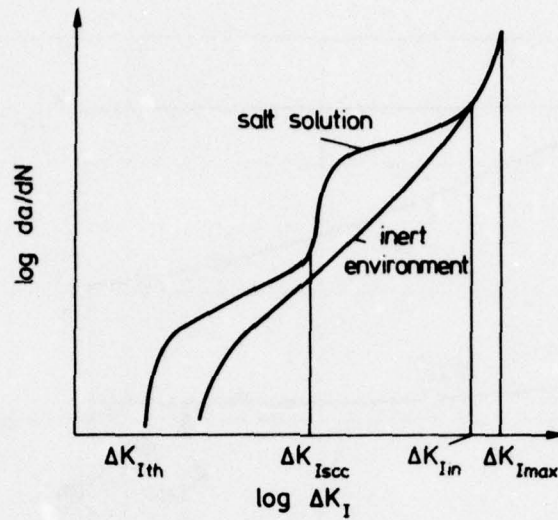


Fig.3 Schematic  $\log da/dN$ - $\log \Delta K_I$ -curves for titanium alloys

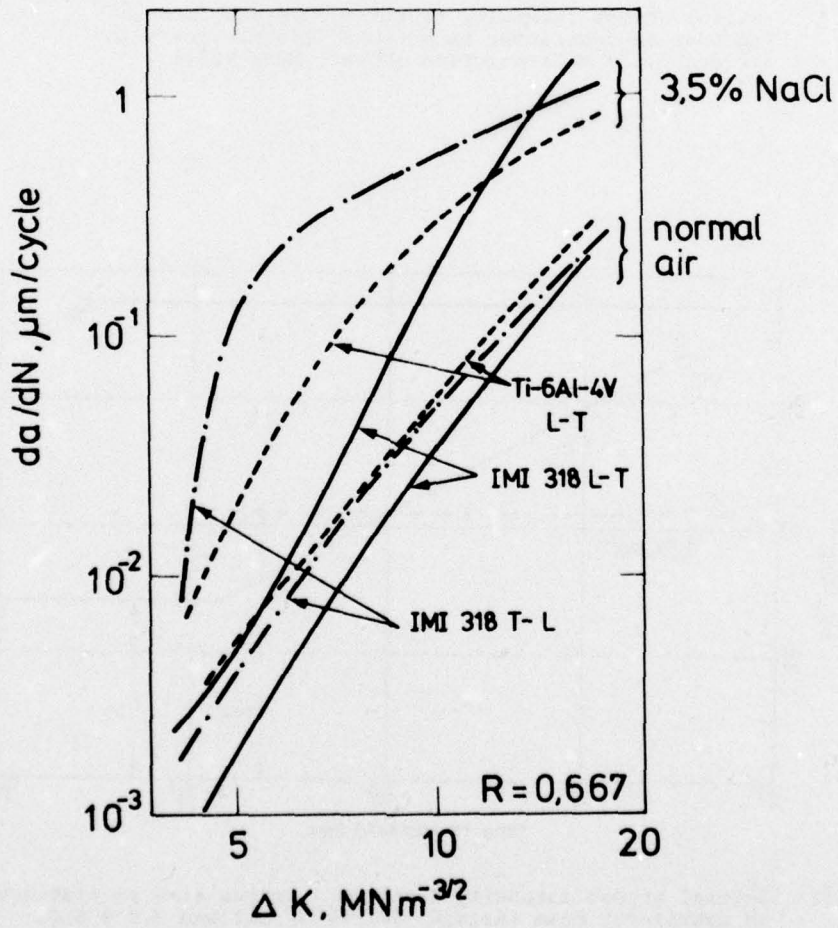


Fig.4  $da/dN$ - $\Delta K_I$ -curves for two different Ti-6Al-4V alloys in air and 3,5 % NaCl solution (Wanhill [42])



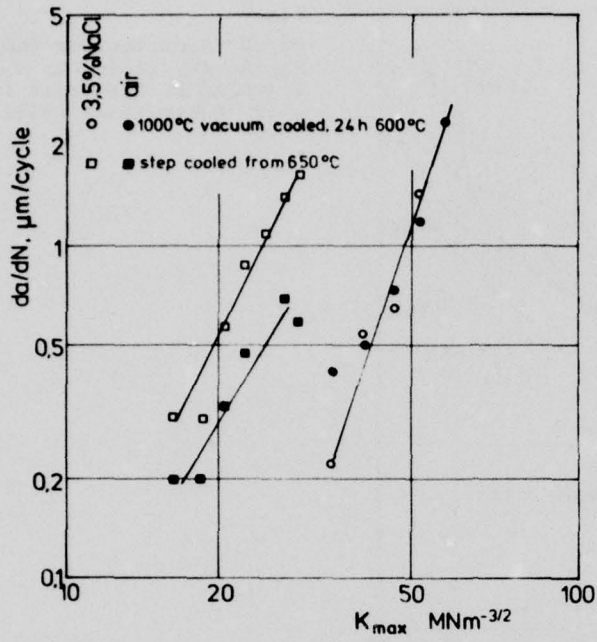


Fig. 5  
da/dN versus  $K_{max} = \Delta K / (1-R)$   
for two different heat treatments of the alloy Ti-6Al-2Sn-4Zr-6Mo (Williams et.al.[47])

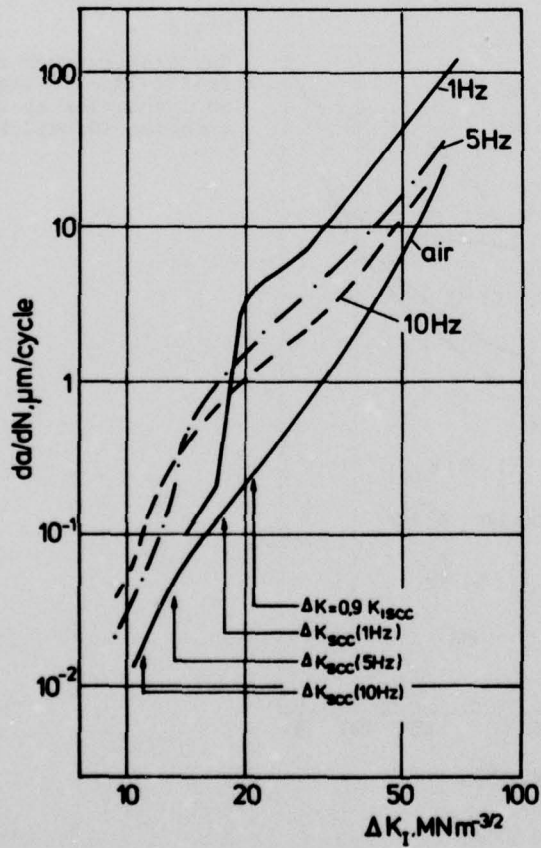


Fig. 6  
da/dN- $\Delta K_I$ -curves for the alloy Ti-6Al-4V, tested in 3,5 % NaCl at different frequencies (Dawson, Pelloux [42])

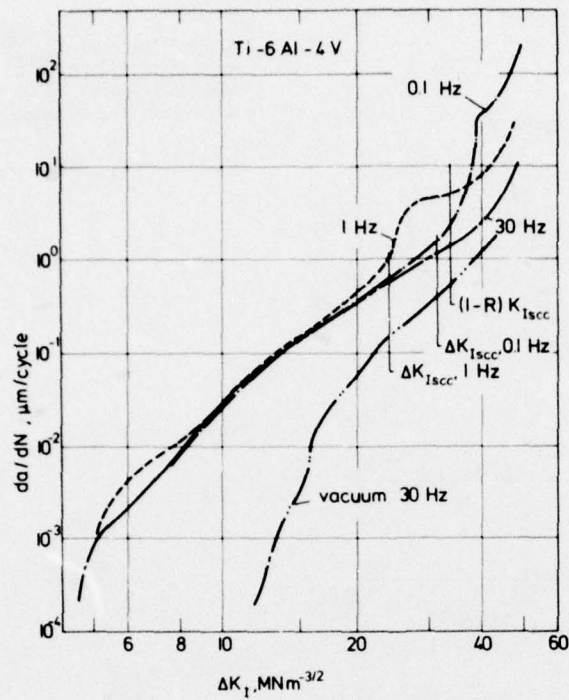


Fig. 7

$da/dN-\Delta K_I$  curves for the alloy Ti-6Al-4V, tested in 3,5 % NaCl solution at different frequencies (Döker, Munz [33])

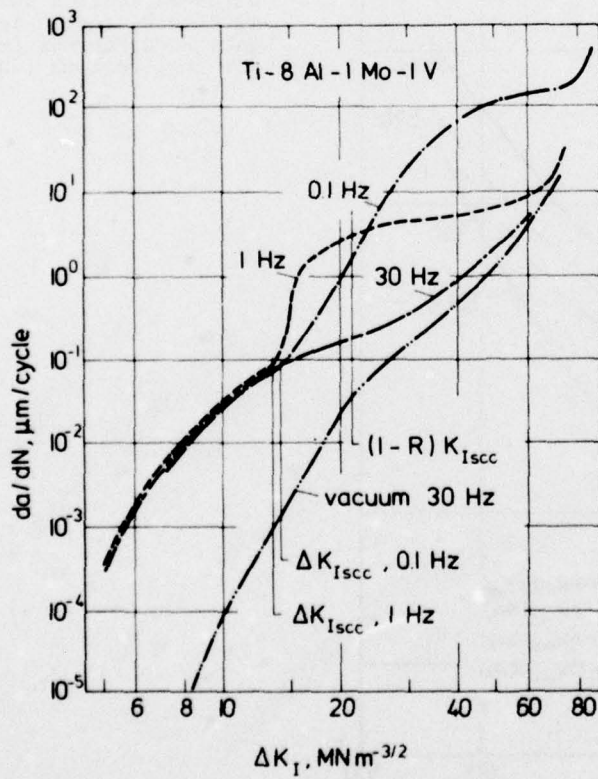


Fig. 8

$da/dN-\Delta K_I$  curves for the alloy Ti-8Al-1Mo-1V, tested in 3,5 % NaCl solution at different frequencies (Döker, Munz [33])

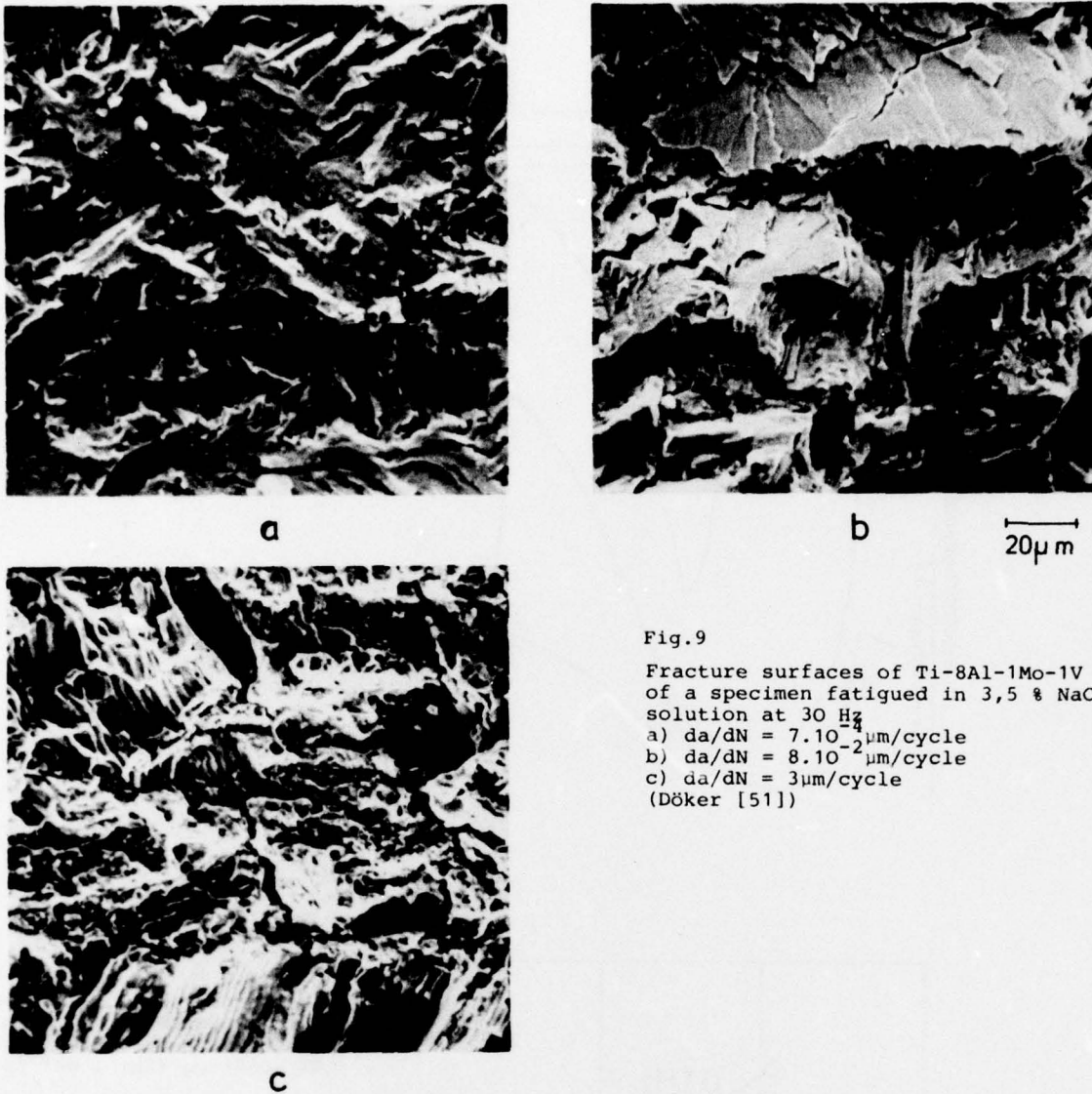


Fig.9

Fracture surfaces of Ti-8Al-1Mo-1V  
of a specimen fatigued in 3,5 % NaCl-  
solution at 30 Hz  
a)  $da/dN = 7 \cdot 10^{-2} \mu\text{m}/\text{cycle}$   
b)  $da/dN = 8 \cdot 10^{-2} \mu\text{m}/\text{cycle}$   
c)  $da/dN = 3 \mu\text{m}/\text{cycle}$   
(Döker [51])

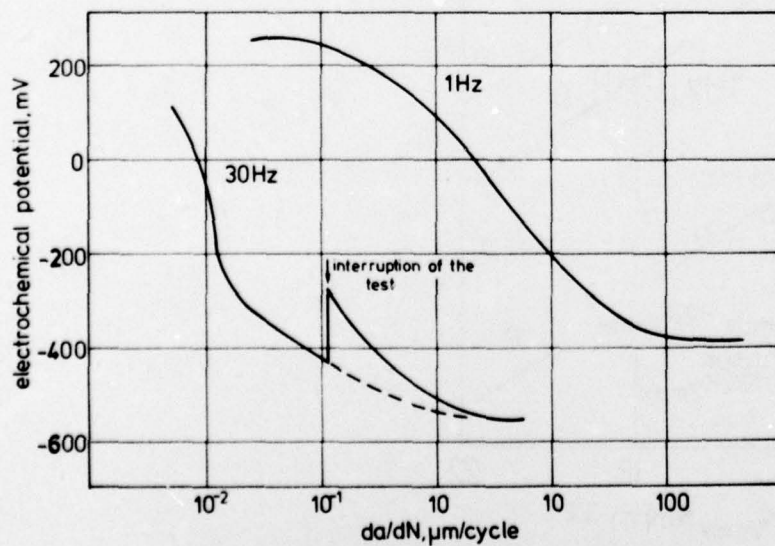


Fig.10

Electrochemical potential,  
measured against Ag/AgCl-  
electrode for Ti-6Al-4V as  
a function of crack growth  
rate (Döker [51])

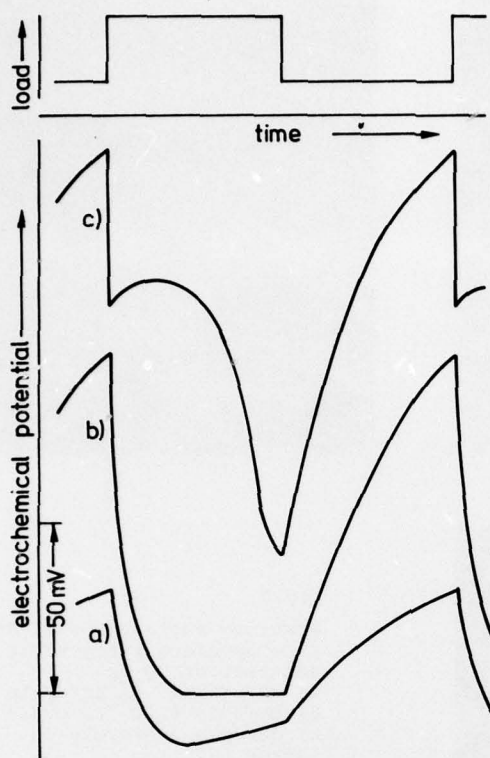


Fig. 11

Change of electrochemical potential during one cycle, Ti-8Al-1Mo-1V,  $f = 0.1$  Hz, 3,5 % NaCl solution (Döker [51])

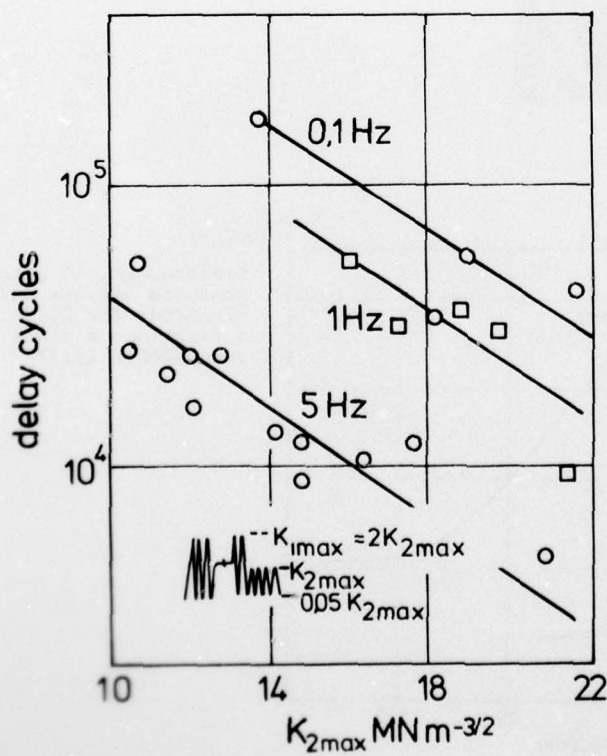


Fig. 12

Number of delay cycles after 500 high cycles at  $K_{1max} = 2K_{2max}$  versus  $K_{2max}$ , Ti-6Al-4V in 3,5 % NaCl solution (Shih, Wei [44])

## CORROSION FATIGUE OF ALUMINUM ALLOYS

by

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The environment plays an important role in the fatigue of metals, particularly in the fatigue of high-strength aluminum alloys used in the aircraft industry. Inert gases and vacuum are known to increase fatigue strength, while humidity and aggressive media reduce the endurance limit and increase the crack propagation rate. Since the knowledge about corrosion fatigue seems essential to evaluation of the service behavior of materials, the usefulness of such knowledge is unquestionable.

In this respect, it should be kept in mind that a specific environment might have no effect on a given alloy in the absence of cyclic stresses and that, generally, corrosion fatigue of aluminum alloys takes place at lower stresses than those required for stress corrosion.

Therefore, we are not safe from corrosion fatigue even under conditions precluding stress corrosion. Furthermore, it should be remembered that cyclic stresses are much more frequently encountered during service than monotonic stresses, particularly in aircraft structures, so that the understanding of corrosion fatigue behavior seems to provide a better tool for evaluation of service behavior of materials than does reference to stress corrosion.

In practice, a large part of the aircraft's structure is exposed to humid environment owing to vapor condensation, rain or even sea-water splashes in the case of aircraft stationed at airports located close to the seashore. This also applies to helicopter blades which are particularly exposed to the weather.

In the following paragraphs we concentrated on the main results obtained for the most widely used aluminum alloys. Our efforts were mainly devoted to the study of crack propagation on which environmental influence is inevitable, while many measures may be adopted to avoid crack initiation by protecting the surface. In the first part we shall discuss the mechanical behavior of aluminum alloys, while the second part is devoted to the study of mechanisms.

2. CORROSION-FATIGUE CRACKING IN ALUMINUM ALLOYS

Several studies (1,2,3) have already shown that among aluminum alloys, those belonging to series 7000 are particularly susceptible to the environment, so that most of the works indeed dealt with these alloys. The various alloys investigated for corrosion-fatigue cracking are shown below. Tables I and II list their chemical composition and mechanical properties.

7175	T 7351
7175	T 651
2618 A	T 851
2024	T 351

2.1 Experimental conditions

The reference environment was dry Argon containing less than 5 ppm of water. Taking into account the strong influence of vapor on crack propagation, we have carried out tests in air 95 % saturated with water vapor, while laboratory humidity was only 50 %. Other tests, supplementing the first series, were carried out in salt water containing 3 g salt per liter and having a pH-value of 8. In order to avoid any galvanic coupling effect, the aggressive medium was contained in a microcell composed of two shells cemented to the specimen, while the medium was continuously renewed by forced circulation.

Besides the aggressiveness of the medium, a detailed study has been made of the state of stress and strain, which influences the start of the embrittlement process that seems to be the source of corrosion fatigue.

Under these conditions we have tested 1 to 40 mm-thick specimens in which cracking is produced either by plane stress or by plane strain. Since aluminum alloy cracking is very sensitive to the mean stress, we correlate the tests carried out with different R ratios used.

Several authors (3,4) indicated that, besides the aggressiveness of the test medium itself and the triaxial state of stress, the crack-tip opening rate depends also on the strain rate.

All our tests have been carried out on solid specimens with  $W = 75$  mm and on specimens of a size of  $600 \times 200$  mm with a central notch arranged in such a way that the crack will propagate in the direction of rolling

## 2.2 Results and discussion

For the aluminum alloys investigated, the environmental effect depends on the crack-propagation rate range. Before going into details, it should be noted that in the case of alloys 7175 and 2618, not very sensitive to stress corrosion, the environmental effect is generally marked at rates between  $10^{-4}$  and  $5 \cdot 10^{-3}$  mm/cycle. At higher rates, the environmental effect tends to disappear. As against this, alloy 2024, susceptible to stress corrosion especially in the transverse direction, is sensitive to corrosion fatigue for rates higher than  $10^{-3}$  mm/cycle (Figures 1 to 3).

Furthermore, it was found that crack propagation was faster in salt water than in vapor-saturated air and that, in the worst case, the crack propagation rate is twenty times higher than the reference rate obtained in dry argon.

When the ratio  $R = \frac{K_{min}}{K_{max}}$  is close to zero, it was found that, in the strain-rate range between  $10^{-4}$  and  $10^{-2}$  mm/cycle, the influence of the specimen's thickness between 1 and 40 mm does not exceed a factor of four.

It is worth noting that, whatever the alloy, the influence of thickness on the crack propagation rate is almost nul for a rate of  $10^{-4}$  mm/cycle in dry argon or humid air. On the other hand, in the salt water solution, the influence of specimen thickness is observed for all rates investigated.

In other words, the effect of thickness on crack propagation rate in corrosion fatigue is meaningful only for very aggressive media. However, on a practical level, it is interesting to note that corrosion-fatigue cracks may develop also in sheets as thin as 1 mm.

Aluminum alloys are very sensitive to mean loads given by the ratio  $R$ . While in the ambient air this is confirmed by many results, the phenomenon is almost nonexistent in controlled media (5).

In order to verify the simultaneous effects of the environment and the ratio  $R$ , we have carried out tests on 1.6 and 10 mm-thick specimens, with the results given in Figures 8 to 11.

Whatever the thickness, it was found that the influence of environment on the crack propagation rate diminishes for a ratio of  $R = 0,5$  as compared to an  $R$  ratio close to zero. This is more noticeable in the 2618, than in the 7175 alloy. Consequently, these observations show that the influence of the  $R$  ratio is maximal in non-aggressive environments and that the corrosion fatigue of aluminum alloys becomes more important when the mean load is low.

It should be stressed that for a given  $\Delta K$ , the increase in propagation rate caused by the combined effects of an aggressive environment, a high ratio  $R$  and a large thickness is not the result of a linear accumulation obtained by the addition of the effects of each parameter.

Finally, we should note that the stress frequency plays an important role in the case of aluminum alloys, as recognized by various authors (3,4 and 6). We found that the influence of frequency is the more pronounced the more sensitive the material to corrosion fatigue and the more aggressive the medium. At a frequency of 25 Hz, crack-propagation rates differ by a factor of no more than three, according to the environment. On the other hand, low stress frequencies increase the propagation rate spread in such a way that at a frequency of about 1 Hz, their influence is maximal. With even lower frequencies : 0,1 or 0,01 Hz, no noticeable effect seems to be detectable, except for high rates of over  $5 \cdot 10^{-3}$  mm/cycle or else stress corrosion phenomena may appear. Taking into account previous observations (3,4), it would appear that the corrosion fatigue effect continues to increase when frequency decreases below 1 Hz. But for very low frequencies of less than 0,1 Hz, the development of corrosion fatigue seems to be stationary, and the range seems to be the one corresponding to stress corrosion.

## 2.3 Comparison of the corrosion-fatigue behavior of various aluminum alloys.

Aluminum alloys containing copper and magnesium are less sensitive to corrosion-fatigue cracking than those containing copper and zinc (Figures 1 to 3).

The fact that sensitivity of aluminum alloys to corrosion fatigue does not particularly depend on their toughness, is remarkable. Alloys such as 2024 and 2618, being of very different toughness, crack in salt water at the same rate in the range of  $10^{-4}$  and  $10^{-2}$  mm/cycle. As against this, alloy 7175 T 7351, with higher toughness than 2618 T 651 cracks faster, due to corrosion fatigue, than does the latter.

In the case of alloy 7175, the T 7351 treatment increases resistance to corrosion fatigue within the range of  $10^{-3}$  and  $10^{-2}$  mm/cycle. Furthermore, there is no noticeable difference between T 651 and T 7351 treatments. Thus, we can see that a treatment improving the stress corrosion resistance of aluminum alloys has not necessarily the same effect on corrosion fatigue.

## 3. MECHANISMS OF CORROSION FATIGUE

Corrosion-fatigue cracking involves its own mechanisms which are different from those of stress corrosion. In aluminum alloys, fatigue cracks propagate by formation of striations, quasi-cleavages and pits and sometimes, but less frequently, along grain boundaries. The environment affects each of these mechanisms and not only the formation of striations which is typical for cyclic stressing.

Several reviews (6,7) show that the increase in the crack propagation rate is due to an embrittlement process during which hydrogen plays the principal role, even in the case of aluminum alloys. This embrittlement occurs mainly in the form of fragile striations (3,8) in the alloys of series 7000. Such brittle striations do not occur in the alloys of series 2000.

Other studies would be necessary to establish the operating mechanisms.

A more mechanical approach, comprising a comprehensive study of the crack-tip opening displacement allows us today to give additional data.

For alloy 7175 T 651 we have obtained displacement measurements for cracks in air and salt water which show how the medium influences the process of crack-tip opening displacement.

We have measured the opening displacement of a V-notch using a double cantilever beam gage placed on the tip of the notch. Load vs. crack-displacement diagrams were analyzed using the differential method proposed by ELBER, following point-by-point the elastic component of the total displacement (10).

In general, three stages can be distinguished in these diagrams. With increasing load, the first stage corresponds to a change in the configuration of the crack tip, the second to the elastic behavior with constant configuration and the third stage to the plastic behavior.

According to ELBER, the crack tip commences to extend at the beginning of the second stage. This transition permits determination of  $K_0$ , the difference between  $K_{max}$  and  $K_0$  being equal to  $K$  effective. In this study,  $K_{Ieff}$  is the stress-intensity factor corresponding to the end of the second stage.

First, we can observe that, all other parameters being the same, V-opening displacement is somewhat larger in salt water than in the air. However, the difference is small enough to conclude that the change in propagation rate cannot be explained by the difference between the displacements (Fig.12).

Furthermore, it can be noted that the force which initiates the crack-tip opening displacement, in other words,  $K_0$ , is larger in salt water than it is in the air (Fig.13). But even then, the difference is so small that one can conclude that, in all probability,  $K_0$  is almost the same in both media investigated. This observation agrees with that of SCHIJVE.

Moreover, it seems that the elastic behavior with constant configuration corresponding to  $K_1 - K_0$  is exactly the same in both the air and the salt water.

From these remarks it appears that the environment does not affect the stress-intensity factor  $\Delta K_{eff}$ , since when  $K_{max}$  is given,  $K_0$  remains the same, whatever the medium (Fig.14). Thus, the only point remaining is that displacement in salt water is slightly higher than in the air.

It is interesting to point out that, in spite of these similarities, there are important differences in the opening mode in regard to the medium.

In salt water, the displacement curve is different for the increasing part of the cycle and its decreasing part. With a given loading and crack length, the displacement at the closure is the same at all points, both in the air and in salt water. As against this, displacement at the crack tip is different, in salt water, in two ranges (Fig. 15 and 16).

At the beginning, the change in the configuration of the crack in salt water seems a little longer than in the air. At the end, the plastic behavior is such that everything takes place as if consolidation were higher in the air than in salt water. In this range, with the same force, relative displacement is larger in the air than in salt water. These observations still remain to be confirmed.

Since, for all practical purposes, crack propagation in salt water does not depend on  $\Delta K_{eff}$  and the elastic behavior of the crack has no influence, it seems that the environment effect is decisive in the crack configuration change and the last extension stage. The embrittling effect of the aggressive medium, therefore, has no noticeable effect on the elastic extension and on  $\Delta K_{eff}$ .

#### 4. CONCLUSIONS

1. The aluminum alloys we have studied are sensitive to corrosion fatigue within the rate range  $10^{-4}$  and  $10^{-2}$  mm/cycle, even then the alloy is not sensitive to stress corrosion.
2. It was verified that alloys belonging to the 7000 series are particularly vulnerable, even in the T-7351 condition and in thicknesses as low as 1.6 mm.
3. It was found that certain low-toughness alloys such as 2618 show good resistance to corrosion fatigue, and consequently it does not seem that either high toughness or good stress-corrosion resistance, ensure good corrosion-fatigue resistance.
4. The effect of the environment increases when stress frequency decreases, and when ratio R and thickness increase. In practice, the influence of each one of these parameters can not be linearly added.
5. The mechanism of corrosion fatigue is largely connected to embrittlement of the crack tip.
6. Measurements of the crack-tip opening displacement show that environment has practically no effect on  $\Delta K_{eff}$  and a very slight one on COD.
7. The elastic displacement of the crack is not modified by the environment. On the other hand, the displacement stages preceding and following the elastic displacement do depend on the environment.

Table 1: Chemical composition of aluminum alloys

ALLOYS	Si	Fe	Cu	Mn	Mg	Cr	Ni	Zn	Ti
7175 T 651	0,08	0,20	1,40	0,04	2,56	0,18	-	5,60	0,03
7175 T 7351	0,08	0,20	1,61	0,04	2,52	0,20	-	5,90	0,07
2618 A T 651	0,22	1,11	2,62	0,08	1,52	0,01	1,23	0,06	0,10
2024 T 351	0,11	0,25	4,15	0,68	1,48	0,06	-	0,05	0,05

Table 2: MECHANICAL PROPERTIES OF ALUMINUM ALLOYS

ALLOYS	$\bar{\sigma}_y$ (MPa) T.L	$\bar{\sigma}_u$ (MPa) T.L	e%	KIc (MPa $\sqrt{m}$ )T.L	KIsc (MPa $\sqrt{m}$ )S.L
7175 T 651	483	555	10	24	7,5
7175 T 7351	428	501	8	26	21
2618 A T 651	417	460	6	22	17
2024 T 351	321,5	477,5	16,8	32	7,5



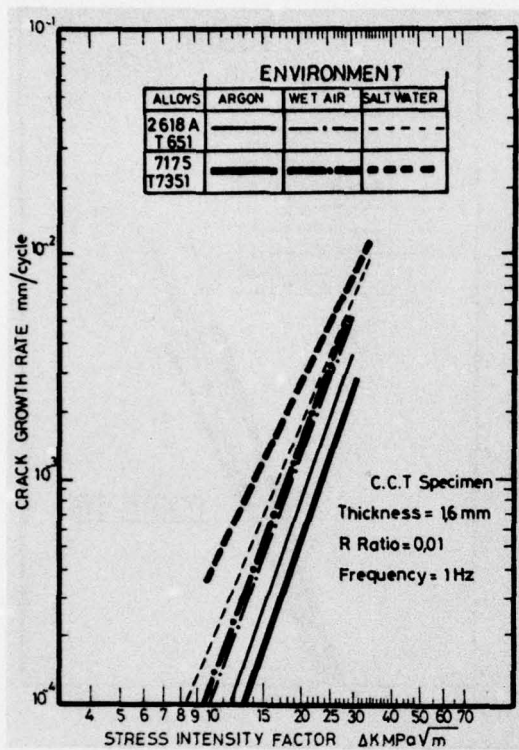


Fig-1 CORROSION FATIGUE CRACK GROWTH RATE FOR SEVERALS ALUMINUM ALLOYS

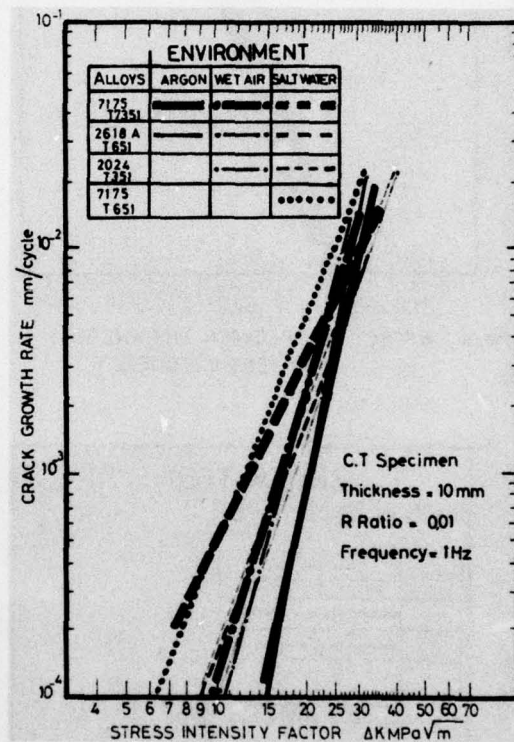


Fig-2 CORROSION FATIGUE CRACK GROWTH RATE FOR SEVERALS ALUMINUM ALLOYS

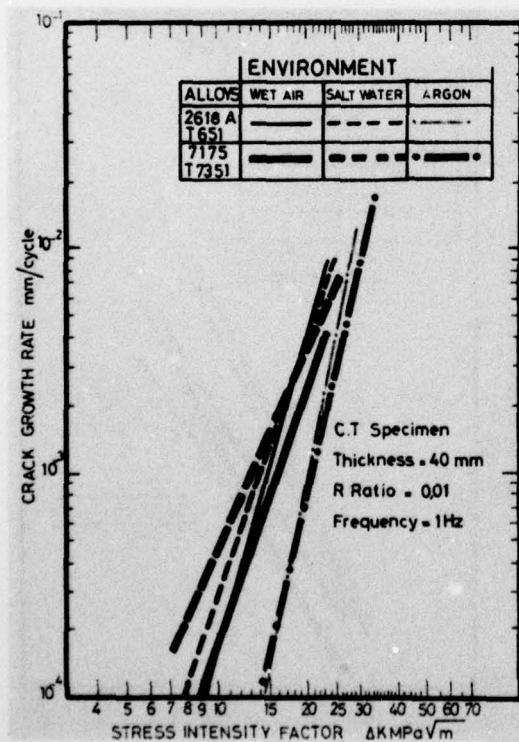
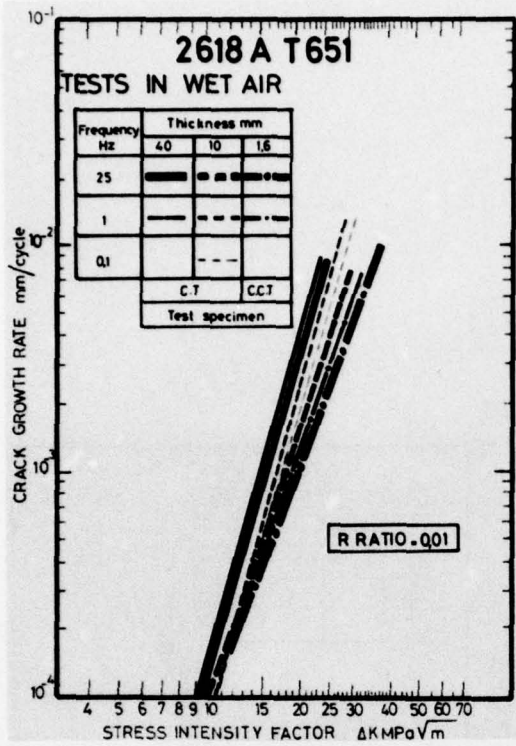
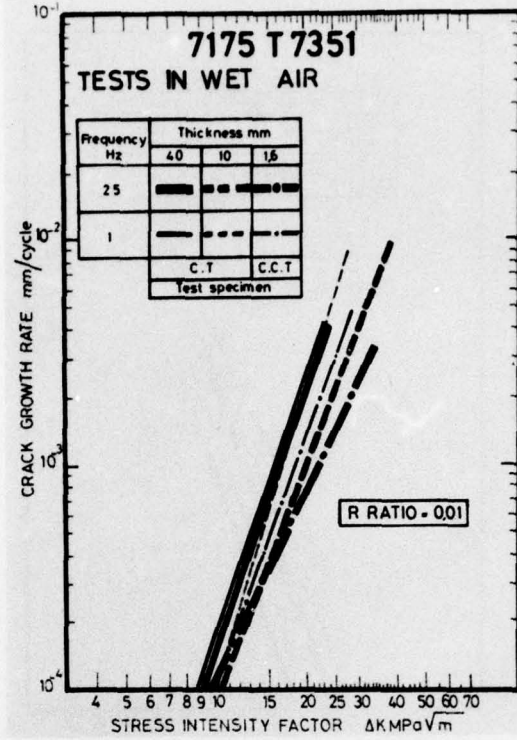


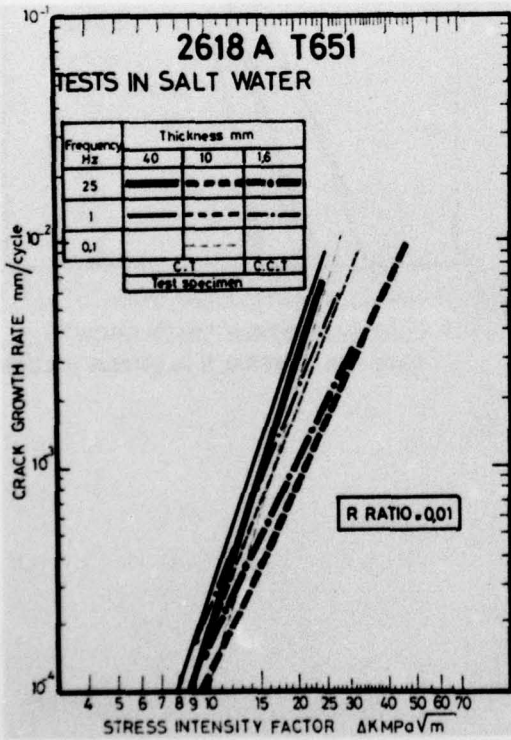
Fig-3 CORROSION FATIGUE CRACK GROWTH RATE FOR SEVERALS ALUMINUM ALLOYS



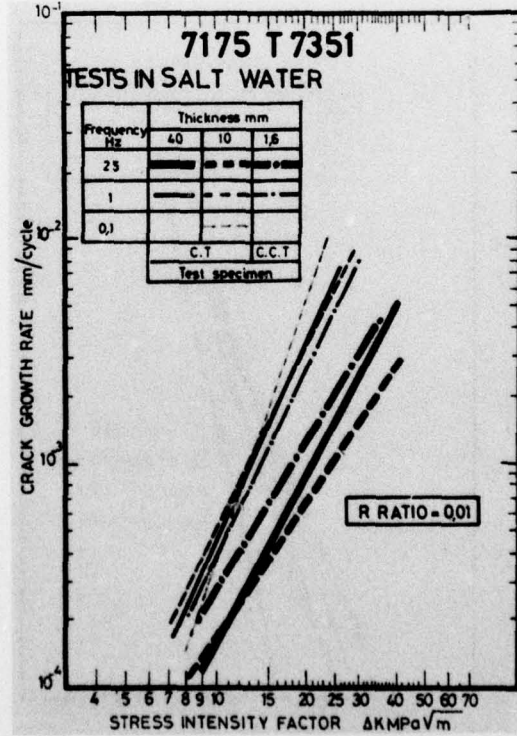
Eig-4 EFFECT OF SPECIMEN THICKNESS AND TEST FREQUENCY



Eig-5 EFFECT OF SPECIMEN THICKNESS AND TEST FREQUENCY



Eig-6 EFFECT OF SPECIMEN THICKNESS AND TEST FREQUENCY



Eig-7 EFFECT OF SPECIMEN THICKNESS AND TEST FREQUENCY

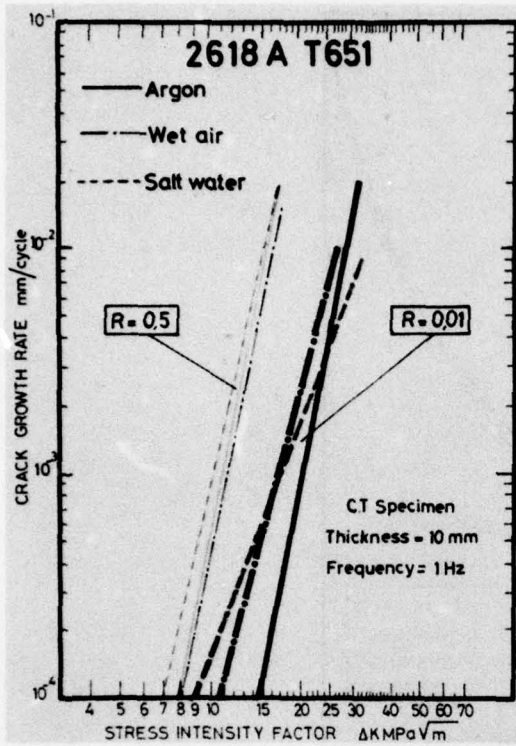


Fig- 8 EFFECT OF R RATIO AND ENVIRONMENT

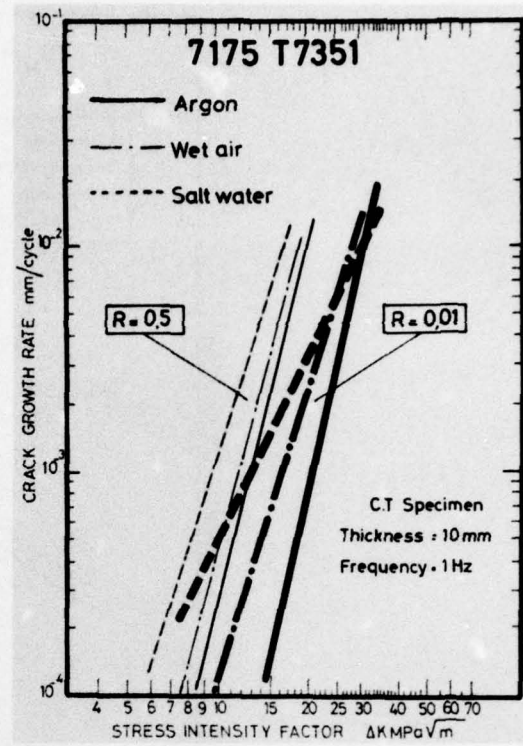


Fig- 9 EFFECT OF R RATIO AND ENVIRONMENT

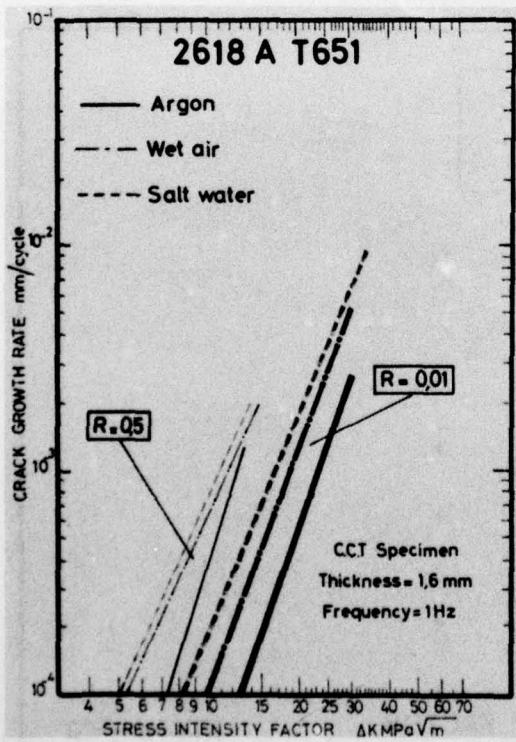


Fig- 10 EFFECT OF R RATIO AND ENVIRONMENT

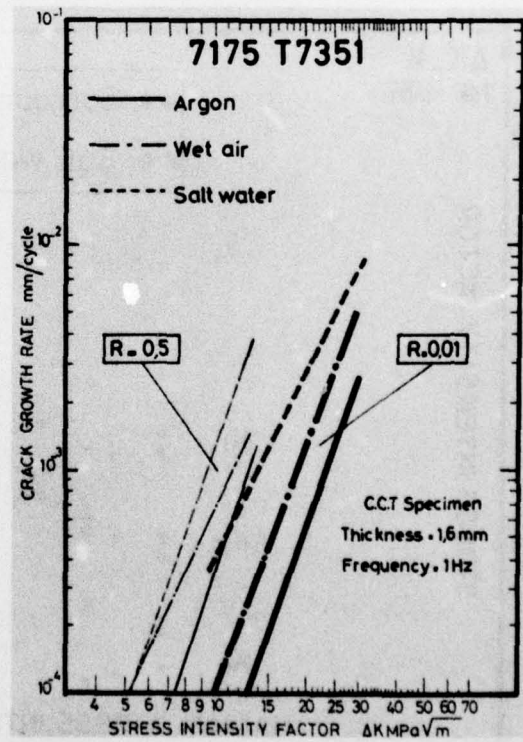


Fig- 11 EFFECT OF R RATIO AND ENVIRONMENT

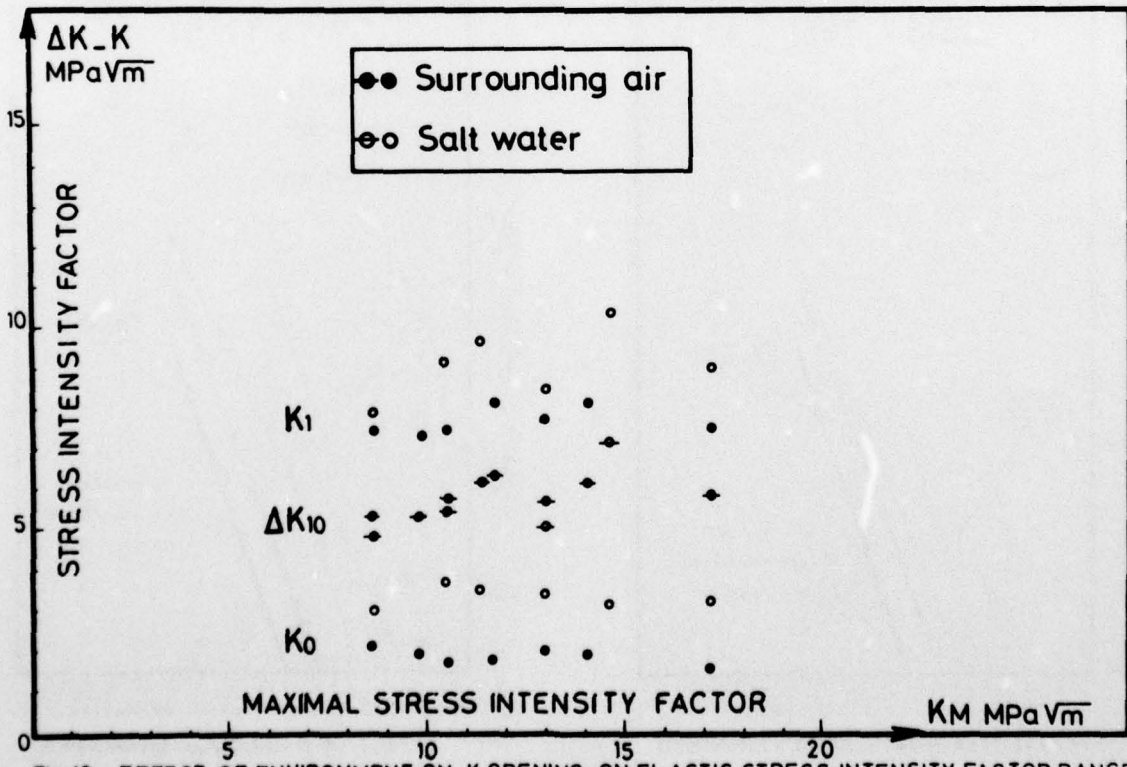
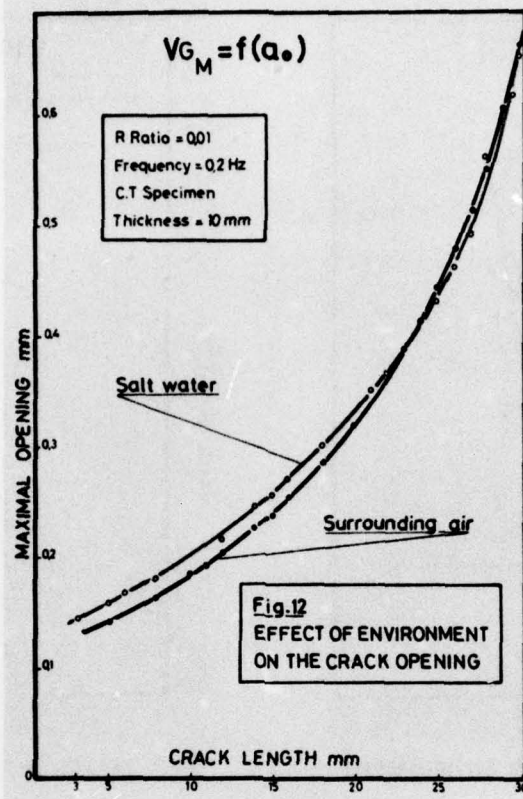


Fig.13 EFFECT OF ENVIRONMENT ON K OPENING, ON ELASTIC STRESS INTENSITY FACTOR RANGE FOR SEVERALS K MAX

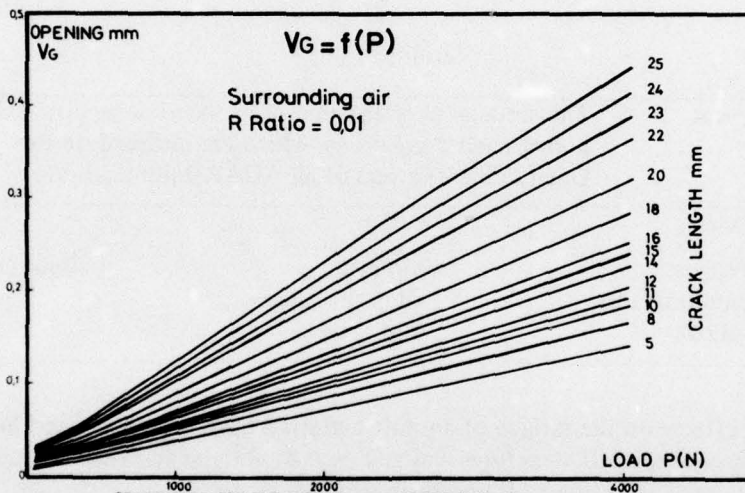
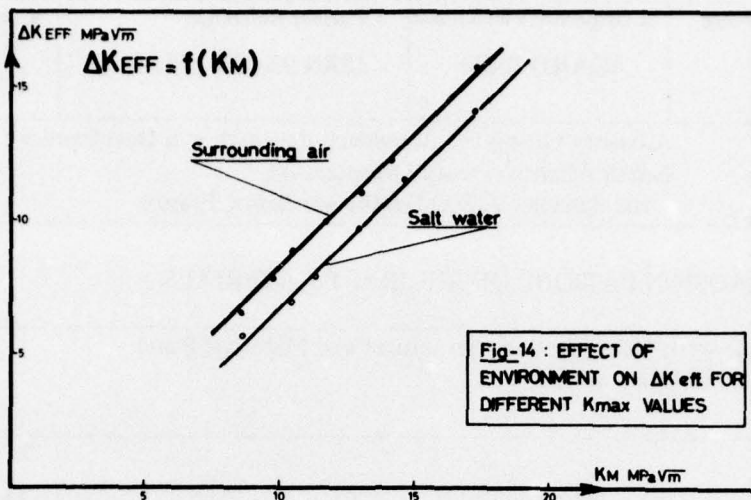


Fig. 15 CRACK EDGE OPENING VERSUS LOADING FOR 7175 T 651 SPECIMEN IN SURROUNDING AIR

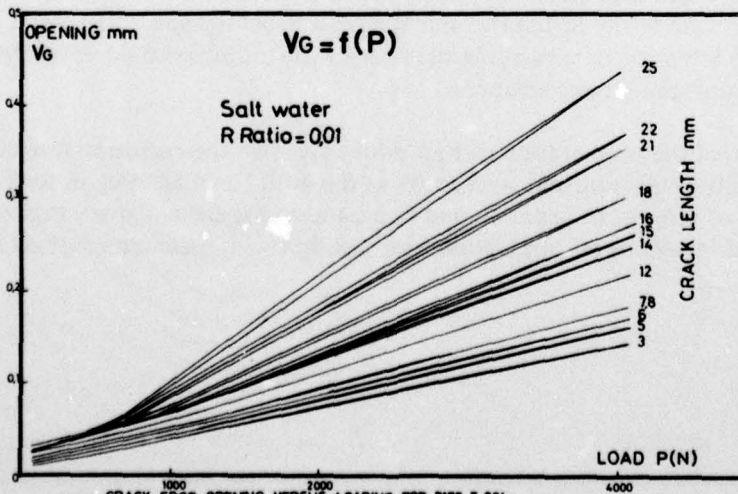


Fig. 16 CRACK EDGE OPENING VERSUS LOADING FOR 7175 T 651 SPECIMEN IN SALT WATER

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