Modelling the fissuring flat-fracture mode of crack growth in Zr–2.5Nb Candu pressure-tube material

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Modelling of the fissuring mode of fracture in Candu pressure-tube material and, in particular, Stage 1 crack growth (essentially flat J_R curve) as observed in some irradiated compact toughness specimens, is reported. A preliminary attempt has been made at modelling the characteristics of the fracture process zone associated with a crack that tunnels at the specimen mid-section. Based on the experimentally determined J_R curve, both the fracture-process zone size and the crack opening displacement at the trailing edge of the zone have been predicted, and their values are seemingly not unreasonable. The initial considerations enable specific issues to be highlighted which need to be addressed before a complete picture is obtained of the fissuring mode of crack growth. A theoretical analysis has also shown that the non-tunnelling material at the specimen side surfaces exerts very little restraint on the cumulative (tunnelling) mode of crack propagation, a prediction that is consistent with the experimental finding that Stage 1 crack growth in irradiated material is associated with an essentially flat J_R curve.

1. Introduction

The general objective of the Atomic Energy of Canada Limited fracture toughness variability programme, which was initiated several years ago, was to attain an understanding of the microstructural factors responsible for the experimentally observed variability in fracture toughness. The intention was that such an understanding might lead to an improvement in the fracture properties of future pressure tubes and an extension of their service lives. The experimental programme first involved [1] the testing of material from the centres of several pre-selected pressure tubes. The J_R crack-growth resistance behaviour of curved compact toughness specimens (17 mm wide) was found to be highly variable, and this variability was related to the formation of a fractographic feature, called fissures [1, 2], along the crack plane and in the crack direction. It has been argued that the fissures were responsible for a lowering of the J_R curve. Further work [2] has involved the testing of material from the same pressure tubes, but with the material having been irradiated. Again, there was a wide variation in the J_R curve behaviour of curved compact toughness specimens, and this variation was again attributed to fissure formation. Most importantly, there was a clear correlation between the fracture behaviour of material before and after irradiation.

Fissures clearly have a very important effect on the fracture behaviour, and it is highly desirable to quantify, by micro-mechanical modelling, how they affect the shape of the J_R curve. The present paper reports a preliminary attempt at modelling the fracture process zone associated with the fissuring flat fracture mode of crack growth in irradiated pressure-tube material, and also shows that the non-tunnelling material exerts very little restraint on this mode.

2. Relevant experimental observations

With unirradiated material [1], curved compact toughness specimens were tested at room temperature and 240 °C, the specimen geometry in relation to a pressure tube being such that crack growth occurred in a plane containing the axial and radial directions of a pressure tube. The J (deformation integral)–crack-growth (Δa) resistance curves (J_R curves) of material from different tubes exhibited a wide variation at both room temperature and 240 °C. Examination of the fracture surfaces suggested that there was significant crack-front tunnelling at the mid-thickness position, with the extent of tunnelling increasing and through-thickness yielding decreasing with decreasing level of J_R curve. Scanning electron fractography revealed the presence of fissures along the fracture surface (a plane containing the axial and radial directions of a pressure tube), and the fissures were parallel to the direction of crack propagation. They were parallel to the axial or extrusion direction of a tube and delineated strips of

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low-energy fracture (SLEFs) [1] which lie in a plane containing the circumferential and axial directions. The distance between the fissures generally increased, and the fissure length decreased, with increasing level of J_R curve. The regions between the fissures consisted of equiaxed voids. It has been argued [1] that the fissures (SLEFs) divide a specimen into a series of thin ligaments which can then fail easily; in the absence of fissure formation, a high J_R curve is achieved. Experimental evidence [1] has shown that the SLEFs are strips of decohesion on the circumferential–axial plane, this decohesion being associated with the microsegregation of chlorine and carbon (as carbide).

With irradiated material [2], curved compact toughness specimens were again tested at 240 °C; the material was from the same tubes as used for the unirradiated material tests described in the preceding paragraph, and was irradiated to a fast neutron fluence of 2×10^{24} neutrons m^{-2}, prior to testing. The tubes that showed a wide variation in J_R curve behaviour before irradiation also showed a wide variation in behaviour after irradiation with the ranking being similar. While J_R curves for unirradiated material at 240 °C generally exhibited a steep rise followed by a steadily decreasing slope, curves for irradiated material at 240 °C showed a stepped appearance, with an initial sharp reduction in slope close to the onset of crack extension (Stage 1) being followed by a sharp increase in slope (Stage 2) and then a steadily decreasing slope (Stage 3). Discrete increments of crack growth or "crack jumps" ($\gtrsim$0.1 mm) were observed with some specimens especially in Stage 1. As with the unirradiated material, examination of the fracture surfaces suggested that for the majority of specimens there was extensive tunnelling, there being a central tunnelled region of flat fracture with shear lips developing at the specimen surfaces. As with the unirradiated material, the extent of tunnelling increased and through-thickness yielding decreased with decreasing level of J_R curve. The central fracture zone had a very "woody" appearance with the fissures found in unirradiated material again being observed with irradiated material, where their appearance was more enhanced. The general trend was that a lowered J_R curve was associated with increasing fissure length and decreasing fissure spacing. Furthermore, a comparison of irradiated and unirradiated material showed [2] that at the same test temperature, 240 °C, and with material from the same tube, the fissure length was generally greater and the fissure spacing less with irradiated material. With irradiated material, the regions between the fissures again consisted of dimples, but the dimple size was smaller than for unirradiated material.

3. Modelling the effect of fissures

It has been recognized [1, 2] that the failure of material displaying fissures can be broken down into two stages: (a) the formation of strips of low-energy fracture (SLEFs) lying in planes containing the axial and circumferential direction, these SLEFs manifesting themselves as fissures lying in the axial direction in the plane of macroscopic crack growth, (b) the failure of the ligaments between these low-energy fracture strips. From this basis, the fissuring mode of fracture can be considered within Cottrell's categorization [3] of fracture modes.

Cottrell [3] categorized plastic fracture modes into two types: non-cumulative and cumulative. The former, in its extreme form, is typified by the behaviour of a sharp slit in a fully ductile solid opening up under plane strain tension, when the behaviour can be represented by the injection of edge dislocations into the material from the tip along slip lines at about 45 °C to the tensile loading. The plastic notch widens as it deepens, and each dislocation, of Burgers vector $\boldsymbol{b}$, injected into a slip line from the tip, contributes an increment $\boldsymbol{b}/2^{1/2}$ to both the width and depth of the notch. The plastic zones spread much more readily than the notch across the section, and we therefore have a "non-cumulative" situation in which each increment of growth of the plastic notch requires the injection of more dislocations into the slip lines, and these push the existing dislocations further across the section. The plastic zone increases and soon spans the whole load-bearing cross-section, so becoming general yield. Therefore, we have a stable situation provided the applied stress is below general yield levels. With the extreme form of non-cumulative fracture, the slope of the J_R curve is very high, and is manifested by a crack-tip opening angle (CTOA) of 90 °.

Cottrell [3] also recognized that there are other modes of ductile fracture in which the notch can spread in an unstable manner without the general yield stress ever being reached. Such unstable growth can be understood by considering "cumulative" modes [3], so called because the contribution to fracture, made by a dislocation moving through the material ahead of the crack, accumulates continuously with the distance travelled by the dislocation. Cottrell provided two idealized simple examples: (a) sliding-off by a group of screw dislocations and (b) plane stress necking by a group of edge dislocations, both in thin sheets. In both cases, the original group of dislocations needed to start fracture has merely to run through the material without adding more dislocations to its number, and the fracture is unstable; this behaviour contrasts sharply with the "non-cumulative" model, where the fracture is stable. Rapid ductile tearing in thin sheets, e.g. aluminium foil, occurs by such processes. At its extreme form, the J_R curve is perfectly flat when crack propagation proceeds via this mode.

The authors suggest that when fissuring is pronounced, i.e. at the specimen mid-section position, fracture can proceed in this cumulative manner, with the failure of the thin ligaments joining the low-energy fracture strips proceeding by a mixture of the cumulative mechanisms described in the preceding paragraph. For example, narrow ligaments in irradiated material can easily fail by shear or sliding-off when the local crack-tip opening displacement is of the order of the fissure spacing, due to the low work hardening and strain localization. On the other hand, additional necking and/or void nucleation and coalescence is required before failure of wider ligaments. If the whole thickness of a cross-section were able to fail via this cumulative mode, then fracture would be unstable and proceed at a constant J value, with a zero CTOA. Instabilities (discrete increments of unstable crack growth) have, in fact, been observed with irradiated material (see Section 2). Failure of material in a cumulative manner at a constant J value and a zero CTOA is best displayed near the onset of crack extension with irradiated material that exhibits pronounced fissuring. In the next section, we give particular consideration to modelling quantitatively the fracture process zone associated with this cumulative mode of crack extension.

4. Quantitative modelling of the fracture process zone associated with the cumulative mode

Fig. 1 relates the specimen geometry, fracture plane and crack propagation direction to the pressure-tube geometry. In modelling the fracture process zone associated with a crack propagating via the cumulative mode, we consider the model of a process zone associated with a semi-infinite crack in a remotely loaded infinite solid; this is a reasonable basis for our considerations if the process zone is small compared with the test specimen's characteristic dimension (we will return to this point later). The process zone, which is of length R_c, extends from the position where the SLEFs first form, i.e. where the material first fractures, to the position where the ligaments between the SLEFs are completely broken and there is then a complete loss of cohesion (Fig. 2). In modelling the zone, we average the behaviour of the ligaments in such a way that the stress is p_c at the leading edge of the process zone where the SLEFs first nucleate and grow, and where the relative displacement, u, of the material within the zone is zero, while the stress, p, is zero at the trailing edge of the process zone where the ligaments are completely broken, and where the relative displacement, u, is equal to a critical value, u_c. The relation between p and u throughout the process zone will depend on the way in which the ligaments between the SLEFs actually break. Without knowing the precise failure mechanism, we will assume that the relation between p and u is linear (Fig. 3).

Remembering that we are dealing with the case of a semi-infinite crack in a remotely loaded infinite solid, it is easily shown by use of Rice's J-integral methodology [4] that the value of the J integral, i.e. J_c, associated with the propagation of the crack together with its process zone, assuming a general p–u law, is

$$J_c = \int_0^{u_c} p(u)\mathrm{d}u \qquad (1)$$

Unfortunately, for a general p–u law, there is no similarly simple relation giving the size R_c of process zone. However, a detailed analysis [5], which has been substantiated by some more recent as yet unpublished work by one of the authors [6], has shown that for a general p–u law, though R_c is very dependent on the

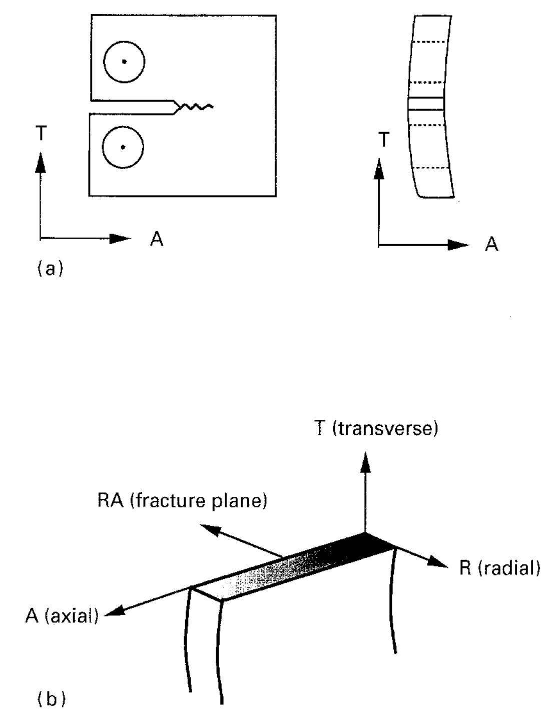

Figure 1 The relation between the specimen geometry, fracture plane and crack propagation direction to the pressure-tube geometry.

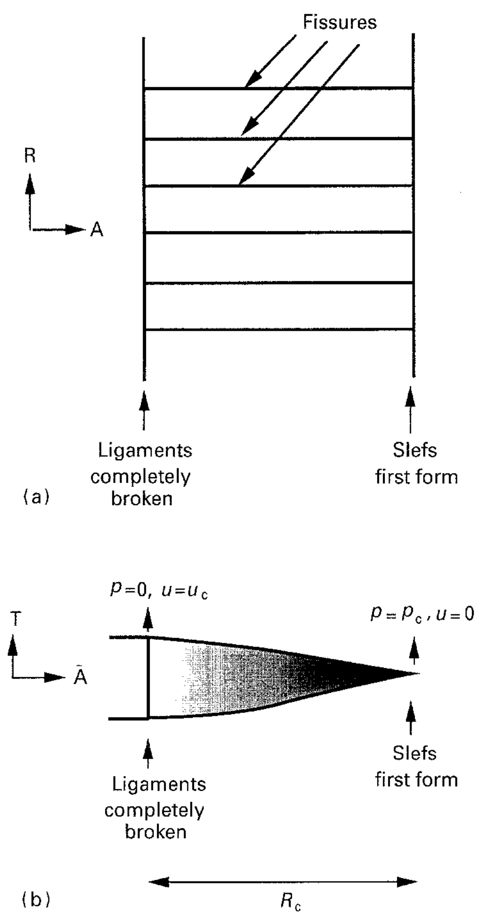

Figure 2 View of the cumulative mode fracture process zone: (a) looking down on fracture surface, and (b) side view. R is the radial or through-wall direction, T is the transverse or loading direction, and A is the axial or crack growth direction. The figure shows the position where SLEFs first form, i.e. where the material first fractures, and where the ligaments are completely broken and there is then a complete loss of cohesion.

maximum stress p_c and the maximum displacement u_c, it is not overly dependent on the details of the p–u law, except when the area under the p–u curve is $\ll p_c u_c$, say less than $0.2p_c u_c$. The relation [7] appropriate to the case where p retains a constant value p_c within the process zone is then a good enough working estimate for R_c, i.e.

$$R_c = \frac{\pi E_0 u_c}{8 p_c} \quad (2)$$

where $E_0 = E/(1 - \nu^2)$, E being Young's modulus and ν being Poisson's ratio. If the area under the p–u curve is $mp_c u_c$ where m is a constant whose value is less than unity ($m = 0.5$ for the linear variation in Fig. 2), then J_c is given by Equation 1 as

$$J_c = mp_c u_c \quad (3)$$

whereupon elimination of u_c between equations 2 and 3 gives

$$R_c = \frac{\pi E_0 J_c}{8mp_c^2} \quad (4)$$

Now an appropriate value for J_c, the J value associated with the cumulative mode of crack propagation, is the value ($\sim$ 15 kJ m^{-2} at 240 °C) associated with Stage I growth in irradiated material showing pronounced fissuring [2], and an appropriate value for E_0 at 240 °C is 95×10^3 MPa. With regard to selecting a value for p_c, the tensile stress within the process zone at the leading edge where the SLEFs first form, we will choose the tensile yield stress of the material, though the p_c value might be less; if p_c were to be greater than the tensile yield stress, then following Cottrell's arguments [3], crack propagation would not be able to proceed via a cumulative mode. Thus with $p_c = 850$ MPa, a typical value for the transverse tensile yield stress of irradiated material at 240 °C [2], $J_c = 15$ kJ m^{-2}, $E_0 = 95 \times 10^3$ MPa and $m = 0.5$ (the linear p–u variation shown in Fig. 3), Equation 4 gives R_c as $\sim$ 1.5 mm; though this is a large value, it is not an unreasonable value. To have an appreciably lower value, we would need a much larger m value (we cannot have a value of $m > 1$) or a higher p_c value, i.e. one in excess of the tensile yield stress, which would, as indicated earlier, present difficulties as regards the ability of a cumulative mode to operate. With $J_c = 15$ kJ m^{-2}, $p_c = 850$ MPa and $m = 0.5$, Equation 3 gives u_c as $\sim$ 35 μm. Because a typical mean fissure spacing, s, is 60 μm for irradiated material [2] showing pronounced fissuring, then $u_c/s \sim 0.6$, which is not an unreasonable value when a thin ligament fails by a ductile fracture mechanism. The R_c value of 1.5 mm is small compared with the initial crack depth or initial ligament width of the test specimen ($\sim$ 8.5 mm), and consequently use of the results from the analysis of a semi-infinite crack in a remotely loaded infinite solid is vindicated. Interestingly, an R_c value of 1.5 mm is of the same order as the maximum fissure length [2].

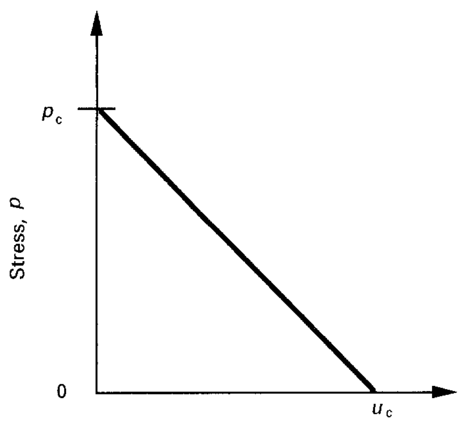

Figure 3 The relation between the stress, p, and the displacement, u, within the fracture process zone.

5. The restraining effect of the non-tunnelling material on cumulative crack propagation

As indicated in Section 2, with irradiated material that displays extensive fissuring, the experimental evidence suggests [2] that the central section of a specimen fractures by the cumulative mode, with the crack front tunnelling forward and leaving regions of material, at the specimen sides, that do not fail so readily because fissuring is less pronounced, or not present at all, because of insufficient through-thickness tensile stress. That being the case, it is necessary to demonstrate that the non-fracturing material exerts little or no restraining effect on the cumulative mode of crack propagation; this is the objective of the present section's analysis.

Consider the idealized model shown in Fig. 4, where there is a semi-infinite crack in an infinite solid; the tunnelled region is as shown, with the non-fracturing material being shown by the shaded regions. The tunnelled region is of width $2w$ and length $2l$, the

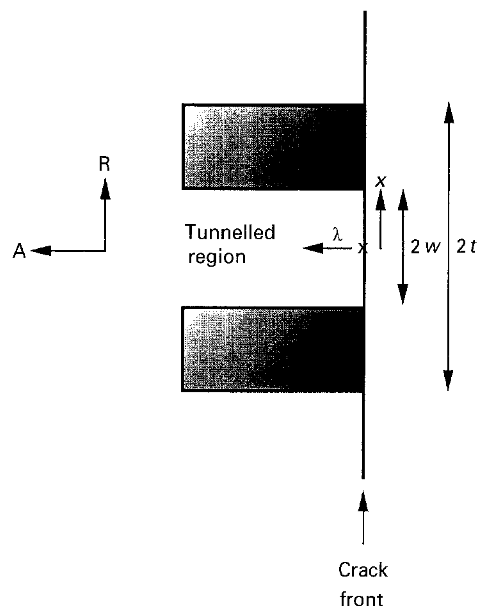

Figure 4 The model used to quantify the restraining effect of the non-tunnelling material on cumulative crack propagation. R is the radial or through-wall direction, and A is the axial or crack growth direction.

non-fracturing regions are each of width $(t-w)$ and length $2l$, with $2t$ simulating the specimen thickness. We will calculate the stress intensity at X due to the restraining effects of the non-fracturing regions, assuming that these regions exert restraining stresses of magnitude Y on the faces of the semi-infinite crack. This stress intensity is viewed to be a measure of the restraining effect of the non-fracturing material, and when converted into a corresponding J value, we will show that it is small compared with the value of J ($\sim 15\,\mathrm{kJ\,m^{-2}}$) associated with the cumulative mode of crack propagation. (With this approach we are collapsing the fracture process zone associated with the cumulative mode, a not unreasonable assumption when the process zone size is small in comparison with the specimen dimensions).

We use the solution [8] for point forces of magnitude P applied to both faces of the crack at a vertical distance x and a horizontal distance λ from X

$$K = \frac{2^{1/2}P}{\pi^{3/2}}\frac{\lambda^{1/2}}{(\lambda^2 + x^2)} \tag{5}$$

The restraining stress intensity, K_R, at x due to the non-fracturing material consists of two contributions: a positive contribution, K_1, due to a stress, Y, acting over an interval of width $2t$ and length $2l$, and a negative contribution, $-K_2$, due to a stress, $-Y$, acting over an interval of width $2w$ and length $2l$. On the basis of Equation 5

$$K_1 = \frac{2^{3/2}Y}{\pi^{3/2}}\int_{\lambda=0}^{2l}\int_{x=0}^{t}\frac{\lambda^{1/2}\,d\lambda dx}{(\lambda^2 + x^2)} \tag{6}$$

whereupon

$$K_1 = \frac{2^{3/2}Y}{\pi^{3/2}}t^{1/2}\int_{\theta=\phi}^{\pi/2}\theta\operatorname{cosec}^2\theta(\tan\theta)^{1/2}\,d\theta \tag{7}$$

with $\tan\phi = t/2l$. Evaluation of the integral leads to the result

$$\frac{\pi^{3/2}K_1}{4\times 2^{1/2}Y t^{1/2}} = F(\phi) \tag{8}$$

with $F(\phi)$ being given by the expression

$$F(\phi) = \frac{\pi}{2^{1/2}} + \frac{\phi}{(\tan\phi)^{1/2}} - \frac{1}{2^{3/2}}\ln\left[\frac{\tan\phi + (2\tan\phi)^{1/2} + 1}{\tan\phi - (2\tan\phi)^{1/2} + 1}\right] - \frac{1}{2^{1/2}}\{\tan^{-1}(2\tan\phi)^{1/2} + 1] + \tan^{-1}[(2\tan\phi)^{1/2} - 1]\} \tag{9}$$

Proceeding along similar lines, we have

$$\frac{\pi^{3/2}K_2}{4\times 2^{1/2}Y t^{1/2}} = F(\psi) \tag{10}$$

with $\tan\psi = w/2l$ and $F(\psi)$ being given by the expression

$$F(\psi) = \frac{\pi}{2^{1/2}} + \frac{\psi}{(\tan\psi)^{1/2}} - \frac{1}{2^{3/2}}\ln\left[\frac{\tan\psi + (2\tan\psi)^{1/2} + 1}{\tan\psi - (2\tan\psi)^{1/2} + 1}\right] - \frac{1}{2^{1/2}}\{\tan^{-1}(2\tan\psi)^{1/2} + 1] + \tan^{-1}[2\tan\psi)^{1/2} - 1]\} \tag{11}$$

The restraining stress intensity K_R, at x is given by the expression

$$K_R = K_1 - K_2 \tag{12}$$

with K_1 and K_2 being given respectively, by Equations 8 and 10.

Table I shows K_1 for a range of $2l/2t$ values. With $2t$ simulating the specimen thickness (~ 4 mm) and $2l$ being the extent of tunnelling, a typical value of $2l/2t$ is 0.160 (the top figure in the first column), a value that is appropriate to a crack which has extended ~ 0.5 mm (an approximate limit [2] to the end of Stage 1), whereupon $K_1/Yt^{1/2} = 0.840$. With $w = t/4$, i.e. 25% of the cross-section consists of non-fractured material at each surface, and noting that Equation 11 is of the same form as Equation 9, it follows from the results in Table I that $K_2 Y w^{1/2} = 1.388$. It therefore follows that

$$\begin{aligned} K_R &= K_1 - K_2 \\ &= 0.840\,Yt^{1/2} - 1.388\,Yw^{1/2} \\ &= 0.146\,Yt^{1/2} \end{aligned} \tag{13}$$

with $w = t/4$. Thus with Y being the tensile yield stress (850 MPa) and with $E_0 = 95\times 10^3$ MPa, $t = 2$ mm, and noting that the J_R value corresponding to K_R is $J_R = K_R^2/E_0$, it follows that $J_R = 0.32\,\mathrm{kJ\,m^{-2}}$ which is more than an order of magnitude less than the J_c value of $\sim 15\,\mathrm{kJ\,m^{-2}}$ for the operation of the cumulative mode, as measured experimentally. We therefore predict that the non-fracturing material exerts very little restraining effect on the cumulative mode of crack propagation (Stage 1), a prediction that is consistent with the experimental findings. The limited restraining effect is due primarily to the fact that the non-fracturing material deforms plastically, thereby relaxing the restraining stresses.

TABLE I K_1 for a range of $2l/2t$ values

$2l/2t$	ϕ	$F(\phi)$	$K_1/Yt^{1/2}$
0.160	1.263	0.827	0.840
0.318	1.004	1.092	1.110
0.636	0.666	1.366	1.388
1.274	0.374	1.601	1.627
2.544	0.194	1.779	1.808

6. Discussion

This paper has been concerned with modelling the fissuring flat-fracture mode of crack growth in irradiated pressure-tube material against the background of the experimental results obtained by Davies *et al.* [2]. The basic starting point is Davies *et al.*'s recognition [1, 2] that failure of material displaying fissures can be broken down into two stages: (a) the formation of strips of low-energy fractures (SLEFs) lying in planes containing the axial and circumferential directions, and (b) the failure of the ligaments between the low-energy fracture strips. In this paper, it has been argued that Stage 1 (essentially zero J_R curve slope) crack growth in irradiated material proceeds by an essentially cumulative mode, using Cottrell's terminology [3], whereby non-linearity of material behaviour is confined to a thin fracture process zone ahead of the crack tip, and the analyses in Sections 4 and 5 have provided quantitative underpinning for this picture.

Firstly, based on the experimentally measured value of the J value, J_c, associated with the cumulative mode, we have made (Section 4) predictions of the fracture process zone size and the crack opening at the trailing edge of the process zone, and have argued that the predicted values are not unreasonable. However, the considerations in Section 4 do put the focus on several issues which are worthy of further consideration.

1. In the analysis in Section 4, it has been assumed that the tensile stress, p_c, within the process zone at its leading edge, where the SLEFs presumably first form, is the material's tensile yield stress.

2. Related to issue (1) is the criterion for formation of a SLEF, because this might impact upon the p_c value. Is plastic deformation important with regard to SLEF formation? If it is not important, then how do we explain why SLEF formation and fissuring is more pronounced with irradiated material? The enhanced yield stress of irradiated material may, of course, be an important factor.

3. In Section 4, it has been assumed for the sake of simplicity that the relation between the stress, p, and the relative displacement, u, within the process zone is linear. There is a need to underpin (or modify) this assumption by, say, inputting the detailed way in which the ligaments break.

4. In Section 4 we have concentrated on the behaviour of irradiated material. We must find an explanation as to why unirradiated material, though exhibiting a tunnelling behaviour, does not show a Stage 1 flat J_R curve; presumably it is because a cumulative mode is unable to operate in the same way as it does with irradiated material. Is this because fissuring is less pronounced, or is it because ligament rupture with unirradiated material cannot proceed without there being some element of non-cumulative crack extension? What role does flow localization play with regard to this issue?

These are all issues which have to be addressed before we have a complete picture of the fracture process zone associated with cumulative crack propagation.

The second important aspect of Stage 1 crack propagation examined in the paper is the restraining effect of the non-tunnelling material on cumulative crack propagation. Here there would appear to be less uncertainty, because the analysis in Section 5 has shown that the non-tunnelling material exerts very little restraint on the cumulative model of crack propagation, a prediction that is consistent with the experimental finding that Stage 1 crack growth in irradiated material is associated with an essentially flat J_R curve.

Before concluding this paper, it should be emphasized that the paper has focused on the crack-growth resistance behaviour of irradiated material that displays fissuring, and more especially, Stage 1 crack growth. After a limited amount of Stage 1 growth, there follows a sharp increase in the J_R curve slope (Stage 2) which, in turn, is followed by a steadily decreasing J_R curve slope (Stage 3) [2]. Stage 1 changes over to Stage 2 presumably because the constraint at the crack tip falls sufficiently to prevent continued fracture of the fissuring material by a purely cumulative mode and/or because continued crack growth then involves non-cumulative growth of the non-tunnelling material away from the specimen mid-section. Stage 2 presumably continues until the proportion of this material fracturing increases to such an extent that the failure mode becomes completely slant, and then we have an essentially cumulative crack growth mode of the type described in Section 3 which is associated with a zero J_R curve slope. As regards the importance of the J_R curve in the context of determining the critical axial crack length for a Candu pressure tube, it is likely that it is primarily the Stage 2 slope and the transition to Stage 3 that have the major effect. However, the various stages are coupled in the sense that the existence of Stage 1 and tunneling facilitates (in the authors' view) a short Stage 2 and, consequently, a low maximum J value; such a coupling is the driving force for understanding the fissuring mode of failure and its impact upon the overall J_R curve behaviour.

7. Conclusions

1. This paper has provided a preliminary attempt at quantitatively modelling the fracture process zone associated with the fissuring cumulative flat fracture mode of crack growth in compact toughness specimens (irradiated pressure tube material). Based on the experimentally determined J_R curve, the fracture process zone size and the crack opening displacement at the trailing edge of the zone have been predicted and their values are seemingly not unreasonable.

2. A theoretical analysis has shown that the non-tunnelling material at the specimen side surfaces exerts very little restraint on the cumulative (tunnelling) mode of crack propagation at short crack extensions, a prediction that is consistent with the experimental finding that Stage 1 crack growth in irradiated material is associated with an essentially flat J_R curve.

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