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FRACTURE OF PVC PIPES

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ABSTRACT

PVC pipe is the major application of PVC in Thailand, utilising the mechanical properties and low cost of the material. However, unstable failure at low stresses is a problem that can occur in material that has been approved by standard quality control tests. This paper demonstrates the fracture of PVC pipe under creep deformation. Fracture mechanics concepts have been utilised to assess the stability of artificial cracks in tested samples and thus establishing criteria of failure in PVC and design data for long term predictions. The value of fracture toughness of materials can provide an understanding of the time-dependent failure mechanisms in PVC and can be used to rank the quality of different grades of PVC with respect to the processing conditions, thus providing a means of optimising the manufacturing process.

In long term testing at 20°C in air, the fracture toughness of the PVC decreased from a value of 4.7 to 2.0 MPa$\sqrt{m}$ over two years, highlighting the marked time-dependence. The experimental data can be described by a simple Power Law, the index value was found to be in excellent agreement with the indices from the modulus and yield stress data.

INTRODUCTION

The fracture of pipelines involves creep deformation and cracking, in addition to fast fracture and crack arrest. Unstable failure at low stresses is a consequence of great concern, particularly, for material that has been approved by standard quality control tests. The phenomenon of failure by catastrophic crack propagation in unplasticised polyvinyl chloride (uPVC) pipelines, under service conditions, poses problems for design and analysis for the specification of material production. In most situations, catastrophic failure is initiated by inherent flaws which arise either during manufacture or during end-use. Since a cold water uPVC pipe is normally designed for a lifetime of 50 years, fracture mechanics concepts have been utilized, to assess the stability of artificial cracks in tested samples and thus establishing criteria of failure in uPVC and design data for long term predictions.

This work was part of a larger study to investigate the failure behaviour of uPVC under static loading conditions for up to 2 years, so that information for design practices and material quality control could be obtained.[1] By applying a fracture mechanics approach, the values of fracture toughness of materials can provide an understanding of

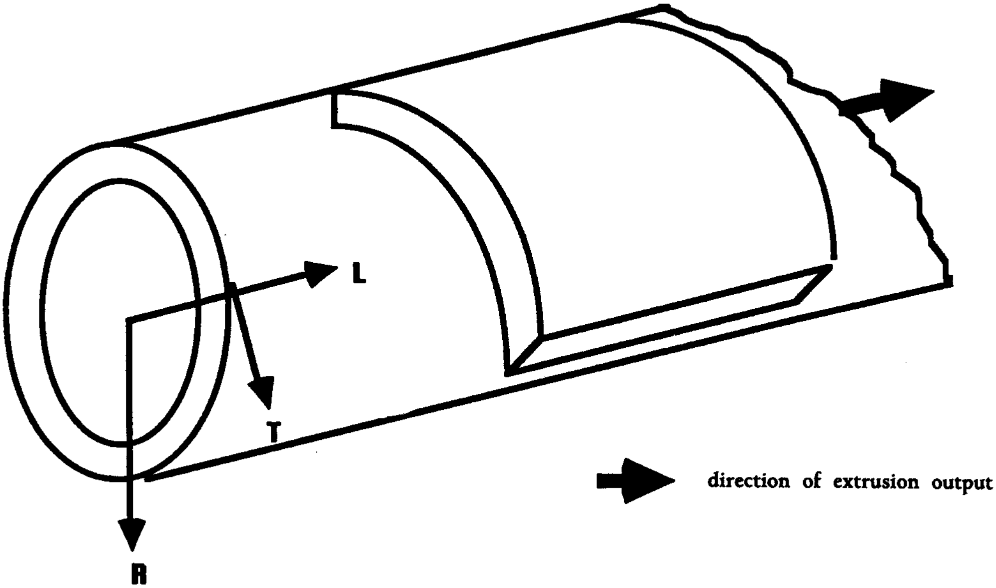
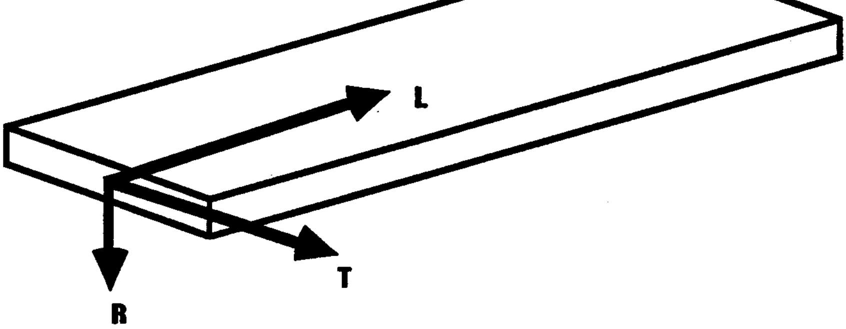


Fig. 1. Schematic diagram, showing the selection of specimens from a pipe section.



The arbitary directions in a three dimensional geometry.

the time-dependent failure mechanisms in uPVC, and to rank the quality of different grades of uPVC with respect to the processing conditions, thus providing a means of optimising the manufacturing processes.

The long-term tests were conducted to elucidate whether crack propagation was controlled by net section yielding or by the stress intensity of the crack tip.

MATERIALS

uPVC materials were prepared from 12″ "watermains" pressure pipe using a typical formulation with 2.5 phr filler. The sheet thickness, i.e. the pipe wall thickness was approximately 20 mm, it deviated within 3% for each sample sheet.

EQUIPMENT

The apparatus consisted essentially of a batch of single-lever systems incorporating three-point bending and constant-load devices. A fixed span of 120 mm was used. The lever arm ratio was 10:1, allowing for a weight to be applied at high stress. The dead-weight was loaded at a uniform rate by means of a hydraulic jack.

SEN Specimens and Preparation

SEN specimens were prepared from sections of pipe being pressed into flat sheets. Artificial cracks were formed in the same plane, either RT or LT direction (see Figure 1).

The initial crack was introduced by notching through the pipe thickness or specimen width. Notching is an important part of the specimen preparation. The sharpness of the notch is a major factor in assessing the fracture behaviour of the polymer. The influence of notch tip sharpness has been reported.[2, 3, 4] It is not possible to introduce an initial flaw to this tough material by means of the sharp razor blade method. The crack has to be formed by slow and gradual machining with sharp fly cutters having tip radii of 10 μm or less. Slow machining is essential to avoid any local heat accumulation. The temperature can rise above the T_g of PVC (80°C) with high speed machining. This causes a transition of material deformation from a glassy to rubbery state and results in stress relaxation. The material was removed from the specimen surface at about 200 μm depth at intervals to ensure a perfect sharp crack tip for each individual specimen. For a deeper notch of $a/w > 0.3$, a pre-crack can be formed by a saw-cut.

Power Law Relationships

Most polymers are time-dependent or viscoelastic. Fracture of this type can be modelled quite accurately by using time-dependent elastic parameters in elastic solutions. A full analysis has been described using a simple tension stress system for a number of distinct features of the time-dependence.[5] In a creep test, the relationship of strain and time generally gives a time-dependence of the form t^n, where n is a number generally less than unity. The creep modulus and yield stress of many polymers can be represented according to power law relationships of the form:

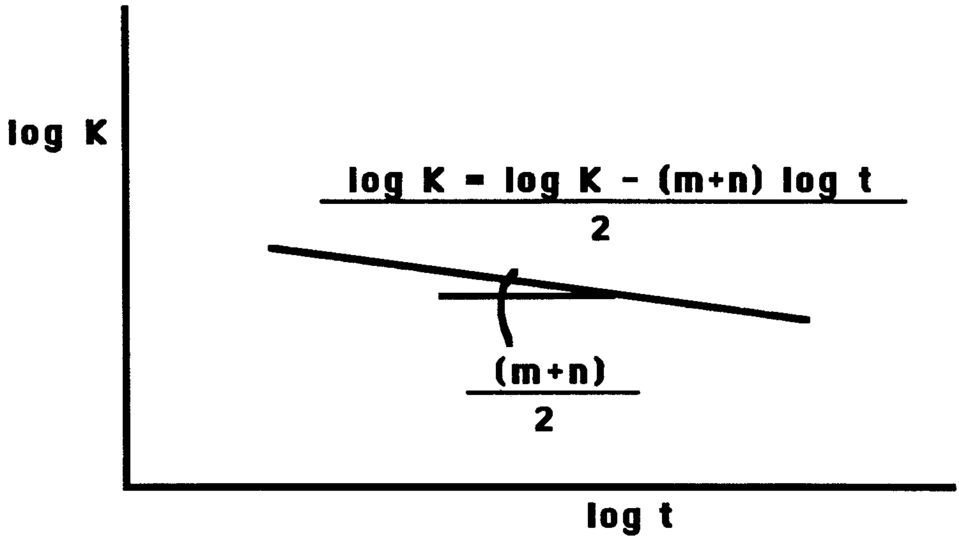


Fig.2 Schematic diagram of log K versus log t.

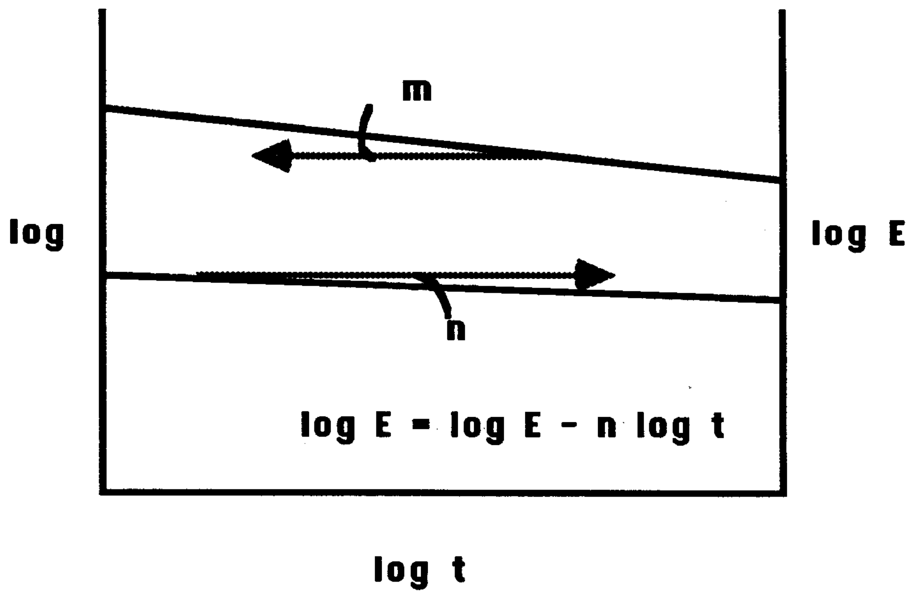


Fig.3 Time-dependent yield stress and elastic modulus.

$$E = E_o t^{-n} \qquad [1]$$
$$\sigma_y = \sigma_o t^{-m} \qquad [2]$$

where E_o and σ_o are the values of E and σ_y for $t=1$. They can be determined from conventional data in short term tests.

Using the Dugdale analysis and a constant COD criterion ($\delta = \delta_c$ at fracture), the time-dependence of fracture toughness K_c can be obtained by substituting equations [1] and [2] into [3]:
Under LEFM conditions, the Dugdale analysis gives:

$$\delta = n\sigma^2 a/\sigma_y E = G/\sigma_y = (K_1)^2/E\sigma_y \qquad [3]$$
$$K_c^2 = [\delta_c \sigma_o E_o t]^{-(m+n)} \qquad [4]$$
$$K_c = K_o t^{-(m+n)/2} \qquad [5]$$

where K_o is the value of K_c at $t=1$ and is equal to $\sqrt{\ }[\delta_o \sigma_o E_o]$

The long-term toughness variation can be assessed by plotting the applied K_c versus fracture time on a log-log basis. A straight line is obtained with a slope of $(m+n)/2$ [see Figure 2]. The prediction of the time-dependent fracture toughness can be estimated through data from equations [1] and [2] as shown in Figure 3. The predicted and experimental values can then be compared.

RESULTS AND DISCUSSION

Time-dependent data

It was found that a plot of K_c versus time-to-failure and K_c versus time to initiation fitted the simple Power Laws very well [Figure 4]. There was no distinct transition in the data that deviated from the laws. Some scatter was observed, this was a result of the inherent inhomogeneity of the polymer mass, rather than any analytical/ experimental inaccuracies. Walker *et al.* [6] gave strong evidence on the effect of anisotropy on the yield point and post-yield behaviour of the PVC matrix. This was associated with the formation of a characteristic diamond shape of cavities in cold drawn PVC in a simple shear deformation. Nevertheless, the failure behaviour can be predicted on a long-term basis for engineering design. The initiation time is a significant portion of the life. All macroscopic properties can be determined. The time-dependent fracture toughness index $(m+n)/2$ was determined from the creep loading for up to 2 years. The value was compared with those obtained individually from conventional short term tests for the uniaxial and flexural yield stress and elastic modulus determination. The results are tabulated in Tables 1 and 2.

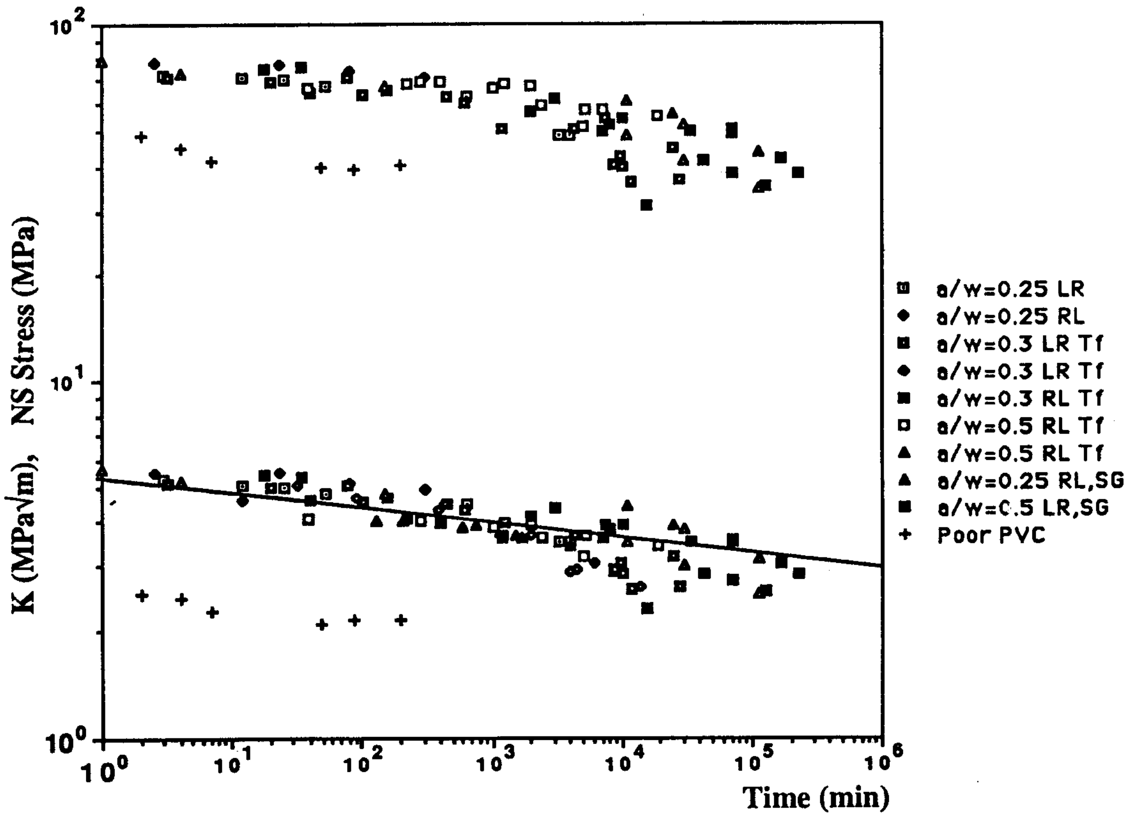


Fig. 4. Time-dependent fracture of uPVC at 20C

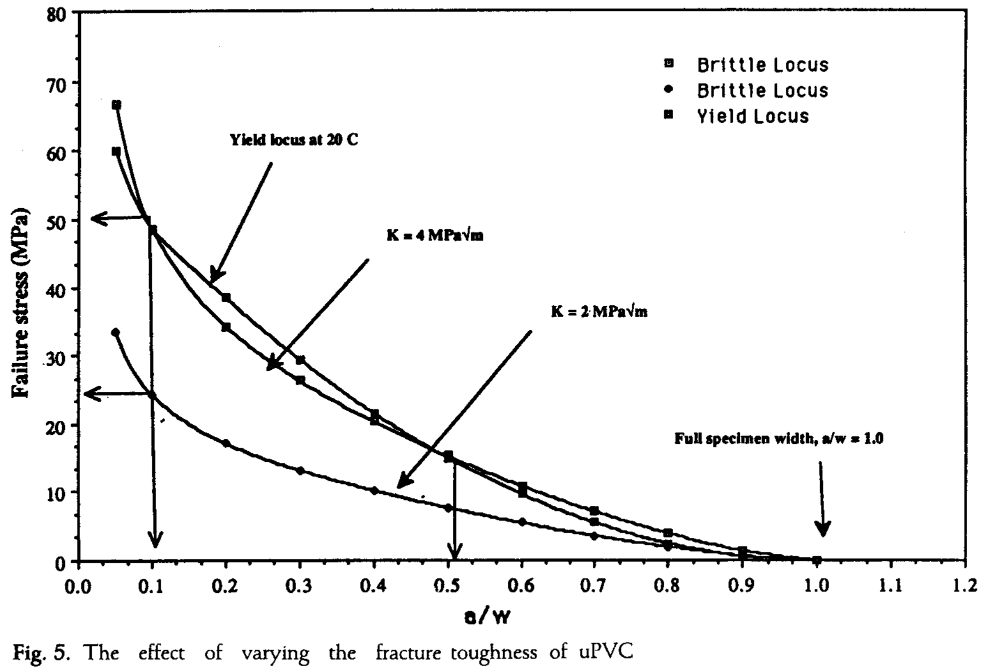


Fig. 5. The effect of varying the fracture toughness of uPVC

TABLE 1: The unit properties t= 1 minute

σ_o [MPa]	54
E_o [GPa]	3.05
K_o [MPa$\sqrt{m}$]	4.7
δ_o [μm]	134

TABLE 2: The dimensionless functions

		Marshall's data
m	0.04	0.045
n	0.027	0.008
(m+n)/2	0.034	0.026

The slope (m+n)/2 described the data reasonably well, but was 25% greater than Marshall's value of 0.026.[7] It was also pertinent to note that $m \neq n$, and that threw doubt upon the constant COD criterion. Clearly, there must be some other mechanism responsible for the failure of different types of uPVC. Previous work on low toughness uPVC has reported that fracture can occur at the achievement of a critical K_c or crack speed such that:
K α [crack speed]$^{(m+n)/2}$ and that the slow crack growth mechanism is responsible.[7] A distinct "knee" in the K/$\dot{a}$ curve was reported for a high toughness pipe having a K_c level of 3.5 MPa$\sqrt{m}$, indicating that there is a change in the crack tip zone.[8] The mechanics of plastic zone development at the crack tip potentially control the failure mechanism. Yielding results in a plastic blunting at a sharp crack which serves to resist crack growth by magnifying COD, rather than promoting crack extension and thus decelerating the crack. Final failure is caused by the rising COD at the onset of ductile tearing as the crack accelerates to cause plastic collapse. The failure mechanism is predominantly controlled by net section yielding rather than the achievement of a critical K_c or crack speed because the rate of acceleration is controlled by the rate of increase in σ_y with strain rate at the crack tip.

Net section stress control

Figure 5 illustrates the residual strength diagram of two levels of K_c values (K=4 MPa$\sqrt{m}$, and K=2 MPa$\sqrt{m}$,). It demonstrates two types of controlling criteria in the failure of uPVC according to its fracture toughness, and the temperature and strain rate at which yield stress was a controlling factor.

The net section yield locus was drawn to be approximately close to the yield stress of PVC at room temperature (σ_y=55 MPa). The line forms tangential points to the locus at which crack propagation controls failure according to the form: $\sigma + \alpha 1/\sqrt{a}$.

As yield stress decreases with creep loading time, the net section yield line would always be lower than the brittle locus for K=4 MPa$\sqrt{m}$, and would entirely control

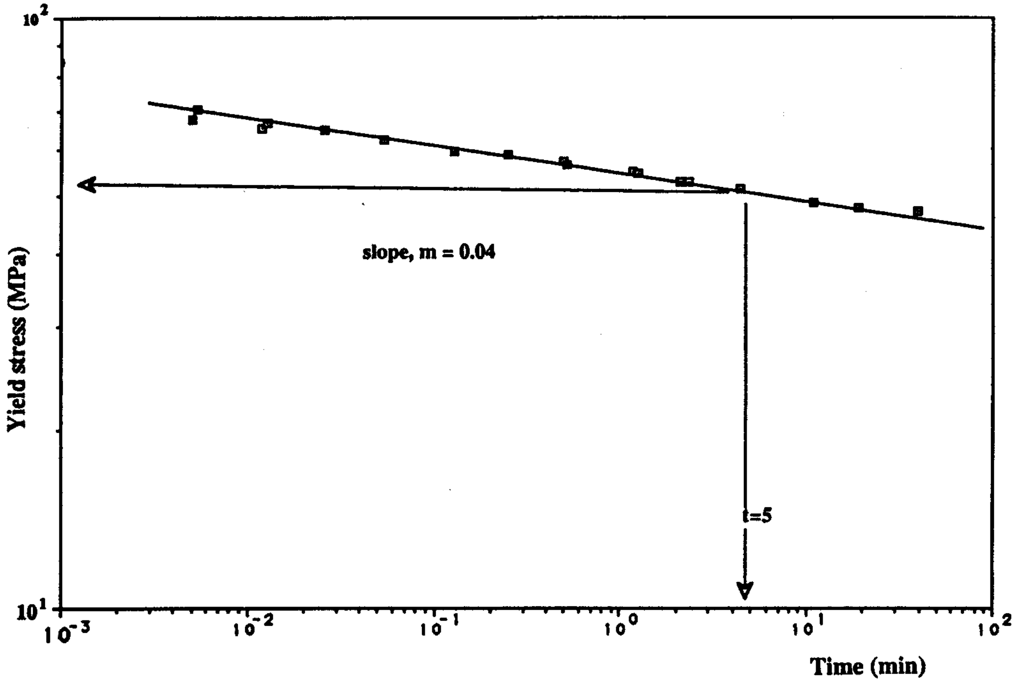


Fig. 6. Tensile yield stress versus time at 20C

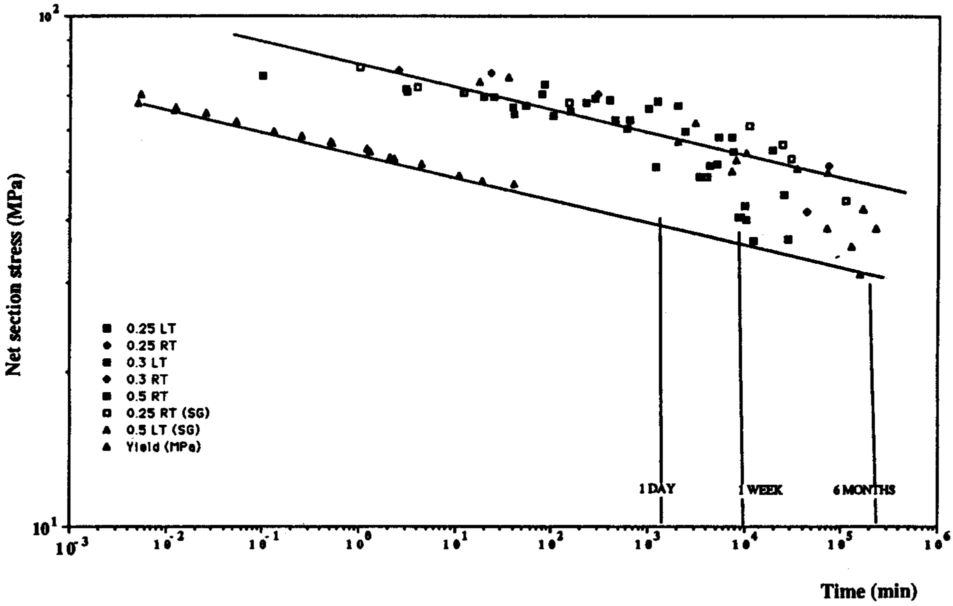


Fig. 7. Time-dependent yield stress and applied stress

the failure for all crack lengths because the notched ligament reaches a value close to the yield stress, σ_y, at the onset of initiation. The log-log plot of yield stress versus time in Figure 6 shows that the slope m is equal to 0.04. In bending, the material on the net section will not be fully yielded at the net section yield line. Post-yielding occurs via plastic collapse when the bending moment is 1.7 times the bending moment to cause first yield because of the non-linear stress distribution and some plastic constraint caused by the triaxiality that develops at the crack tip in plane strain such that the stress 'across' the tip, $\sigma_{yy} \approx (\sqrt{3})\ \sigma_y$. For a fully hydrostatic stress system, the Von Mises Yield criterion predicts an increase of a factor of 3 in yield stress at the crack tip.[9, 10] This is unlikely to occur with a visco-elastic polymer such as uPVC.

Clearly, the failure of high toughness uPVC follows a simple net section yield criterion and its time-dependency obeys the Power Laws. The applied net section correlates well with the yielding criterion in that it is always 1.5 times greater than the uniaxial yield stress to cause first yield at the onset of cracking. This is demonstrated by the log-log plot of the time-dependent uniaxial yield stress and applied net section stress in bending (see Figure 7). The failure is controlled by a critical stress or macroscopic yielding at the crack tip, and not by a stress intensity factor. Crack tip blunting increases as the yield stress in the plastic zone length decreases until the net section stress exceeds the stress being carried by the plastic zone calculated using simple uniaxial values. Failure occurs as K increases with a reduction in ligament area during the ductile tearing process. The mechanism is accompanied by the microscopic shear yielding in the formation of crazes. Considerable energy was absorbed as shear lips developed. Shear bands were observed at the outer plane stress skin at 45°. No level of valid K_c could be acceptable because the ASTM criterion limits were $\sigma_n \leqslant 0.8\sigma_y$. This is essentially a criterion of failure of uPVC.

The effect of increasing fracture toughness

The resistance to brittle fracture of uPVC can be increased by optimising the processing conditions during pipe extrusion. The yield stress of uPVC is insignificantly different for material that has been classified as 'poor' or 'good', with respect to their fracture toughness values, whose gelation levels were determined quantitatively by rheological techniques, and qualitatively by dark-field microscopy and methylene chloride tests.[11, 12] The well processed PVC would give a K_c value of 3.5 MPa$\sqrt{m}$, or higher, resulting in the brittle locus line K_2, being higher than the yield locus Y_1. Furthermore, the effective yield stress, σ_y, at the crack tip will increase to $C\sigma_y$ due to the constraint caused by the triaxial stress system in plane strain condition. C is the plastic constraint factor and varies from 1-3. For a fully hydrostatic stress field, C=3. C is much lower for a viscoelastic polymer. It is apparent that the plastic collapse line, P, is intersecting with or tangential to the K_c curve at a deep crack length, $a \geqslant a_2$. EPFM may be applied in the region between the points of tangency or intersection. The J integral concept can be used to characterise fracture in this situation of high constraint. The ASTM procedure specifies the conditions for deep crack length to width ratio of 0.5 to 0.6. The test is of

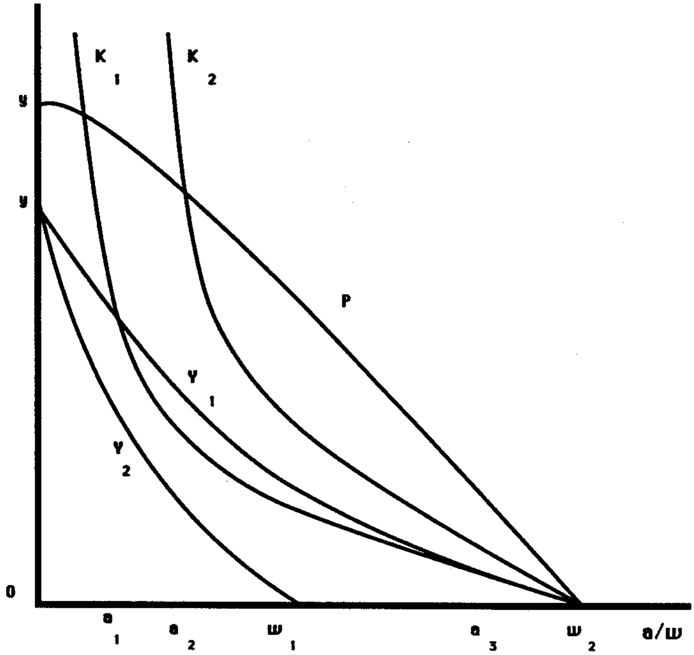


Fig. 8. Schematic Yielding Diagram for Low and High Fracture Toughness Materials in Bending

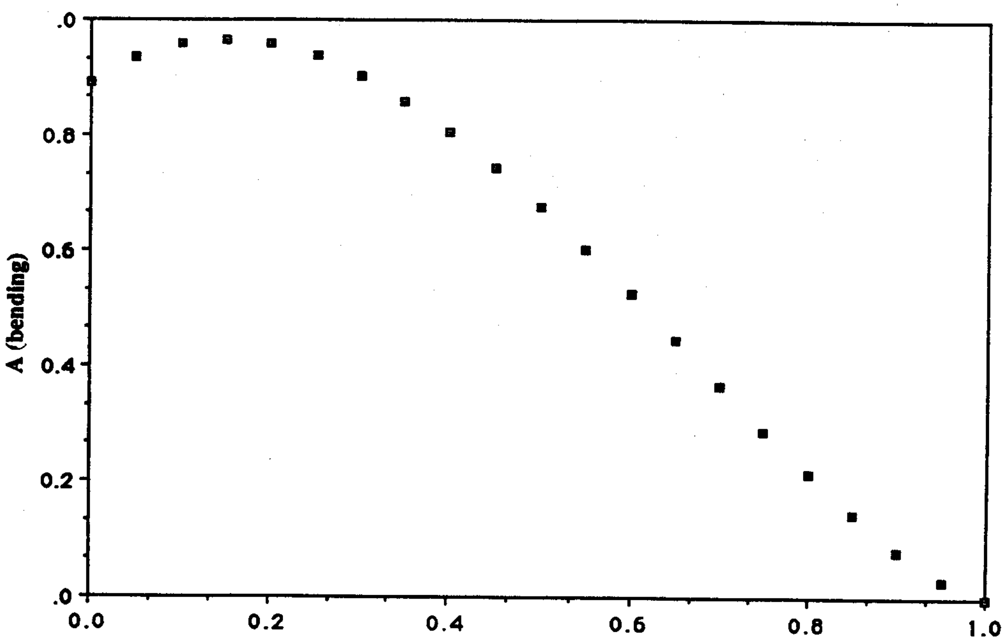


Fig. 9. Theoretical values for 'A' (Ref. Kendall 1983)

prime importance with respect to practical applications where cracks occur in high pressure parts such as thick-walled vessels and pipes. For a short crack size, $a < a_2$, the specimen failed by plastic collapse which was applicable to a post-yield theory. For a thin section, the unconstrained yield stress, σ_y, was reached in the uncracked ligaments well before the critical stress intensity factor K_c. The material failed in a ductile manner since the yield locus (line Y_2) was always lower than the K_c curves (K_1 and K_2). Fracture was controlled by general yielding in the plane stress state, neither LEFM nor EPFM are applicable.

The criteria of failure in uPVC

Unlike most thermoplastics, uPVC is formulated with at least three or more additives, giving variable properties ranging from high to low molecular weight grades according to the ISO Standard (e.g. ISO K-72, K-67, K-56). The mechanical properties in terms of strength and stiffness are influenced by different processing conditions, and the formulation of a PVC composition have a pronounced effect on the time and strain dependence of the creep function.[13] Applying a fracture mechanics approach, each grade of PVC can be expressed in terms of the stress intensity factor K for a certain processing condition. Hence, K cannot be a single value. It varies from a value of 1.5 to 4.0 MPa$\sqrt{m}$, depending on the level of gelation. This results in two different types of failure mechanism.

(i) The brittle mode of linear-elastic material.
(ii) The ductile mode of non-linear elastic material.

If the gelation of uPVC is low, the material behaves linearly elastically with low crack resistance. A craze may form at the loaded crack tip and the voids may coalesce followed by slow crack growth leading to unstable brittle failure. The time-dependency of stabilized craze and void generation is governed by the degree of particle fusion.[8] In poorly gelled uPVC, the craze becomes unstable leading to an unstable catastrophic brittle failure at low stresses. The crack tip plasticity is severely limited and an 'apparent valid K_c is obtained without any slow crack growth. Clearly, the analysis of crack initiation and propagation mechanism in different 'types' of uPVC is not a straightforward process because of the variation in fracture toughness K_c. Thus, there is no unique K-$\dot{a}$ relationship.

The more satisfactory criterion of acceptability is that the critical stress that controls fracture on the ligament area is greater than the uniaxial yield stress, σ_y, calculated from the uniaxial tension determination, such that σ_{net} in bending is greater than $\alpha\sigma_y$, where α is a plastic collapse factor having a value varying from 0.8 to 1.5. All failures must occur via net section yielding.

The criteria of failure can be demonstrated schematically in the residual strength diagram [see Figure 8]; which illustrates two levels of brittle curves (K_1 and K_2), two different widths of net section yielding (Y, and Y_2), and a plastic constraint effect in bending (P).

The fundamental principle of the onset of the ductile-brittle transition lies on the position of the K and Y curves. For the low toughness material, the failure stress is determined by the stress intensity factor on the brittle locus (line K_1). The K curve always lies well below the yield locus (line Y_1) for full width section (a/w=1), W being greater than the limiting size criterion. The failure stress decreases with increasing crack size as $a_1 < a_2 < a_3$. In this regime, brittle failure is predominant and LEFM applies according to an equation of the form:

$$\sigma_c = A.\ K_c \div \sqrt{(\pi a)} \qquad [5]$$

Where A is the theoretical constant and has analytical values shown graphically (Fig.9). Equation 5 is modified from the original Griffith equation that is limited to perfect elastic materials for infinite strips. The theoretical constant. A, was determined for an edge crack length, a, of a finite strip. 'A' diminishes towards zero as the crack propagates through the entire width, W. Kendall[14] showed that the bending strength computed from the equation [5] lay within 11% of the Griffith curve for values of a/w up to 0.3. Furthermore, the experimental data for bending stress of elastic material (such as PMMA) are in excellent agreement with the theoretical values for $a/w > 0.2$. Stress analysis showed that long cracks have a plane strain hydrostatic stress field, stimulating brittle behaviour. For very short cracks or inherent flaws ($a \ll a_1$), the fracture stress is determined by the yield stress of the material. The plastic zone size becomes large and can no longer be neglected. The net section yielding criterion is usually obeyed. Brittle failure could not occur as it was not possible for the gross stress to be greater than the yield stress. LEFM broke down in this region, and EPFM could be applied. Fracture is invariably ductile.

CONCLUSIONS

The fracture toughness of uPVC pipe is not a unique value. It varies from 2.0 MPa$\sqrt{m}$ to 4.7 MPa$\sqrt{m}$ for pipe grade, depending on the level of gelation and hence the degree of fusion of uPVC grains. In poorly gelled uPVC, the material behaves linearly elastic with low crack resistance. The crack tip plasticity is severely limited and a low K_c value ($<$3.0 MPa$\sqrt{m}$) is obtained. In 'good' uPVC, the gelation level is high, the material behaves non-linearly elastic with high crack resistance. The crack tip plasticity becomes more pronounced as crazes and voids are generated.

The time dependence of craze formation is a function of the time-dependent yield stress. The failure behaviour of uPVC can be predicted on a long term basis for engineering design. The $K-t_i$ and $K-t_f$ curves of creep fracture at two different temperatures verified the simple Power Laws very well. There was no distinct transition that represents the onset of ductile-brittle fracture. The experimental slope (m+n)/2 showed excellent agreement with the calculated values, where m and n were determined independently from standard short term tests.

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