STABLE DUCTILE CRACK GROWTH IN REACTOR PRESSURE VESSEL STEELS

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Summary

It is generally believed that steels, such as A533B, currently used in nuclear reactor pressure vessels, have a high resistance to ductile fracture initiation when they are tested in the upper-shelf region which includes reactor operating temperatures. Furthermore, the steels exhibit stable crack growth, and this is usually viewed as a bonus, from the viewpoint of pressure vessel integrity. A careful consideration of the behaviour of A533B, however, suggests that ductile fracture could be initiating at fairly low K_I values, and that the fracture resistance of A533B in the upper-shelf region might in fact be primarily due to a very high resistance to crack growth at very small crack extensions. This conclusion is reached mainly as a result of comparing the behaviour of A533B in the upper-shelf and transition regions, and making use of the fact that the macroscopic deformation characteristics are essentially temperature independent over a temperature range that spans the boundary between these two regions.

These arguments focus on the importance of stable crack growth, and it is clearly important to see whether there is some parameter that characterizes the growth behaviour of a crack, and which has a constant value during growth; this will enable stable crack growth to be incorporated within a rational methodology for predicting the mechanical response of a vessel containing a crack. Consideration is given to this problem using appropriately modified versions of the Dugdale-Bilby-Cottrell-Swinden strip yield model; the analysis suggests the desirability of using crack tip opening angle as a characterizing parameter. This is in accord with the conclusion reached in the EPRI ductile fracture research programme, which was based on the matching of experimental, load point displacement-crack extension data and finite element results, with various local and global parameters being assessed for their suitability as characterizing parameters.

The models used in the present investigation are characterized by three regions of material behaviour: (a) an elastic region, (b) a plastic region, and (c) a crack tip end region or fracture process zone within which fracture processes operate. Use of the simple models allows one to focus on the fracture process characteristics, and the comparative behaviours of unirradiated and irradiated material are examined, with particular regard to their relative tendencies for unstable fracture. In this context, the role of flow concentration in irradiated material is given special consideration; it is shown that there is likely to be a greater tendency for unstable fracture in irradiated material.

1. Introduction

It is generally believed that steels, such as A 533 B, currently used in nuclear reactor pressure vessels, have a high resistance to ductile fracture initiation when they are tested in the upper-shelf region which includes reactor operating temperatures. Furthermore, the steels exhibit stable crack growth, which is usually viewed as a bonus from the viewpoint of pressure vessel integrity. A careful consideration of the behaviour of A 533B, however, suggests that ductile fracture could be initiating at fairly low K_I values, and that the fracture resistance of A 533 B in the upper-shelf region might in fact be primarily due to a very high resistance to crack growth at very small crack extensions. These arguments underscore more than ever the necessity for incorporating stable crack growth within a rational methodology for predicting the mechanical response of a vessel containing a crack. This paper surveys the main points arising from the author's recent theoretical research which addresses this problem.

2. Appraisal of Some Recent Experimental Results on Crack Growth in A 533 B

The upper-shelf region for unirradiated A 533 B commences at around +50F, and laboratory specimen tests and also a limited number of model vessel tests have shown that plane strain stable crack growth can proceed from a pre-existing crack in the upper-shelf region. The fracture mode is ductile, electron microscopic observations showing that fracture involves the formation of voids at dispersed particles, followed by their growth and linkage with the crack tip; this results in a fracture surface that is covered with dimples.

The occurrence of this stable growth focusses attention on the value of K (or equivalent J) at the onset of crack extension, here being defined as to exclude stretch zone extension, which is essentially a manifestation of crack tip deformation and not a true fracture process. If this K value is high, the stable crack growth occurring at higher K values can be viewed as being a bonus, if this K value is used as input to safety assessment procedures; on the other hand, if this K value is low, stable crack growth is being relied upon to provide protection against failure in service. It is generally believed that K values for the initiation of plane strain crack growth within the upper-shelf region are high ($\gtrsim$ 150 Ksi $\sqrt{\text{in.}}$) these values have been obtained by a variety of methods including the measurement of J as a function of the crack extension Δa, and then regarding J_{Ic} as the limiting value of J as Δa tends to zero.

It is interesting to see how these results for the initiation of plane strain ductile crack growth within the upper-shelf region compare with experimental results obtained at slightly lower temperatures. With the aid of compact tension specimens, Wessel(1) has measured K_{Ic} for A 533 B in the temperature region (-50F to +50F) and found a rapid increase in K_{Ic} with increasing temperature. Metallographic observations on the fracture surfaces show a stretch zone at the fatigue starter crack tip, followed by a ductile fracture zone as is evidenced by the presence of dimples, and this zone is followed by a cleavage region. There are no reported metallographic observations for partially loaded K_{Ic} type specimens tested within this temperature range, and one can only speculate as to the precise nature of this ductile fracture. There would seem to be four likely possibilities(2): a) it is unstable and is followed by unstable cleavage, K_{Ic} referring to the onset of unstable ductile crack extension; b) it is stable, followed by unstable ductile crack extension and then by unstable cleavage, K_{Ic} referring to the onset of unstable ductile crack extension; c) it is stable, followed by unstable cleavage crack extension, K_{Ic} referring to the onset of unstable cleavage crack extension; d) cleavage fracture is initiated ahead of the fatigue crack and stretch zone, with the intervening ligament failing by a ductile mechanism, and the region ahead of the initiation site failing by cleavage, K_{Ic} in this case referring to the formation of a cleavage crack. Ductile fracture, associated with void formation and growth, is very dependent on the material's plastic deformation characteristics; these are essentially temperature independent, at least in a macroscopic sense, within the transition region(1), and it is therefore difficult to accept that the measured K_{Ic} values, being very sensitive to temperature in this region, refer to the onset of unstable ductile crack extension. On this reckoning, possibilities a) and b) can be ruled out. Possibility c) would appear to be more likely than d), and it would thus seem that plane strain ductile crack growth can initiate within the transition region at low K values ($\sim$ 50 Ksi $\sqrt{\text{in.}}$)

If this interpretation is correct, there are important implications, since the macroscopic plastic deformation characteristics of A 533 B are essentially temperature independent well into the upper-shelf region(1). On the basis of the transition temperature range results, some stable crack growth ought to occur within the upper-shelf region at similarly low K values, yet as indicated earlier in this section, there are no reported observations of such growth at low K values. This apparent incompatibility between the experimental results, or rather the author's interpretation of these results, for the transition and upper-shelf regions focusses attention on what is actually meant by the onset of stable ductile crack extension in the context of J_{Ic} measurements, and the necessity of carefully studying whether there is indeed any stable ductile crack extension, aside from the stretch zone extension, at low K values in the upper-shelf region. If it did transpire that stable crack extension does occur at low K values, albeit with J increasing extremely rapidly with Δa at small Δa, the implication would be that the high fracture resistance of A 533 B in the upper-shelf region relies on the occurrence of stable crack growth.

3. Characterising Parameters for Plane Strain Ductile Crack Growth

From the previous section's considerations, it is quite clear that plane strain ductile crack growth plays an important role in the fracture of pressure vessel steels; it is therefore essential to develop a methodology for dealing with ductile crack extension in order to predict the mechanical response of a structure containing a crack. When a crack extends in a ductile mode, there are three regions of material behaviour: a) an elastic region, b) a plastic region, and c) a crack tip end region (fracture process zone) within which fracture processes operate. The characteristics and size of this end region depend on the material under consideration, and are governed by the nature and dispersion of the material inhomogeneities that are responsible for the material losing its cohesion; as regards the materials of interest in this investigation, these inhomogeneities are inclusions and second phase particles. It is sometimes assumed(3) that the end region state is autonomous during stable growth, in that the local deformation pattern referred to the moving crack tip remains constant. At first sight, this is an attractive assumption, since it suggests that crack growth ought to be characterised by some parameter, that describes the end region, and which has a constant value during growth; however, it has been suggested (4) that the validity of the autonomy assumption should not be taken for granted.

Against this background, the author has undertaken a preliminary examination of this question of crack tip autonomy using, in the first instance, the Dugdale-Bilby-Cottrell-Swinden (DBCS) type model[5,6] to describe the growth process. Although such a model is highly idealised and any conclusions might therefore be dubious when viewed in isolation, the results should have value when considered in conjunction with results from more sophisticated models. When the simple version of the DBCS model, which does not include a fracture process zone, is applied to plane strain mode I and small-scale yielding situations, the plastic zone size w is independent of both the prior load pattern and crack length increments; it is therefore often believed that the DBCS model cannot be used to discuss stable crack growth. However, as Rice has emphasised[7], stable growth effects can be introduced into the DBCS model by arguing[8] that only the relative displacement Φ_c accumulated while a material point is within some critical distance Δa from the crack tip should enter the failure criterion (the failure criterion generally used in the DBCS model is based on the crack tip displacement Φ_f); Δa is supposed to be appreciably smaller than the plastic zone size w, and can be regarded as the size of the crack tip end region within which fracture processes operate. With this new failure criterion, it is easily shown that the state of the crack tip end region is not autonomous. Cottrell[9] has also used the DBCS model to represent crack growth and has distinguished between cumulative growth where no additional crack tip displacement is associated with continued growth, and non-cumulative growth where there is additional displacement. If the fracture process is such that the crack tip displacement increases by $\delta \Phi_f = \lambda \delta c$ for a crack growth increment δc where λ is a constant, it can again be demonstrated that the end region state changes during crack growth.

Consequently, by using two plausible fracture criteria in conjunction with the DBCS model, it has been shown that the region in the immediate vicinity of a stably growing crack tip need not necessarily be autonomous. In appraising this conclusion, it must again be emphasised that the model is extremely idealised since it neglects, for example, the wake of plasticity behind a moving crack tip and the induced incompatible elastic strains, which are features of more realistic crack growth models[7] that allow for plastic deformation away from the crack plane. Nevertheless, the results do suggest that the autonomy problem should be investigated via these more realistic models; furthermore, use of the present simple model does focus attention on the crucial importance of the fracture process characteristics in stable growth. Most stable crack growth discussions consider the region away from a crack tip and then work inwards towards the tip, whereas the fracture process zone behaviour controls the overall crack growth behaviour; recognition of this important feature is central to the present investigation's approach.

The conclusion that the crack tip end region need not necessarily be autonomous during stable growth, could be taken to imply that one might be unable to develop a stable crack growth methodology based on some crack tip characterising parameter having a constant value; however, this need not be the case. Thus, following Rice[7], and regarding $d\Phi_f/dc$ as the crack tip opening angle, this remains constant with both fracture criteria considered so far in this investigation, even though the crack tip deformation pattern is not autonomous. This, of course, points to the possibility of using crack tip opening angle as a characterising parameter for ductile crack growth; it is therefore interesting to note that the EPRI ductile fracture research programme concludes[10] that crack tip opening angle is the most promising parameter for characterising plane strain stable crack extension in the ductile fracture regime. The EPRI programme was based on the matching of experimental load point displacement-crack extension data with finite element results, various local and global parameters being tested for their suitability as characterising parameters. The present investigation's conclusions are in accord with those of the EPRI investigation, and are complementary in that they have been obtained for the small-scale yielding condition, whereas the EPRI results refer primarily to the widespread yielding **situation.**

A feature of the present investigation's approach is that crack growth criteria are expressed in terms of parameters which are related to the microstructural fracture processes; the critical parameter (Φ_c or λ) is the local deformation associated with failure between inhomogeneities whose spacing is Δa. This approach thereforr provides a basis for interpreting effects of, for example, temperature and neutron irradiation. It should also be noted that crack tip opening angle is a parameter which is local in an average sense; this could be an important reason why it might be a more satisfactory characterising parameter than one which is sensitive to the precise deformation pattern in the immediate vicinity of a crack tip. This particular point, with regard to the crack tip separation energy rate parameter, has been examined by the author.

In formulating the crack tip separation energy rate approach, Kfouri and Miller[11] recognise that there is a finite size scale associated with the crack tip separation process and the approach computes, at the continuum level, the work done in quasi-statically and proportionally reducing to zero the stresses that act over a finite crack growth step Δa; this distance is a material characteristic and could be the fracture process zone size. If ΔW is the work (per unit length of crack front) of unloading the initially stressed segments of crack wall, the separation energy rate $G^{\Delta} \equiv \Delta W/\Delta a$. For crack growth, G^{Δ} must equal the separation work G_c^{Δ} which is also considered to be a material characteristic. Kfouri and Miller used finite element procedures to study the growth process, but Kfouri and Rice[12] later employed the DBCS model, comparing their analytical results for G^{Δ} with those of Kfouri and Miller; the two sets of results were mutually reinforcing in that the general trend of G^{Δ} with Δa, as predicted numerically, is well-matched by the analytical calculations. There are major assumptions inherent in the crack tip separation energy rate approach; these concern: a) the decoupling of the energy transfers between the fracture process zone and the surrounding plastic region; b) the calculation of the work done by quasi-statically and proportionally reducing the stresses acting over a finite crack growth step Δa. In view of these assumptions, it is desirable to assess the accuracy of the approach; with this objective in mind, the author has given consideration to the crack tip separation energy rate approach.

Again consider the DBCS type model in which a crack tip is subject to the mode I stress intensification factor K. There is a fracture process zone of length Δa at the crack tip, and beyond this zone there is a plastic region of length $(w - \Delta a)$ within which the yield stress is Y; the cohesive stress within the fracture process zone is qY where q is a constant, while the displacement where the fracture process and plastic zones join is Φ_s, and the crack tip displacement is Φ_f. The DBCS model, as employed here, is clearly more elaborate than the model considered earlier in this section. When small-scale yielding conditions are operative and the fracture process zone size is small in comparison with the plastic zone size, the fracture process zone is autonomous, and the plastic zone size and crack tip stress intensifications have unique values. The DBCS model, as used in this manner, refers to the steady state situation. The deformation pattern within the process zone, the plastic zone size, and crack tip stress intensification K were also found to be independent of q. The major difference between the author's expression for K and that obtained by Kfouri and Rice is that the present investigation's analysis gives a crack growth condition that is insensitive to the detailed characteristics within the fracture process zone (i.e. it is independent of q), whereas the crack tip separation energy rate approach as developed by Kfouri and Rice gives a condition that is very sensitive to these characteristics (i.e. it is dependent on G_c^{Δ}).

The present study of the DBCS model, modified to include a fracture process zone, therefore suggests that even in the steady state situation the crack tip separation energy rate approach might only be approximate as regards its predictions of the crack extension condition. Furthermore, since the earlier work described in this section has indicated that the deformation pattern in the immediate vicinity of a crack tip might not be autonomous in the non-stationary state, the general applicability of crack tip separation energy rate as a characterising parameter for stable crack growth must be seriously questioned.

4. Some Comments on the Behaviour of Irradiated Material

In Section 2 it was proposed that the high fracture resistance of unirradiated A 533 B in the upper-shelf region could be due either to a high K for initiation with an additional contribution from resistance to crack growth, or to the crack growth resistance being especially high in the early stages of growth with the initiation K value not necessarily being high. From a practical viewpoint, the difference between these two viewpoints is probably not important. However, it is important to know whether irradiated material follows a similar behaviour pattern. If it is argued that unirradiated material has a relatively low K value for ductile fracture initiation, and that its fracture resistance within the upper-shelf region is dependent primarily on the resistance to crack growth, and if it is additionally assumed that irradiated material behaves similarly and also has a relatively low K value for ductile fracture initiation, in presuming that irradiated material has a high fracture resistance, one is implying that crack growth is very difficult in irradiated material.

Such considerations focus attention on the importance of assessing the ductile crack growth resistance of irradiated material in the upper-shelf region. In the first instance, crack growth resistance depends on whether or not cumulative propagation is possible. Examination of the conditions for continued cumulative crack propagation has led the author to the conclusion that the critical parameter is $(E/Y)\ d\Phi/dc$, where Y is the yield stress, E is Young's modulus, and $d\Phi/dc$ is a measure of the deformation increment associated with a crack growth increment, and is equivalent to the crack tip opening angle. Continued cumulative fracture is preferred if $(E/Y)\ d\Phi/dc$ is small, whereas there is likely to be a transition to a non-cumulative mode if the parameter is large. Although it is difficult to ascertain the critical value of the parameter at which the transition occurs, the relative behaviour of irradiated and unirradiated material can be discussed and one immediately notes the higher yield stress of irradiated material. In considering possible differences in $d\Phi/dc$ values, the detailed crack growth mechanisms must be examined. As indicated already in this paper, the fracture mode of unirradiated A 533 B within the upper-shelf region is dimpled rupture; microscopic voids form at dispersed phases, and the voids then grow and link with the crack tip. The fracture surface in fact exhibits a duplex dimple size distribution(10). Large dimples are due to voids which nucleate at the large 5 to 10 μ size manganese sulphide and aluminium oxide inclusions; this type of void nucleation is particularly difficult in A 533 B as compared with high strength steels, and this has been suggested as the most likely reason why A 533 B is so tough. The voids around these inclusions grow with increasing plastic strain, but before they impinge upon each other, smaller voids are nucleated at very small precipitates within bands that form between the larger voids; this void sheet failure mechanism is not peculiar to A 533 B but has been observed in other materials(13). The fracture process therefore consists of three stages: void nucleation, void growth and void sheet coalescence, the latter truncating the void growth process. Irradiated material is likely to be more susceptible to the void sheet failure mechanism than unirradiated material, since flow concentration within bands is known to occur in many materials that have been subject to neutron irradiation; this is due to the leading dislocations removing the debris created by neutron irradiation and thereby facilitating the passage of subsequent dislocations. $d\Phi/dc$ should therefore be lower for irradiated material. Coupling this effect with that of the yield stress, $(E/Y)\, d\Phi/dc$ should be lower for irradiated material than unirradiated material, and consequently there is likely to be a greater tendency for unstable fracture in irradiated material. This conclusion has been reached by considering the behaviour of a propagating cumulative crack and the possible transition to a non-cumulative mode; it is important that the arguments be extended to the ductile fracture initiation stage, which will be influenced by the deformation associated with stretch zone formation. This further investigation should also be coupled with an assessment of the differences between contained yield and general yield situations, as there are good reasons for expecting different behaviours in the two cases.

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