International Journal of Fracture, Vol. 16, No. 6, December 1980

Alphen aan den Rijn, The Netherlands

Fatigue crack growth in polymers

J.C. RADON

Department of Mechanical Engineering, Imperial College of Science & Technology, London SW7 2BX, UK

(Received August 22, 1979)

ABSTRACT
A number of fatigue crack propagation laws applied in the study of polymers is described. Consideration of the stress field distribution at the crack tip leads to the application of fracture mechanics. It is shown that a simplified relationship of the form $da/dN = F\phi^{\alpha}$, where ϕ is a function of K_{IC}, K_{max}, K_{min} and K_{TH} appears to be a convenient expression for cyclic crack growth. The effect of mean stress is more complicated than that in the field of metals, the compressive component of cyclic stress may delay the crack growth. Cyclic tests in tension performed on PMMA and PVC are dependent on ΔK and its mean value, K_m. The threshold value, K_{TH}, is also influenced by K_m but a more complicated behaviour due to strain rate effects may be observed. Other differences, such as the position of upper and lower transition points and growth rate changes with frequence, are noted. The effect of biaxial cyclic loading of PMMA and PVC plates is compared and some differences highlighted. The results available so far indicate little effect of the crack curving on its growth. However, it is shown that, while the increasing biaxiality can substantially retard the crack growth in PMMA, no such effect was recorded in PVC. Finally, it is shown that at very high stress levels (region III), the cyclic crack growth consists of two propagation modes, namely, a pure cyclic propagation, together with slow growth. At lower stress levels, slow growth disappears and the crack propagates in pure fatigue (region II). In region I, the propagation is very slow, without the usual correspondence between cycles and striations. The results recently obtained on glass reinforced plastics (GRP) are also presented and differences highlighted.

1. Introduction

The fatigue failure phenomenon has been recognised by technologists for well over a century. Indeed, the first attempt to study the failure of a machine component subjected to repeated loading was made by Albert in 1829, who studied the fatigue strength of welded mine hoist chains. However, the first original systematic testing method is due to Wöhler [1] who, in 1852, published the results of his studies of the failure of railroad axles.

Based upon his test data, Wöhler proposed the two fundamental laws of fatigue:
1. That iron and steel may fracture under a stress not merely less than the static rupture stress, but also less than the elastic limit, if the application of the stress is repeated a sufficient number of times; and
2. No matter how many times the stress cycle is repeated, rupture will not take place if the range between the maximum and the minimum levels of stress is less than a certain limiting value.

These two laws have formed the basis of the major part of the conventional methods of fatigue failure analysis, i.e. in the determination of the endurance limit and fatigue strength of materials.

The Wöhler type rotating bending fatigue test is one of the cheapest most widely used simple methods for comparing the fatigue lives of different materials. This test is used in both the commercial and research fields to establish the relation between the applied loading stress and the number of cycles to failure for unnotched specimens.

0376-9429/80/060533-20 $00.20/0

The results are usually presented in the form of S–N curves. The use of unnotched specimens makes the results difficult to analyse or interpret because of the unknown stresses induced by the inherent flaws in the material. It would therefore appear more logical to induce initial cracks or notches of predetermined size, rather than to measure the behaviour of undetermined flaws.

It is now generally accepted that the concepts of fracture mechanics, which are based on the assumption that the material is flawed, can be used in studying crack behaviour in different polymeric materials. The simplest application of fracture mechanics in Wöhler's rotating bending tests of polymers was that by Constable *et al.* [2]. They showed that, for PVC sharp-notched specimens, in the rotating bending test, representing the results in a K_0–N curve rather than the conventional S–N curve would give a direct correlation between K_0, the initial stress intensity factor, and N, independent of notch size, since K_0 effectively includes notch size representation.

In the subsequent study of El-Hakeem [3], the environmental effects on two thermoplastics using similar equipment were investigated. Notched specimens of PMMA in methanol and high density polyethylene (HDPE) in Adinol were used and a correlation between K_0 and the number of cycles to failure, similar to that observed by Constable, was established. It was observed that, in the low speed tests (2 RPM) on PMMA, considerably longer lives were obtained in the methanol environment than in air. However, at higher cyclic speeds (2000 RPM), this increase in life occurred only at high K_0 values. A similar increase was recorded in the high speed tests performed on HDPE in Adinol.

S–N type data, while being of some qualitative use in aiding material selection, do not indicate the effect of discontinuities or stress concentrations on overall fracture resistance. For this purpose, other test methods have been developed and those using fracture mechanics are discussed later.

Our interest in fatigue failures centres now on those in polymers and a few brief comments regarding these materials seem appropriate.

Polymeric materials are now frequently used by engineering designers, wherever they can suitably replace the conventional materials. The comparatively favourable strength to weight ratio, excellent damping properties, good wear resistance and the facility of economic manufacture of intricately shaped components, are some of the advantages of polymeric materials, highly attractive in the design of components such as light duty gears, bellows, bearings, bushes, dampers, etc. Thus, in recent years with the continuous development of industrial plastics with improved mechanical properties, a tremendous amount of interest has been created in the behaviour of these materials under monotonic and cyclic loading conditions.

Static fracture has received by far the largest attention, especially in the case of glassy thermoplastics, where concepts of linear elastic fracture mechanics (LEFM) have been successfully utilised in the analysis of craze growth and the measurement of such parameters as the critical stress and flaw size for fracture under various loading conditions [4].

Attempts to analyse the fatigue failure process, however, have been relatively scarce, mainly due to the inherent complexity of the fatigue failure mechanisms as well as to the complexity of the polymer molecular structure and the subsequent lack of sufficient understanding of the effects of structural changes on physical behaviour. The initial studies have, inevitably, been carried out on elastomers [5].

In the case of thermoplastics, fatigue failure studies were initially based upon the conventional techniques of empirical analysis used for metals, i.e. the determination of the total fatigue life of specimens of such materials from plate-bending, rotational bending and reversed shear (torsion) tests. Results obtained from such tests were normally presented in the form of the S–N curves, as was the case for metals. However, the total fatigue life studies have, in practice, often proved to be of limited importance in the analysis of service failure problems. The structural inhomogeneity of polymeric solids inevitably results in the presence of a number of stress concentration points within the bulk of the material. Thus, the cyclic crack growth initiation life in these solids is often relatively short [6], and hence the importance of the information on the effects of various parameters influencing the cyclic rate of crack propagation becomes apparent.

It is pertinent to remark here that, due to the wide variations in the viscoelastic and mechanical properties of polymeric solids, attempts to generalise the results obtained from tests on one material must be cautiously treated. The quasi-brittle glassy thermoplastics behave in many ways similarly to metals. For example, the fatigue behaviour of PMMA and its fracture surface appearance were studied by Feltner [7] from rotating bending tests. His results demonstrated that the failure occurred in a cycle dependent manner with surface striations similar to those in metals. McEvily *et al.*[6] studied the high stress, short life fatigue crack propagation in polycarbonate (PC) and polyethylene (PE), reaching generally similar conclusions. However, as will become clear later, important differences in material behaviour due to basic structural properties may also be observed.

The studies in the fatigue of polymers, now in progress at Imperial College, are too extensive to be dealt with in one paper alone. Two problems have been chosen as being of particular interest here: the first describes recent efforts to develop a convenient form of a fatigue crack propagation law; and the second deals with the influence of biaxial stress on crack growth in large plates. A limited description of the relevant work originated at other research establishments is also included.

2. Fatigue crack growth

Fatigue, in general, and of polymers in particular, is the most common type of structural failure caused by the cyclic or the random application of loads. Consider first a simplified situation involving a cyclic life of a polymer. The fatigue failure of a component may start with the formation of a crack at a point of high stress concentration. During subsequent loadings, the polymer will exhibit either thermal softening or cyclic crack growth, depending on the frequency and stress amplitude applied; this paper deals with the latter form of fatigue failure.

The life of a structure subjected to cyclic loading can be divided into two parts, crack initiation and crack propogation leading to final failure. Fatigue crack propagation (FCP) frequently occupies the larger part of the total fatigue life; this is simply because structures invariably contain defects or inhomogenities acting as nuclei for crack growth. Macroscopically, the propagation of a fatigue crack appears to be a continuous process, thus a direct relationship between crack growth and loading has always been an attractive proposition because of its apparent simplicity. A relationship suggested as long ago as 1953 by Head [8a] in the form:

$$\frac{\mathrm{d}a}{\mathrm{d}N} = \frac{C\sigma^3 a^{3/2}}{(\sigma_y - \sigma)r^{1/2}} \tag{1}$$

may serve as an early example of a semi-empirical form. Here, σ is the applied stress, σ_y is the yield stress, a is a half crack length, r is the plastic zone, and C is a constant.

Another relationship, in an even simpler form, was preposed by Liu, who reformulated the original dimensional analysis suggested by Frost *et al.* [8b] using the concept of total hysteresis energy absorption to failure in the equation:

$$\frac{da}{dN} = C\sigma^2 a \tag{2}$$

Other laws formulated on the basis of fatigue damage modelling, dislocation theories, or using such methods as dimensional analysis, are too numerous to be described in detail here [4, 8b, 9].

Because of the successful use of linear elastic fracture mechanics (LEFM) in the analysis of fracture in statically loaded structures, it was logical to investigate the applications of these concepts to situations involving cyclic loading. As in earlier brittle fracture considerations, it was assumed that the structure already contained a crack, so that flaw initiation and stage I propagation (slip-line cracking) were not treated. Only stage II, the macropropagation phase, was studied. Within these terms, a relationship was sought between the cyclic crack growth rate, da/dN, and crack characterising parameters.

The first evidence for such a relationship was presented by Paris and Erdogan [8c] while investigating the validity of several crack propagation theories using data available in the literature, including the results of Frost mentioned above. It was concluded that the general validity of any theory could only be assessed by analysing a wide range of fatigue crack growth rates (FCGR). Plots of ΔK against FCGR were constructed and these appeared to unify all the results available, mostly from load cycling tests on metals, and could be represented by the equation:

$$\frac{da}{dN} = C(\Delta K)^m \tag{3}$$

where ΔK is the cyclic range of K_I, and m was found to be about 4.

In polymers, more direct evidence followed from constant load and constant ΔK tests on PMMA (polymethylmethacrylate) [10] with $K_{min} = 0$; m for this polymer was found to be higher than for metals, reaching at least 5.7. As will be seen later, other modifications were developed, in particular that for instability, as K_{max} tends to K_{IC} [11] and at least for metals, that for a minimum growth rate cut-off at about one lattice spacing per cycle [8b]. The range of K appeared to be the dominant variable, although it is now generally accepted that the mean value of K also seems to have some effect on da/dN. This is specially so for soft metals and polymers, where further modifications have to be incorporated.

The observation that no single analytical da/dN "law" can adequately represent the data in general can be partially explained by an examination of a typical data curve derived from a very large number of tests [8b, 10]. Region I, Fig. 2, shows considerable scatter, possibly due to starter notch effects or loading conditions. Other curve discontinuities are associated with minor changes in the fracture model and with the influence of environment. It can be seen that the analytical representation of such a curve is difficult.

Simple energy considerations may be used to confirm the influence of threshold, K_{TH}, and critical stress intensity, K_C, values which, in turn, depend on frequency, temperature and environment. The introduction of these factors will provide an improved representation of experimental data over the original equation (3). Unfortunately, because of inherent difficulties, the inter-dependence of various parameters has only been established for very few materials. It is for these reasons that the continuation of, and experimentation with, new FCG relationships is of relevance to further progress in fatigue. Some of these equations, together with the related research work at the Imperial College, will be discussed next.

As mentioned above, the influence of mean stress is now well established, although it was not taken into account in the construction of the original equation (3). Two extensive reviews on this subject are now available; the compilation carried out by the ASTM Task Force E24.04 [9] and the literature review by Maddox [11].

The Forman equation [11], used with success in many applications (cf. work by Hudson, Crooker and Lange, etc.), was developed to consider the acceleration of crack growth rate in region III in the form:

$$\frac{\mathrm{d}a}{\mathrm{d}N}=\frac{C(\Delta K)^p}{K_C(1-R)-\Delta K} \tag{4}$$

where R is the stress ratio $\sigma_{min}/\sigma_{max}=K_{min}/K_{max}$, and C and p are constants.

Erdogan and Ratwani [12] extended the above equation to include region I behaviour and obtained:

$$\frac{\mathrm{d}a}{\mathrm{d}N}=\frac{C(1+\beta)^p(\Delta K-\Delta K_{TH})^q}{K_C(1+\beta)\Delta K} \tag{5}$$

where $\beta=(K_{max}+K_{min})/\Delta K$, and C, p and q are constants.

Pearson [13] investigated the effect of mean stress using thick specimens. He replaced the value of K_C by K_{IC} and modified Forman's equation (4) to give:

$$\frac{\mathrm{d}a}{\mathrm{d}N}=\frac{C(\Delta K)^p}{[(1-R)K_{IC}-\Delta K]^{1/2}} \tag{6}$$

where C and p are constants.

Similar relationships were recently developed by Chu [14] who endeavoured to describe the stress ratio effect and predict the fatigue crack growth rate in the entire sub-critical range from ΔK_{TH} to K_C. One of these has the form:

$$\frac{\mathrm{d}a}{\mathrm{d}N}=\frac{C(\Delta K^p-\Delta K_{TH}^q)}{[(1-R)K_C-\Delta K]^r} \tag{7}$$

Elber [15] has shown that a fatigue crack may close at its tip for a part of the loading amplitude. The stresses necessary for crack opening were obtained experimentally and it was assumed that crack propagation can only occur when the crack is fully open. The crack closure phenomenon was attributed to the presence of compressive residual stresses as a result of the zone of tensile yielding (plastic zone) at the crack tip, left in the wake of the propagating crack. Crack closure stresses were measured for different applied mean stresses and it was observed that the proportion of the stress range during which the crack was open, U, was a linear function of R. An effective stress intensity factor, ΔK_{eff}, was then defined as $\Delta K_{eff}=U\,\Delta K$, and the crack growth rate expressed as:

$$\frac{\mathrm{d}a}{\mathrm{d}N}=C(\Delta K_{eff})^p=C(U\,\Delta K)^q \tag{8}$$

Since U increases with the stress ratio, Eqn. (8) is able to explain the increase in crack growth rate in terms of R.

The crack closure approach was also used to explain the experimental results obtained by Katcher and Kaplan [16], especially the existence of an R value, designated R_{cut}, above which $\mathrm{d}a/\mathrm{d}N$ *versus* ΔK data do not shift to faster rates with increasing stress ratios. The R_{cut} is the R value above which the crack is always open ($U=1$). Hence, for $\Delta K>\Delta K_{TH}$ and $R>R_{cut}$, ΔK_{eff} is the same irrespective of R, and $\mathrm{d}a/\mathrm{d}N$ will no longer depend on R. Using a finite element analysis of an elastic-perfectly-plastic material, Newman [17] computed the crack opening stresses under constant amplitude loading and found good agreement with Elber's results.

The equations quoted so far were designed primarily to describe the crack growth data in aluminium alloys which are highly sensitive to mean stress. Although FCP in steels appeared to be more complicated than in Al-alloys, steels were not considered mean stress sensitive. This was particularly so for mild steels. A simple relation used by Barsom [18] for all steels has the form:

$$\frac{da}{dN} = C\Delta G \tag{9}$$

where ΔG is the range of the crack extension force, G.

The effect of deformations on FCP in the zone adjacent to the crack tip has also been considered and two models may be mentioned. Krafft's [19] tensile ligament instability model (TLIM) assumed that the instability necessary for FCP was caused by the separation of microstructural tensile ligaments at the tip of the crack. The second, recently proposed by Proctor and Duggan [20], used a process zone of length l situated ahead of the crack tip and completely enclosed by the reversed plastic zone. It was proposed that the average plastic strain in the process zone should determine FCGR and that the active length of the process zone depends on a material structural constant j (for example, corresponding to the dispersion spacing in Al-alloys). A relation was obtained:

$$\frac{da}{dN} = f\left[(\epsilon)\left(\frac{\Delta K}{\sigma_y}\right)(l - j)\right]^{\alpha} \tag{10}$$

where α is the Coffin/Manson coefficient.

It is appreciated that the models discussed above, and in particular those of a semi-empirical nature, do not assist in clarifying the FCP process. They offer, however, a simple way of calculating approximate crack growth values. Consequently, they stay in use even when more advanced analysis, necessarily more complicated, is available.

It is therefore not surprising that the analysis of experimental results available in this department, together with the data from the open literature initiated a further investigation of the FCG relationship to other functions of K, COD and J-integral. A dependency on parameter K^2 was considered sufficiently simple and convenient, and it was shown by Arad [10] that the equation:

$$\frac{da}{dN} = \beta\lambda^n = \beta(K_{\max}^2 - K_{\min}^2)^n = \beta(2\Delta K K_m)^n \tag{11}$$

(where β and n are constants) could be applied successfully in the analysis of region II crack propagation data, both in high strength aluminium alloys and steels. Plotting $\log \Delta K$ against $\log (da/dN)/\beta'$, where $\beta' = \beta 2^n$ for a range of K_m values, and using a hypothetical value of $n = 1$, we obtained a linear relationship (Fig. 1) showing that this predicted mean stress effect corresponds to the experimental results for region II. The use of the parameter λ for the FCG in metals had the effect of bringing all the experimental data onto a single curve at least in region II, and partly in region III, although in region I the factor λ was dependent on K_m.

The application of Eqn. (11) to some polymers appeared to be satisfactory and, in some instances, even the data in region I showed much reduced scatter. Tests on Nylon 66 at a range of frequencies, cycled in air at 50% RH and 21°C, are shown in Fig. 2. The specimens were in the form of large CN plates and the effects of ΔK, K_m, orientation (crack speed decreased when cracks propagated perpendicularly to the extrusion direction of the Nylon sheet), loading waveform and frequency were investigated. Apart from Nylon 66, other polymers included here were PMMA,

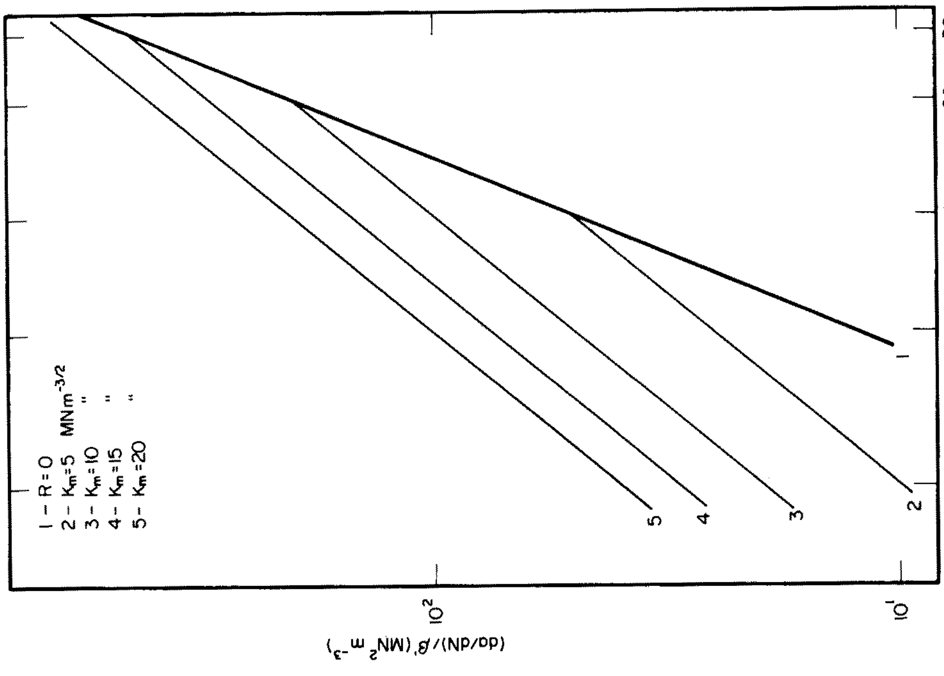

Figure 1. Theoretical lines of $(da/dN)/\beta'$ versus ΔK for a value of $n = 1$.

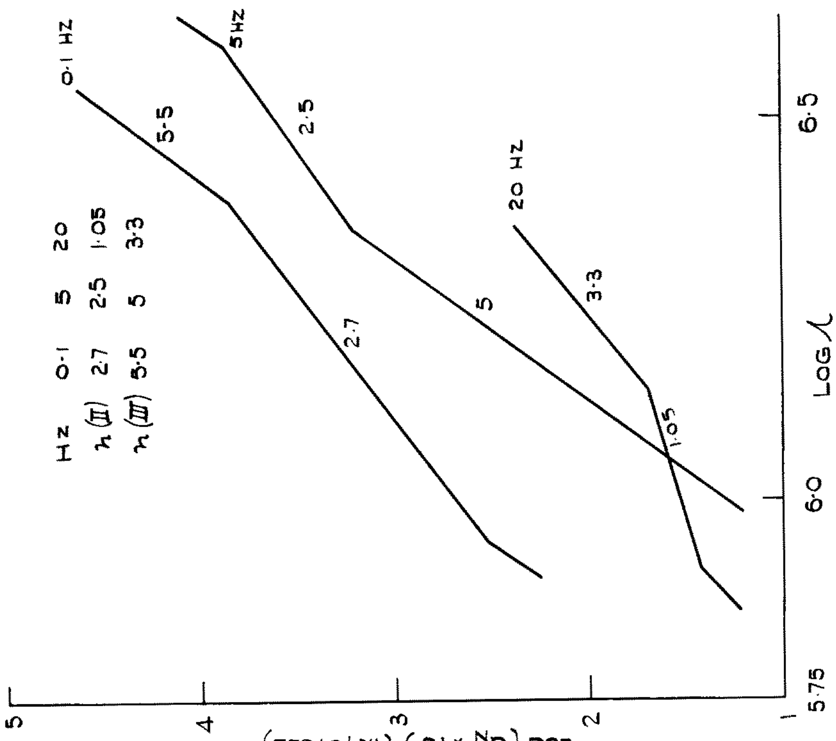

Figure 2. Dry Nylon 66. Crack growth versus $l(=K^2_{max} - K^2_{min})$ at 21°C and various frequencies.

polycarbonate (PC) and polyacetal copolymer [21]. Equation (11) represented the experimental data of the tested polymers well, except in the case of PC, where the effect of K_m disappeared at the loading rates ($\mathrm{d}K/\mathrm{d}t$) exceeding 4000 $\mathrm{lbf/in^{3/2}/s}$. A general tendency of FCGR to decrease with increasing frequency was observed in all polymers, although a cross-over between 5 Hz and 20 Hz (Fig. 2) indicates that a faster crack growth at higher frequency and in limited ΔK regimes is possible.

To predict behaviour in regions I and III, and also to normalise crack growth data in the entire sub-critical range, a modification of the Arad equation (11) was suggested by Branco [22] as follows:

For lower crack growth rates ($<2.5\times10^{-5}$ mm/cycle), λ is strongly dependent on K_m and its threshold value, $\lambda_{TH}=2\Delta K_{TH}K_m$, influences crack growth in this region. Further, a sharp increase in growth rate occurs at higher values of K_{max} approaching the critical value K_C. These two values provided the limiting conditions of the model:

$$\lim_{\lambda\to\lambda_{TH}}\frac{\mathrm{d}a}{\mathrm{d}N}=0 \quad\text{and}\quad \lim_{K_{max}\to K_C}\frac{\mathrm{d}a}{\mathrm{d}N}=\infty \tag{12}$$

Hence, in the modified model, the term $(\lambda-\lambda_{TH})$ figures in the numerator and (K_C-K_{max}) in the denominator. A generalised equation was formulated as:

$$\frac{\mathrm{d}a}{\mathrm{d}N}=A_C\Phi^{\alpha} \tag{13}$$

where:

$$\Phi=\frac{\lambda-\lambda_{TH}}{K_C^2-K_{max}^2}=\frac{2K_m(\Delta K-\Delta K_{TH})}{K_C^2-K_{max}^2} \tag{14}$$

and A_C and α are constants.

Using simple LEFM conversion, the form of (13) may be rewritten in terms of the crack opening displacement, COD(v), and crack extension force, G, as:

$$\frac{\mathrm{d}a}{\mathrm{d}N}=A_C\left(\frac{\Delta v-\Delta v_{TH}}{v_C-v_{max}}\right)^{\alpha}=A_C\left(\frac{\Delta G-\Delta G_{TH}}{G_C-G_{max}}\right)^{\alpha} \tag{15}$$

where $\Delta v=v_{max}-v_{min}$, $\Delta G=G_{max}-G_{min}$, v_{max} and v_{min}, G_{max} and G_{min} are the maximum and minimum values of v and G, respectively, Δv_{TH} is the threshold value of Δv for crack extension, v_C and G_C are the critical values of v and G for unstable fracture, and ΔG_{TH} is the threshold value of ΔG. For plane stress conditions, ΔG_{TH} is given by:

$$\Delta G_{TH}=\frac{(K_{max}^2)_{TH}-(K_{min}^2)_{TH}}{E} \tag{16}$$

where $(K_{max})_{TH}$ and $(K_{min})_{TH}$ are the maximum and minimum values of K at the threshold point. ΔG_{TH} represents the variation in cyclic potential energy with crack length, assuming the crack growth is still obtained at the threshold point (which is approximately true since $\mathrm{d}a/\mathrm{d}N<10^{-7}$ mm/cycle).

The parameter Φ is an energy or crack opening displacement ratio. It can be shown that it has the same dimensionless value whether expressed in terms of K, G or COD (v). Crack growth rate increases as $(\Delta v-\Delta v_{TH})$ increases and as v_{max} approaches v_C. These relations (13) and (15) imply that the crack growth will occur when a certain level of $\Delta v>\Delta v_{TH}$ is attained, and also an unstable fracture will be obtained when $v_{max}\to v_C$ or $G_{max}\to G_C$. Figure 3a shows results obtained on rigid PVC plates, CN, 6 mm thick, cycled at 0.2 Hz and plotted as a function of Φ, together with experimental points for two higher frequencies. Figure 3b shows the similar plot in terms of ΔK.

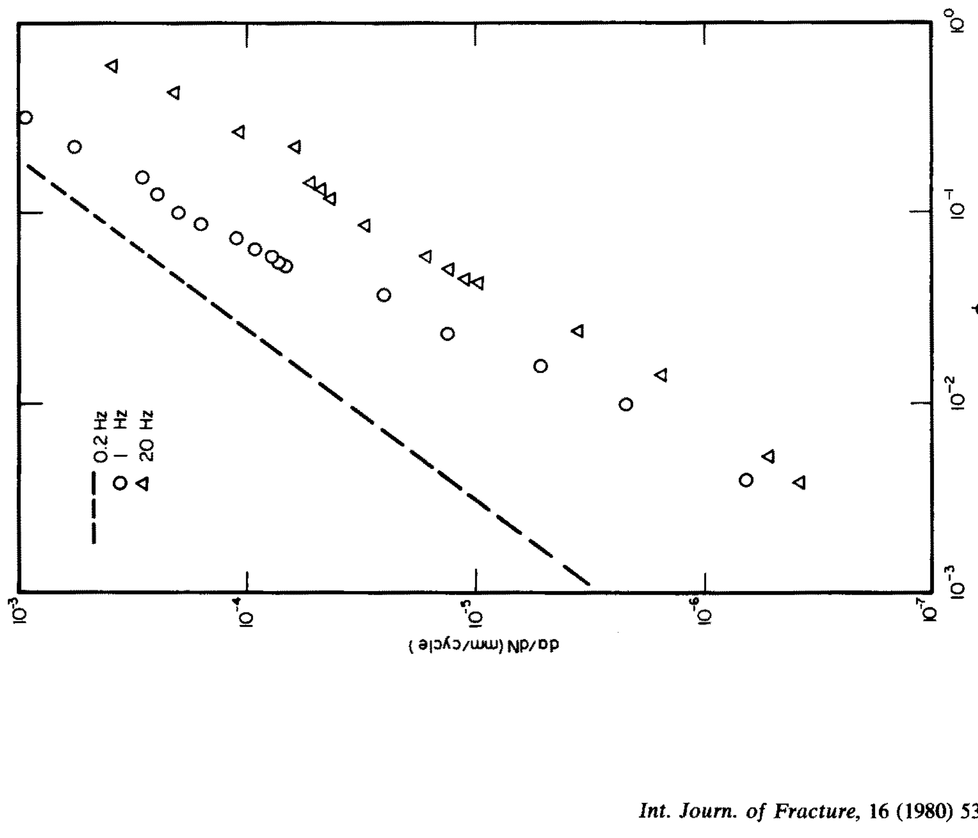

Figure 3. (a) PVC. Φ versus da/dN.

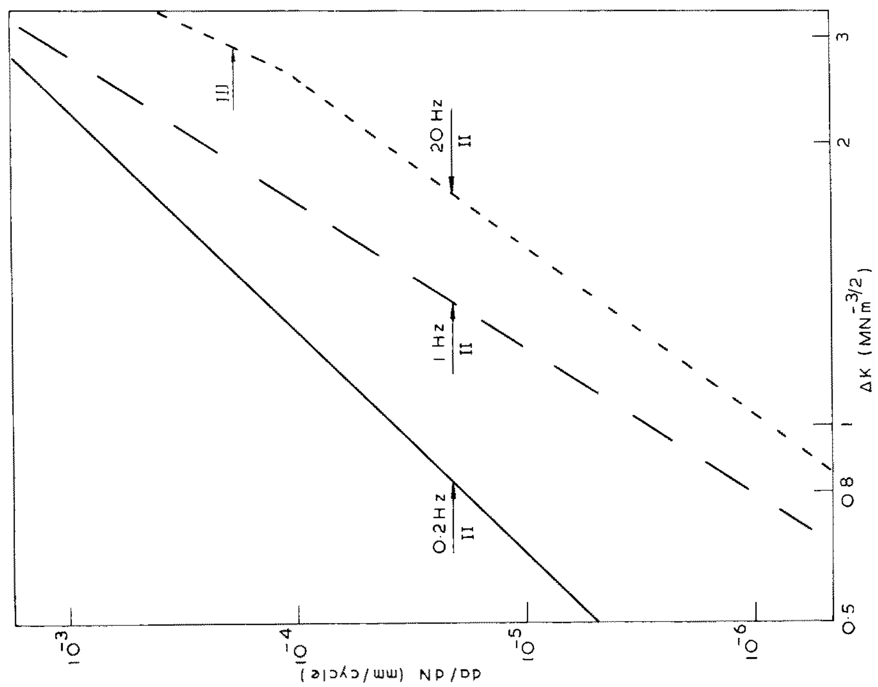

Figure 3. (b) PVC. ΔK versus da/dN.

An analysis of FCG in epoxy resin using (11) was reported by Sutton [23], who expressed the growth in terms of ΔG and thus met the total energy criterion, discussed elsewhere [24, 25, 26].

A further generalisation of Arad's law was proposed by El-Hakeem [3] in the form:

$$\frac{\mathrm{d}a}{\mathrm{d}N} = B(K_{\max}^2 - K_{\min}^2)^{\alpha}(K_m^2)^{\beta} = B(\psi)^L \tag{17}$$

the parameters B, α and β being functions of frequency, environment, test conditions and material properties, and:

$$\begin{aligned} \psi &= (K_{\max}^2 - K_{\min}^2)^{\alpha/L}(K_m^2)^{\beta/L} \\ L &= \alpha + \beta \end{aligned} \tag{18}$$

or, using stress ratio R:

$$\frac{\mathrm{d}a}{\mathrm{d}N} = B\psi^{\alpha+\beta} = B\left\{(\Delta K)^q\left[(0.25)^{\beta}\left(\frac{1+R}{1-R}\right)^{\alpha+2\beta}\right]^{1/(\alpha+\beta)}\right\}^{(\alpha+\beta)} \tag{19}$$

Since Eqn. (19) was based on Arad's law, it retained its dependence on K^2. Hence, it can be expressed in terms of COD or the plastic zone size, r_p, and this transformation is described in [27]. An example representing FCGR in dry Nylon 66 tested at 5 Hz, in air, 20°C and 50% RH, is shown in Fig. 4. Results obtained in a range of polymers tested at various frequencies and environments have not been satisfactorily analysed. It was shown that the proposed modification of Arad's law provides a unique relationship between $\mathrm{d}a/\mathrm{d}N$ and the parameter ψ for any particular test conditions.

A close resemblance between fatigue and environmental effects was discussed recently by Williams [28]. A simple model was proposed, according to which a

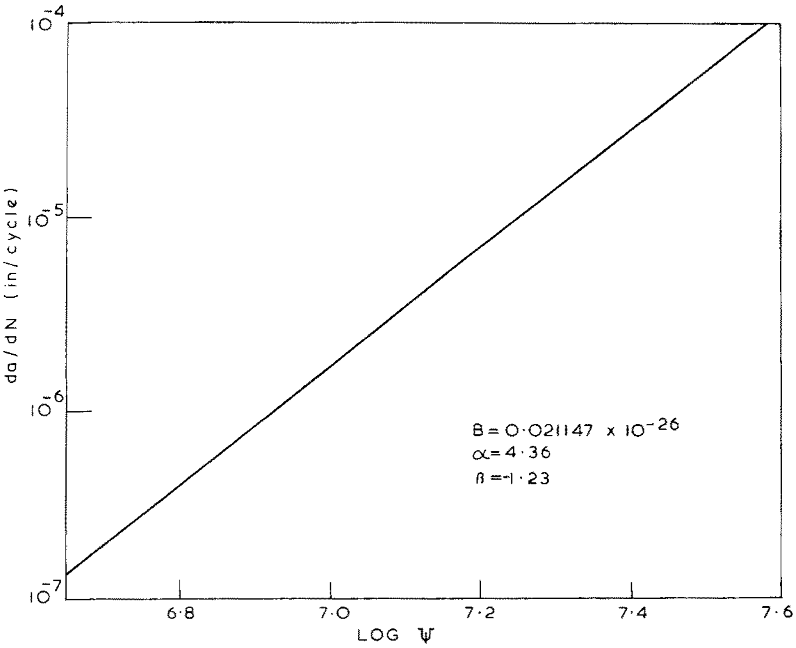

Figure 4. Dry Nylon 66 at 21°C and 5 Hz

two-stage craze zone was formed in front of the crack tip during each cycle. The original length of this zone was:

$$r_{p_0} = \frac{\pi}{8} \frac{K_{IC}^2}{\sigma_C^2} \tag{20}$$

As the process repeats itself, the stress on this original length will be reduced to $\alpha\sigma_C$. In order to maintain equilibrium some new craze must be formed with a craze stress of σ_C. A crack growth relationship, analogous to the Paris equation, is thus derived in the form:

$$\frac{da}{dN} = \frac{r_C}{(1-\alpha)^2}\left(\frac{K_I^2}{K_{IC}^2} - \alpha\right) \tag{21}$$

3. Biaxial fatigue

The important role played by LEFM in the investigation of FCP in large engineering structures was shown in the first part of this paper. The magnitude of this problem, exemplified by the number of fatigue failures reported in the literature, warrants detailed investigation. In the studies of service failures, it is often realised that, unfortunately, the available information is not adequate and the data incomplete. A useful approach, therefore, is to simulate a real structure by using simplified models or specially designed test pieces, such as are considered here.

Large plates, uniaxially stressed, containing a crack propagating perpendicularly to the applied stress are a suitable model for such experimental studies. The introduction of a complex stress field by a superposition of a transverse stress will complicate this situation substantially.

Two factors are of particular interest, the prediction of the crack trajectory and the velocity, or growth rate, of a propagating fatigue crack. Many examples of flawed structural components are available. Both factors, involving fatigue life predictions, are encountered in a rotor disc, such as is used in gas turbines. The modelling of a crack initiating from a turbine blade root using PMMA plates as a suitable model material was discussed by Arad [29], and the analysis of a crack path in gear discs using a finite element program by Urriolagoitia [30].

Before discussing some recent biaxial fatigue results, the effect of biaxiality on fracture toughness, as described in the preliminary static tests, will be considered. Also, some brief comments on craze formation are included.

(a) *Fracture toughness*

The influence of transverse stress on fracture has been studied in the last few years with renewed interest and a number of theoretical models have been proposed. However, on careful perusal of published data, it appears that only a few of those studies are relevant to the present work. This is mostly due to the different testing methods used, the test conditions, such as velocity or temperature range, and the varying specification of the investigated materials.

In 1920, Griffith [31] originally considered biaxially-stressed plate in the energy balance calculation, but found that the transverse stress had no effect. Not all subsequent investigators agreed with his results, and, in fact, some have found otherwise. Swedlow's work [32] is of particular interest here; he calculated G on identical premises to Griffith, using a different procedure, and showed that the transverse stress appeared in it. He concluded that the transverse stress could either strengthen or weaken the plate, depending on the Poisson's ratio and the stress state. Other recent models were based on elastic-plastic finite element analysis and on dislocation mechanics [33]. As already mentioned, the existing experimental work is rather limited, but the following reports are closely related to the present study.

Kibbler and Roberts [34] studied PMMA and aluminium alloy 6061 plates subjected to transverse dead loads, and also to lateral constraint, the latter situation creating an in-phase biaxiality ratio of about one-third. They used 3.2 mm thick CN plates of a cruciform shape. Changes in the biaxiality, B, from 0 to 2 increased the fracture toughness of both materials by about 25%. It is not possible to analyse the cause of these high values of K_{IC} at present, but the contribution of the load interaction effects on a cruciform geometry should not be excluded. Using the same specimen shape, Ueda *et al.* [35] carried out fracture tests on PMMA. It is important to note that, in this case, no effects of biaxiality were observed.

Turning now to the tests performed by Leevers [36], it should first be mentioned that they were very similar in all respects to both of the previous investigations. It should also be remembered that the testing facility, specimen design and other particulars were unchanged for the subsequent work on fatigue crack growth and crack path stability discussed later. In the first series of tests, cruciform PMMA specimens, 4 and 6 mm thick, were precracked to obtain sharp and straight crack starters, and then tested in a specially built hydraulic biaxial fatigue machine at a constant loading rate of 0.2 $MNm^{-3/2}/s^{-1}$, at 20°C and at a range of biaxiality B between 0 and 3.

For the quoted loading rate and biaxiality ratio, the calculated fracture toughness, K_{IC}, was nearly constant and equal to 1.6 $MNm^{-3/2}$. This value agrees well with the fracture toughness investigated earlier by Johnson [37a] over a wide range of loading rates and temperatures. In this work, some features of the toughness curves were found to be subjected to time-temperature shifts and were explained in terms of relaxation motions of parts of the polymer molecule. The coincidence of the fracture mode transition with the peak of the β relaxation was also observed, and the recorded stable crack growth studied analytically and fractographically. One of the conclusions reached was that this stable flaw growth, which occurred only above a certain K-level, could substantially influence the critical toughness values and thus this observation is of relevance.

The K_{IC} values derived from the biaxial static tests described above were calculated from the load at fracture using the crack length measured on the specimen surface. Corrections of the "time to fracture" corresponding to the delay caused by the slow crack growth and also allowance for a slight crack tip curvature (out of the linear path of the crack) were included in the calculation of the K_{IC} values. However, these corrections had only a minimal effect. The final results showed a slight decrease of K_{IC} with increasing biaxiality B amounting to less than 5% over the range tested.

The results available from the second series of tests conducted on rigid PVC plate using the same experimental procedure appear to be more complicated. They suggest that this material fractures at a value of K_q equal to 9 $MNm^{-3/2}$, only marginally increasing with the biaxiality increase from 0 to 3. These tests were performed at 20°C and at a slightly higher loading rate of 0.4 $MNm^{-3/2}/s^{-1}$ than the previous series on PMMA plates. All specimens fractured in a ductile mode, hence the results could not be treated by rigorously applying LEFM analysis.

The ductility of PVC was investigated in three-point bend tests at the same loading rate using Charpy sized specimens, and the transition temperature obtained at −50°C [38]. However, for a fast crack, such as was induced by impact, this transition shifted towards +50°C with a consequent danger of brittle fracture around room temperature. At present, there are not many toughness data available on this particular grade of PVC (Darvic, grade 110, ICI), but earlier tests indicate that the dynamic toughness could be substantially lower than its static counterpart in the whole regime from room temperature down to −200°C. It was also noticed that the strain rate sensitivity increased with decreasing temperature. Other data available on PVC, albeit of a different provenience, confirmed the previous results [39]. The static toughness of this grade at 21°C was only 5.5 $MNm^{-3/2}$, decreasing substantially at higher strain rates. In fact, at impact rates, K_{IC} was as low as 2.4 $MNm^{-3/2}$. It was noted that the toughness peaks, arising from the increased molecular mobility at the β relaxation, were moving to higher temperatures with increasing testing speeds. To demonstrate this trend, a prominent peak K_{IC} was established in the slow bend test (8.5 mm/s) at −50°C. At higher loading rates (565 mm/s), this peak shifted to −15°C. However, because of the proximity of the major α relaxation, the resolution of minor features of the K_{IC} curve was more difficult to establish than in the case of other polymers.

(b) *Crazes*

The formation of a zone of crazed material ahead of the crack tip has been observed in many glassy polymers, polycarbonate, polystyrene and others, as well as those discussed here. Some recent studies of the mechanism of craze initiation and growth under static loading conditions were mentioned in [40a]. A brief account of pertinent research in fatigue is included here.

It is known that cyclic loading at low stresses and frequencies may cause variations in modulus and internal damping. It has been generally accepted that, in metals, the increase in modulus and the decrease in damping is the result of the restriction of dislocation motion. However, the mechanisms involved in polymers are different and models leading to the craze initiation stage are not yet available. The process of the formation of microvoids, varying in size from 1 to 100 mm was discussed by Zhurkov [41], who used the low-angle X-ray scattering method. Subsequently, Bouda [42a] carried out a detailed investigation of some mechanical properties early in cyclic life. From the analysis of his results, it was suggested that low stress cyclic loading induced a non-uniform volume contraction, creating internal tension within the structure which then relaxed by yielding and microvoid formation. In particular, it was observed that certain mechanical properties changed their characteristics during the early fatigue life and this occurred long before the appearance of crazes. Thus, internal damping of PMMA at γ peak decreased; this reduction was associated with the corresponding increase of damping at the β_{II} peak. Also, an increase of the shear modulus over a wide temperature range (93°K − 310°K) was recorded. Although the full explanation of these processes preceding craze initiation is not yet available, it could be speculated that the bulk contraction of the amorphous domains may cause changes, not only in the internal damping, but also in the activation energy and a shift of the glass transition temperature.

The second step in the fatigue damage process is craze initiation, discussed by Kitagawa, who proposed a method for the determination of the critical craze stress under combined stresses in PVC, PMMA and other polymers. Subsequently, Argon [43] suggested that craze nuclei develop from the microcavities at a rate dependent on the hydrostatic stress level, this process being followed up by the craze growth. As the craze matter resembles bundles of filaments, it appeared that the transverse stress will exercise only a limited influence on the growth rate. Its influence on the craze initiation is, however, more difficult to estimate. Recently, the attention of many workers has been drawn to the shape of the crazed zone, which was successfully correlated with Dugdale's plastic zone for the static crack growth. Apart from the mechanical and structural changes mentioned above, strain hardening occurs during the initial stage of the fatigue process; a number of models accounting for the increased load carrying capacity have recently been developed. Kausch and Becht [40b] suggested that during the second and subsequent cycles an increased load is carried by partly extended chains, and also by unextended ones which do not undergo chain scission. Another model was proposed by Peterlin [40c], who suggested that the sliding motion of the fibrils and microfibrils will similarly increase the load bearing capacity. Experimental evidence is at present incomplete.

Also, more work is needed in pure fatigue, briefly indicated as follows. It has been suggested that the fatigue crack growth could be represented by cyclic fracturing of singular filaments in the zone of crazed material adjacent to the crack tip, and this model has been discussed by various workers [40, 42b]. Using the experimental observations available, Hertzberg [44] suggested splitting the crazed zone into two parts: the process of primary craze formation from the bulk polymer occurring at the tip of the actual zone; and the stretching and cyclic hardening of the bundles of fibres close to the crack tip until fracture. This is a particularly suitable model, which can also account for the discontinuous growth observed in PVC. A similar two-part division of the craze, together with the Dugdale configuration, was used by Williams [40a], who expressed the length of the zone after one cycle in the form of Eqn. (20). In this model, a local balance between the broken fibres close to the crack tip and the new craze formed at the craze tip appears to be a reasonable proposition. FCG was expressed as in Eqn. (21) and a good qualitative agreement with experiments was obtained [40d]. It could be argued that the stresses at the craze tip of the two-stage zone are too high and should be reduced. A gradually decreasing stress from the maximum at the craze tip into the elastic field ahead of the zone and the introduction of the strain hardening are two obvious refinements which could be added to the model. Before evaluating the merits of the above models, further experimental evidence is needed, in particular on the size and shape of the craze zones.

4. Effect of biaxial stress on FCG

The discussion of several models in the previous paragraphs indicates the complexity of the local mechanisms of fatigue crack extension on the microscale and, until these are better understood, little theoretical analysis of any effects of stress biaxiality seems possible. The continuum mechanics approach of Rice [45] may, however, provide some insight. In Rice's "boundary layer" approach, necessarily simplified, continuum processes are seen as setting boundary conditions on the microstructural crack extension processes in the near-tip region. If this region can be regarded as small compared to the overall component geometry, a cyclic history of near-field fracture mechanics parameters, such as K_I, must determine the fatigue crack growth rate. Since LEFM leads to an expression for plastic zone size in terms of K_I, a relationship of the form of Eqn. (3) seems to point to a critical "hysteresis energy absorption rate" criterion for which the plastic zone volume provides a measure. Knowledge of the effect of stress biaxiality on the plastic zone configuration could then be used to estimate the change in propagation rates.

Since biaxial stress reduces the plastic zone size, it must be expected by this argument to reduce the fatigue crack growth rate. The experimental evidence for such an effect is scanty, and what little exists seems to indicate the opposite: the decrease in plastic zone size appears to cause increased cyclic crack growth.

In a continuation of their static work, Kibbler and Roberts [34] studied PMMA and aluminium alloy CN plates subjected to fatigue. They restrained lateral extension to produce a B value equal to Poisson's ratio. The results showed slight reductions in both the exponent m and the coefficient C of the Paris expression with increasing biaxiality.

Joshi and Shewchuck [46] carried out bulge-bending tests on simply-supported round and elliptical plates, producing various stress biaxiality ratios at their centres. They noted increasing fatigue crack growth rates with increasing transverse stress up to the equibiaxial state, covering the range of $0.75 < B < 1.0$. Also, Pook and Holmes [47] investigated cruciform specimens, this time made of Ni-alloy, in a special rig using two servo-hydraulic actuators. It appeared from the tests conducted so far that a biaxiality $B = 2$ was necessary to cause substantial deviation from the original plane. For crack which did not deviate, it was noted that the stress parallel to the crack had little effect on fatigue crack growth rates.

Miller [48] has examined fatigue crack growth under biaxial loading conditions, where the transverse stress was both stationary and alternating. A finite element examination of plastic zone size revealed that the zone was smallest for equibiaxial tension and a maximum for pure shear. A two parameter description was required to describe fatigue crack growth for such loading conditions. Only a slight decrease of FCG was recorded for positive B values.

The biaxial fatigue tests reported here are similar to those described in the previous paragraphs. Three materials were investigated: PMMA and PVC, mentioned earlier, and a glass reinforced resin, Atlac 382–05A. All materials, supplied by ICI, UK, in the form of plates between 3 and 6 mm thick, differed substantially in their respective mechanical properties. At room temperature, PMMA is brittle, but also isotropic, homogeneous and transparent, well suited as a model material. PVC is highly ductile in slow load applications; however, when loaded at high strain rates, it may suffer catastrophic brittle fracture. It is also one of the largest-used industrial thermoplastics, in particular in the manufacture of pipelines. Glass reinforced plastic (GRP) is also ductile and, like PVC, has many industrial applications. It is highly corrosion resistant and often used in combination with PVC as a structural laminate in chemical plants. So far, it has been tested in only one thickness of 3 mm as 30% glass chopped-strand-mat reinforced plate. The mechanical behaviour and fracture properties of PMMA are well known, while those of PVC are not well documented. Very little work has been done on GRP.

The tests were carried out in the biaxial rig mentioned earlier [36] in air and at 21°C, using wholly tensile cycles of approximately 0.2 Hz frequency and a range of biaxiality ratios B between 0 and +2. The equipment and testing method used, as well as the geometry of the cruciform specimens, have been described in [49]. Apart from the FCG tests, crack closure and changes of the crack trajectory caused by the introduction of the biaxial stress field were also investigated; they are the subject of separate reports [50].

In what follows, the FCG rates are expressed in the form of the relationship da/dN *versus* ΔK. Uniaxial fatigue tests performed on an Instron type TT-C 45 kN machine on SEN specimens were closely comparable with the results obtained in the biaxial rig. In these tests, and in the biaxial tests carried out at $B = 0$ and $R = 0$, large differences were observed in the value of the exponent, m, in Eqn. (3), namely, 8 for PMMA, 2.6 for PVC and 3.2 for GRP. The results on PMMA (Fig. 5) show a distinct, but not large, reduction of the growth rates with increasing biaxiality over the whole range from 10^{-7} to 10 mm/cycle. The tendency to faster growth rates for the uniaxial results appears to be stronger in region III. The results from the test performed on PVC, using the same range of biaxiality ratios, are substantially different. No influence of the biaxiality on the FCG was detected (Fig. 6). The range of FCG rates

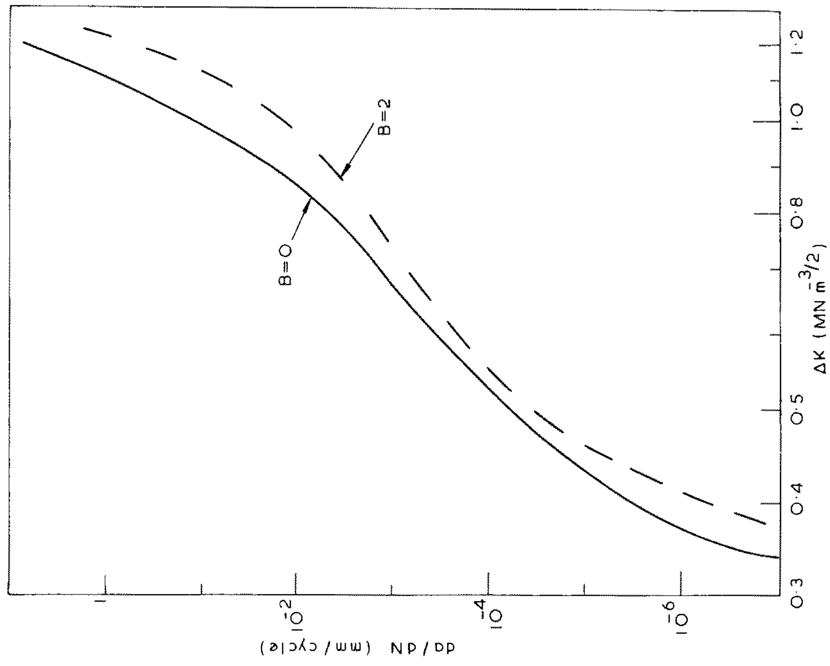

Figure 5. PMMA. ΔK versus d*a*/d*N*. 0.2 Hz.

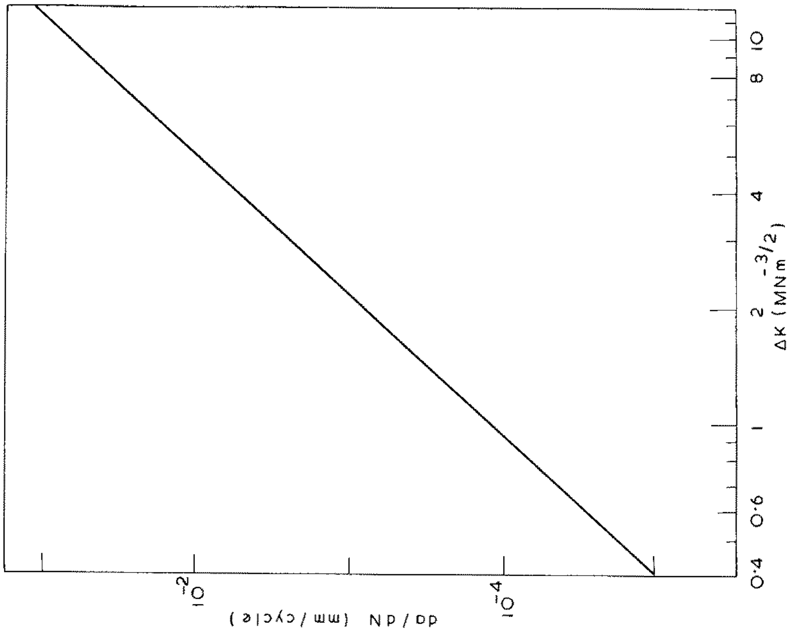

Figure 6. PVC. ΔK versus d*a*/d*N*, B = 0 and 2. 0.2 Hz.

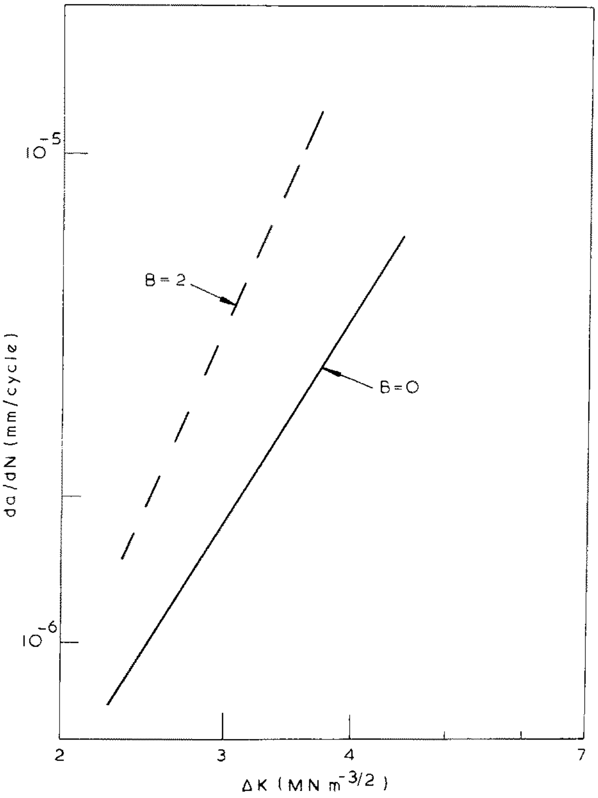

Figure 7. GRP. ΔK versus da/dN. 0.2 Hz.

covered so far is somewhat smaller than for PMMA, namely, from 10^{-5} to 10^{-1} mm/cycle. The results, which are within a narrow scatter band, appear to be on a straight line of slope $m = 2.6$. The limited results on GRP available at present (Fig. 7) suggest increasing growth rates with higher biaxiality ratios. As the rates investigated cover not much more than one order of magnitude, these results should be treated with care; this work is in progress.

The results available so far may be summarised as follows. PMMA shows a distinct reduction of FCGR with increasing biaxiality. Further tests at lower growth rates would provide the designer with useful data for low stress cycling. The tests on PVC showed no influence of biaxiality in the investigated regime and a much wider range will have to be covered. The same applies to GRP, where results for both higher and lower crack growth rates will be sought. However, as the present tests have established the trend of increasing FCG rates with higher biaxiality ratios, further crack growth prediction may not be too difficult.

A further study of fractured surfaces is also necessary. This will facilitate a precise evaluation of slow crack growth in PMMA. Slow growth in PVC has not been observed in the investigated range and it appears that the crack propagates by the true fatigue mechanism. It is expected, however, that the slow stable extension may occur at much higher values of K than those applied so far. In some cyclic tests performed at ΔK values of 5 MNm$^{-3/2}$ and higher, a macroscopic hole formation ahead of the crack tip was observed. This tearing process may assist in explaining the discrepancy between the number of cycles and striations reported in the literature. It is accepted that one-to-one correspondence between cycles and striations is usually observed, and this represents a continuous, step-by-step, movement of the crack front. In the situations where a discontinuous growth of the crack occurs, a small number of striations correspond to a much larger number of cycles. This process was first reported by Elinck *et al.* [51], who also proposed a discontinuous formation, and subsequent fracture, of the craze zone. Such similarities on a microscale were investigated by other researchers and Hertzberg [44] presented evidence that the formation of the zone takes up the first 10% of its life.

Discontinuous crack growth may also be induced externally, by sudden changes in the applied load, ΔK or the biaxiality ratio. The resultant accelerations, retardations or crack arrests are described elsewhere [50, 52].

5. Conclusions

a. Some modifications of an empirical fatigue crack growth model proposed by Arad are presented, together with experimental results for a range of polymers.
b. The introduction of the stress biaxiality may influence FCG rates in polymers. It was found that in tensile fatigue tests at 0.2 Hz, in air and at room temperature, the increased load biaxiality ratio from 0 to 2 caused (1) a decrease of FCGR in PMMA, (2) no change of FCGR in PVC, and (3) an increase of FCGR in GRP.

REFERENCES

[1] A. Wöhler, English abstract in Engineering, 11 (1871) 199.
[2] I. Constable, L.E. Culver and J.G. Williams. *International Journal of Fracture Mechanics*, 6 (1970)
[3] H.M.A. El-Hakeen, Environmental Fatigue of Polymers - A Fracture Mechanics Approach, Ph.D. Thesis, University of London (1975).
[4] S. Rabinowitz and P. Beardmore, *Journal of Materials Science*, 9 (1974) 81.
[5] E.H. Andrews, *Journal of Applied Physics*, 32 (1961) 542.
[6] A.J. McEvily, R.C. Boetner and T.L. Johnston in Proceedings of the 10th Sagamore AMR Conference, Syracuse (1963) 95.
[7] C.E. Feltner, *Journal of Applied Physics*, 38 (1967) 3576.
[8a] A.K. Head, *Philosophical Magazine*, 44 (1953) 925.
[8b] N.E. Frost, K.J. Marsh and L.P. Pook, *Metal Fatigue*, Oxford University Press, London (1974).
[8c] P.C. Paris and F. Erdogan, *Journal of Basic Engineering*, Transactions of ASME, D85 (1963) 528.
[9] ASTM Task Force E24.04, Report No. 1 (1975).
[10] S. Arad, J.C. Radon and L.E. Culver, *Journal of Mechanical Engineering Science*, 13 (1971) 75.
[11] S.J. Maddox, *International Journal of Fracture*, 11 (1975) 389–408.
[12] F. Erdogan and M. Ratwani, *International Journal of Fracture Mechanics*, 6 (1970) 379–392.
[13] S. Person, *Engineering Fracture Mechanics*, 4 (1972) 9.
[14] H.O. Chu in Fracture Toughness and Slow Stable Cracking, ASTM STP 595 (1974) 245.
[15] W. Elber in Damage Tolerance in Aircraft Structures, ASTM STP 486 (1971) 230.
[16] M. Katcher and M. Kaplan in Fracture Toughness and Slow Stable Cracking, ASTM STP 559 (1974) 264.
[17] J.C. Newman in Mechanics of Crack Growth, ASTM STP 590 (1976).
[18] J.M. Barson in Damage Tolerance in Aircraft Structures, ASTM STP 486 (1971) 1.
[19] J.M. Krafft, *Applied Materials Research*, 3 (1964).
[20] M.W. Proctor, Report F333, Portsmouth Polytechnic, UK (1977) in print.
[21] S. Arad, J.C. Radon and L. Culver, *Engineering Fracture Mechanics*, 6 (1974) 195.
[22] C.A.M. Branco, J.C. Radon and L.E. Culver, *Journal of Testing and Evaluation*, 3 (1975) 195.
[23] S.A. Sutton, *Engineering Fracture Mechanics*, 6 (1974) 587.
[24] E.H. Andrews, *Journal of Materials Science*, 9 (1974) 887.
[25a] J.C. Radon, *Journal of Macromolecular Science-Physics*, B14(4) (1977) 511.
[25b] J.C. Radon, *Journal of Applied Polymer Science*, 17 (1973) 3515.
[25c] J.C. Radon, *Journal of Polymer Science (Chemistry)* 13 (1975) 383.
[26] V.I. Aleshin, E.L. Aero, M.F. Lebedeva and E.V. Kuvshinski, *Mekhanika Kompozitnykh Mat.*, 1 (1979) 15.

[27] S. Arad, J.C. Radon and L.E. Culver, *Engineering Fracture Mechanics*, 4 (1972) 511.
[28] J.G. Williams in Advances in Polymer Science, Springer-Verlag, Berlin (1978) 67.
[29] S. Arad, Fracture Mechanics Analysis of Fatigue Crack Growth in Viscoelastic Solids, Ph.D. Thesis, University of London (1972).
[30] G. Urriolagoitia, Directional Stability of Cracks Under Biaxial Stress, Ph.D. Thesis, University of London (1976).
[31] A.A. Griffith, *Philosophical Transactions of the Royal Society of London*, A221 (1920) 163.
[32] J.L. Swedlow, *International Journal of Fracture Mechanics*, 1 (1965) 210–216.
[33] H. Riedel, *Journal of Mechanics and Physics of Solids*, 24 (1976) 277.
[34] J.J. Kibler and R. Roberts, *Journal Eng. Ind.*, 92 (1970) 727.
[35] Y. Ueda, K. Ikeda, T. Yao, M. Aoki, T. Yoshe and T. Shikakura in Fracture 1977, Waterloo, 2 (1977) 173.
[36] P.S. Leevers, J.C. Randon and L.E. Culver, *Polymer*, 17 (1967) 627.
[37a] F.A. Johnson and J.C. Radon, *Journal of Polymer Science (Chemistry)* 11 (1973) 1995.
[37b] E.J. Kramer, *Journal of Materials Science*, 14 (1978) 1381.
[38] J.C. Radon, *Colloid and Polymer Science*, 252 (1974) 117.
[39a] F.A. Johnson and J.C. Radon, *Journal of Polymer Science (Chemistry)* 13 (1975) 495.
[39b] H. Oberst and W. Retting, *Journal of Macromolecular Science-Physics*, B5(3) (1971) 559.
[40a] J.C. Radon and J.G. Williams in Fracture Mechanics, University Press, Virginia (1978) 279.
[40b] H.H. Kausch and J. Becht, *Rheologica Acta*, 9 (1970) 137.
[40c] A. Peterlin, *Journal of Macromolecular Science-Physics*, B8 (1975) 761.
[40d] Y.W. Mai and J.G. Williams, *Journal of Materials Science*, 14 (1979) 1933.
[41] S.N. Zhurkov, V.S. Kuksenko and A.I. Slutsker in Fracture 1969, Chapman and Hall, London (1969) 531.
[42a] V. Bouda, V. Zilvar and A.J. Staverman, *Journal of Polymer Science* (Polymer Physics) 14 (1976) 2313.
[42b] M. Ishikawa, I. Narisawa and H. Ogawa, *Journal of Polymer Science* (Polymer Physics) 15 (1977) 1791.
[43] A.S. Argon, J.G. Hannoosh and M.M. Salama in Fracture 1977, Waterloo, 1 (1977) 445.
[44] R.W. Hertzberg, M.D. Skibo and J.A. Manson, *Journal of Applied Polymer Science*, 17 (1973) 987.
[45] J.R. Rice in ASTM STP 415 (1967) 247.
[46] S.R. Joshi and J. Shewchuck, *Journal of Experimental Mechanics*, 10 (1970) 529.
[47] L.P. Pook and R. Holmes in Proceedings of the International Conference on Fatigue Testing and Design, London (1976).
[48] K.J. Miller, *Journal of Materials Science*, 11 (1977) 432.
[49] J.C. Radon, P.S. Leevers and L.E. Culver, *Journal of Experimental Mechanics*, 17 (1977) 228.
[50] P.S. Leevers, L.E. Culver and J.C. Radon, *Engineering Fracture Mechanics*, 11 (1979) 487.
[51] J.P. Elinck, J.C. Bauwens and G. Homes, *International Journal of Fracture Mechanics*, 7 (1971) 277–287.
[52] F.J. Pitioniak, A.F. Grandt, L.T. Montuilli and P.F. Packman, *Engineering Fracture Mechanics*, 6 (1974) 663.

RÉSUMÉ

On décrit les diverses lois de propagation des fissures de fatigue appliquées à l'étude des polymères. En considérant la distribution de champs de contrainte a l'extremité de la fissure, on est conduit á appliquer la mécanique de la rupture. On montre qu'une relation simplifiée de la forme $da/dN = F\phi^a$, où ϕ est une fonction de K_{IC}, K_{max}, K_{min} et K_{TH} apparait être une expression convenable pour la croissance cyclique d'une fissure. L'effet de la contrainte moyenne est plus complexe que dans le domaine des métaux et la composante de compression du cycle de contraintes peut différer la croissance de la fissure. Des essais cycliques en traction exécutés sur du PMMA et du PVC dépendent de ΔK et de la valeur moyenne K_m. La valeur de seuil K_{TH} est également influencée par K_m mais un comportement plus complexe associé aux effets de vitesses de déformation peut être observé aux effets de vitesees de déformation peut être observé. D'autres differénces, telles que la position des points de transition supérieurs et inférieurs ainsi que les changements de vitesse de croissance avec la fréquence ont été notées. L'effet d'une mise en charge cyclique biaxiale d'un PMMA ou d'un PVC sous forme de plaque est comparé et on met en avant certaines des différences observées. Les résultats disponibles jusqu'ici indiquent un effet modéré de la courbure de la fissure sur sa propagation. Cependant, on montre que si une biaxialité croissante peut retarder d'une manière substancielle la croissance d'une fissure dans du PMMA, aucun effet de ce genre n a été enregistré dans le cas d'un PVC. Enfin, on montre que pour des niveaux de contrainte très élevés (région III) la croissance cyclique d'une fissure consiste en deux modes de propagation, à savoir une propagation purement cyclique accompagnée d'une croissance lente. A des niveaux de contrainte plus faible, la phase de croissance lente disparait et la propagation de la fissure s'effectue en fatigue pure (région II). Dans la région I, la propagation est très lente sans que se présente la correspondance usuelle entre les cycles et les striures. Les résultats récemment obtenus sur des plastiques renforcés de verre (GRP) sont également présentés et les différences en sont mises en évidence.