STP426-EB/Dec. 1967

General Discussion

A. C. Roberts[1]—In one group of papers presented in this symposium we heard a great deal about the role played by helium in reducing the postirradiation ductility of austenitic steels at elevated temperatures. Lithium, which is the other product of the ^{10}B transmutation, has been ignored and I would like to give some justification for this omission. Higgins and Roberts[2] at Atomic Energy Research Establishment, Harwell, have irradiated 0.001-in.-thick tension specimens of 20Cr-25Ni,Cb stabilized steel with lithium and helium ions, respectively. The specimens were irradiated from both sides and the beam energy was reduced in steps so that 50 layers of the impinging ion were produced in the specimens. Control and irradiated specimens were then tested in argon at 750 C at an initial strain rate of 4×10^{-4} sec^{-1}.

Figure 1 shows the effect of increasing helium and lithium concentrations on the tensile ductility at 750 C. Some tensile results of Barton and Pfeil[3] on the effect of helium produced by thermal neutron transmutation of ^{10}B in the same steel with the same grain size are included for comparison.

The ion irradiation results show that lithium in the absence of helium has no significant effect on the tensile ductility at 750 C. Helium, however, causes a similar decrease in ductility to that observed after neutron irradiation and appears to be primarily responsible for the reduced ductility. Work is in progress to determine the effect of lithium ion irradiations on the ductility of steel previously irradiated with helium ions because it is possible that lithium may alter the bubble matrix interfacial energy and allow the bubble size to increase more readily.

G. J. C. Carpenter[4]—Are you satisfied that the results quoted for specimens 0.001 in. thick are typical of results in bulk material.

Mr. Roberts—The grain size (3 μ) of the foil tension specimens allows a minimum of 8 grains through the thickness of the specimen. This number is sufficient to give excellent reproducibility in specimens given the same treatment. Therefore although the absolute value of the ductility may not be the same as in a conventional tension specimen the work

[1] Metallurgy Division, Atomic Energy Research Establishment, Harwell, Berks., England.

[2] Higgins, P. R. B., and Roberts, A. C., unpublished work.

[3] Barton, P. J., and Pfeil, P. C. L., private communication.

[4] Cambridge University Metallurgy Department, Cambridge, England.

clearly illustrates the relative effect of small amounts of helium and lithium in stainless steel.

R. W. Nichols[5]—A particularly rewarding aspect of this symposium has been the use in a number of the papers of fracture mechanics concepts as a basis for the engineering interpretation of irradiation effects data. I support fully the viewpoint that the important step in the avoidance of catastrophic fracture is that of preventing the onset of unstable growth of a crack, rather than attempting to provide conditions under which such a crack will arrest. Tests have been made[6] by the

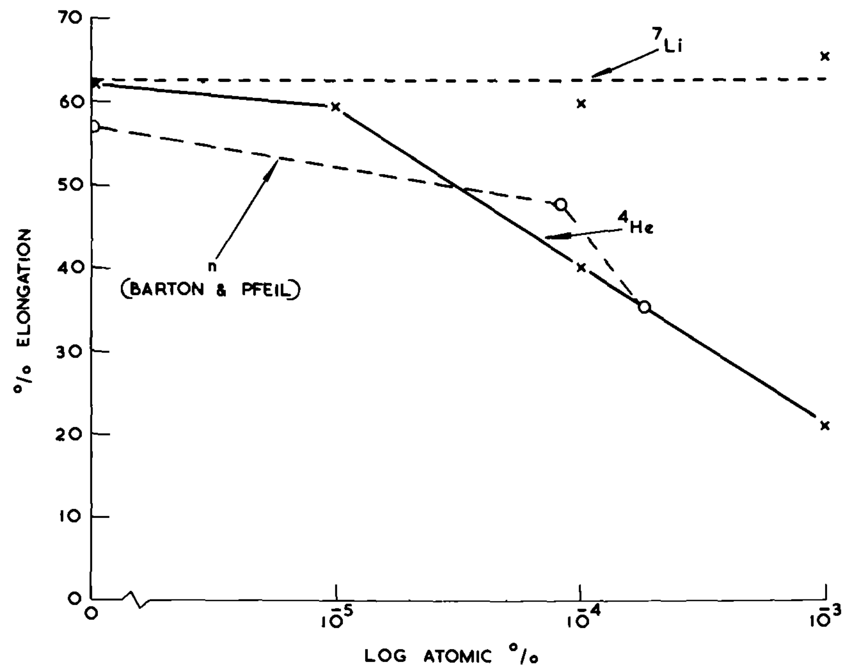

FIG. 1—*Effect of ^4He or ^7Li on the tensile ductility of austenitic steel at 750 C.*

United Kingdom Atomic Energy Authority on internally pressurized 5-ft.-diameter 1-in.-thick steel cylinders in which unstable cracks were initiated from artificial defects, and it was found[7] that, while a crack arrest temperature effect could be demonstrated with fully hydraulic pressurization, it was not possible to arrest an unstable crack even at very high temperatures when pressurized by a compressible fluid.

[5] Deputy head of laboratory, United Kingdom Atomic Energy Authority, Reactor Materials Laboratory, Culcheth, Nr. Warrington, Lancashire, England.

[6] Bevitt, E., Cowan, A., and Stott, A. L., *Journal of the British Nuclear Energy Society*, 1964, Vol. 3, No. 1.

[7] Irvine, W. H., Quirk, A., and Bevitt, E., *Journal of the British Nuclear Energy Society*, 1964, Vol. 3, No. 1.

We believe that crack arrest occurs only under conditions where extension of the crack can cause some relaxation of the stress across the crack, and this will not be the case for a reactor pressure vessel, pressurized by gas, boiling or pressurized water (this last case in practice containing pockets of steam). An implication of this work is that the isothermal crack arrest temperature does not have the same significance with respect to pressure vessel failure as was previously claimed, although of course the conditions for initiation do change somewhat over a range of temperatures around that of the crack arrest temperature. Similar

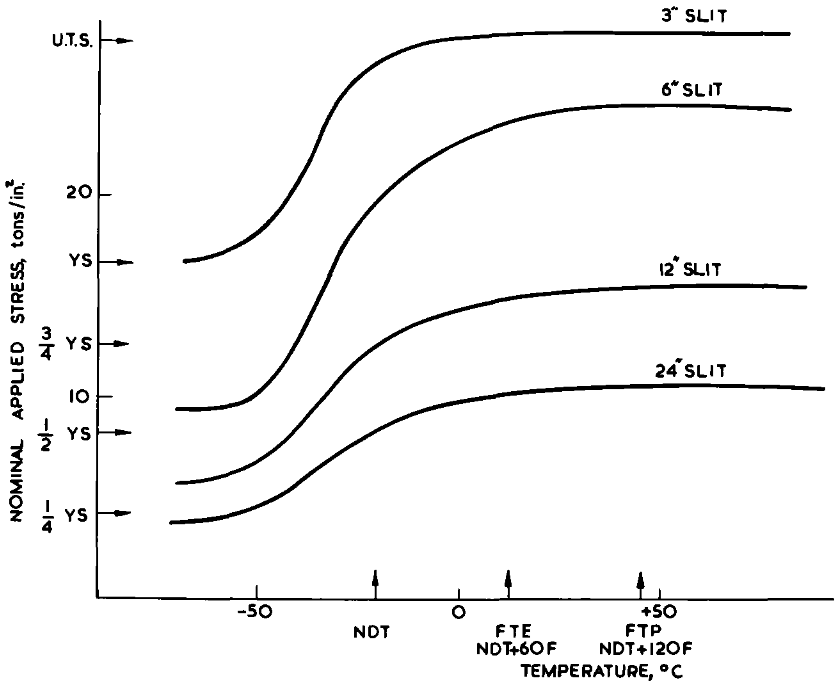

FIG. 2—*Conditions for initiation of unstable crack aluminum treated mild steel.*

remarks apply to the Pellini "NDT + 60" concept, insofar as this temperature is associated with crack arrest behavior, and indeed our work does suggest that the fracture analysis diagram for reactor pressure vessels does not normally take the form usually drawn by Pellini. Comparison of experimental results with the usual form of the diagram[8] suggests that the latter results from the combination of curves representing the condition for initiation of unstable cracking and those for crack arrest. Since the latter would not occur for internally pressurized reactor vessels, the relevant fracture analysis diagram for this case is that describing the conditions for unstable cracking which can take the form shown in Fig. 2.

[8] Nichols, R. W., *Nuclear Engineering,* May 1966.

It is worth drawing attention to a number of features of this diagram. First, with sufficiently large defects, failures can be produced even at high temperature with "nominal" stresses below those for general yield. Secondly, the curves are made up of three parts. At low temperatures, roughly those below NDT, the fractures are plain strain; at temperatures above NDT one in practice gets mixed fractures (with shear lips) and increasing fracture strengths until at the highest temperatures one gets essentially plain stress failures. However, there is not an order of magnitude change with temperature, but only a change by a factor of three or four in critical defect size for a particular stress. Thirdly, the experimental results suggest that these curves show some correlation with the Charpy energy at corresponding temperatures, the relationship being of the type:

$$f_g^3 l^2 = f_y^d f_u^e \,(C + D\phi)$$

where:

f_g = general applied stress at failure,
f_y = yield strength of material,
f_u = appropriate failure strength,
l = length of defect,
ϕ = Charpy V-notch energy, and
d, e, C, D = constants for particular conditions.

On this basis it is possible to make use of the wealth of information on the effect of irradiation on Charpy energy to predict its effect on the failure stress/defect size/temperature relationships. The general effect will correspond to a shift of the curves along the temperature axis to higher temperatures, so that the maximum embrittling effect at a particular temperature corresponds to the shift from the failure conditions at the top leg of the curve to those at the bottom leg. Even this difference is greater than what in fact will occur, since irradiation hardening will cause an increase in yield strength, and thus an increased failure strength for a given notch toughness. A further point arises as a result of this increased yield strength. In addition to satisfying the appropriate energy balance, before unstable cracking actually occurs, there is the need for some mechanism to be available to enable the initial extension of a crack of stable length. Under most practical conditions, this involves local yield at the end of a defect. Thus it can be argued that defects which have not led to failure before irradiation will not lead to failure after irradiation even to a dose where the energy balance criterion would suggest instability, because irradiation hardening would prevent local yield. It follows that it is necessary to consider carefully any proposal to recover the properties of an irradiated vessel by thermal annealing, since this could lead to a less safe situation in some circumstances.

It will be seen that much of the argument above parallels the approaches made by the advocates of linear elastic fracture mechanics.

However, it must be pointed out that at present the latter can only be applied readily to conditions when failure occurs in plane strain, and when the defect size is less than about twice the plate thickness. For thinner reactor vessels, such conditions do not appear to occur over most of the temperature range of interest to operators, and the effect of thickness remains to be demonstrated. It has been argued that the "plane strain" approach is bound to be pessimistic (that is, predict failure under conditions less severe than would be needed in practice to produce failure) and indeed diagrams such as Fig. 2 suggest this. If this were so, then the only criticism of such techniques is that they could call for unnecessarily expensive solutions in selection of materials or operation conditions. Again it can be argued that the existence of defects greater than plate thickness is so unlikely that it can be ruled out on grounds of engineering experience.

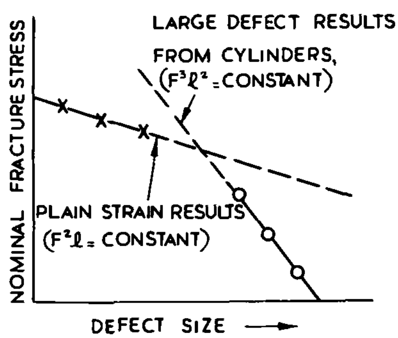

FIG. 3—*Schematic illustration of possible danger of extrapolating plain strain results to large defect size.*

One has thus changed the problem from one of defining what fracture toughness is tolerable to one of defining what defects must be considered to exist. The difficulty of proving absolutely by normal inspection techniques that large defects do not exist lead us to develop[9] the "limit approach" to pressure vessel safety which relies on the successful pressure test to define the maximum credible defect size. One can then make a crack growth analysis based upon this.

It is at this point that the use of linear elastic fracture mechanics can be misleading. One must have realistic maximum credible crack lengths information in order to select the relevant fatigue and creep data for the crack growth analysis. Extrapolation of results obtained from plane strain specimens to long crack lengths could be dangerously misleading as indicated in Fig. 3.

Thus we believe that the analysis of safety of pressure vessels can

[9] Nichols, R. W., et al., *Proceedings,* International Fracture Conference, Sendai, 1965.

best be based at the present time on the limit approach, making use of information from actual vessel tests and the associated small specimen correlations. The use of general yield fracture mechanics as developed by Wells et al[10] provides a possible approach for further improvement, and in the paper on zirconium at this conference,[11] we have indicated that the use of crack opening displacement techniques shows considerable promise as fracture mechanics tools for intermediate and lower strength materials. It is to be hoped that such techniques, as well as direct tests to failure, be used in the examination of the actual irradiated pressure vessel referred to in another paper to this symposium.[12]

[10] Wells, A. A., Burdekin, F. M., and Stone, D. E. W., "Panel Discussion," *Fracture Toughness Testing and Its Applications, ASTM STP 381,* American Society for Testing and Materials, pp. 400–405.

[11] Nichols, R. W., and Watkins, B., ASTM 1966. Symposium on Irradiation Effects Preprint No. 45.

[12] Monahan and Halpine, ASTM 1966 Symposium on Irradiation Effects, Preprint No. 77A.