where $t$ is the total time required to load the specimen to failure. But the intense strains associated with blunting extend only over a distance of size roughly $2\delta$ [11], at which point the strains have decayed to the order of an initial yield strain. Further, this decay has occured over a distance which is of order yield strain times plastic zone dimension. Thus, in by far the bulk of the plastic zone, a typical strain rate would be of order $10^{-3}$ times the crack tip strain rate. The ASTM specifications for $K_{Ic}$ testing [23] prescribe loading rates such that the total fracture test time for this material is of order one minute. Thus a typical strain rate in the plastic zone is of order $10^{-4}$/sec, and, in the temperature range studied, this strain rate occasions a yield stress which is essentially the same as quasi-static.

An estimate of an irradiated fracture toughness transition temperature can be made in the same manner as for the unirradiated material. Using the maximum achievable stress intensification for non-hardening materials, 2.97, gives a transition yield stress of 570 $MNm^{-2}$ (82 ksi). Depending on the particular irradiating temperature and total neutron fluence, the temperature corresponding to this yield stress could range from 94°C (200°F), to nearly 261°C (500°F), again complicated by the flatness of the $\sigma_0$ versus temperature curve. Although valid $K_{Ic}$ measurements on irradiated material are scarce, there are indications that an irradiated fracture transition temperature is near 94°C(200°F), [27] where the yield stress is roughly 590 $MNm^{-2}$.

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# DISCUSSION

**J. M. Krafft[2]**

About five years ago, R. A. Gray and the writer conducted measurements of fracture toughness $K_{Ic}$ on an A533 B Class II pressure vessel steel, after its irradiation with $3 \times 10^{19}$ n/cm² of energy > 1 MeV.[3] Both loading rate $\dot{K}$ and temperature was varied. As with the unirradiated material, a strong, intricate dependency of $K_{Ic}$ upon $\dot{K}$ was observed after irradiation. Selecting a $\dot{K}$ region of low $\dot{K}$-sensitivity ($8 \times 10^3$ ksi $\sqrt{\text{in}}$/sec), temperature was varied between +50 and −200°C. Values of $K_{Ic}$ de-

[2]Naval Research Laboratory, Washington, D. C. Mem. ASME.

[3]Krafft, J. M., and Gray, R. A., "Effect of Neutron Irradiation on Bulk and Micro Flow-Fracture Behaviors of Pressure Vessel Steels," *Practical Applications of Fracture Mechanics to Pressure Vessel Technology*, Inst. Mech. Engrs., London, 1971, pp. 93–102.

creased continuously with temperature down to about −50°C, but then rose sharply around −80°C before continuing in decline at still lower temperatures. The $K_{Ic}(T)$ data (for comparable irradiation dosages) shown in Figs. 5, 6, and 7 of Parks' paper appears to exhibit a corresponding upturn in $K_{Ic}$. Our attempts to explain this upturn in terms of models, such as the RKR model, in which the controlling variable is the yield strength, were unsatisfactory. The yield strength varies continuously through the −80°C temperature region, hence the predicted $K_{Ic}(T)$. We reasoned alternatively that the tensile instability strain of irradiated carbon ferrite, the matrix constituent of the steel, could control $K_{Ic}$ even as low as −80°C. Its value shows a distinct and closely corresponding upturn in this temperature region. If a model associated with Knott is desired, his paper with G. Green in this Symposium (Paper 75-MAT-10) utilizes such a criterion at temperatures where "a small amount of fibrous "dimpling" preceded a cleavage fracture." One wonders whether such a precursor could occur and control the fracture at lower temperatures, such as −80°C, but be fractographically obliterated by the subsequent separation process, cleavage.

## Author's Closure

First, I would like to thank Dr. Krafft for drawing to my attention his previous work. It is interesting to note that the data cited here for irradiated fracture toughness versus temperature show what may correspond to a local maximum, near −80°C, as Dr. Krafft has suggested. His data may be similarly interpreted. However, the magnitude of the postulated "hump" would seem rather small compared with, say, the magnitude of the scatterband in the unirradiated condition, where a considerably larger number of tests have been performed. That scatter must be considered in the irradiated condition as well is evident from Fig. 7, where two tests run at −59°C give fracture toughnesses differing by twenty per cent, while two tests run at −4°C obtain toughnesses differing by fifteen per cent. In view of the considerable degree of scatter present in the unirradiated tests, it would seem plausible that a similar degree of scatter could be expected in the irradiated state, providing that the same fracture mechanism is operative within a given toughness regime. A complete analysis of the micromechanics of fracture initiation in low to medium strength steels must necessarily include an appropriate statistical treatment of this scatter.

I have some reservations concerning the applicability of the tensile ligament instability strain model to the low temperature fracture of low-to-medium strength steels. Typically such models are applied to the initiation of microscopically ductile fracture, where the fracture process involves the nucleation and coalescence of microvoids, often associated with such second phase particles as cracked precipitates or non-wetting inclusions. In such cases the "process zone size," $d_T$, of the tensile ligament instability model can often be related to the spacing of these void-nucleating particles, such as in the case of manganese sulfide inclusions in 4340 steel [28].[3] In the application of the tensile instability strain model to the low temperature fracture behavior of A533B and other low-to-medium strength steels, based upon the properties of the matrix ferrite, Krafft infers a value of $d_T \approx 1\ \mu$, which is of the order of the spacing of the fine carbide precipitates in this material (see Hester, reference [24]), and likely considerably less than the spacing of manganese sulfide inclusions. As Krafft notes, it might be difficult to obtain fractographic evidence of the occurrence of such a fracture process, especially if the $\approx 1\mu$ ligaments were created due to void sites formed at carbide precipitates, which are of sub-micron size. But their small size would seem to make them unlikely void initiation sites. If the void initiation sites for the postulated low temperature tensile ligament instability mechanism are the manganese sulfide inclusions, however, then fractographic evidence should be obtainable, as Hellerich and Hunter [29] were able to do using scanning electron fractography on higher temperature Charpy specimens of A302B, a similar steel. I am unaware of a definitive fractographic study of low-temperature fracture initiation from a sharp (fatigue) crack in A533B, wherein it can be concluded that cleavage microcrack initiation or a microscopically ductile mechanism is operative. However, fractographic evidence available from low temperature fracture toughness tests of spheroidized carbon steels [30] and low carbon alloy irons [31] often supports the concept of internal microcleavage crack initiation a small distance ahead of the crack tip, and back propagation to the fatigue crack front in a cleavage mode, essentially as in the RKR model. The evidence consists primarily of the so-called "river markings" on cleavage facets, from which the local direction of crack propagation can be inferred.

At higher temperatures, a fracture surface appearance transition takes place; increasing amounts of plastic dimpling are observed, and then a microscopically ductile fracture initiation mechanism, such as the tensile ligament instability model, perhaps, would presumably be applicable.

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