Impact energy and rapid crack propagation in plastic pipes

Witold Brostow
Department of Materials Engineering, Drexel University, Philadelphia, PA 19104, USA

and Willi F. Müller
*Abteilung Kunststoffe, Hoechst Aktiengesellschaft, 6230 Frankfurt am Main 80,
B.R.D./West Germany*
(Received 4 March 1985)

Theoretical, computational and experimental work has demonstrated that there is for each material a limit to its chain relaxation capability (*CRC*). Thus, there is also a limit U_l to the impact energy U that a material can absorb per unit weight at a given temperature. At the same time, experimental studies of rapid crack propagation (RCP) have shown the existence of a one-to-one relationship between the crack length L and U. In this work we have developed an equation between U and L. The validity of our $L(U)$ relation has been proven by experimental data for polyethylene pipes. A formula for the *CRC* limit U_l follows immediately. Thus, we have the first model of RCP in macroscopic as well as molecular terms, with evident consequences for our capabilities of prevention of RCP occurrence.

INTRODUCTION

Instead of traditional metal pipes, the gas industry, as well as other industries, is strongly interested in conversion to plastic pipes. Advantages of such a conversion are evident, including complete elimination of corrosion problems and low costs of laying and installation. However, to date fuel gas transmission between cities is still predominantly via metallic pipes. Local gas distribution within localities also involves metal pipes for large diameters, and plastic pipes only for relatively small diameters.

One of the key reasons for these limitations in the use of plastic pipes is the danger of rapid crack propagation (RCP). The phenomenon consists of the growth of a crack at high speeds. Velocities of 300 to 1400 feet/second, that is of 100 to 400 m/s, have been observed in polyethylene (PE) pipes[1]. Given the importance of the conversion to plastic pipes, fairly extensive investigations of pipes under internal pressure are conducted[2] including for instance studies of fatigue of such pipes[3]. However, internal pressure testing normally induces either ductile type failure (at room temperature) or brittle slow-crack-growth failure (at elevated temperatures)[9,10]. Neither of these two resembles RCP, whether one takes into account the fracture speed or morphology.

Current research on RCP consists largely of empirical studies coupled with utilization of fracture-mechanical concepts; the objective here is to provide gas utilities with operating guidelines for prevention of RCP occurrence in the short term. As noted elsewhere[11], fracture mechanics has been developed for metals first, and transfer of the same concepts to polymer science and engineering has only limited applications; this applies in particular to highly viscoelastic and time-dependent polymers such as PE. Thus, much more ought to be done; material improvements are necessary to prevent RCP occurrence, to allow larger pipe diameters to be used in gas distri-

bution, and finally to possibly enable a switch to plastic pipes in gas transmission. To achieve these objectives, we first need a better understanding of the mechanism of the rapid crack propagation phenomenon. The understanding should then be coupled with development of quantitative relationships between RCP and materials variables: molecular weight, degree of crystallinity, comonomer content, etc. While recently American manufacturers have successfully modified piping materials from the point of view of the resistance to slow crack growth, it is not at all certain how these changes affect RCP.

Work conducted by Gaube and one of us[2,6,8] on rapid crack propagation and arrest in PE pipes has lead to an important conclusion: there exists a one-to-one relationship between notch impact energy U and crack length L. The latter was determined by a procedure proposed by Greig and Smith[12] such that a knife is pushed through a pressurized pipe by a falling weight. One particular objective in the present paper will be to develop a model of RCP taking into account the existence of a $L(U)$ relationship.

We have already noted the limited usefulness of fracture mechanics (FM) in dealing with polymeric materials. Indeed, Fearnehough and his colleagues working on steel pipes[13] and using FM have developed an equation for the critical stress σ_c in the pipe wall: if σ_c is exceeded, the crack will propagate. Experiments on PE pipes have shown[6,8] that these pipes have much higher σ_c values than those calculated from the equation of Fearnehough, Jude and Weiner[13].

MODEL OF RAPID CRACK PROPAGATION

We shall utilize results of earlier work on mechanical properties of polymers, impact behaviour in particular, as reviewed[14]. It has been concluded[15] that a given specimen

(or a given unit weight of polymer) can dissipate by chain relaxation a certain maximum of energy per unit time. The excess energy furnished by external mechanical means which cannot be so dissipated goes into competitive processes, including bond breaking and crack propagation.

The above imposes a condition on a model of RCP that we want to develop. Not only must each such model lead to a relation between U and L, but each also has to provide the limiting value U_1 such that

$$L(U \geqslant U_1) = 0 \qquad (1)$$

While in the present case it is $L = 0$ which determines U_1, we write the relation in this form so as to show that in general the crack length L is a function of the energy U absorbed on impact, and not *vice versa*.

Consider now a limiting case when the material cannot absorb any energy at all. Then

$$L(U = 0) = \infty \qquad (2)$$

Thus, we have one more criterion that a $L(U)$ relation has to fulfill.

Between limits provided by relations (1) and (2) we have finite values of the crack length L. Consider the energy provided externally to a pipe of specified dimensions during an impact event. This energy can go into a number of 'competitive' processes, as discussed in texts on materials science and engineering; see for instance Chapter 12 in ref. 16. However, it has been stressed[15] that the chain relaxation has priority. The chain relaxational capability (CRC) of a given polymeric material depends on chain conformations and on free volume v^f available. If we are dealing with one type of a material, with relatively minor differences in density between various specimens, then CRC will be approximately constant.

Thus, we now have to analyse processes other than relaxation. To do this, the presence of microcracks ought to be taken into account. If microcracks are absent, plastic deformation is a good candidate[15] for the second dominant process; fracturing the material will require still further expenditure of the energy, and the impact energy U will be high. By contrast, if microcracks have been already built-in, either during manufacture or during subsequent handling, then stress concentrations will take away energy from processes of deformation type. The fracture will occur at a lower value of the impact energy U. Pertinent here is also the experimental finding that higher U values are related to rough fracture surfaces characteristic for partly ductile failure; conversely, lower U values in impact, as well as higher crack speeds, are connected with relatively smooth brittle surfaces[6,8]. This finding also fits well with the remainder of our just developed model. Clearly, the relation between L and U is of the type

$$L \sim \frac{1}{U} \qquad (3)$$

Thus, the higher the energy-absorbing capability of the material — a measure of which is the value of U determined in the Charpy test — the shorter is the crack length L in the corresponding pipe determined by the Greig–Smith test.

The simplest possible formula which satisfies all the criteria listed above can be written as

$$L = L_0 + \frac{L_1}{U} \qquad (4)$$

where L_0 is a constant with dimensions of length; L_1 is another constant with dimensions of length × energy. A formula for the limiting energy value U_1 required by relation (1) follows immediately from equation (4):

$$U_1 = -\frac{L_1}{L_0} \qquad (5)$$

It should be noted that equation (4) applies to a *class* of materials, with values of free volume, and thus of CRC, within a relatively narrow range. If all materials were completely identical, we would have just one value of U, and a single corresponding value of L. On the other hand, if one would deal simultaneously with several classes of pipes, for instance made of PE and made of a polymer of a widely different chemical nature, then values of L_0 and L_1 would be different for different classes. Applicability of equation (4) will be tested against experimental data in the next section.

CALCULATIONS

We have noted in the Introduction the existence of L and U data for PE obtained by Gaube and one of us[6,8]. The Greig–Smith tests to determine crack length L were made for 5 m long pipes, of 110 mm diameter, 10 mm wall thickness, at $0 \rightarrow 5°C$. The Charpy tests were made for specimens with notch depth $h = 1$ mm at $0°C$.

The density range of the pipes was between 0.938 and $0.958 \, \mathrm{g \, cm^{-3}}$. As discussed in ref. 15, there is quite a variety of definitions of v^f. The statistical–mechanical theory of the liquid state developed by Flory[18,19] provides a definition of v^f which is widely used, mostly in its original form although modifications exist also[20]. Now

$$v^f = v - v^* \qquad (6)$$

$$\tilde{v} = \frac{v}{v^*} \qquad (7)$$

where v can be taken as the specific volume per 1 gram, v^* as the corresponding characteristic ('hard-core') volume, and $\tilde{v}$ is the reduced volume. Shih and Flory[21] have tabulated values for v^* for a number of amorphous polymers. Their value for polymethylene is $v^*(25°C) = 1.000 \, \mathrm{g \, cm^{-3}}$. While this value has been obtained for a material of lower density, it is very convenient for our comparative purposes. We find therefrom that the range of $\tilde{v}$ values for the pipes studied extends from 1.044 to 1.066. Thus, we are well justified in neglecting changes in the chain relaxation capability within the range.

The set of experimentally obtained pairs of U and L values was used to solve an overdetermined system of equations (4) for L_0 and L_1. A Marquardt-type regression procedure was used (ZXSSQ subroutine from IMSL, Houston, Texas). The results were $L_0 = -896$ mm and $L_1 = 269$ J mm. Both calculated and experimental values are shown in *Figure 1*.
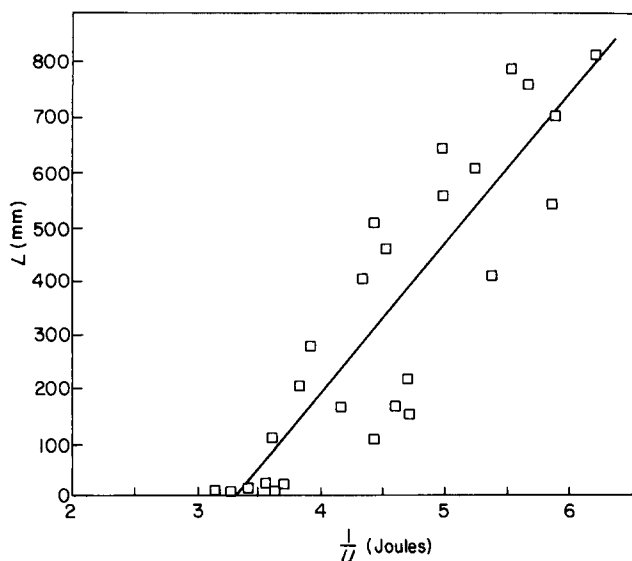
Figure 1 Length L of the crack in PE pipe determined by the Greig–Smith test vs. U^{-1}, where U is the notch impact energy measured in the Charpy test

The U_I value from equation (5) is 0.300 J. Thus, if the impact energy determined in the Charpy test is higher than this value, no crack propagation will occur in the respective pipe. Of course, since this value pertains to the entire class, a safety factor slightly larger than unity can be applied.

ANALYSIS OF RESULTS

Since we are dealing here with two types of experiments, let us now consider each of them separately.

We begin with the Charpy test. Fortunately, a theory of the phenomenon already exists, as reviewed in detail in ref. 14. The energy U provided to the specimen in an impact event can be represented at a time after the event as follows:

$$U_b = U_r + U_n - U \tag{8}$$

Here U_b is the part of energy already spent in a destructive way (bond breaking and crack growth leading to fracture), U_r the part spent in a non-destructive way (chain relaxation, plastic deformation) and U_n (denoted by U in ref. 14) the part of impact energy that at a given time has not yet gone into either. If one assumes that both ways of spending the impact energy constitute first-order time processes with respect to the unused energy U_n:

$$\frac{dU_n}{dt} = -(c_b + c_r)U_n \tag{9}$$

where t is time while c_b and c_r are appropriate rate constants, and we obtain

$$U_b = \frac{c_b U}{c_b + c_r}[1 - e^{-(c_b + c_r)t}] \tag{10a}$$

$$U_r = \frac{c_r U}{c_b + c_r}[1 - e^{-(c_b + c_r)t}] \tag{10b}$$

The same theory[14] provides a connection with the

stress concentration factor K_I defined as

$$K_I = 1 + 2\left(\frac{h}{l}\right)^{1/2} \tag{11}$$

where h is the depth of the notch and l the radius of curvature at the tip of the crack. While K_I is a purely geometric quantity, for a well-defined Charpy test there is a corresponding ductile–brittle transition temperature T_i, called[22] impact transition temperature. Assuming now the validity of the Doolittle[23] equation

$$\ln \eta = \ln A' + B\frac{v^*}{v^f} \tag{12}$$

where η is the viscosity and A' and B some constants, and invoking the temperature shift factor

$$a_T = \frac{\eta T_{ref}\rho_{ref}}{\eta_{ref}Tp} \tag{13}$$

where ρ is the mass density and the subscript ref pertains to a reference parameter, one obtains

$$K_I = F \exp[-B/(\tilde{v} - 1)] \tag{14}$$

Here F is related to c_b, c_r and A'. Since $\tilde{v}$ depends on temperature, equation (14) constitutes an implicit formula between K_I and T_i; it has been tested against experimental data for low-density PE[22]. The theory summarized in equations (8)–(14) has produced good agreement between predicted and experimental $T_i(K_I)$ values.

We now analyse the Greig–Smith test which provides us with values of L. Of course, a part of this analysis has been already conducted in the section dealing with the rapid crack propagation model and led us to equation (4). However, now we would like to have a model somewhat similar in its structure to that outlined above for the Charpy test.

An analogue of the bond-breaking energy U_b in the Charpy test will be for the Greig–Smith test denoted by U_d (d for destruction). In view of relation (1)

$$U_d = U_1 - U \tag{15}$$

By using equations (4) and (5) after some algebra we obtain

$$U_d = -\frac{L}{L_0} \tag{16}$$

We recall that our computations for PE have provided a negative value of L_0.

Equation (15) is similar to equation (8). We presume that the time dependence of chain relaxation and other non-destructive processes in the Greig–Smith test can be represented by an equation of the form of equation (10b). However, the destructive process in RCP is different than in the Charpy test and equation (10a) does not seem to be applicable. The kinetics, while not quite instantaneous (see micrographs of fracture surfaces in ref. 6 or ref. 8) is different; clearly more work on this aspect of the problem is needed.

Instead of eliminating L_1 between equations (4) and (5), we can eliminate L_0. The result can be written as

$$L = L_1\left(\frac{1}{U} - \frac{1}{U_1}\right) \qquad (17)$$

The last result improves further the perspicuity: we have a different explicit way of seeing the interplay between the Charpy impact energy U and the energy absorption limit U_1 necessary to stop the crack.

If $U > U_1$, with positive L_1 equation (17) leads to negative values of L. It is tempting to interpret such values as an indication of healing capability. However, healing can hardly occur in the presence of an internal pressure. Therefore, we simply interpret negative values of L as crack arrest.

Thus, our analysis of the material response at the molecular level to an external mechanical force has led to equation (14) with all its consequences. We have a relation between the crack length L in a pipe subjected to the Greig–Smith test and the impact energy U determined in the Charpy test. The equation has been tested against an extensive set of experimental data for PE pipes with entirely satisfactory results. The distribution of points around the straight line in *Figure 1* does not reflect experimental accuracy; each point is an average of five tests. Rather, it is a consequence of the minor differences between pipes, as noted above when discussing the reduced volume range. The U_1 value obtained truly represents the entire class of PE pipes.

While the main framework of connections between macromolecular chain behaviour, results of macroscopic tests and rapid crack propagation has largely been constructed, much more remains to be done. In future work the pipe diameter and the pipe thickness have to be taken into account as variables and also related to responses at the polymeric chain level. Along with improved understanding, the use of large-diameter plastic pipes can then become a realistic option.

REFERENCES

1 *Plastic Pipe Line* 1982, 3(2), 1
2 Gaube, E. and Müller, W. F. *Kunststoffe* 1980, **70**, 72
3 Nowak, M. 'Badania wytrzymałości zmeczeniowej rur z tworzyw sztucznych', Wyd. Politechniki Wrocławskiej, Wrocław, 1980
4 Prabhat, K., Maxey, W. A., Hulbert, L. E. and Mamoun, M. M. *Plastic Fuel Pipe Symp. Proc.* 1983, **8**, 128
5 Vancrombrugge, R. *Plastic Fuel Pipe Symp. Proc.* 1983, **8**, 133
6 Gaube, E. and Müller, W. *Plastic Fuel Pipe Symp. Proc.* 1983, **8**, 137
7 Wolters, M. and Ketel, G. *Plastic Fuel Pipe Symp. Proc.* 1983, **8**, 141
8 Gaube, E. and Müller, W. *3R International* 1984, **23**, 236
9 Lustiger, A. and Markham, R. *Polymer* 1983, **24**, 1647
10 Lustiger, A., Chapter 16 in 'Failure of Plastics' (Eds. W. Brostow and R. D. Corneliussen), Hanser, Munich–Vienna–New York, 1986
11 Brostow, W. and Corneliussen, R. D., Chapter 1 in 'Failure of Plastics' (Eds. W. Brostow and R. D. Corneliussen), Hanser, Munich–Vienna–New York, 1986
12 Greig, J. M. and Smith, T. R., paper 3 at the Conference on Designing to Avoid Mechanical Failure, Plastics Institute of London, Cranfield Institute of Technology, 8–10 January 1973
13 Fearnehough, G. D., Jude, D. W. and Weiner, R. T., paper C42/71 at the Conference on Practical Applications of Fracture Mechanics to Pressure-Vessel Technology, May 1971
14 Brostow, W., Chapter 10 in 'Failure of Plastics' (Eds. W. Brostow and R. D. Corneliussen), Hanser, Munich–Vienna–New York, 1986
15 Brostow, W. *Mater. Chem. Phys*, 1985, **13**, 47
16 Brostow, W., 'Science of Materials', Wiley, New York, 1979; Brostow, W., 'Introduccion a la ciencia de los materiales', Limusa, Mexico, D.F. 1981; Brostow, W., 'Einstieg in die moderne Werkstoffwissenschaft', Carl Hanser, München–Wien, 1985
17 Brostow, W. *Polymer* 1980, **21**, 1410
18 Flory, P. J. *J. Am. Chem. Soc.* 1965, **87**, 1833
19 Flory, P. J. *Disc. Faraday Soc.* 1970, **49**, 7
20 Roszkowski, Z. *Mater. Chem. Phys.* 1981, **6**, 455
21 Shih, H. and Flory, P. J. *Macromolecules* 1972, **5**, 758
22 Brostow, W. and Corneliussen, R. D. *J. Mater. Sci.* 1981, **16**, 1665
23 Doolittle, A. K. *J. Appl. Phys.* 1951, **22**, 1741