An examination of the role of flaw size and material toughness in the brittle fracture of polyethylene pipes

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At low temperatures and hoop stresses, polyethylene pipes fail by the time-dependent propagation of a crack. These brittle, fissure-like failures have been observed to initiate from adventitious flaws, and the concepts and methods of fracture mechanics indicate that flaw size should determine stress rupture lifetime. A number of controlled model experiments have therefore been undertaken to assess the influence of flaw size and material toughness on the stress rupture lifetimes of polyethylene pipes. To two different pipe grade polyethylene resins (one shorter, one longer lifetime resin) flaws of varying sizes have been added. For the shorter lifetime resin small flaws were, in addition, purposely excluded by the use of fine melt filtration techniques. Pipes containing added flaws or pipes where flaws were excluded were then stress rupture tested under those conditions designed to induce brittle failure by slow crack growth. The stress rupture lifetimes of the various pipes are then correlated with flaw size. The results of the tests using the shorter lifetime resin show that flaw size does have a significant influence. It is particularly interesting to note that melt filtration, which removes large inherent flaws, substantially improved the stress rupture lifetime. With respect to material toughness, the longer lifetime pipe grade polyethylene resin showed a healthy tolerance to included flaws. In respect of the stress rupture test preferred resins can therefore be identified in terms of their tolerance to included flaws.

1. Introduction

Thermoplastic polymers are finding an increasing number of applications in areas where components are subjected to constant or fluctuating stresses for extended periods. Both types of loading, static [1–3] and dynamic [4], are capable of initiating a "brittle" crack, and causing that crack to propagate in a slow stable manner, either until a critical crack size for fast fracture is attained, or until the net section stress is high enough for ductile failure. The kinetics of the growth of these stable cracks, under both static and dynamic loading, has received considerable attention, particularly in respect of analysing crack growth rates using the concepts and methods of fracture mechanics [2–4]. However, what has received less attention is the size and nature of the flaws that initiate fracture. This can be illustrated by considering the PRI Churchill College conferences [5] on Yield, Deformation and Fracture; of 140 papers presented to date, 36 have encompassed some aspect of fracture mechanics, yet no presentation has been devoted to the nature of the flaws that initiate the cracks. This is surprising since fracture mechanics incorporates the concept of a flaw to the extent that the "brittle strength" of a plastics component can be controlled by the size of the flaw [6, 7]. This paper reports the results of experiments designed to identify the importance of the flaws that initiate fracture, and relate their importance to the toughness of the matrix material. At the same time the experimental approach has included the use of equipment that can effectively remove flaws from plastics melts, so offering a method of improving the strength of components manufactured from thermoplastic materials.

In order to examine the role of flaws in the fracture of thermoplastic polymers, the authors have confined themselves to a study of the failure of polyethylene pipes This particular combination of material and artefact was selected because of the extensive use of polyethylene pipe systems for gas distribution [8], an application where there is the additional requirement of a minimal number of or no failures within the design life of the pipe system (usually 50 years). The results of the tests reported here show clearly that in "low toughness" (a relative term) polyethylene resins, flaw size can be critical, with large flaws inducing failure at an early stage. However, in the "high toughness" gas pipe grade polyethylene resin studied, the tolerance to flaws increased, when measured by the stress rupture test. This confirms that the approach of maximizing resin toughness is the correct one, and this gives rise to pipe that can be used with confidence for gas distribution.

The role of the initiating flaw has been examined in polyethylene pipes by inducing the pipes to rupture by the propagation of stable cracks which grow in

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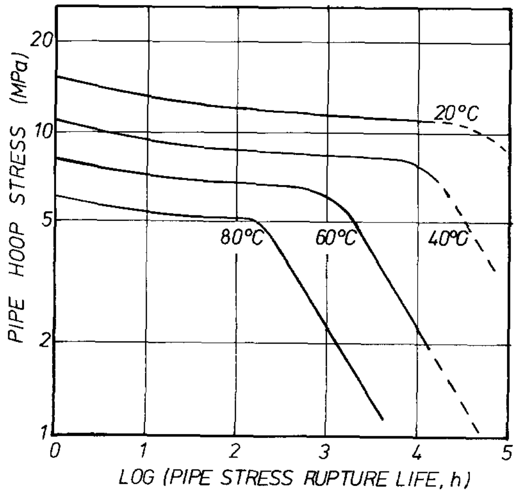


Figure 1 Schematic presentation of typical stress rupture curves for HDPE/MDPE copolymer pipes. The temperatures refer to the test temperature and indicate clearly the advantage of accelerated testing at 80° C to identify the brittle performance of pipe.

response to a constant stress. Flaws have either been purposely added or adventitious flaws have been removed by the use of ultra-fine melt filtration. The results of these tests have been compared to the results of tests on pipe extruded using a standard screen pack. In addition, the influence of resin toughness has been explored by extruding pipe from two different pipe grade polyethylene resins. The results of the tests are analysed using the methods of linear elastic fracture mechanics (LEFM), an approach which helps shed light on the fracture process, and in particular on the observed lifetime scatter that is encountered with the testing of polyethylene (or other polymers) pipes in the brittle region [1].

2. Failure of polyethylene pipes due to static stress loading

2.1. Failure modes

The characteristic shape of the stress rupture curves for medium (MDPE) and high (HDPE) density polyethylene pipes are shown schematically in Fig. 1. The curves can be divided into three failure regions: ductile, brittle and intermediate [2]. In the shallow region of the curve the high hoop stresses induce short time ductile failures, the pipe exhibiting extensive plastic deformation prior to failure. In the region of the curve where there is a steep decline, the lower hoop stresses favour long lifetimes and fissure-like "brittle" failures, see Fig. 2a. In the intermediate region, the knee of the curve, either failure mode can be observed. In general the stresses that MDPE or HDPE pipes systems experience in service are in the brittle region. All of the tests undertaken and reported here have been conducted using those conditions which induce slit-like brittle failure.

The "brittle" failures of polyethylene pipes and materials referred to above have two characteristic features. First, a scanning electron microscope examination of the "brittle" fracture surfaces reveals that there is significant microductility [3, 9], evidence of which is shown in Fig. 2b. Second, these "brittle" cracks grow in a slow stable manner at, typically, less than $10^{-8}\,\mathrm{m\,sec^{-1}}$ [2, 3]. These "brittle" failures are therefore not to be confused with the fast fracture of polymers, often observed with polystyrene.

2.2. The kinetics of slow stable crack growth

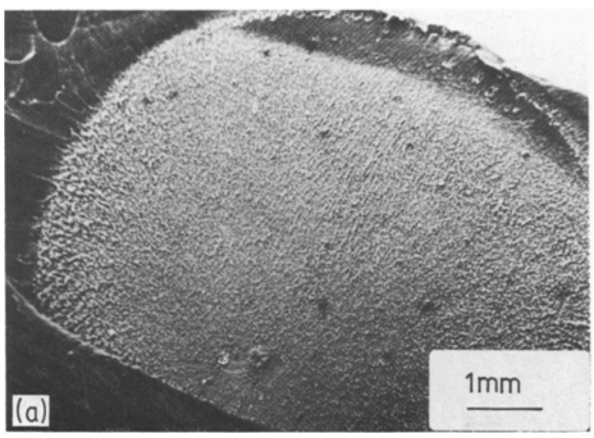
The kinetics of slow stable crack growth in polymers have been examined using the concepts and methods of fracture mechanics [2, 3, 7]. The stress intensity factor, K, which is given by

$$K = Y\sigma(\pi a)^{1/2} \qquad (1)$$

(where Y is a geometric correction factor to account for crack orientation and shape in finite sized structures, σ the gross applied stress and a the crack length) is a measure of the stress intensification at the crack tip. For specimens of sufficient depth and width, a critical value of K, referred to as K_c, can be used to identify the crack kinetics. For polyethylene materials, Chan and Williams [3] and Gray *et al.* [2] have demonstrated that the rate of growth of a stable crack, $\mathrm{d}a/\mathrm{d}t$, can be given by

$$\frac{\mathrm{d}a}{\mathrm{d}t} = \beta(K_c)^b \qquad (2)$$

where β and b are material constants. This equation infers that a double logarithmic plot of $\mathrm{d}a/\mathrm{d}t$ against K_c yields a straight line. This has been observed [2, 3, 7], although it should be recorded that an incubation period often exists prior to the start of crack growth. This point will be returned to at a later stage.



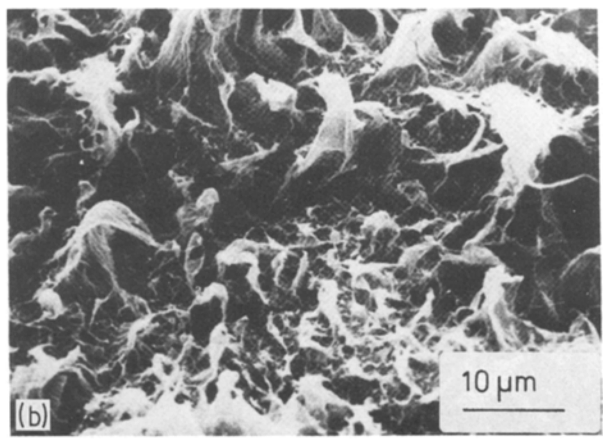


Figure 2 (a) Photomicrograph of a fracture surface from a polyethylene pipe that failed by slow crack growth. (b) Scanning electron micrograph showing the micro-ductility associated with failure of polyethylene pipes by slow crack growth.

2.3. Prediction of the lifetimes of polyethylene pipes

For a polyethylene pipe subjected to that combination of stress and temperature such that the failure will be brittle, it should be possible to predict the pipe stress rupture lifetime, τ_{SR}, by combining Equations 1 and 2 and integrating between the limits of the initiating flaw size, a_0, and the pipe wall thickness, h. In order that derived equations are valid, the following assumptions must hold:

1. There is no significant incubation period prior to the start of crack growth; this is discussed later in the text.
2. Equation 2 holds for all reasonable values of K_c with β and b invarient with respect to K_c; this holds for selected resins [2, 3].
3. By integrating between the limits of the flaw size and pipe wall thickness it is assumed that the crack grows in a stable manner for the significant fraction of the wall thickness. For both MDPE and HDPE pipes the slowly propagating brittle cracks have been observed to occupy at least 90% of the pipe wall thickness, with only final rupture taking place by a ductile process.
4. Y remains essentially constant for the significant fraction of the stress rupture lifetime. Since for small initial flaw sizes the majority of the time is spent with the crack small ($a/h < 0.5$), this may be regarded as reasonable as Y does not vary significantly for small cracks of the form and orientation found in pipes [10].

If the above assumptions hold, and $b \neq 2$, then for a pipe which fails when a stable crack opens in response to the hoop stress, σ_H (as is usually observed), the lifetime is given by

$$\tau_{SR} = \left[\frac{2(\pi^{1/2}Y\sigma_H)^{-b}}{\beta(2-b)}\right][h^{1-(b/2)} - a_0^{1-(b/2)}] \quad (3)$$

If $h \gg a_0$ and b sufficiently large, then Equation 3 simplifies to

$$\tau_{SR} = \left[\frac{2(\pi^{1/2}Y\sigma_H)^{-b}}{\beta(b-2)}\right] a_0^{1-(b/2)} \quad (4)$$

where the terms in the large square bracket are constant for a given temperature and hoop stress. Both Equations 3 and 4 infer that flaw size can have an influence on stress rupture lifetime of polyethylene pipes failing in the brittle mode by slow, stable crack growth, as illustrated schematically in Fig. 3.

The influence of flaw size on lifetime was explored experimentally by purposely adding or removing, by fine melt filtration, flaws from polyethylene pipes. The influence of resin toughness was examined by extruding pipes from different pipe grade polyethylene resins. The performance of the pipes was assessed by stress rupture testing using conditions designed to induce brittle failure. The results are interpreted using the concepts and methods of LEFM.

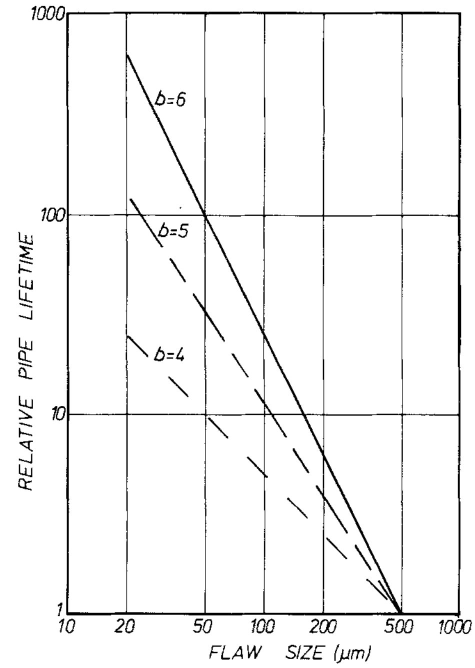


Figure 3 Illustration of the influence of flaw size (a_0) on pipe stress rupture lifetime. The improvement in lifetime with decreasing flaw size is related to the lifetime for a flaw size of 500 μm. b is the exponent in Equation 2, and the influence of flaw size on lifetime depends critically on the value of b.

3. Experimental details

3.1. Materials

Single batches of two pipe grade resins, an HDPE copolymer and a higher toughness MDPE resin, were used in this programme. Details of material density, melt index, molecular weight (by GPC) and melting point are given in Table I. These data are for material taken from the centre of the wall of pipe extruded using a standard screen pack. Fine melt filtration prior to pipe extrusion did not induce any significant change in melt index, M_w, or oxidation induction time.

3.2. Extrusion of pipe

Both 32 and 63 mm SDR*11 pipes have been

TABLE I Physical properties of the pipe grade polyethylene resins

Material	Density ($kg\,m^{-3}$)	Melt index (g/10 min)*	$\bar{M}_w$†	$\bar{M}_w/\bar{M}_n$†	T_m (K)‡
HDPE Copolymer	958	0.29	154 500	24.0	402
MDPE Copolymer§	940	0.22	176 500	14.5	402

*190° C, 2.16 kg.
†By GPC at the Polymer Characterization Centre, RAPRA, Shawbury, UK.
‡Perkin Elmer DSC2.
§BP Chemicals LTD, Rigidex 002-40.

*SDR, standard dimension ratio, is the average outside diameter divided by the minimum specified wall thickness.

extruded, the HDPE and MDPE resins as 32 mm pipe, the MDPE resin only for 63 mm pipe. The 32 mm pipe was extruded at Brunel University using a Samafor 45 mm single screw extruder with a screw of length to diameter ratio (L/D) of 28 : 1. The 63 mm MDPE pipe was extruded at Wavin Plastics, Durham, England, on a Francis Shaw 60 mm single screw extruder with an L/D ratio of 25 : 1. The pipe in both exercises was vacuum calibrated and the water bath was typically in the temperature range 12 to 15° C.

The effect of flaw size on stress rupture lifetime was modelled for the 32 mm HDPE copolymer pipe by:

1. Adding near spherical aluminium particles to the feedstock via a masterbatch made using the same HDPE copolymer resin. The aluminium particles were sized, prior to masterbatching, by sieving. The amount of masterbatch added was less than 0.5 wt %. The particle size was within the range 180 to 212 μm;
2. Filtering the melt, in separate experiments, to 150 and 45 μm. The 150 μm melt filtration was via the use of a standard stationary screen pack; the 45 μm filtration was effected by the use of an automatic twin screen, filter unit manufactured by Process Developments Ltd, London. The unit fits at the head of the extruder prior to the pipe die.

The influence of resin toughness was explored by extruding 32 and 63 mm MDPE pipe containing added flaws, and comparing the stress rupture behaviour of the MDPE pipes to that of the various 32 mm HDPE copolymer pipes. The 32 mm MDPE pipe contained aluminium particles of (425 to 600) μm diameter. The 63 mm pipe was extruded containing either (100 to 200) μm diameter glass spheres or (250 to 300) μm aluminium particles. As with the work on HDPE copolymer pipes, the particles were added via a masterbatch which constituted, typically, less than 0.5 wt % of the feedstock.

3.3. Stress rupture testing

The various pipes were stress rupture tested at a temperature of 79 ± 1° C and two internal pressures, 9.3 and 6.0 bar gauge, conditions that induced brittle failure. The pipes were unconstrained in the axial direction and met the dimensional requirements of ASTM D1598. Only brittle failures at least one pipe diameter away from the end fittings were recorded. Pipe failure was detected by an electrical conduction technique previously described [11], with lifetime automatically monitored.

In the extrusion of the small diameter pipes some variations occurred in the wall thickness and diameter. For valid failures local values of wall thickness (h_l) and outside diameter (D_l) were measured using a screw micrometer and vernier calipers, respectively. The local hoop stress, σ_{Hl}, was calculated using:

$$\sigma_{Hl} = \frac{P(D_l - h_l)}{2h_l} \quad (5)$$

where P was the internal gauge pressure. All stress rupture lifetimes have been correlated with the local hoop stress.

4. Experimental results and discussion

4.1. Pipe failure mode

All of the data presented here are for pipes which failed in the usual brittle manner [1, 2], the crack propagating in a slow stable fashion for the significant fraction (> 90%) of the pipe wall thickness. The cracks lay parallel with the pipe axis and therefore opened in response to the hoop stress. A close examination of all the fracture surfaces revealed evidence of the microductility that is associated with slow stable crack growth [3, 9]. Figs 2a and b are therefore typical photomicrographs and scanning electron micrographs, respectively, of the fracture surfaces observed for the brittle failure of these polyethylene pipes.

In general the fractures tended to initiate towards the pipe bore. This is the usual site [12] and is due to the higher local hoop stresses. The higher stresses result from the pipe being thick walled; using the Lame analysis for SDR11 pipe the hoop stresses are approximately 25% higher at the bore compared to the outside [13]. In addition the method of production gives rise to residual stresses in the wall of the pipe which, for outside cooled pipe, gives tensile residual stresses on the bore and compressive on the outer surface [14]. The combination of these two stresses leads to considerably higher hoop stresses at the pipe bore, hence the preferred initiation site. The exceptions to this rule were the cases where large particles lay towards the centre of the wall of the pipe.

4.2. The influence of flaw size

The influence of flaw size in controlling the stress rupture lifetime of polyethylene pipes failing by slow crack growth is examined by reference to the 32 mm SDR11 HDPE copolymer pipes. Initially three forms of the HDPE copolymer pipes will be considered:

1. Pipes produced from melt filtered to 45 μm using the balanced pressure melt filtration unit;
2. Pipes produced from melt filtered with the standard 150 μm screen pack;
3. Pipes produced from a feedstock to which was added masterbatch with 180 to 212 μm equivalent diameter aluminium particles.

It is assumed here and in the following discussion of these results that the 150 and 45 μm filters acted to effectively remove flaws larger than the screen size. The maximum flaw size is then given by the screen size.

The role of flaw size in controlling the stress rupture lifetime of the 32 mm SDR11 HDPE copolymer pipes is illustrated in Fig. 4 as a plot of σ_{Hl} against τ_{SR} on logarithmic axes. Four major points emerge for discussion from the data contained in Fig. 4.

4.2.1. The stress dependency of pipe lifetime

The stress rupture data for the three HDPE copolymer pipes yield straight lines when plotted as $\log \sigma_{Hl}$ against $\log \tau_{SR}$, see Fig. 4. This is consistent with previous studies on polyethylene pipes [1, 2] and is as expected from Equations 3 and 4. The slopes of these curves allow the calculation of the exponent of Equation 2, since the slope is equal to $-b^{-1}$ A least squares fit of the data in Fig. 4 was undertaken for the three forms

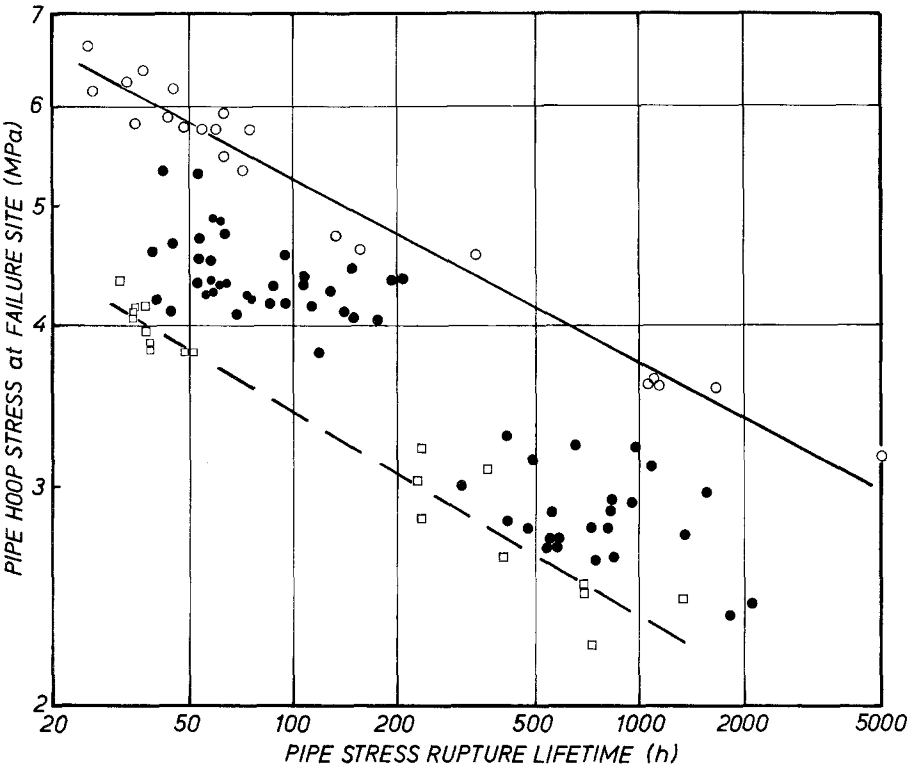


Figure 4 Stress rupture data for three forms of HDPE copolymer pipe; □ pipes with 180 to 212 μm aluminium flaws added; ● pipe filtered to 150 μm using the standard stationary screen pack; ○ pipe filtered to 45 μm using the balanced pressure Autoscreen from Process Developments Ltd. The test temperature was 79° C.

of pipe, and the value of b is recorded in Table II together with the determining coefficient.

An examination of Fig. 4, and the data in Table II shows the three curves are approximately parallel. This implies that fine melt filtration, which improved performance (see next section), did not deleteriously alter the resin, as is evidenced by the constant b, so that fine melt filtration does not significantly alter the stress dependency of the lifetime of the polyethylene pipes.

4.2.2. The influence of flaw size on lifetime

At any value of σ_{HI} below about 6 MPa (for 79° C testing) the steps of removing the coarse particles, and then the fine filtration of the melt prior to pipe extrusion substantially improved the stress rupture performance of the HDPE copolymer pipes. Flaw size therefore appears to influence stress rupture lifetime, a result consistent with a previous study [12]. In this present study, for the same value of σ_{HI}, pipe extruded from the melt filtered to 45 μm was typically six times better (in lifetime) than the shortest lifetime pipe extruded from melt filtered with the standard stationary screen pack. In comparison with the pipe dosed with the aluminium particles the improvement is of the order of × 20. Since the slopes of the stress rupture curves for the three forms of pipe are approximately parallel, then the improvement in pipe performance with fine melt filtration should persist to lower stresses. Similarly, with regard to temperature it would be expected that the improvement seen here with elevated temperature testing should be evident at typical service temperatures (5 to 30° C), since the stress rupture curves at different temperatures are, for polyethylene pipes, considered parallel [1, 2].

The antithesis to the statement that fine melt filtration improved the performance of the HDPE copolymer pipes is that the presence of the aluminium flaws reduced pipe performance and led to early failure. In previous studies on the infuence of flaw size on the stress rupture lifetime of "off the shelf" polyethylene pipes [12], large inherent flaws were observed; Fig. 5 reproduces the data on the observed range of flaw

TABLE II Stress rupture data characteristics for the various HDPE copolymer pipes

Pipe extrusion conditions	Calculated value for exponent b*†	Coefficient of determination*
45 μm filtered melt	6.67	0.97
150 μm screen pack	6.27	0.935
180 to 212 μm Al particles	6.25	0.95

*Both are dimensionless.
†The constant b is calculated from the slope of the stress rupture curve (see text).

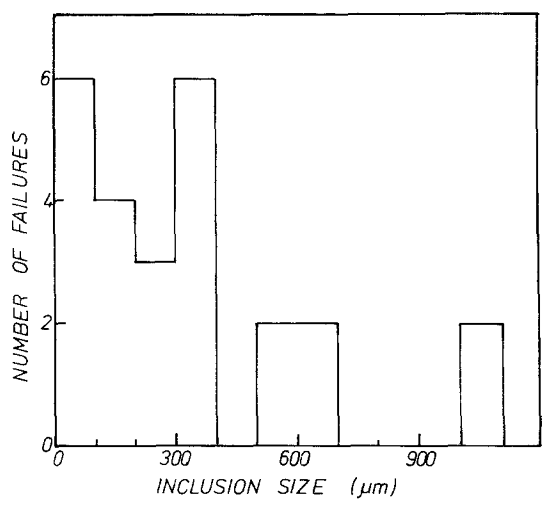


Figure 5 Histogram of the maximum size, on the fracture surface plane, of the flaw initiating the fracture of Continental European-produced HDPE pipe. The data are taken from [12].

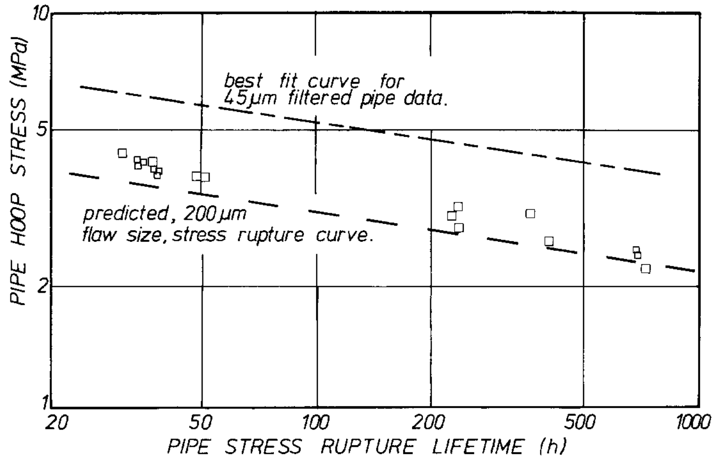


Figure 6 A comparison of the predicted 200 μm flaw size stress rupture curve with the experimental data (□) for flaws within the size range 180 to 212 μm. The test temperature was 79° C.

sizes. The study by Barker *et al.* [12] showed a statistically significant correlation between adventitious flaw size and pipe lifetime, with flaws up to 400 μm present and initiating fracture. Thus large included particles can arise from various sources, and these can reduce pipe performance, a result confirmed in this present study. It is therefore important to fit suitable melt filtration equipment and ensure that it is always operative.

4.2.3. Crack growth incubation times

Equation 2, which describes the kinetics of crack growth, can be used to predict pipe stress rupture lifetime provided that, among other factors, there is no significant incubation period prior to the start of crack growth. One way to test if a significant incubation period exists is to see if the stress rupture curve for one flaw size can be predicted from the stress rupture curve for a different flaw size. Provided the exponent b from Equation 2 is sufficiently large, which it is [6], Equation 4 can be used; for a given hoop stress and testing temperatures, Equation 4 is recast as:

$$\log(\tau_{SR})_1 + [1 - (b/2)]\log(a_0)_1 = \log(\tau_{SR})_2 + [1 - (b/2)]\log(a_0)_2 \qquad (6)$$

where the subscripts 1 and 2 refer to the different flaw sizes. Taking $(a_0)_1$ and $(a_0)_2$ equal to 45 and 1/2(180 + 212 μm), respectively, and letting b equal 6.46 (the approximate mean value for the 45 and 180 to 212 μm pipes), then knowing $(\tau_{SR})_1$, the value of $(\tau_{SR})_2$ for the larger flaw size can be calculated. This task was undertaken and the (180 to 212) μm stress rupture curve, which was predicted from the 45 μm stress rupture curve, is compared to the actual (180 to 212) μm results. The agreement between the predicted curve and the measured lifetimes is good as is shown in Fig. 6.

The good agreement between the predicted curve and the measured lifetimes infers that either the incubation period prior to the start of crack growth was small or that the incubation period itself was flaw-size dependent. Gray *et al.* [2], from their studies on notched pipes manufactured from a non-pipe grade polyethylene resin, support the observation that the incubation period may be small. There is, however, evidence put forward with other pipe grade polyethylenes, tested with pipe fittings, that the incubation period may be very extensive at typical service conditions [15]. A similar observation was reported for plaque samples by Chan and Williams [3] and Bragaw [16]. Thus, the evidence from testing short lifetime pipes is of a small or no incubation period, whereas with plaques the results highlight the presence of an incubation period.

Equation 6 can also be used for a separate calculation of the exponent b, provided that the flaw size associated with the fine melt filtration unit, $(a_0)_1$, is 45 μm. Calculating b from Equation 6 gives a value of 5.93 (at $\sigma_{HI} \simeq 3.8$ MPa), a value close to that measured from the $\log \sigma_{HI}$ against $\log \tau_{SR}$ plots. This is, of course, expected from the agreement between the predicted 180 to 212 μm flaw size data and the measured lifetimes for that flaw size. The agreement is encouraging in persuading us that fracture mechanics may be applied to describe the fracture of selected polyethylene pipes by slow crack growth.

4.2.4. Lifetime scatter

It is recognized and has been reported that significant scatter can exist in the lifetimes of polyethylene pipes failing in a brittle manner when tested at the same hoop stress and temperature [1]. It is assumed that these pipes were extruded using a standard 100 or 150 μm filter pack. Equations 3 and 4, and previous studies have all indicated that for slow stable crack growth either under a constant [2, 7] or alternating stress [6], flaw size can control lifetime, so that the scatter may have been due to flaw size variation. If that is the case, then limiting the flaw size, either by fine melt filtration (which would remove the large particles) or by adding flaws of a narrow size range would have the effect of reducing the lifetime scatter.

On the basis of the limited number of tests undertaken here, the evidence of Fig. 4 and the calculated coefficient of determination (Table II) shows that lifetime scatter can be reduced by either fine melt filtration or the purposeful additon of flaws within a narrow size range. The seemingly inherent scatter in the lifetime of polyethylene (and polypropylene) pipes failing in the brittle mode may therefore be related to a distribution in the size of the adventitious flaws that initiate fracture. The use of fine melt filtration has the effect of not only improving pipe performance, but also of reducing lifetime scatter. Pipe extrusion combined with fine melt filtration therefore gives a more consistent product.

In examining the scatter in the lifetime of the fine melt filtered pipe (using the Process Developments twin screen filter unit) one other aspect should be recorded. The data for the pipe produced with the twin screen unit has no unexpected early failures, see Fig. 4. From this we infer that the filter unit was very efficient in removing the naturally occurring particles. Using pipe samples to assess the influences of flaw size is interesting because all of the pipe (bar that material close to the end fittings) sees a similar stress. Hence, if just one particle gets through the filter it will initiate fracture. The evidence thus suggests that this method and equipment for melt filtration is extremely efficient at removing those particles naturally present in plastics melts.

4.3. The influence of matrix toughness

The performance of the 32 and 63 mm SDR11 pipes produced from the MDPE gas grade resin is recorded in Table III for a range of flaw sizes. No data are included on the influence of the filter since the performance with the added flaws was so good that these additional experiments were not undertaken. The data in the table show that this Rigidex 002-40 resin is tolerant of large included flaws. This observation is particularly true for the 63 mm pipe, and shows that the combination of good extrusion expertise combined with a tough resin gives rise to a good product.

To assess the influence of resin toughness on the brittle failure of polyethylene pipes the performance of the various MDPE pipes is now compared to the 32 mm HDPE copolymer pipes. This comparison shows clearly that the MDPE pipes out performed the HDPE copolymer pipes, with the MDPE pipes more tolerant to large included flaws. The MDPE resin (BP Chemicals Rigidex 002-40) is one of the newer pipe resins that show outstanding stress crack resistance, such that the generation of $da/dt - K_c$ data proves extremely difficult [3].

The improvement in the resin toughness led to a tolerance of large included flaws for testing under simple internal pressure loadings that were designed to induce brittle failure. However, while these tests indicate very long lifetimes in the elevated temperature stress rupture tests, service conditions will clearly differ from these; point loadings, for instance, are often present in service and pipes laid under roads may be subjected to complex loadings from traffic. Therefore, it is desirable always to exclude from the pipe material those flaws that will lead to a high local stress concentration. Some suitable form of melt filtration should therefore be fitted and in operation even with the new high toughness polyethylene pipe resins. The preference will be for a continuous screen changer to ensure good filtration without the build-up of significant back pressures which may induce some form of degradation into the melt.

TABLE III Stress rupture performance of the MDPE copolymer pipes

Pipe diameter (mm)*	Nature of added flaws	Equivalent diameter (μm)	Lifetime (h) of pipe†
32	Al particles	425–600	> 2 200
63	Al particles	250–300	> 2 000
63	Glass spheres	100–200	> 10 500

*SDR11 rated pipe.
†All tests conducted at 79° C with an internal pressure of 9.3 bar gauge (equivalent pipe hoop stress of about 4.6 MPa).

5. Conclusions

A study was undertaken to assess the influence of flaw size and material toughness on the stress rupture lifetimes of polyethylene pipes failing in a brittle manner. For a particular group of HDPE copolymer pipes it is shown that flaw size has an influence on pipe stress rupture performance. These model experiments therefore agree with previous work in which the influence of flaw size was examined by seeking correlations between flaw size and lifetime for "off the shelf" polyethylene pipes [12]. Both studies therefore conclude that large included flaws can reduce pipe lifetime. The present study shows that compared to pipe produced using the standard stationary screen pack, pipe performance can be enhanced by ultra-fine filtration of the melt prior to final extrusion. The fine filtration effectively removes those large flaws that would initiate fracture. In addition, fine filtration gives rise to a product with a more uniform flaw size distribution and hence less scatter in the lifetime.

In addition to these points, an analysis of the differences in lifetimes between the various pipes (different flaw sizes) leads to the conclusion that for the HDPE copolymer pipes studied the incubation period prior to the start of crack growth may be small or flaw-size dependent. The agreement Gray *et al.* [2] obtained between the measured lifetimes of notched polyethylene pipes and that predicted from crack growth measurements supports the contention that in some polyethylene pipes the incubation period may be small. Such a conclusion does not, however, apply to all pipe grade polyethylene materials [3, 15, 16].

With respect to the toughness of the matrix material the evidence is clear that the newer "tougher" pipe grade polyethylene resins (such as BP Chemicals Ltd, Rigidex 002-40) show a tolerance to flaws not exhibited by other materials studied in this programme. This gives confidence to the use of the material for the production of pipes and fittings for arduous applications, such as gas distribution networks. However, the authors note that this conclusion is derived from a series of tests on pipe where the loading was via simple internal pressure. The more complex loadings present in the field lead us to say that the pipe must be produced to a high standard, and therefore that melt filtration equipment should be fitted and working on the various extruders used in the production of the pipe.

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References

1. E. GAUBE, G. DIEDRICH and W. MULLER, *Kunststoffe* **66** (1976) 2.
2. A. GRAY, J. N. MALLINSON and J. B. PRICE, *Plast. Rubb. Process. Appl.* **1** (1981) 51.
3. M. K. V. CHAN and J. G. WILLIAMS, *Polymer* **24** (1983) 234.
4. R. W. HERTZBERG and J. A. MANSON, "Fatigue of Engineering Plastics" (Academic Press, London, 1980).
5. Plastics and Rubber Institute, "Triannual Conference on Deformation, Yield and Fracture of Polymer", (1973, 1976, 1979, 1982).
6. S. H. JOSEPH and P. S. LEEVERS, *J. Mater. Sci.* **20** (1985) 237.
7. R. J. YOUNG, "Introduction to Polymers", (Chapman and Hall, London, 1981).
8. J. M. GREIG, *Plast. Rubber Prcoess. Appl.* **1** (1981) 43.
9. C. S. LEE and M. M. EPSTEIN, *Polym. Eng. Sci.* **22** (1982) 549.
10. D. P. ROOKE and D. J. CARTWRIGHT, "Compendium of Stress Intensity Factors" (HMSO, London, 1976).
11. G. C. FORD, M. B. BARKER, S. BENTLEY, K. BATCHELOR and J. BOWMAN, *Polymer Testing* **3** (1983) 161.
12. M. B. BARKER, J. BOWMAN and M. J. BEVIS, *J. Mater. Sci.* **18** (1983) 1095.
13. E. J. HEARN, "Mechanics of Materials", Vol. 1 (Pergamon Press, Oxford, 1977).
14. J. M. HODGKINSON and J. G. WILLIAMS, *Plast. Rubber Process Appl.* **3** (1983) 37.
15. C. G. BRAGAW, Proceedings of the 4th International Conference on Deformation, Yield and Fracture of Polymers, Cambridge (April 1979) (The Plastics and Rubber Institute, London) Paper 1.
16. C. G. BRAGAW, Proceedings of the American Gas Association, VIth Plastics Pipes Symposium, Ohio, USA (1978).

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