Fatigue crack propagation in unplasticized poly(vinyl chloride): 1. Effect of mean stress

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The effect of mean stress on fatigue crack growth in an unplasticized poly(vinyl chloride) (uPVC) pipe material was investigated using single-edge-notched (SEN) specimens and the results analysed in terms of fracture mechanics. Three series of experiments were conducted. In the first the fatigue crack growth rates da/dN were collected as a function of the applied stress intensity factor range $\Delta K(=K_{max}-K_{min})$ by keeping the stress ratio $R(=K_{min}/K_{max})$ constant. In the second series of experiments da/dN were obtained as a function of ΔK when the mean stress intensity factor K_m was constant; and in the third series ΔK was maintained constant, da/dN were obtained with increasing K_m. All experimental results were obtained and processed using computerized data acquisition systems. When the data were analysed in terms of the conventional log da/dN versus log ΔK and log K_m plots the effects of R and K_m on fatigue crack growth rate were insignificant. However, there was a strong dependence of da/dN on R when analysed in terms of Williams' line zone model for fatigue crack growth.

(Keywords: poly(vinyl chloride); fatigue crack growth; mean stress; crack closure; plane strain–plane stress transition)

INTRODUCTION

Unplasticized poly(vinyl chloride) (uPVC) is increasingly finding use as a material for pressure pipes in water reticulation. In such an application, the uPVC is often subjected to a cyclic pressure or fatigue environment in which a variable pressure is superimposed on a static mean pressure. Although there are a number of studies on fatigue of uPVC[1–8], most of these are based on a cyclic pressure or stress which varies from zero to some controlled maximum value. This could of course be considered as a case in which the mean stress is half the peak stress and the stress ratio is zero. However, this mean stress variable is rarely included in the analysis of the fatigue data. In this paper we report some systematic work on fatigue crack propagation rates in a grade of a uPVC pipe material which have been measured as a function of both stress amplitude and mean stress (stress ratio).

Most modern fatigue studies on polymers have been based on a fracture mechanics approach which correlates the fatigue crack growth rate per cycle, da/dN, with the applied crack tip stress intensity factor range, ΔK $(=K_{max}-K_{min})$, using the Paris power law equations[9,10]:

$$\frac{da}{dN}=A(\Delta K)^m \tag{1}$$

where A and m are constants and K is related to the applied stress σ and the crack length a by:

$$K=\sigma Y\sqrt{(\pi a)} \tag{2}$$

and Y is a geometry correction factor. Equation (1) has been found to adequately describe the fatigue crack growth rate in uPVC pipe materials[2–5,8]. Since the mean stress intensity factor K_m is related to the stress ratio R $(=K_{min}/K_{max})$ by

$$K_m=\frac{\Delta K}{2}\left(\frac{1+R}{1-R}\right) \tag{3}$$

the effect of mean stress on da/dN can be easily studied by comparing the crack growth rates at the same ΔK but varying R. Such an effect is also reflected by the different magnitudes of the constants A and m in equation (1) as the mean stress or R ratio changes.

Attempts to relate explicitly da/dN with mean stress intensity factor K_m have been made by Arad *et al.*[11–13] and Radon *et al.*[14] This follows the observation that the mean stress effect on fatigue crack growth data in some polymers[11–15] can be rationalized by using either a parameter λ $(=K_{max}^2-K_{min}^2)$ or the cycle potential energy release rate ΔG, which are equivalent, since $\Delta G=\lambda/E$ and E is the Young's modulus, i.e.:

$$\frac{da}{dN}=B\lambda^n=(BE^n)(\Delta G)^n \tag{4}$$

where (B,n) are constants. The physical basis of using ΔG to characterize fatigue crack growth has been given earlier by Mai[16,17] in terms of the quasi-static fracture energy approach[10,18,19]. Because $\lambda=2\Delta KK_m$ we can rewrite equation (4) as:

$$\frac{da}{dN}=B^*(\Delta K\cdot K_m)^n \tag{5}$$

with $B^*=2B$. When the stress ratio $R=0$, $K_m=\Delta K/2$ and

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equation (4) is reduced to equation (1) so that $m = 2n$ and $A = B^*/2^n$. There have been doubts as to whether equation (4) and hence equation (5) can be used to uniquely describe the fatigue crack growth rates in all polymers irrespective of the mean stress intensity factor[20]. Polycarbonate and polystyrene are two examples[20,21] for which (B,n) in equation (4) are dependent on mean stress. However, if equations (4) and (5) can indeed normalize all fatigue crack growth rates into a single master curve, then the effect of increasing stress intensity K_m is to increase crack growth rate da/dN for the same ΔK. Conversely, if K_m is kept constant increasing ΔK increases da/dN.

Recently, a modified form of equation (4) has been suggested which is claimed to give a unique fatigue crack propagation curve independent of mean stress and material type. This is given by[22]:

$$\frac{da}{dN} = \frac{\gamma(\Delta G)^{\beta}}{G_c - G_{max}} \tag{6}$$

where γ and β are constants and G_c is the fracture toughness. Only data for aluminium alloys and PMMA are given to support this so-called unified equation (6). Hence it would be useful to see if it can be extended to other polymeric materials such as uPVC studied here.

It is pointed out by Williams and coworkers[20,23–26] that whilst the Paris power law equation and its various versions, i.e. equations (1), (4)–(6), are most convenient for use in engineering design against fatigue failure, they do not offer any physical insight into the fatigue fracture processes taking place at the crack tip. In an attempt to relate fatigue in polymers to fracture parameters such as crack tip stress σ_c and crack opening displacement δ_c Williams[23] has developed from a line zone model the following fatigue crack growth equation:

$$\frac{da}{dN} = \frac{\pi}{8(1-\alpha)^2\sigma_c^2}(K_{max}^2 - \alpha K_c^2) \tag{7}$$

where α is a stress reduction factor in the fatigue damaged zone. This equation has been shown to give a good description of fatigue crack growth rates in poly(methyl methacrylate) (PMMA), nylon 66, polycarbonate (PC) and GPS[20,23]. To account for R-ratio effects and crack closure associated with residual compressive stress generated at the crack tip due to unloading, equation (7) has been modified to[24–26]:

$$\frac{da}{dN} = \frac{\pi}{8(1-\alpha)^2\sigma_c^2}\{F(R)[K_{max}^2 - K_{cl}^2] - \alpha K_c^2\} \tag{8}$$

where $F(R)$ depends on the crack tip fatigue mechanism, i.e. crazing and/or shear yielding, and K_{cl} is that stress intensity factor due to closure stresses. Notice that the fatigue model of Williams is only applicable to steady state continuous crack growth[27]. Equation (8) has already been applied to account for R-ratio effects in PMMA and poly(vinyl chloride) (PVC) by Williams and Osorio[24,26].

Previous work by Mills and Walker[2] showed that in extruded PVC for given ΔK, da/dN initially decreased when R was increased from 0.05 to 0.20 and it subsequently increased for larger R-ratios of 0.33 and 0.50. However, the changes in da/dN are insignificant. Hertzberg and Manson[9] also reported a decrease in da/dN with R-ratio from 0.1 to 0.5 for extruded sheets but moulded PVC samples of different molecular weights showed an increase in da/dN over the same range of R-ratio. It appears, therefore, that manufacturing processes, either compression moulding or extrusion, have produced opposite mean stress effects on fatigue crack growth rates in PVC. A detailed study on effects of processing methods seems desirable. In this paper we have focused our efforts on an extruded uPVC pipe material. The fatigue crack growth laws given by equations (1), (5), (6) and (8) are used to investigate the effects of R-ratio and K_m on da/dN.

EXPERIMENTAL

The fatigue specimens were cut from sheets prepared by flattening uPVC pipe samples. The pipes used were taken from normal production 150 and 100 mm class 12 (wall thickness ~9 and ~6.3 mm respectively). The pipes were made from unmodified, lead stabilized PVC resin with a K value of 67, corresponding to an approximate average molecular weight of 170 000. The pipes showed no attack on immersion in methylene chloride for 15 min at 20°C and were thus considered well processed. The pipes were slit and warmed in an oven at 100–105°C for 15 min before they were opened out and pressed between metal plates to produce a flat sheet. Three series of fatigue crack propagation experiments were conducted in the ICI Research Laboratory and the Sydney University Materials Laboratory. Single-edge-notched (SEN) samples, $50 \times 170\ mm^2$ (ICI) and $70 \times 210\ mm^2$ (Sydney University) in sizes, were cut from the flattened sheets so that the crack would propagate in the pipe extrusion direction.

For the two series of fatigue experiments performed at Sydney University a computerized data acquisition system based on a screen printed electrically conductive surface grid was used to monitor fatigue crack growth on a computer controlled Shimadzu hydraulic servo-pulser. The details and accuracies of the method have been described in our previous work[28,29]. In one series the R-ratio was varied from 0.03 to 0.5 and the frequencies chosen were 5 Hz for the 150 mm class 12 and 10 Hz for the 100 mm class 12 pipes. In another series ΔK was kept constant and K_m was increased. All experiments were conducted at 22 ± 1°C and ~55% relative humidity. The cyclic load waveform was sinusoidal. For the fatigue crack propagation experiments carried out in the ICI Reseach Laboratory crack growth was monitored by a conducting gauge on the side of the sample (Krak-gauge, KG-B25, made by TTI Division, Hartrun Corporation). The surfaces of the SEN samples were roughened using 40 grit wet and dry paper and were then carefully cleansed and neutralized before the Krak-gauges were fixed to the samples using an epoxy adhesive. A number of fatigue tests were stopped during crack propagation and careful inspection of the sample indicated that the crack in the Krak-gauge faithfully followed the position of the crack in the specimen. The fatigue data were collected on a computer controlled servo hydraulic Instron, model 8031. The output from the Krak-gauge was fed to one of the strain channels of the Instron where it could be calibrated to read crack length directly. The computer logged data at preset increments of crack growth, usually

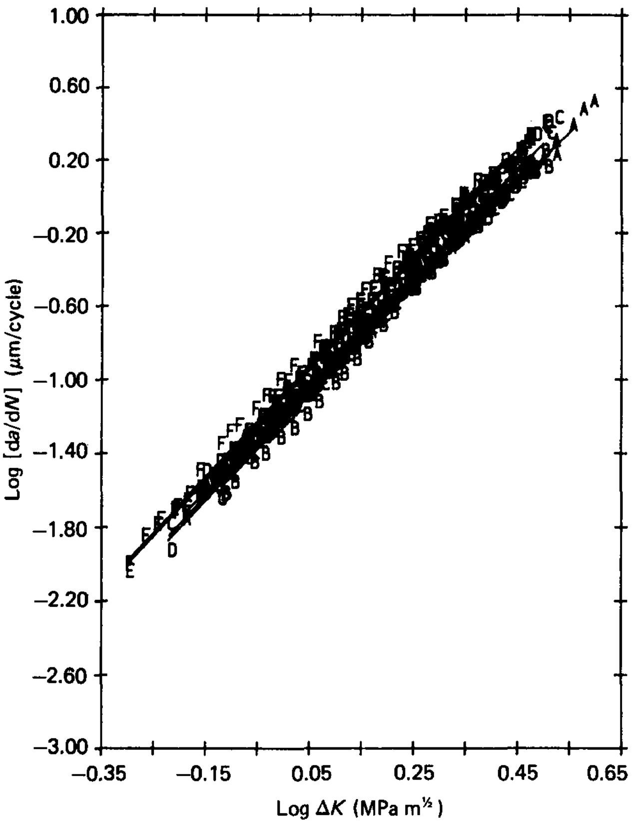


Figure 1 Log–log plot of da/dN versus ΔK for 150 mm class 12 uPVC pipe at 5 Hz. R-ratio: A, 0.03; B, 0.1; C, 0.2; D, 0.3; E, 0.4; F, 0.5. Results of three specimens at each R-ratio are given

0.5 mm. The machine was operated on closed loop load control with the computer being programmed to update the mean load and load amplitude after each logging point. This updating of the control parameters was designed to maintain K_m constant while incrementing ΔK or vice versa.

All the ICI tests were conducted on the 100 mm class 12 pipes with a frequency of 5 Hz and a sinusoidal load waveform. The tests were performed in a constant temperature room maintained at 22.5°C and a nominal relative humidity of 50%. Two series of experiments were conducted. The first maintained ΔK constant and incremented K_m; the second maintained K_m constant and increased ΔK.

RESULTS

Figures 1 and *2* show the fatigue crack growth data plotted against ΔK for the 150 and 100 mm class 12 pipes according to equation (1). There is good agreement with the Paris equation and values of the constants A and m are given in *Table 1*. Although there is a slight increase in da/dN (a maximum of twofold increase) with increasing R, the mean stress effect is considered insignificant. Despite the difference in the pipe thickness these results show that the two frequencies of 5 and 10 Hz do not produce any appreciable effects on da/dN. The ICI fatigue results on the 100 mm class 12 pipes at 5 Hz are shown in *Figure 3* where da/dN is plotted against ΔK for three values of constant K_m (=0.7, 1.5 and 2.5 MPa m$^{1/2}$). The data seem to be best described by a sigmoidal curve. However, for $10^{-2} < da/dN < 1$(μm/cycle), the results are reasonably described by equation (5). That is, at constant K_m, da/dN increases directly with ΔK; but K_m has relatively little effect on da/dN at constant ΔK (although the data for $K_m = 0.7$ MPa m$^{1/2}$ are somewhat lower than those of the other two K_m values). This latter observation is also confirmed by the constant ΔK incremented K_m experimental results obtained in Sydney and shown in *Figure 4*. Indeed when these data are superimposed in *Figure 2* (data points represented by H and I, E, F and G as well as J) they fall within the scatter band of the experimental results—the higher da/dN data correspond to higher R-ratio—and this further confirms the consistency of the Sydney University data and the accuracy of their experimental technique. Similar experiments with constant ΔK and increasing K_m were conducted on smaller size samples and the data acquired

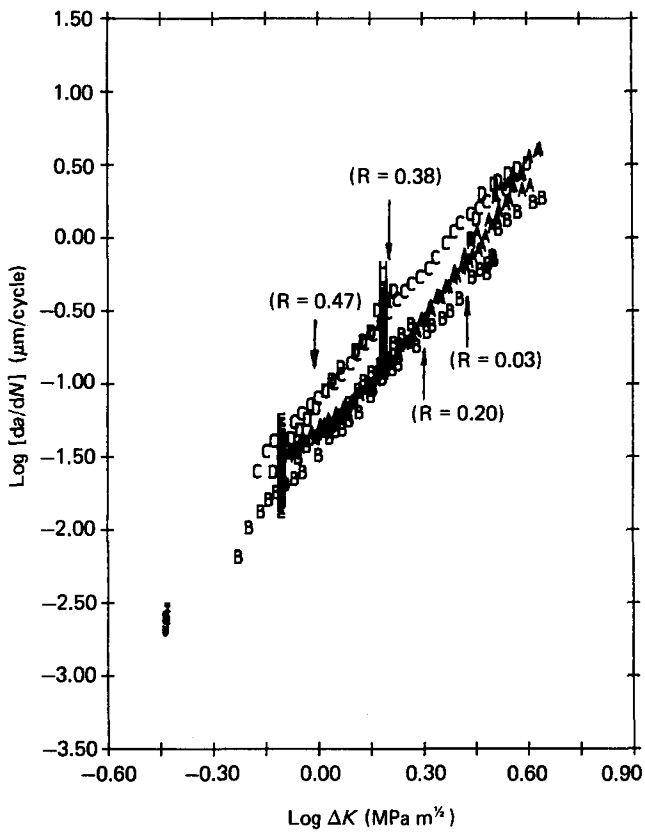


Figure 2 Log–log plot of da/dN versus ΔK for 100 mm class 12 uPVC pipe at 10 Hz. R-ratio: A, 0.03; B, 0.2; C, 0.38; D, 0.47. (Arrows indicate plane strain–plane stress transitions, R-ratio in parentheses)

Table 1 Comparison of the Paris power law equation constants A and m for uPVC

Pipe material	$\log_{10} A$ (μm/cycle)	m	R	Frequency (Hz)
100 mm class 12	−1.40	2.95	0.03	10
	−1.44	2.67	0.20	10
	−1.10	2.74	0.38	10
	−1.12	2.86	0.47	10
150 mm class 12	−1.24	2.94	0.10	5
	−1.21	3.00	0.20	5
	−1.19	2.98	0.30	5
	−1.10	3.00	0.40	5
	−1.08	3.01	0.50	5

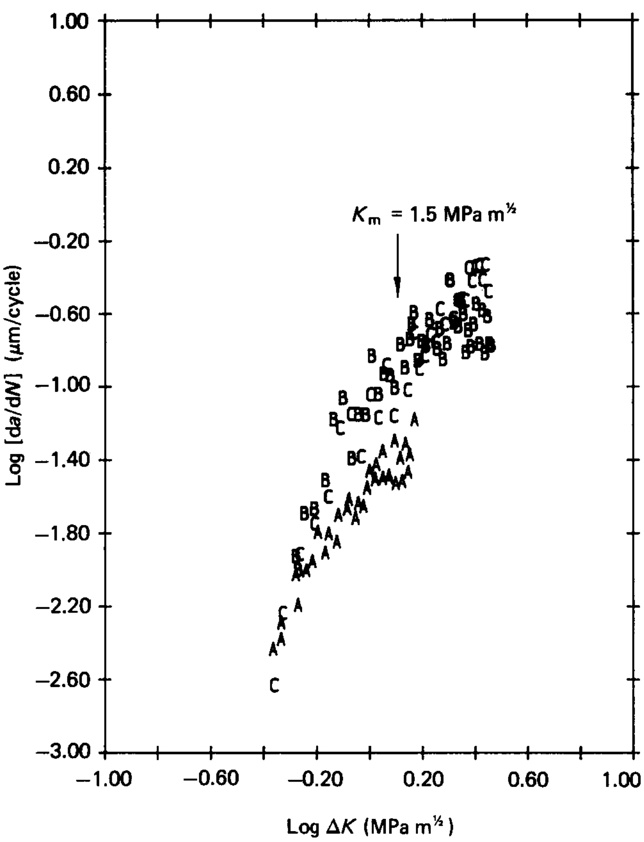


Figure 3 Log–log plots of da/dN versus ΔK for 100 mm class 12 uPVC pipe at three constant mean stress intensity factors: A, $K_m = 0.7$ MPa $m^{1/2}$; B, $K_m = 1.5$ MPa $m^{1/2}$; C, $K_m = 2.5$ MPa $m^{1/2}$

by the Krak-gauge method in the ICI laboratory. Unfortunately, their data shows a marked decrease in da/dN as K_m is increased above ~ 1 MPa $m^{1/2}$, which is independent of the level of applied ΔK. These results are contradictory to those shown in *Figure 4*. Experiments were repeated at Sydney on smaller size samples using both linear and logarithmically increasing K_m techniques and ensuring no overloading effects. The data are consistent with those already shown in *Figure 4*. Since there are many small crazes around the crack tip region in this series of experiments it is suspected that the possible reason is that the Krak-gauge method is not as accurate as the surface conductive grid technique. The Krak-gauge might have overestimated the crack length because of the presence of the crazes. However, it is visually confirmed that the crazes do not break the conducting bars and only the main crack does. Another more likely explanation is that due to overloading causing fatigue crack growth retardation in the ICI data.

DISCUSSION

Normalizing mean stress effects on fatigue crack growth

One of the objectives of this work was to investigate the consistency of fatigue crack growth data collected in the two different laboratories for an identical uPVC pipe material. There have been very few studies of this nature reported in the literature. Another objective was to see if equations (5) and (6) can be used to normalize the fatigue data independent of mean stress. In *Figure 5* the data shown in *Figures 1, 3* and *4* together with others for which

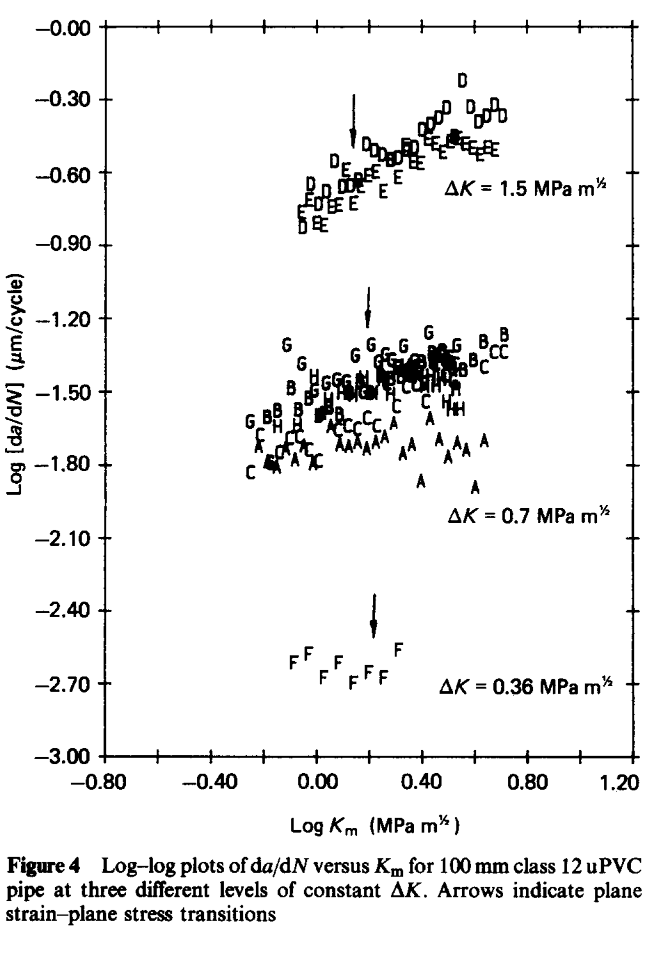


Figure 4 Log–log plots of da/dN versus K_m for 100 mm class 12 uPVC pipe at three different levels of constant ΔK. Arrows indicate plane strain–plane stress transitions

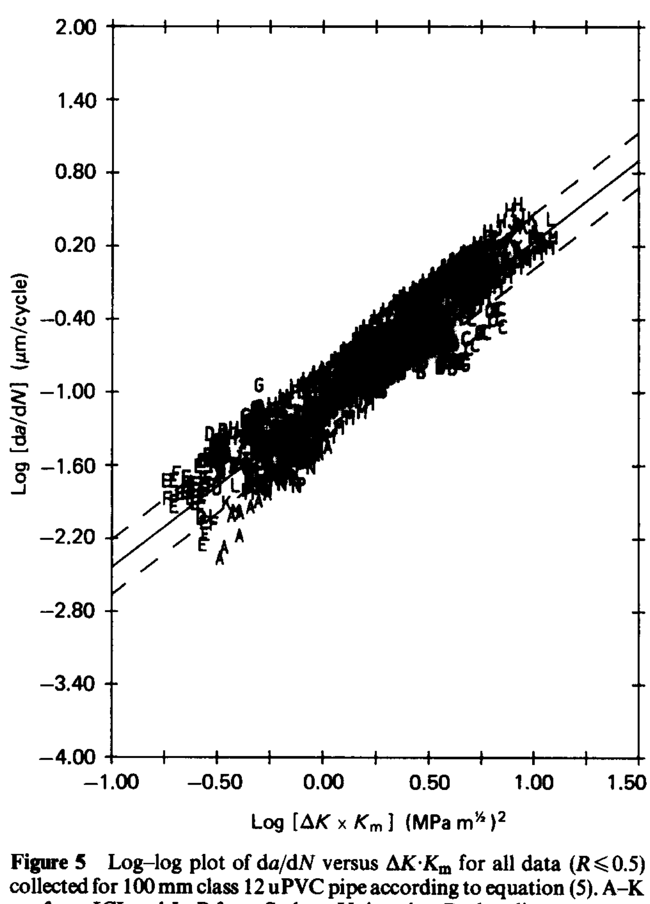


Figure 5 Log–log plot of da/dN versus $\Delta K \cdot K_m$ for all data ($R \leqslant 0.5$) collected for 100 mm class 12 uPVC pipe according to equation (5). A–K are from ICI and L–P from Sydney University. Broken lines represent ± 1 standard deviation from mean line

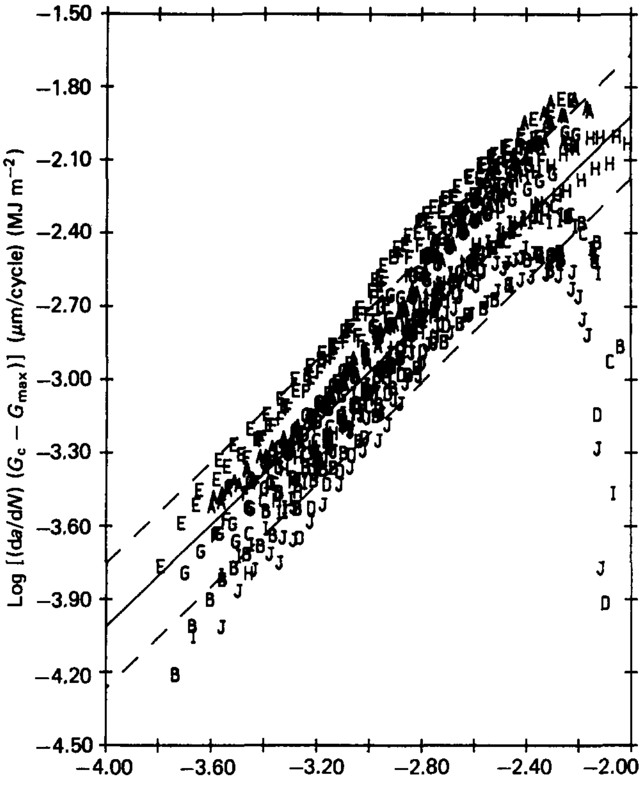


Figure 6 Log–log plot of $(G_c - G_{max})$ da/dN versus ΔG according to equation (6). $G_c = 8.93$ kJ m^{-2} (ref. 8) at fracture instability

the mean stress ratio $R \leqslant 0.5$ are all reanalysed and plotted in accordance with equation (5). It is seen that the experimental results obtained in the two laboratories agree well with each other. For $10^{-2} < da/dN < 10$ (μm/cycle) the mean line of the data can be fitted to equation (5) with $\log_{10} B^* = -1.10\,\mu$m/cycle and $n = 1.34$. The difference in the upper and lower bound da/dN is about three times and such a small scatter band is acceptable in fatigue testing. Thus, a 'master' curve of fatigue crack growth data is obtained if $R \leqslant 0.5$. However, overloading effects and stress ratios larger than 0.5 both cause the da/dN data to deviate from the master plot of *Figure 5*.

Analysis of the fatigue crack growth data in terms of equation (6) is given in *Figure 6*. It is seen that no master fatigue curve can be obtained and the remarkable drop in da/dN for large ΔG is due to the fact that $G_c - G_{max}$ is decreasing at a rate faster than that at which da/dN is increasing. Here G_c is the value at fracture instability for the mixed mode of plane strain–plane stress and is equal to 8.93 kJ m^{-2} (ref. 8). The coefficient of correlation is only 0.85 and the standard deviation is 0.26, which is rather large. If G_c (≈ 3.65 kJ m^{-2}) for plane strain is used the correlation of the data is even worse. Consequently, equation (6) is not capable of fully accounting for mean stress effects on da/dN in this uPVC material. Indeed analysis of additional fatigue data including polystyrene[20], uPVC, PMMA and aluminium alloys[22] according to equation (6) does not provide a master fatigue curve for *all* materials independent of mean stress.

Plane strain–plane stress transitions in fatigue crack growth

A typical failed surface of a fatigued sample shows the presence of a fracture mode transition. Initially, fatigue fracture prevails under essentially plane strain conditions; but as the crack advances there is a transition to plane stress. Final fracture fails under essentially plane stress conditions with a reduction in thickness and the formation of a double shear across the thickness. To maintain plane strain conditions in a fracture mechanics test, the minimum thickness b of the sample must satisfy the ASTM-E399 recommendation:

$$b \geqslant 2.5(K_c/\sigma_y)^2 \tag{9}$$

where σ_y is the yield strength. This condition essentially requires the plastic zone size to be much smaller than the thickness. In a fatigue test the plastic zone size is controlled by the maximum stress intensity factor, K_{max}. Rearranging equation (9) we can obtain K_{max} at which plane strain–plane stress transition occurs, i.e.

$$K_{max} = \left(\frac{b}{2.5}\right)^{1/2} \sigma_y \tag{10}$$

Noting that σ_y decreases with time whilst under load and given the pipe wall thickness it is easy to calculate the K_{max} values for the fracture transitions to occur. These are indicated by arrows in *Figures 2* and *3* with plane stress to the right and plane strain to the left. In *Figure 3* where the experiments were conducted with K_m constant and ΔK increasing the data for $K_m = 0.70$ MPa m$^{1/2}$ are all in plane strain and those for $K_m = 2.50$ MPa m$^{1/2}$ are all in plane stress. In metals the rate of fatigue crack growth is observed to decrease in plane stress compared to plane strain at the same ΔK because of the larger plastic zone size in the former condition[30]. It is, therefore, expected that in polymers the fatigue crack will also decelerate and the data obtained after the transition point will no longer be described by the same fatigue crack growth equation for the plane strain data. However, as shown in *Figures 2* and *3*, there is no such distinction between plane strain and plane stress. All the data can be adequately described by one single equation, i.e. equations (1) and (5), respectively.

There is progressive surface roughening and increasing density of crazes as the plane stress condition is approached. In *Figure 7*, microtomed sections showing the region below the fatigue crack are shown viewed in a transmission optical microscope. At a low value of K_m, less than 1 MPa m$^{1/2}$, and a ΔK of 0.36 M Pa m$^{1/2}$ (i.e. plane strain) little crazing can be seen below the main crack (*Figure 7a*). However, at a K_m of 2.5 MPa m$^{1/2}$ and $\Delta K = 0.36$ MPa m$^{1/2}$ (i.e. plane stress) (*Figure 7b*), a large deformation zone containing microcrazing and cracking surrounds the propagating fatigue crack. This deformation zone increases as K_m is increased to 3 MPa m$^{1/2}$ (*Figure 7c*). In *Figure 7a*, the fatigue crack is quite planar whereas at higher K_m values (*Figures 7b* and *7c*) the plane of the fatigue crack begins to deviate, jumping between the microcracks and crazes present in the plastic zone in front of the crack. Scanning electron micrographs shown in *Figure 8* also support this conclusion. There is increasing roughness of the fracture surfaces as K_m is increased. It is interesting to note at this stage that in none of the specimens tested in this work was there any evidence of discontinuous crack growth which has been the focus of numerous other studies of fatigue in

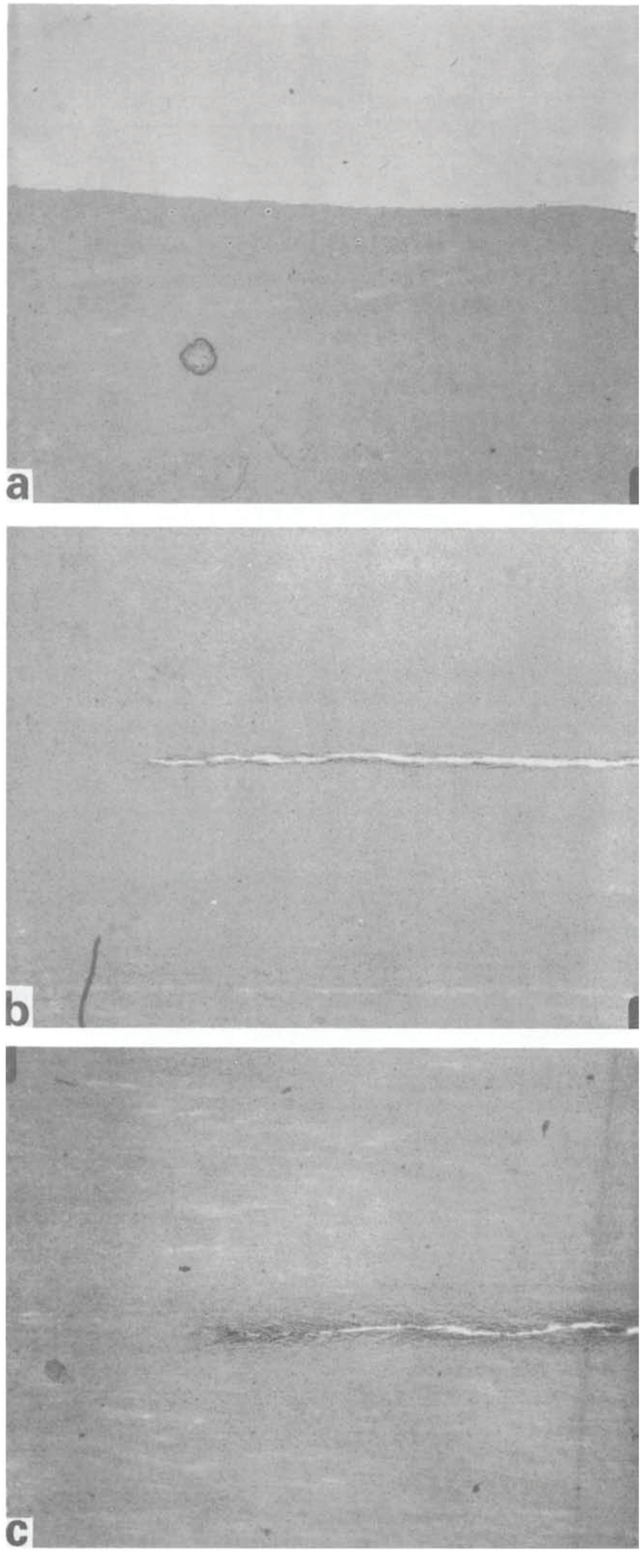


Figure 7 (a) Side view of a fractured surface of a uPVC sample at $K_m = 0.9$ MPa m$^{1/2}$ and $\Delta K = 0.36$ MPa m$^{1/2}$. No crazes and microcracks are seen. (b) A fatigue crack in a PVC pipe at $K_m = 2.5$ MPa m$^{1/2}$ and $\Delta K = 0.36$ MPa m$^{1/2}$ showing crazing and microcracks contiguous to the fracture surfaces. (c) A fatigue crack in a uPVC pipe at $K_m = 3$ MPa m$^{1/2}$ and $\Delta K = 1.5$ MPa m$^{1/2}$ showing more predominant crazes and microcracks contiguous to the fracture surfaces and ahead of the visible crack tip

PVC[3,31,33]. Rimnac, Hertzberg and Manson[32] as well as Mai and Kerr[8] have also noted that fatigue striations are absent from the fracture surface of fatigued PVC if the molecular weight used for pipe extrusion is larger than 110000 (i.e. resin K value >65).

Analysis of fracture parameters for fatigue crack growth

To obtain fracture parameters relevant for fatigue crack growth in uPVC we have replotted the data of *Figure 1* in *Figure 9* in accordance with equation (8) proposed by Williams and coworkers[24-26]. As expected, there is good linearity in this plot but there is now a strong R-dependence for da/dN because K_{max} is used. Only plane strain fatigue crack growth data are shown in *Figure 9*. If the plane stress data are also included there are abrupt transitions giving straight lines of higher slopes (S) and larger intercepts (I) for each R-ratio.

According to equation (8) S and I are related by:

$$S = \frac{\pi F(R)}{8(1-\alpha)^2 \sigma_c^2} \tag{11}$$

and

$$I = K_{cl}^2 + \frac{\alpha K_c^2}{F(R)} \tag{12}$$

since $F(0) = 1$ so that $F(R) = S(R)/S(0)$. K_{cl}^2 can be evaluated independently of R by:

$$SI = \frac{\pi \alpha K_c^2}{8(1-\alpha)^2 \sigma_c^2} + K_{cl}^2 S \tag{13}$$

Figures 10a and *10b* show S plotted against R and SI against S, respectively. From these results we can obtain $K_{cl} = 0.30$ MPa m$^{1/2}$, $\sqrt{\alpha} K_c = 0.346$ MPa m$^{1/2}$, $\alpha = 0.012$ since $K_c = 3.2$ MPa m$^{1/2}$ (ref. 8) for plane strain and $\sigma_c = 2.20$ GPa. The values of these fracture parameters are similar to those obtained by Williams and Osorio[26] for other similar pipe-grade PVC materials. It is possible to estimate from the threshold value αK_c^2 ($= \alpha \sigma_c E \delta_c$) that the critical crack tip opening displacement δ_c is equal to 1.6 μm since $E = 2.8$ GPa. Also, the dependence of S on R shown in *Figure 10a* indicates that the functional form of $(F)R$ is $(1-R)^2$ for $\alpha \ll 1.0$ which, according to Williams' model, suggests that two deformation mechanisms occur during fatigue crack growth[24]. One is a crazing mechanism at the tip of the line zone and the other is either a shear band or multiple crazing mechanism at the advancing crack tip. Such mechanisms have indeed been observed by the authors in this uPVC material.

The intercept of *Figure 9* given by equation (12) may be regarded as the threshold K_{max} value below which da/dN is zero. Since the fatigue threshold stress intensity range ΔK_{th} is $K_{max}(1-R)$, *Table 2* shows ΔK_{th} as a function of R and there is a very moderate decrease with increasing stress ratio. Further results on fatigue threshold and near-threshold crack growth are given in part 2 of this paper[34].

CONCLUSION

The effect of mean stress on fatigue crack growth rates in a uPVC pipe material has been investigated. It is observed that both R and K_m do not produce any significant effects on da/dN in the usual increasing ΔK constant R and constant K_m experiments. Within experimental scatter these results can be fitted to equation (5) (i.e. da/d$N \propto (\Delta K \cdot K_m)^n$), indicating that a master fatigue crack growth curve might exist if $R \leqslant 0.5$.

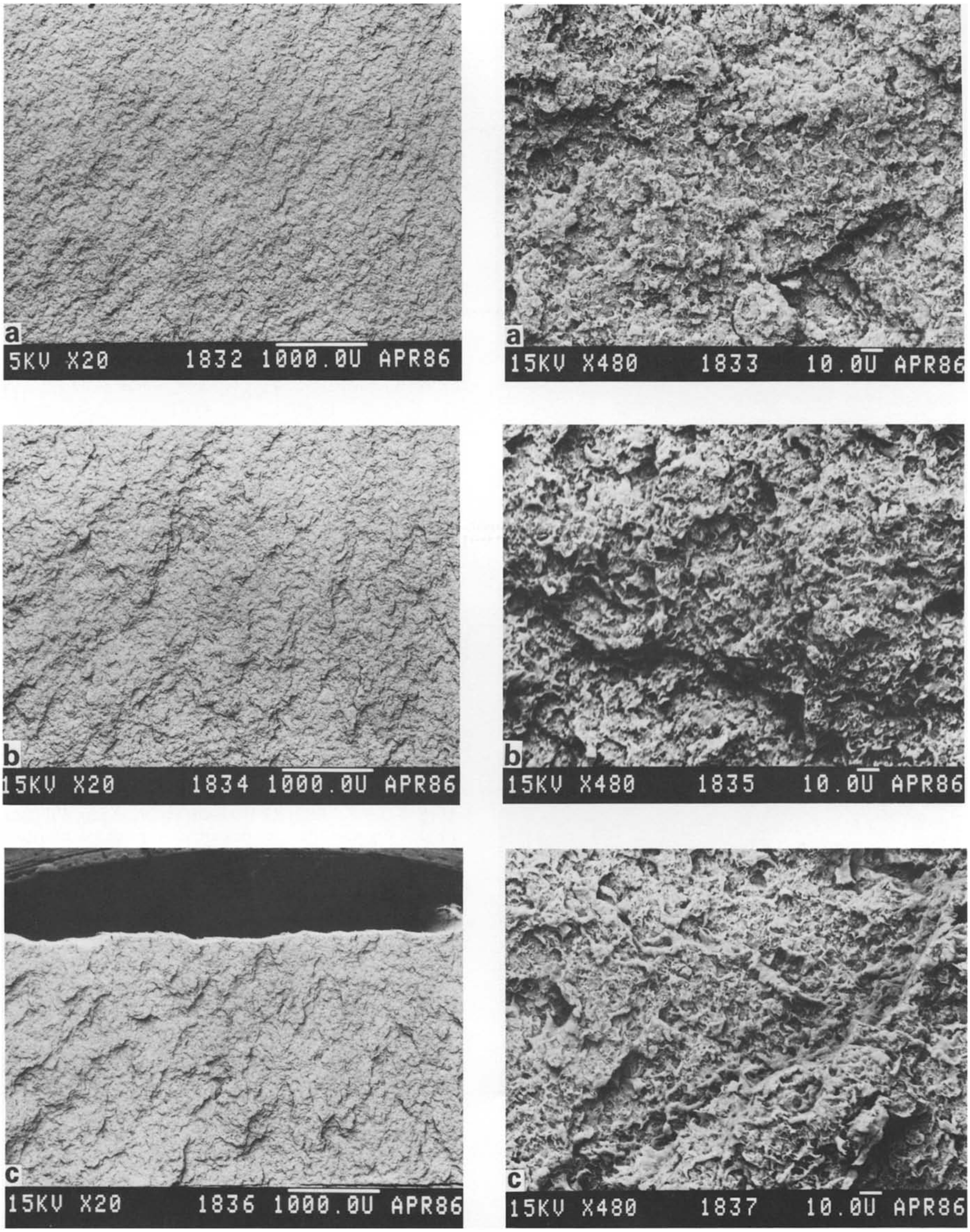
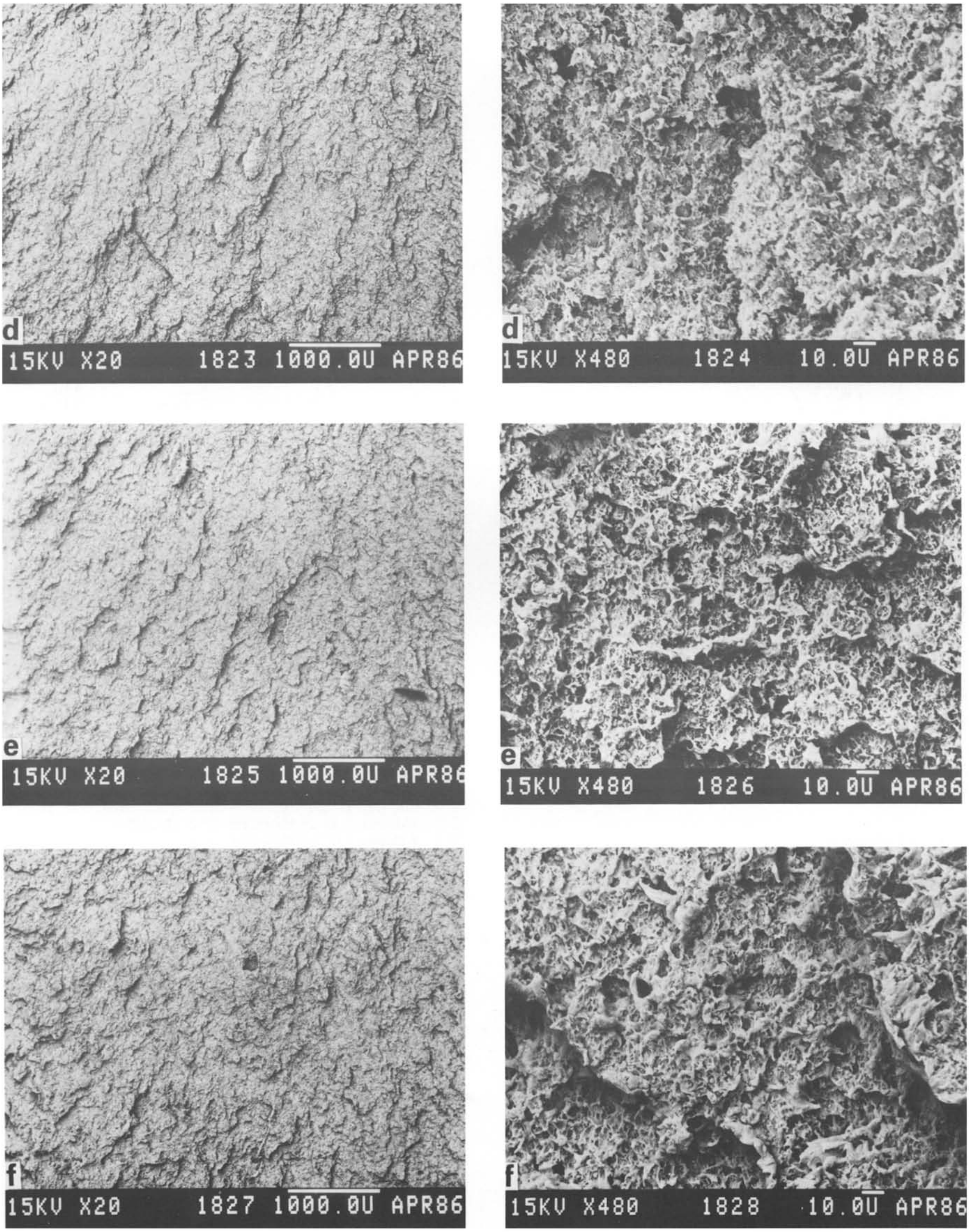


Figure 8 SEM fracture surfaces of a uPVC pipe showing increasing roughness on fracture plane with (a) $K_m = 0.75$, $\Delta K = 1$; (b) $K_m = 1.50$, $\Delta K = 2$; (c) $K_m = 2.25$, $\Delta K = 3$; (d) $K_m = 1.39$, $\Delta K = 1$; (e) $K_m = 2.77$, $\Delta K = 2$; (f) $K_m = 4.16$, $\Delta K = 3$ (all units in $MPa\,m^{1/2}$)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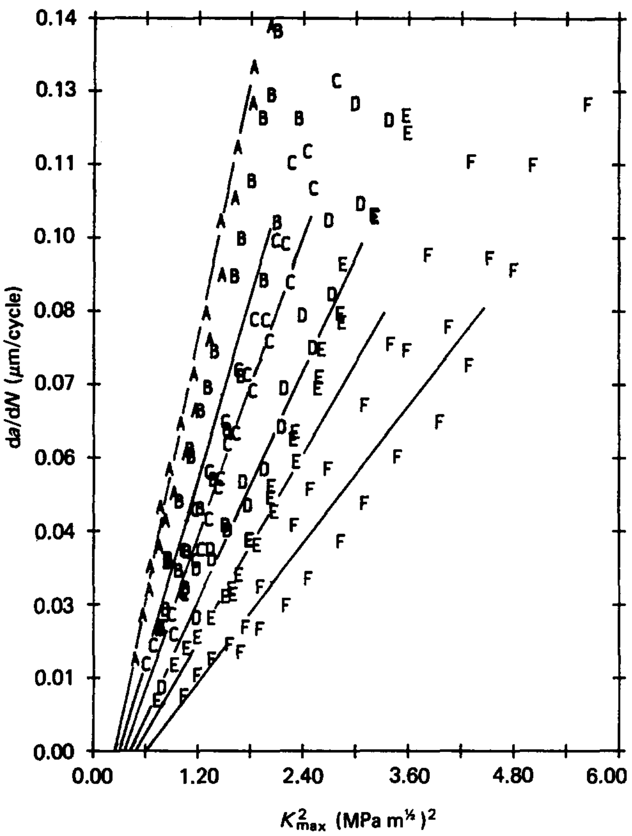

Figure 9 Plot of da/dN against K_{max}^2 in accordance with equation (8). R-ratios: A, 0.03; B, 0.1; C, 0.2; D, 0.3; E, 0.4; F, 0.5

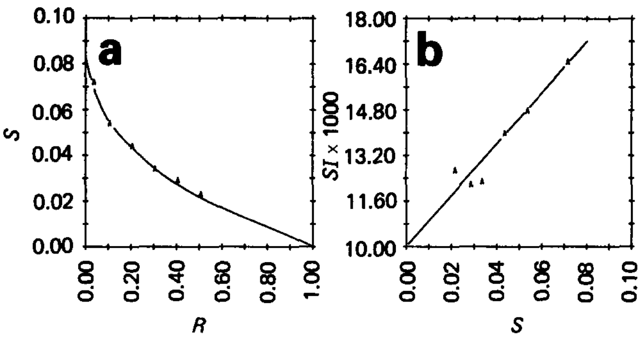

Figure 10 (a) Plot of slope S against stress ratio R. (b) Plot of SI against S

Table 2 Fracture parameters at fatigue threshold based on Williams[24] model

R	K_{cl} (MPa m$^{1/2}$)	K_{max} (MPa m$^{1/2}$)	ΔK_{th} (MPa m$^{1/2}$)
0.03	0.30	0.48	0.47
0.10	0.30	0.52	0.47
0.20	0.30	0.57	0.46
0.30	0.30	0.62	0.43
0.40	0.30	0.67	0.40
0.50	0.30	0.74	0.37

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