

## The Influence of Gaseous Environment and Fatigue Frequency on the Growth of Fatigue Cracks in Some Aluminum Alloys\*

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### ABSTRACT

Measurements of crack growth at fatigue frequencies of 100, 1 and  $\frac{1}{60}$  Hz in some gaseous environments show that changes of crack rate with frequency can be partly or entirely due to environment, depending on the alloy. The effects of oxygen, water vapor alone and water vapor in the presence of oxygen and nitrogen have been studied. The critical gas pressures which are necessary to affect crack growth depend on frequency and crack rate. Mechanisms for the effects observed are discussed and results are interpreted in terms of gas adsorption at the crack tip and the immediate action of adsorbed surface layers on the crack while it is growing.

### Introduction

The variation of fatigue crack rates of aluminum alloys in different environments is now well established [1]–[5]. Crack rates in air can be up to 60 times faster than those *in vacuo* [2] and both water vapor and oxygen can lead to increases in rate. Water vapor has the dominant influence and theories for this effect have been discussed in the references given. The variation of crack rate with the frequency of the applied alternating stress is also well known [3], [5]. In dry air, rates at 0.4 Hz can be up to ten times faster than those at 57 Hz [3].

If the environmental effect is due to the reaction between the surrounding gas and the freshly exposed metal at the crack tip, then the time available for the reaction is important. Thus part of an observed variation in crack rate with fatigue frequency may be due to the environment. Hordon and Wright [5] have considered this effect in their study of aluminum fatigue lives in different residual gas pressures.

We consider here some further measurements of the effects that the constituents of ordinary air may have at different fatigue frequencies on crack rates in aluminum alloys. The results are compared with those of other workers. Theories for the mechanisms operating are discussed and some practical aspects are considered.

### Experimental

Techniques were similar to those previously described [2]. Standard or modified Haigh fatigue machines were used to give gross stressing of  $30.3 \pm 24.8 \text{ M Nm}^{-2}$  ( $4400 \pm 3600 \text{ lb f in}^{-2}$ ) in sheet specimens 57 mm long, 31.7 mm wide, 1.6 mm thick. With this thickness, cracks of up to ~18 mm total length could be grown with fractures face normal to the tensile axis. Above this length the plastic zone at the crack tip was large enough to allow the fracture faces to rotate or break up into planes at  $45^\circ$  to the normal plane. Fatigue cracks were started at both ends of a central transverse slot and were recorded either optically by microscope or automatically by camera. Test frequencies of 100, 1 and  $\frac{1}{60}$  Hz were used. At 100 Hz the load wave form was sinusoidal. At 1 and  $\frac{1}{60}$  Hz the load (provided by a switched d.c. supply) rose exponentially to its maximum and fell similarly to its minimum level; the time constant in both cases and at both frequencies was 0.1 sec. Because of the time involved, testing at  $\frac{1}{60}$  Hz was restricted to rates above  $10^{-4}$  mm/cycle and to a few important cases only.

Fatigue loads were transmitted via bellows to the specimen in an environmental chamber.

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This chamber could be pumped by an oil diffusion pump with a liquid nitrogen trap to give pressures of  $4\mu \text{ Nm}^{-2}$  ( $3 \times 10^{-8}$  torr). Alternatively, water vapor environments could be derived from air free distilled water in a thermostatted side arm. Air and other gases were passed into the chamber through thermostatted water or ice traps to provide selected humidities. The resultant humidities of the gases were measured after they had left the chamber by means of a Casella dew point hygrometer and an Elliott electrolytic moisture monitor. The inaccuracies of both these instruments limited humidity measurements to water vapor pressures  $> 0.3 \text{ Nm}^{-2}$  ( $2 \times 10^{-3}$  torr). The water vapor partial pressure in all the laboratory air tests was monitored and was found to lie between limits of 6.7 and 1.5 k  $\text{Nm}^{-2}$  (5 and 11 torr).

Alloys used were:

- (1) Alclad DTD 5070A: (2.5% Cu, 1.5% Mg, 1.2% Ni, 1.0% Fe, 0.25% Si) 0.2% P.S. =  $400 \text{ M Nm}^{-2}$
- (2) A sheet version of DTD 683: (5.8% Zn, 2.7% Mg, 1.3% Cu, 0.4% Mn, 0.1% Si), cf 7075-T6. 0.2% P.S. =  $530 \text{ M Nm}^{-2}$ .

Some further limited work was done on an S.A.P. having a nominal 10.5% (Wt) of  $\text{Al}_2\text{O}_3$ , (0.2% P.S. =  $270 \text{ M Nm}^{-2}$ ).

Crack rates are plotted against the stress intensity range,  $\Delta K$ , where  $\Delta K = \Delta\sigma(\pi a)^{\frac{1}{2}}\alpha$ .  $\Delta\sigma$  is the alternating stress range ( $49.6 \text{ M Nm}^{-2}$ ;  $1 \text{ M Nm}^{-2} = 910 \text{ lb f in}^{-\frac{3}{2}}$ )  $a$  is the half crack length and  $\alpha$  is Koiter's finite width correction for central cracks [7] (Koiter's correction can be applied to a greater range of crack lengths than the A.S.T.M. correction; the latter is restricted to cracks up to 70% of the sheet width. Up to these lengths the two corrections differ by less than 1% [8]). Specimen reproducibility in terms of stress intensity for a given crack rate was within  $\sim 4\%$  in the 5070A and  $\sim 6\%$  in the 683 alloy.

#### Effect of Fatigue Frequency in Vacuo and in Air

Figure 1(a) shows the rate of cracking per cycle of the 5070A alloy at the different frequencies in laboratory air and *in vacuo*. The differences between the rates *in vacuo* at 100 Hz and 1 Hz are small enough to indicate that time dependent processes such as creep or diffusion are not lead-

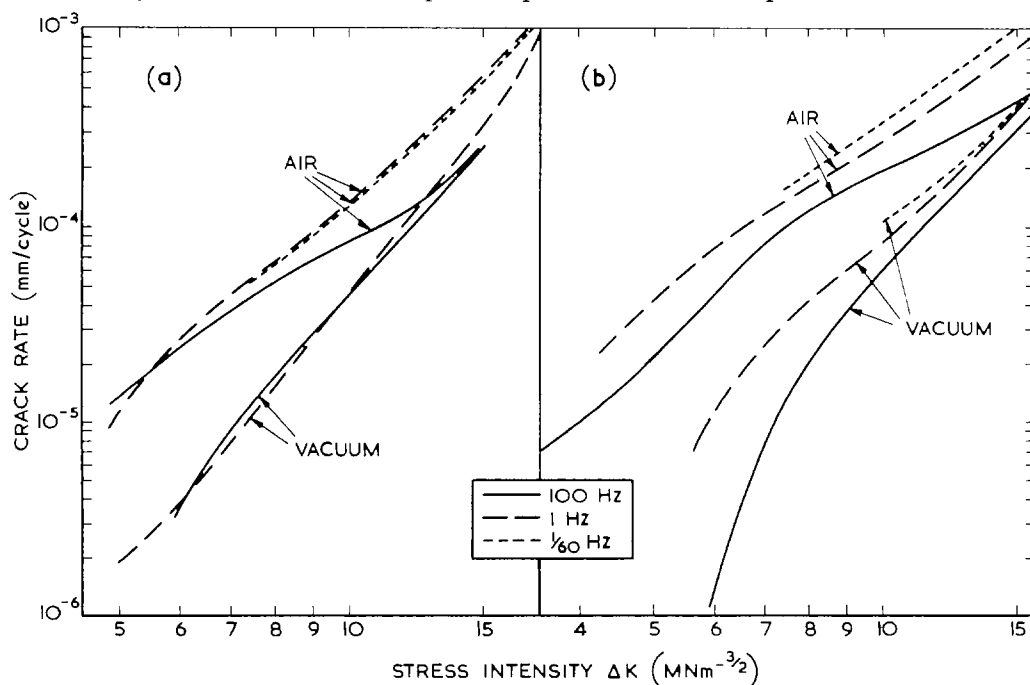


Figure 1. Crack rates in laboratory air and *in vacuo* at different fatigue frequencies; (a) 5070A alloy, (b) 683 alloy.

ing to significant changes of crack rate with frequency, *i.e.* in an inert environment there is no intrinsic frequency effect. The increase in the air rate at 1 Hz over that at 100 Hz is thus likely to be due to the air environment. At  $\frac{1}{60}$  Hz the rate in air is negligibly different from that at 1 Hz. This suggests that *in vacuo* the  $\frac{1}{60}$  Hz rate is not likely to differ greatly from that at 1 Hz and that at 1 Hz the environmental effect may have reached a maximum.

In Fig. 1(b) corresponding curves are given for the 683 alloy. The rates *in vacuo* show that this alloy has an intrinsic frequency effect. The air curves show an environmental frequency effect similar to that of the 5070A alloy with a marginal further increase in rate at  $\frac{1}{60}$  Hz.

Crack rates in S.A.P. were also determined in air and *vacuo* at 1 and 100 Hz. The rates at 1 Hz were not significantly different from those at 100 Hz in either environment [2]. There was a small environmental effect and a negligible intrinsic frequency effect.

The immediate conclusion is that frequency effects *in vacuo* (*i.e.* intrinsic) depend on the alloy and that in air frequency effects can sometimes be caused solely by the environment.

### Effect of Air and Its Constituents

Accelerated cracking in ordinary air is due to the combined effects of water vapor, oxygen and nitrogen. Their separate effects on the two alloys are shown in Figs. 2(a) and 2(b) at a

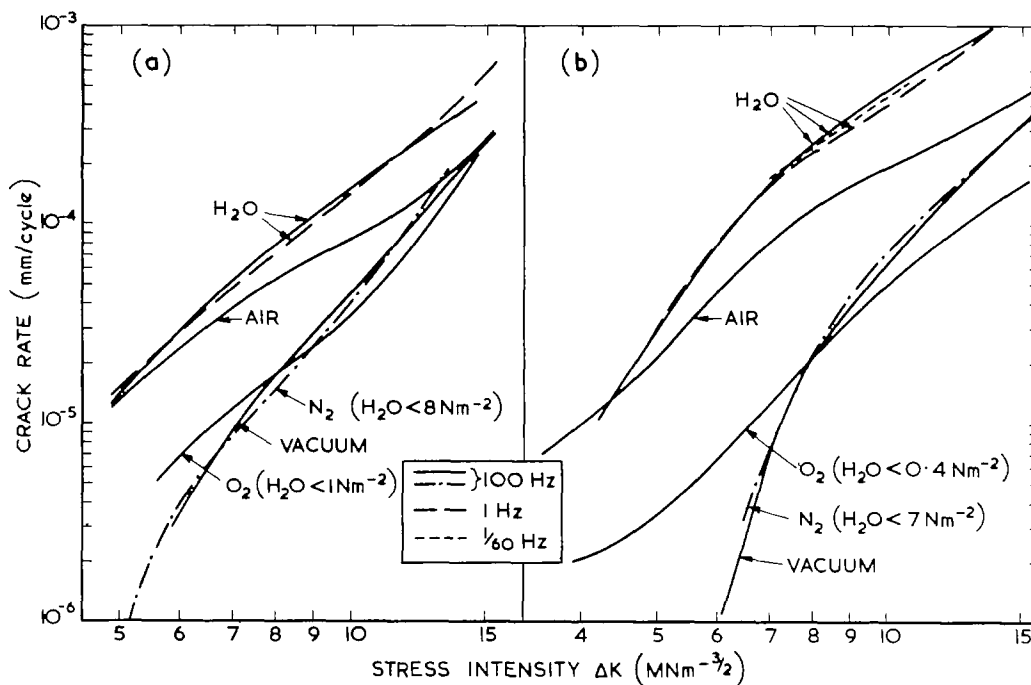


Figure 2. Crack rates in laboratory air, water vapor ( $2 \text{ k Nm}^{-2}$ , 15 torr), oxygen (1 atmos.) nitrogen (1 atmos.) and vacuum at 100 Hz. (a) 5070A alloy, (b) 683 alloy with rates in water vapor at 1 Hz and  $\frac{1}{60}$  Hz.

fatigue frequency of 100 Hz with rates *in vacuo* for comparison. Dry nitrogen gave crack rates negligibly different from those *in vacuo* and can thus be regarded as inert. Oxygen causes a rate increase at low rates and a small decrease at high rates. Hartman [1] on a 2024-T3 aluminum alloy has found similar oxygen effects (taking his dry argon environment as equivalent to our *vacuum*). Wei [4] has found no effect of oxygen on a 7075-T651 alloy. The beneficial effect of oxygen was previously dismissed by Hartman [1] and ourselves [2] as insignificant but our present work confirms its existence; it is considered later. Over the whole range of measurement water vapor alone causes a large rate increase. Air except at the lowest crack rates causes a smaller rate increase than water vapor. With S.A.P. the effects of the same gases were very similar except that the changes in crack rate were smaller (not greater than  $3 \times$ ).

The magnitude of the changes in crack rate appeared to increase with increasing hardness of the alloy. The Vickers hardness numbers were S.A.P. – 106 : 5070A – 136 : 683 – 187. Further measurements were made on a special Al-7% Zn-5% Mg-1% Mn alloy. When aged to a peak hardness of HV220 this alloy showed changes of crack rate in the different environments similar to those of the 683 alloy; when overaged to HV of 100 its behavior was similar to that of S.A.P. Thus, except that the scale of the effects differ, most of the data indicates that different aluminum alloys show similar susceptibilities to the constituents of air.

### Effect of Water Vapor

Crack rates have been determined in humid and dry inert gases [1], but a study of rates in water vapor alone at different frequencies is of interest. The 5070A alloy was used for this as

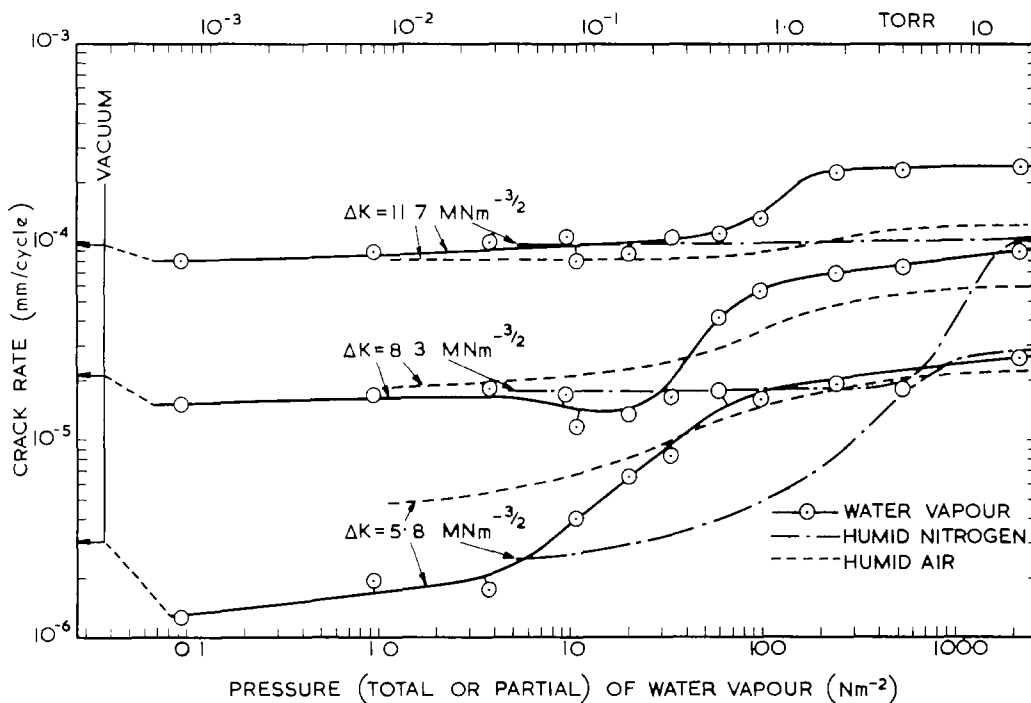


Figure 3. Variation of crack rate of 5070A alloy with water vapor pressure at 100 Hz. In water vapor alone; in humid nitrogen and humid air at one atmosphere.

it had no intrinsic frequency effect and yet showed significant changes in crack rate when in water vapor and *vacuo*.

Some of the basic 100 Hz curves of crack rate in water vapor at different pressures *vs.* stress intensity have been shown in Ref. [2] and will not be repeated here. They were similar in shape to the air curve in Fig. 2(a). More recent measurements have been made at low pressures and it has been found that as the pressure was reduced below  $13 \text{ Nm}^{-2}$  (0.1 torr) the changes of slope with increasing stress intensity became more pronounced. Unlike the behavior at higher pressures, crack rates could, for a while, stay virtually constant or even decrease, despite the increasing stress intensity. The curves were qualitatively similar to those, to be described later, which have been obtained in oxygen (Fig. 6a). This quasi constant transition rate is called the critical rate, with an associated critical pressure.

Figure 3 shows how the crack rate varied with pressure at a frequency of 100 Hz. It includes the results of the earlier and the more recent measurements. The rates (solid lines) are plotted here at three levels of stress intensity; they are virtually constant for vapor pressures  $> 130 \text{ Nm}^{-2}$  (1 torr). Below this pressure the rates decrease; at  $13 \text{ Nm}^{-2}$  (0.1 torr) the rates are vir-

tually the same for high stress intensities as *in vacuo*. The arrows for the corresponding vacuum rates show that the crack rates at low vapor pressures can be significantly less than those *in vacuo*.

Critical crack rates and critical pressures are related and this is shown in Fig. 7. At high water vapor pressures since the transition was much less pronounced the curve is shown dotted. The results of earlier more limited measurements on S.A.P. in water vapor [2] indicated similar critical rates and pressures.

At a fatigue frequency of 1 Hz crack rates at the same stress intensity levels have been determined and are shown in Fig. 4. At the highest water vapor pressures ( $\sim 1.8 \text{ k Nm}^{-2}$ ; 14 torr) the rates were little different from those at 100 Hz, but for a given crack rate critical pressures were one hundredth of those at 100 Hz. Though measurements were more restricted, there was again evidence of a variation of critical pressure with crack rate.

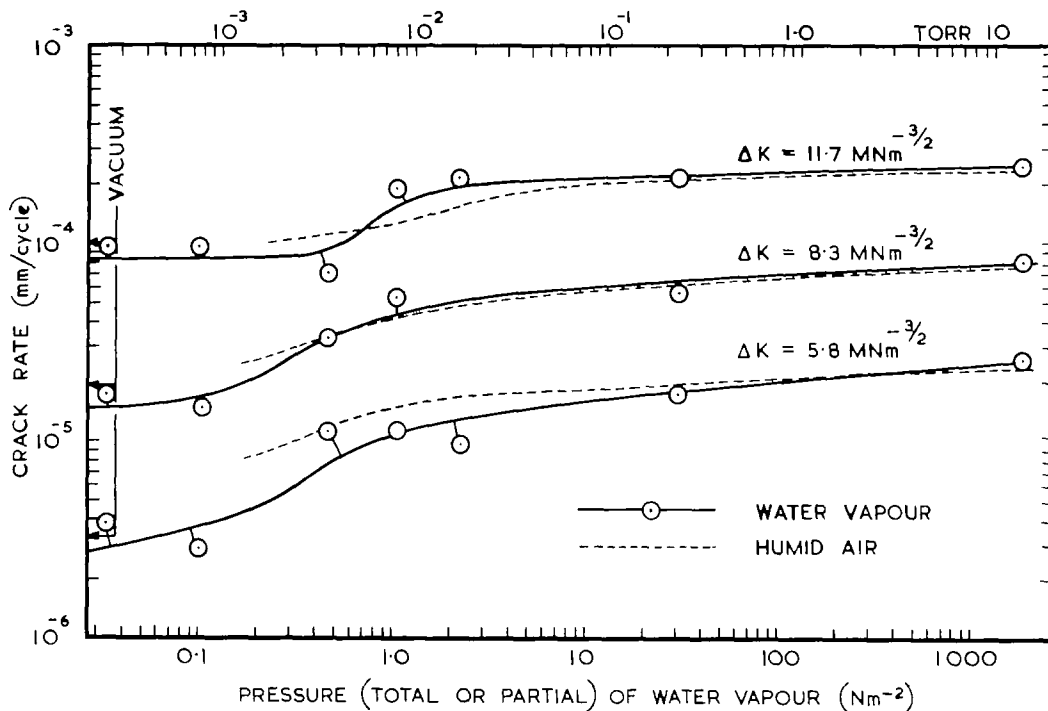


Figure 4. Variation of crack rate of 5070A alloy with water vapor pressure at 1 Hz. In water vapor alone and in humid air at one atmosphere.

Crack rates in the 683 alloy were measured at 1 and  $\frac{1}{60}$  Hz in water at room temperature saturated vapor pressure only. As seen in Fig. 2(b) they are not significantly different from those at 100 Hz. At this pressure the intrinsic frequency effect in this alloy appears to be masked by the greater effect of water vapor.

#### Effect of Water Vapor with Inert Gases

Nitrogen at one atmosphere, while having no effect by itself may modify the access of water vapor to the crack tip. The 100 Hz crack rate/stress intensity curves with different partial pressures of water vapor are shown in Fig. 5 and the variation of rate with partial pressure at the same three stress intensities as for water vapor alone is shown in Fig. 3. The critical pressures are about an order higher than those of water vapor alone for comparable effects, *i.e.* in some way the nitrogen baffles the action of the water vapor. At high crack rates the nitrogen led to more pronounced rate transitions (*cf.* Fig. 5 with Fig. 2 of Ref. [2], and at all rates the dependence of critical pressure on rate was more marked (Fig. 7).

With water vapor at  $1.55 \text{ k Nm}^{-2}$  (11.6 torr) the substitution of argon (carefully purified to remove traces of oxygen) for nitrogen made virtually no difference (Figs. 5 and 7) and measurements on the 683 alloy showed similar effects. However in hydrogen at one atmosphere also with  $1.47 \text{ k Nm}^{-2}$  (11.0 torr) water vapor, the rates were the same as in water vapor alone

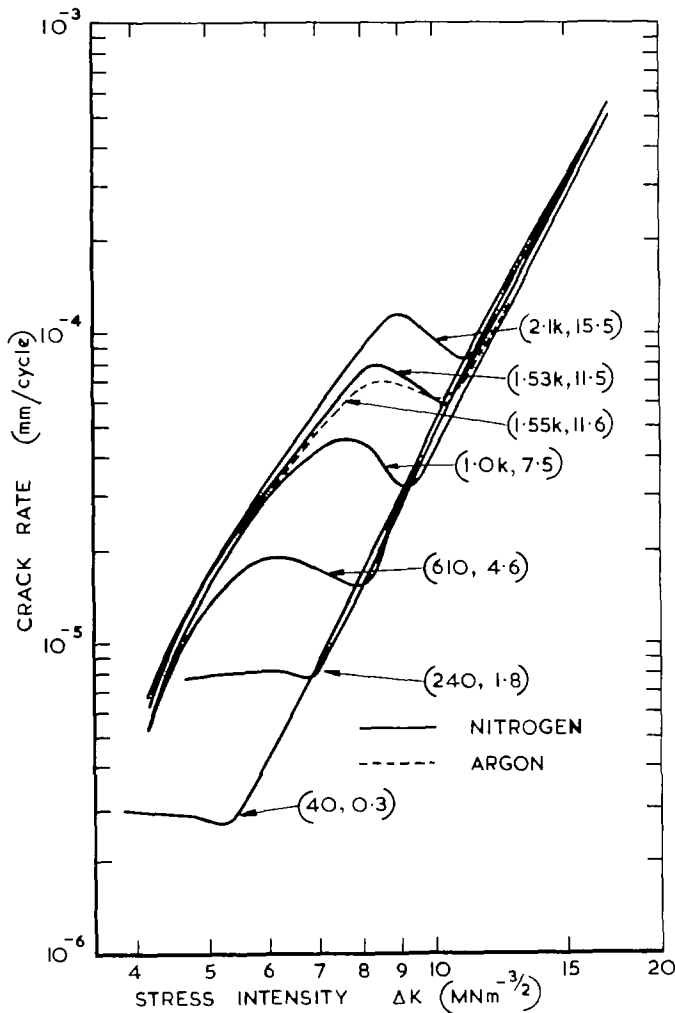


Figure 5. Crack rates of 5070A alloy at different pressures of water vapor ( $\text{Nm}^{-2}$ , torr) in one atmosphere of humid nitrogen, or argon, at 100 Hz.

at the same pressure. Similarly, rates in distilled liquid water were the same, as might be expected, since the aluminum/liquid water reaction generated sufficient hydrogen to lead to an environment at the crack tip consisting effectively of hydrogen (which on its own is inert [2]) reasonably saturated with water vapor.

### Effect of Humid Air

The dependence of crack rate on water vapor pressure in air is shown for the 5070A alloy at 100 Hz and 1 Hz in Figs. 3 and 4 (dashed lines). The presence of oxygen with the nitrogen further alters the effect of water. The crack rate transition between saturated and dry air is now less severe. Hartman's measurements in humid air on 2024-T3 alloy [1] and on 7075-T6 alloy [3] at 57 Hz show results very similar to those in Fig. 3 except that he found with the 2024-T3 alloy little difference between wet air and wet argon.

We may expect that, at small water vapor contents, the air crack rate at low rates will be faster than *in vacuo* because of the oxygen present. This is observed in the 5070A alloy at 100 Hz (Fig. 3); at 1 Hz (Fig. 4) there are indications that it might also have been seen if measurements at lower humidities had been possible. In the same alloy at 100 Hz with air at normal humidities, the air rates, at low rates, are similar to those in water vapor alone but at high rates are less (Fig. 2a). When however the fatigue frequency is 1 Hz (*cf.* Figs. 1a and 2a) the air rate is virtually the same as the rate in water vapor alone over the whole range (and probably also at  $\frac{1}{60}$  Hz). In slight contrast the 683 alloy, while showing similar increases in air crack rate with lowering fatigue frequency (Fig. 1b), at a test frequency of 1 Hz cracks faster in water vapor alone than in air at high rates and more slowly at low rates (*cf.* Figs. 1b and 2b).

### Effect of Oxygen

It is interesting to compare the effect of varying oxygen pressure with that of water vapor. Dry oxygen had its major effect in increasing crack rates at  $\sim 10^{-5}$  mm/cycle and was most obvious with the 683 alloy. This alloy was studied at a frequency of 100 Hz only; at lower frequencies the experimental times would have been inconveniently long. The partial pressure of residual water vapor was  $\approx 10^{-4}$  oxygen pressure.

Resulting crack rates are shown in Fig. 6(a) with the same data cross plotted in Fig. 6(b) at five stress intensity levels. Because of experimental scatter some of the small variations may not be significant. However the following conclusions are thought to be valid: (1) given a small enough initial value of  $\Delta K$ , as this increases with crack growth oxygen at all pressures initially leads to rates higher than those *in vacuo*, but at longer crack lengths and higher stress intensities the rates are less than *in vacuo*; (2) crack rates are not greatly altered as the pressure is reduced from one atmosphere to 27 Nm<sup>-2</sup> (0.2 torr). Further reduction reduces the rate at which oxygen and vacuum rates are equal ( $\approx$  critical rate); (3) the critical rate is proportional to oxygen pressure and is plotted in Fig. 7.

### Discussion

#### *Frequency effects in vacuo*

The fact that the 683 alloy, unlike the 5070A alloy and S.A.P., had a frequency dependent crack rate *in vacuo* suggests that the alloy system and type of dispersion hardening are important. It is not surprising that the two alloys with no frequency dependence are those with good creep properties—the 5070A alloy with a stable Al<sub>2</sub>CuMg hardening phase (usually designated S') and S.A.P. with a stable oxide dispersion and cold worked structure. The measurements by Hartman *et al.* [3] of crack rates at different fatigue frequencies in dry air support this view if it is assumed that their environment was effectively inert, (though apart from the presence of oxygen, the water vapor pressure may have not been low enough for the low frequency rates to be unaffected). Their results on the 7075-T6 alloy are similar to those on our 683 alloy. With the 2024-T3 alloy the frequency effects were greater as might be expected for a naturally aged and thus less stable material.

It is known that fatigue of aluminum alloys can involve re-resolution of precipitates or overaging [9], either of which might be expected to increase crack rate. In the work on the AlZn-MgMn alloy referred to earlier, the crack rate of the alloy when overaged was faster (mainly at low rates) than when in the peak hardness condition. At crack rates of  $10^{-5}$  mm/cycle the material ahead of the crack is inside the plastic zone (size  $\sim 10$   $\mu$ m) for about 10 secs at 100 Hz. If for only a tenth of this time, because of fatigue generated vacancies the solute diffusion,  $D$ , is increased to the same order as that for vacancies *i.e.*  $\sim 10^{-7}$  mm<sup>2</sup>/sec. then the diffusion distance  $2(Dt)^{\frac{1}{2}}$  is of order 1  $\mu$ m. This increase might permit overaging. Wei [4] has found that crack rates in 7075-T651 alloy in dry gases at 100°C are up to twice the rates at room temperature (fatigue frequency 143 Hz); local overaging may similarly be responsible.

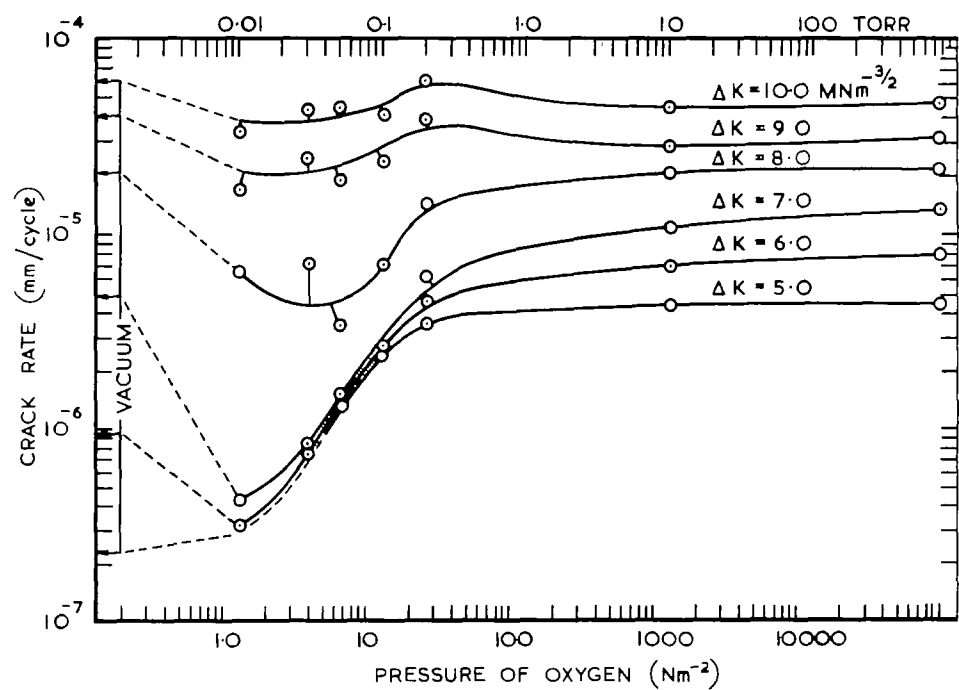
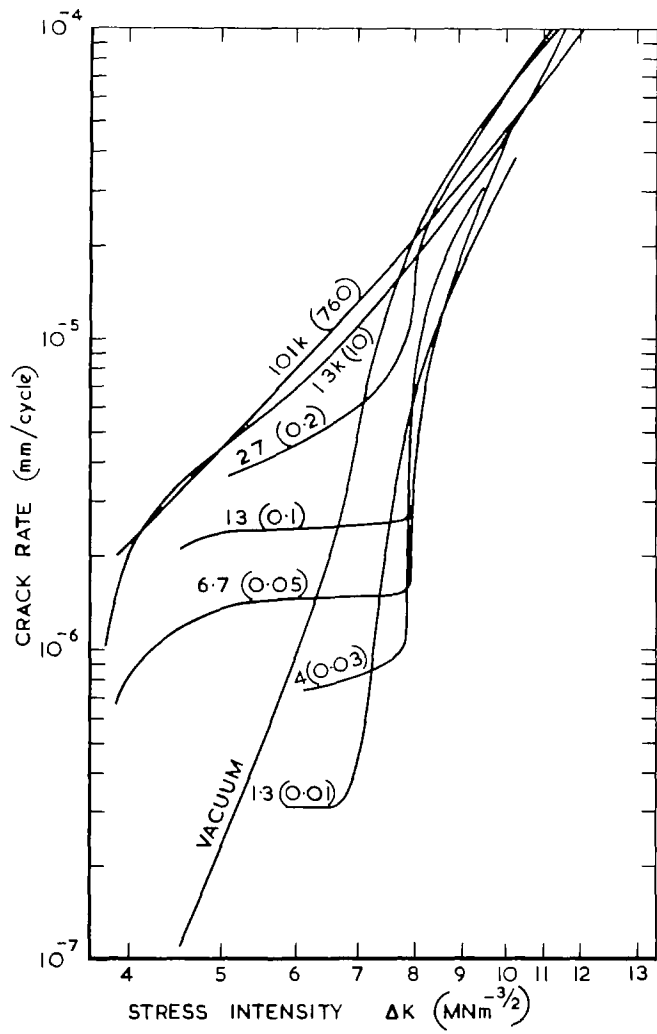


Figure 6. Crack rates of 683 alloy in oxygen at different pressures (Nm<sup>-2</sup>, torr); (a) at fixed oxygen pressures, (b) a fixed stress intensities.



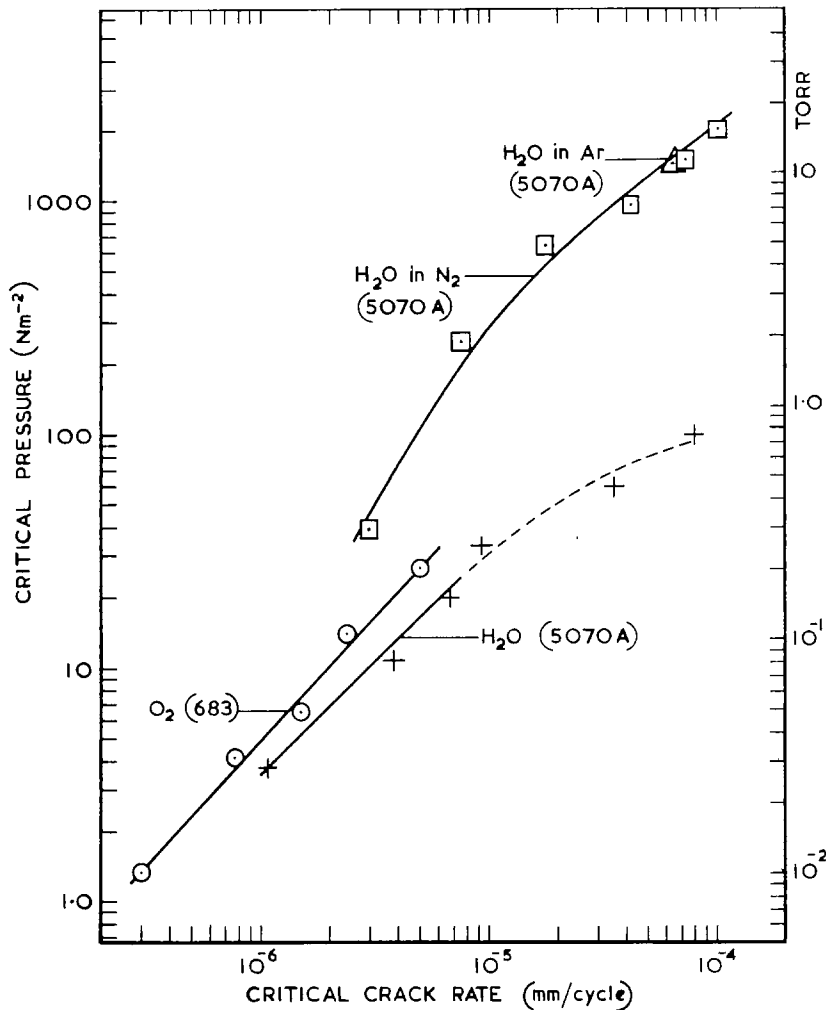


Figure 7. Dependence of critical pressure on critical crack rate.

#### Adsorption of water vapor and oxygen

We could assume that gas can flow down the open crack to the tip for half the fatigue cycle and that the crack rate depends on whether the new surface is covered in this time, before reversed plastic deformation starts. Assuming that a monolayer has to be adsorbed and that the sticking coefficient is unity then the calculated pressures of water vapor or oxygen are  $\sim 40 \text{ m Nm}^{-2}$  ( $3 \times 10^{-4}$  torr) at 100 Hz and  $0.4 \text{ m Nm}^{-2}$  ( $3 \times 10^{-6}$  torr) at 1 Hz. If from Figs. 3 and 4 we take the observed critical pressures for water vapor to be  $\sim 40 \text{ Nm}^{-2}$  ( $3 \times 10^{-1}$  torr) and  $0.4 \text{ Nm}^{-2}$  ( $3 \times 10^{-3}$  torr) then the calculated pressures are  $10^3$  too small. For oxygen (Fig. 6b), taking a critical pressure of  $\sim 13 \text{ Nm}^{-2}$  ( $10^{-1}$  torr), the discrepancy may be rather less but is still more than  $10^2$ . It has been suggested that the impedance of the crack to gas flowing down to the tip may be important, but calculations based on a simple model of crack geometry show that this should be negligible [10] and we have suggested that allowance should be made for the actual sticking coefficients. The work by Huber and Kirk [11], [12] and Jona [13] shows that sticking coefficients at the monolayer level may be  $10^{-2}$  or less and so at our observed pressures not much more than one or two monolayers of water or oxygen may be adsorbed. The pressures necessary for monolayer coverage will be increased if the surface is rough but this increase should be small with sticking probabilities as low as these.

Achter [14] however has taken the argument further and, assuming no crack impedance, has proposed that the crack rate is only altered if, during each cycle while the crack is growing,

the surface up to a distance of one lattice spacing behind the tip is covered with an adsorbed monolayer. This picture might be modified to allow for Laird's "plastic blunting" [15] but the essential points remain: (a) the cyclic features of fatigue crack growth are not directly important and (b) the critical pressures calculated on this basis are higher and should be proportional to crack rate.

The new pressures are simply those calculated above multiplied by the number of lattice spacings moved by the crack tip each cycle. Allowing for sticking coefficients in fitting the data, then it may be necessary to assume that coverages of rather less than one monolayer could affect crack rates or that coverage up to a few lattice spacings behind the tip is sufficient. However the oxygen, and the water vapor results at low pressure (Fig. 7) show clear proportionality between pressure and rate and thus give good support for Achter's model.

It has been assumed that the crack advances uniformly over the whole front every cycle. At crack rates approaching  $10^{-7}$  mm/cycle this is very unlikely and calculation of critical pressures will have to be modified. Experimental data at these low rates would be of interest.

#### *Water vapor with inert gases*

By introducing an inert gas at one atmosphere the migration of water vapor down the crack may be altered from molecular to Poiseuille flow. Assuming the latter, one can look for situations where gas-gas collisions reduce the supply of water vapor to the crack surfaces and thus account for the order of magnitude increase in critical water vapor pressure when nitrogen or argon is present. The results depend on the crack geometry assumed and simple calculations made on the more likely geometries suggest that the inert gas should have a negligible effect. However the experiments indicate one, and also a difference between hydrogen and nitrogen (the ratio of diffusion coefficients is not more than 5:1). Further work is needed to clarify the situation.

#### *Water vapor and oxygen: mechanisms*

At crack rates of less than  $10^{-5}$  mm/cycle the similarity in the critical pressure effects of water vapor and oxygen on crack growth suggests similar mechanisms. Huber and Kirk [11] have concluded that water vapor initially produces a monolayer of oxide only. Jona [13] has found no significant difference in the initial reactions of dry and wet oxygen on an aluminum {111} surface. Our less detailed metallographic observations agree in general with Meyn's very careful studies [16]. Air promoted tensile cracking, vacuum did not. In particular, by using the special hard AlZnMgMn alloy referred to earlier, the large grain size made it possible to see clearly slip band cracking *in vacuo* at low crack rates. Water vapor, air or oxygen prevented this. Figure 8 shows a fatigue crack which was started in air (rate  $\sim 10^{-4}$  mm/cycle, frequency 100 Hz) and then was continued at the same stress *in vacuo*. This led to failure on shear planes and the crack rate to fall to  $\sim 10^{-6}$  mm/cycle; this failure mode persisted until the rate reached  $\sim 10^{-5}$  mm/cycle. A "stereoscan" picture of the fracture surface in Fig. 9 shows that the initial failure mode *in vacuo* was similar to Stage I in contrast with the normal air Stage II mode [17]. Though we have not been able to obtain such clear changes in the other alloys this does show how at low rates water vapor or oxygen can provide an alternative tensile cracking mode.

At medium and high crack rates, oxygen however leads to only small increases or even decreases in crack rate. Even the relatively thick oxide films produced in one atmosphere of oxygen do not appear to promote tensile cracking. At rates where striations should have been visible the fracture surfaces of the 683 alloy, fatigued at 100 Hz, were studied for significant features. We support Meyn in being unable to find any clearly established striations on specimens fatigued *in vacuo*. Similarly in oxygen the striations were poorly defined and had none of the uniformity of those in air which were most well formed (Fig. 10a). Those in water vapor alone ( $1.5 \text{ k Nm}^{-2}$ ,  $\sim 11.0$  torr) appeared in isolated areas (Fig. 10b). The water vapor appeared to promote regions of tensile cracking—small areas of brittle fracture within a grain—rather than to modify each striation.

Returning to the all important question on the role of hydrogen released by the aluminum

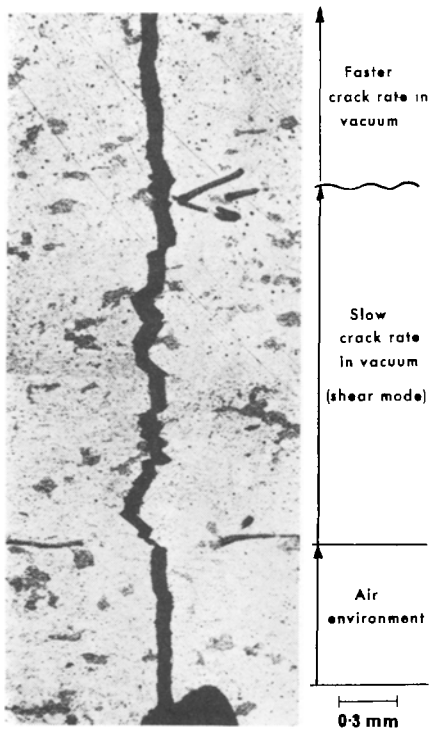


Figure 8. Mode transitions of crack growth in AlZnMgMn alloy. The crack grew from the slot at the bottom of the photograph.

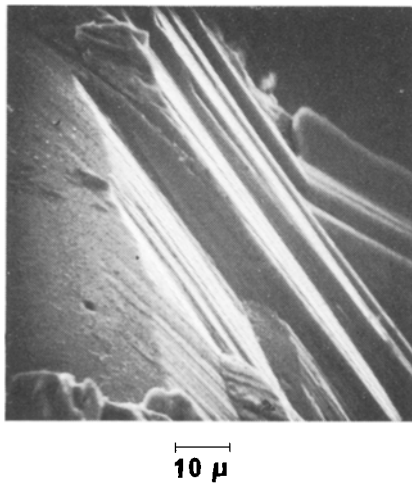
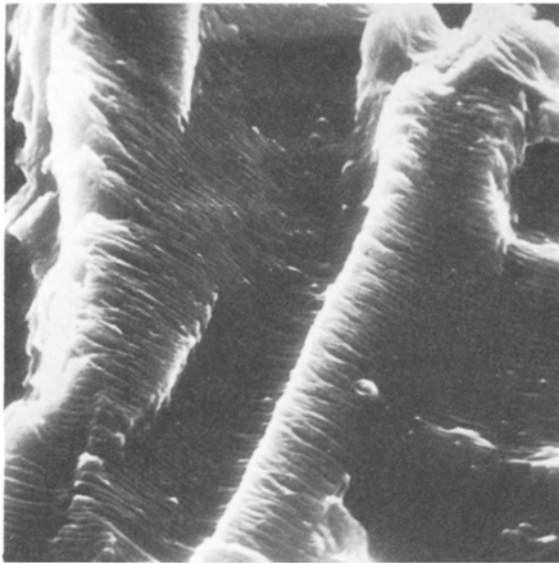


Figure 9. Fracture surface of shear mode failure in the AlZnMgMn alloy.

water reaction, the observed invariance of crack rate with fatigue frequency at high pressures of water vapor implies that if any hydrogen goes into the metal, crack rates are unaffected if the hydrogen diffuses away with time. This could be the same as saying that it is the hydrogen trapped near the surface or in the hydroxide layer which matters. This and the experimental support for Achter's proposal that the crack may be affected while it is actually growing tends to suggest an environmental effect based on the different actions of hydroxide/oxide films.

In air the oxygen appeared to reduce the baffling action of nitrogen. This could be because the oxygen helped to provide the initial oxide layer (which, as Huber and Kirk have found [11], occurs even in water vapor).

The surfaces of the specimens fatigued in low pressures of oxygen showed evidence of fretting debris and surface roughening, which increased as the crack grew through the critical region.



3  $\mu$



4  $\mu$

Figure 10. Striations in 683 alloy; (a) air, (b) water vapor. The crack growth direction is upwards.

It ceased when the crack rate suddenly started to increase. Since this fretting implies a measure of rewelding it could be the fretting at low levels of oxygen adsorption which is responsible for crack rates being less than *in vacuo*. This assumes however that any rewelding *in vacuo* has less effect in slowing down crack growth than the fretting. When crack rates in water vapor became less than *in vacuo* there were some similar though less obvious signs of fretting and thus possibly the same retardation mechanism may be suggested.

#### *Practical aspects*

It is useful to be able to predict crack rates at frequencies down to  $\sim 10^{-5}$  Hz (aircraft flight times) and up to  $\sim 10^3$  Hz (jet noise). The relationship between critical water vapor pressure

and frequency may be extended. Taking high frequencies, at  $10^3$  Hz for example, the critical pressure—between 130 and  $1300 \text{ Nm}^{-2}$  (1 and 10 torr) will be comparable with vapor pressures commonly met in air. Though the nitrogen and oxygen complicate matters one would expect crack rates to be lower at low humidities. Taking low frequencies and assuming no static stress corrosion effects, then for stable alloys such as 5070A whose rates at  $\frac{1}{60}$  Hz in air are not very different from those in water vapor, it seems likely that fatigue at lower frequencies should be little different. With the less stable 683 alloy its intrinsic frequency effect is not simply added to that of the environment. In water vapor the frequency effect is less than *in vacuo* (Figs. 1(b) and 2(b)). Hartman *et al.* [3] have also shown previously that frequency effects in wet air are less than in dry air for 7075-T6 alloy and markedly so for 2024-T3. Thus with such alloys very low frequencies in humid environments may not lead to great increases in crack rate.

At low stress intensities the effects observed are unlikely to be altered by the use of thicker specimens. However for the 5070A alloy in air at values of  $\Delta K \gtrsim 10 \text{ M Nm}^{-\frac{1}{2}}$  (for 683;  $12 \text{ M Nm}^{-\frac{1}{2}}$ ) the normal plane strain mode of failure began to give way to plane stress failure. *In vacuo* the values of  $\Delta K$  were  $\gtrsim 9 \text{ M Nm}^{-\frac{1}{2}}$  and  $8 \text{ M Nm}^{-\frac{1}{2}}$  for the two alloys. Assuming that the material in plane strain is more vulnerable to environmental effects, then at such stress intensities thicker specimens may show greater sensitivity to environment.

## Conclusions

Experiments have shown that fatigue crack rates *in vacuo* were insensitive to changes in fatigue frequency in a 5070A (Al-Cu-Mg) alloy and S.A.P. (10.5% wt  $\text{Al}_2\text{O}_3$ ). In a 683 type (Al-Zn-Mg; cf. 7075-T6) alloy, crack rates were increased at lower frequencies.

All alloys showed similar qualitative changes in crack rate in water vapor, oxygen and air when compared with an inert environment (*e.g.* vacuum or nitrogen). The magnitude of the changes appeared to relate with alloy hardness.

The critical pressures of water vapor necessary to affect crack rates were increased one hundredfold when the fatigue frequency was reduced by a factor of a hundred. This observation supports a dynamic adsorption model. Critical pressures of water vapor in nitrogen at one atmosphere were about an order higher than those of water vapor alone. Oxygen alone had less effect than water vapor and primarily at low rates but showed similar critical pressures. The effect of water vapor in humid air was further modified by the oxygen present; at low rates the oxygen could be regarded as reducing a baffling action of the nitrogen.

Critical pressures of oxygen and of water vapor (alone and in nitrogen) increased with crack rate. This gives support to Achter's proposal that the environment has its effect on the crack at the moment in the fatigue cycle while it is growing. The observations on water vapor and oxygen critical pressure are approximately summarized:

$$\text{critical pressure (Nm}^{-2}\text{)} = 4 \times 10^4 [\text{frequency (Hz)} \times \text{crack rate (mm/cycle)}]$$

At low rates, oxygen and water vapor could encourage tensile instead of shear cracking. At high rates water vapor alone could give rise to regions of tensile cracking free of striations.

The results give some support for models of an environmental effect based on the different actions of surface layers.

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#### ZUSAMMENFASSUNG

Bei Ermüdungsfrequenzen von 100, 1 und  $\frac{1}{60}$  Hz zeigen Messungen des Ribwachstums daß für eine gewisse Legierung die Veränderung des Wachstums als Funktion der Frequenze teilweise oder völlig von der Umgebung abhängt. Man hat versucht den Einfluß von Sauerstoff, Wasserdampf und Wasserdampf im Gegenwart von Sauerstoff und Stickstoff zu untersuchen. Der notwendigerweise das Ribwachstum zu verändernde kritische Gasdruck hängt von Frequenz und Fortpflanzungsgeschwindigkeit ab. Die Vorgänge bei den beobachteten Erfolgen werden erörtert. Die Ergebnisse deuten auf Gasadsorption an den Ribenspitzen und auf die unmittelbare Wirkung der Flächenschichten auf den wachsenden Rib.

#### RÉSUMÉ

La mesure de la croissance de fissure aux fréquences de fatigue de 100, 1 et  $\frac{1}{60}$  Hz dans des environnements gazeux montre pour un alliage défini que la variation de la vitesse de croissance avec la fréquence dépend de l'environnement soit partiellement, soit totalement. L'effet d'oxygène, vapeur d'eau, et vapeur d'eau en présence d'oxygène et azote a été à l'étude. La pression critique de gaz à que la croissance vient de changer dépend de la fréquence et de la vitesse de propagation. Les mécanismes pour les effets observés sont discutés et les résultats s'attribuent à l'adsorption de gaz à l'extrémité de fissure et à l'action immédiate des couches superficielles sur la fissure croissante.