Problem #1

The possibility of subcritical or critical fractures in the reactor vessels and the resulting danger to the general population are topics of utmost concern in the design and operation of modern nuclear power plants. This problem assumes particular significance as the nuclear power plants approach the end of their design life and critical decisions have to be made about the safety, modernization or termination of many currently operating nuclear plants.

In the past several decades, major research programs have been initiated to develop full characterizations of the toughness of nuclear pressure vessel steels as a function of the composition of the alloys, operating temperature of the reactor, neutron irradiation, strain rate and residual element content. Nuclear pressure vessel steels, such as AISI A533 Grade B Class 1 (commonly referred to as A533B–1) steel, exhibit a classical transition in fracture mode from brittle cleavage fracture at low temperatures (lower shelf) to ductile fracture at high temperatures (upper shelf). The brittle–ductile transition temperature is affected strongly by the strain rate and by neutron irradiation. Microscopically, the lower shelf fractures are characterized by a transgranular cleavage mechanism of failure along low energy cleavage planes. The upper shelf fractures, on the other hand, occur by a mechanism of microvoid coalescence involving the nucleation, growth and coalescence of small voids formed at inclusions and second phase particles.

An understanding of the micromechanical origins of fracture resistance in pressure vessel steels and of the effects of various microstructural, mechanical and environmental factors on the macroscopic fracture properties is essential for the design of alloys for nuclear plants and for the control and prevention of catastrophic fracture during the operation of power plants. The objective of this problem is to explore the micromechanistic origins of fracture in a nuclear pressure vessel steel which is currently used in practice.

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The A533B–1 steel has a nominal composition, in weight percent, of 0.23 C, 1.55 Mn, 0.009 P, 0.014 S, 0.2 Si, 0.67 Ni, 0.04 Cr, 0.53 Mo, 0.003 V, 0.16 Cu and the balance Fe. The room temperature mechanical properties of the material are: yield strength = 481 MPa, ultimate tensile strength = 642 MPa and strain to failure = 25.4% (for 50.8 mm gage length tensile specimen). This steel is used in service after the following heat treatment: Normalized for 4 h at 914°C, air cooled, austenitized for 4 h at 900°C, agitated water quenched, tempered for 4 h at 664°C, water quenched, stress-relieved for 40 h at 612°C and furnace cooled. As with many low alloy steels quenched or normalized in thick section plte, the microstructure of this steel is primarily composed of acicular tempered upper bainite with a small fraction of proeutectoid ferrite. The prior austenite grain size is about 25 μm.

As part of the Heavy Section Steel Technology (HSST) program and the Electric Power Research Institute (EPRI) program of the United States Atomic Energy Commision, the A533B–1 steel has been extensively studied to document its mechanical properties. Tensile yield strength data for the steel was obtained in the unirradiated condition from EPRI data bank. Yield strength data as a function of temperature and strain rate were complied, using the strain rate–temperature relationship in terms of $T \ln (A/\dot{\epsilon})$, where T is the absolute temperature, $\dot{\epsilon}$ is the strain rate, and A is a constant equal to 10^8 s^{-1}. Figure 1 shows a plot of such experimental data.

Critical fracture stress values for slip-initiated cleavage in the unirradiated condition was estimated to be 1830 MPa for quasi-static tests (approximate crack-tip strain rate of 10^{-2} s^{-1}) and 2000 MPa for dynamic tests (approximate crack-tip strain rate of 5 s^{-1}). Fracture strain values ($\overline{\epsilon}_f$) as a function of the normalized hydrostatic stress ($\sigma_m/\overline{\sigma}$) were determined from circumferentially notched tensile tests at 24, 71 and 177°C (see Figure 2) on specimens with different notch-root radii. Microscopic observations of upper shelf fracture specimens showed that the spacing between major voids on the fracture surfaces was roughly 50 μm. Figure 3 shows the distribution of longitudinal stress σ_{yy} normalized by the reference yield strength, σ_y for several different values of the strain hardening exponent. Figure 4 shows the distribution of plastic strain ($\overline{\epsilon}_p$) and stress state ($\sigma_m/\overline{\sigma}$) near the crack tip, as a function of the distance x directly ahead of the crack tip (normalized by the crack-tip opening displacement, δ).

(a) From the information provided here, predict the macroscopic fracture initiation toughness K_{Ic} of the A533B-1 pressure vessel steel as a function of temperature for quasi-static and dynamic strain rates of 10^{-2} and 5 s^{-1}, respectively, in the lower shelf region

noting that slip-induced cleavage is the dominant microscopic mechanism of failure. You may assume that the nil-ductility transition temperature (NDTT, below which cleavage fracture occurs) to be approximately −30°C for both the strain rates. If any mechanical property that you deem necessary to do this problem is not provided here, assume the appropriate value for the low strength steel and provide a justification for your assumption.

(b) Predict the macroscopic fracture initiation toughness for the two strain rates as a function of temperature in the upper shelf region by invoking micromechanical failure criteria which you think are appropriate.

(c) If it is desired that all the fracture toughness tests be conducted using a single specimen geometry for the whole range of temperatures examined in parts (a) and (b) above, calculate the minimum specimen size for "valid" measurements of the plane strain fracture toughness values, K_{Ic} and J_{Ic}, using a compact tension specimen. Provide *all* the dimensions of the specimen.

(d) As mentioned earlier, neutron irradiation of the pressure vessel walls is a critical issue in fracture control. In general, neutron irradiation increases the yield strength and reduces the strain hardening exponent, while leaving the critical fracture stress for slip-induced cleavage unaffected. Describe by means of a schematic diagram the variation in the lower shelf fracture toughness as a function of neutron fluence.

(e) From the viewpoint of alloy design, austenitizing temperature is a variable which can, at least in principle, be used to alter the fracture initiation toughness of steels. Speculate on the possible consequences of increasing the austenitizing temperature on the fracture initiation toughness.

Problem #2

In the class, we discussed in detail the micromechanisms of fracture in precipitation hardened aluminum alloys. The purpose of this problem is to help you develop a more comprehensive understanding of the failure mechanisms in aluminum alloys as a function of composition, microstructure and aging treatment.

Precipitation hardenable aluminum alloys are widely used in different components of commercial aircraft. For example, one needs approximately 400,000 pounds of aluminum

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to make a Boeing 747. (Not all of this material ends up in the aircraft; a large portion of it is recycled as scrap material.) Fracture of aluminum alloys is an issue of major concern in this safety-critical application.

(a) Pick any one aluminum alloy of your choice (other than an alloy belonging to the Al–Li system) and find out from the literature the general trends in the variation of fracture initiation toughness as a function of aging treatment. Write a brief summary (one page or less) of such effects. The handout given in class provides a number of references which deal exclusively with the fracture resistance of a wide range of aluminum alloys.

(b) Describe the micromechanisms that are responsible for these variations in fracture toughness, on the basis of information available in the literature and of your critical evaluation of such information.
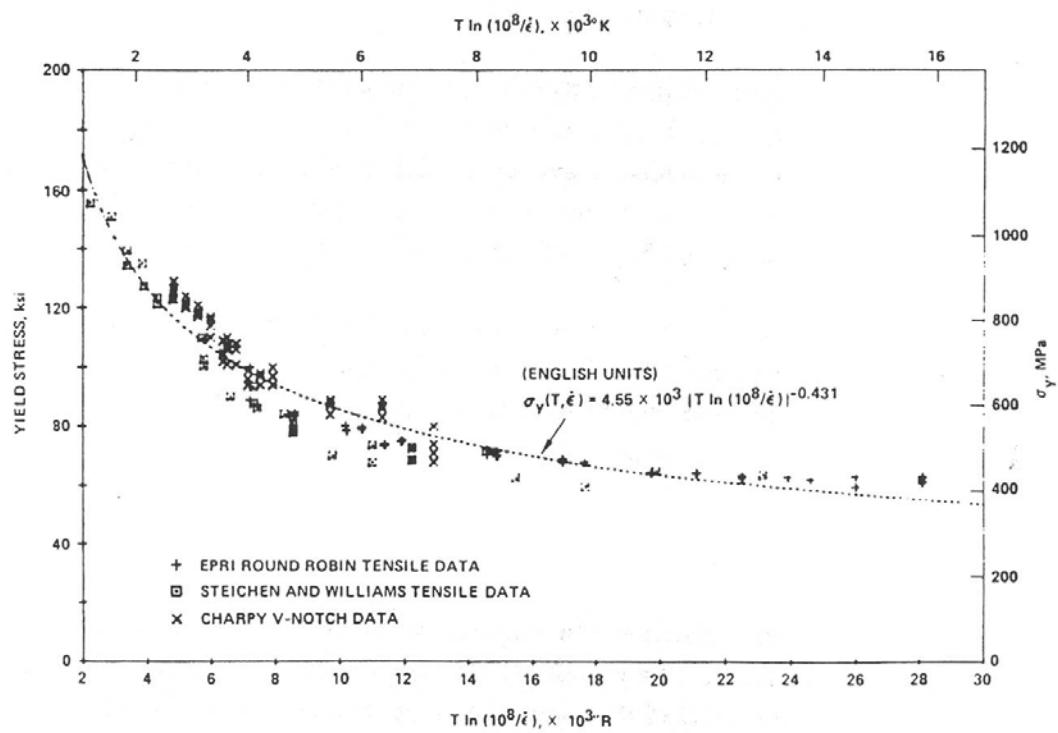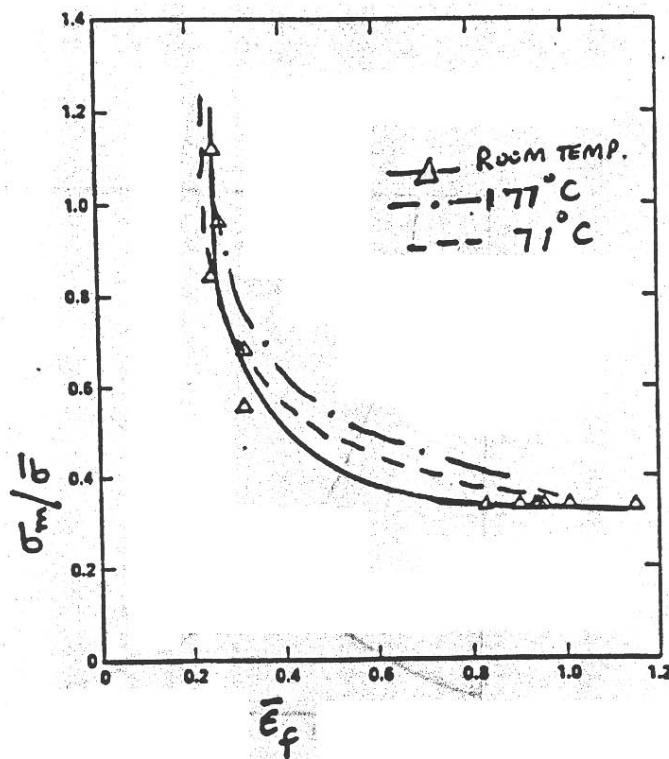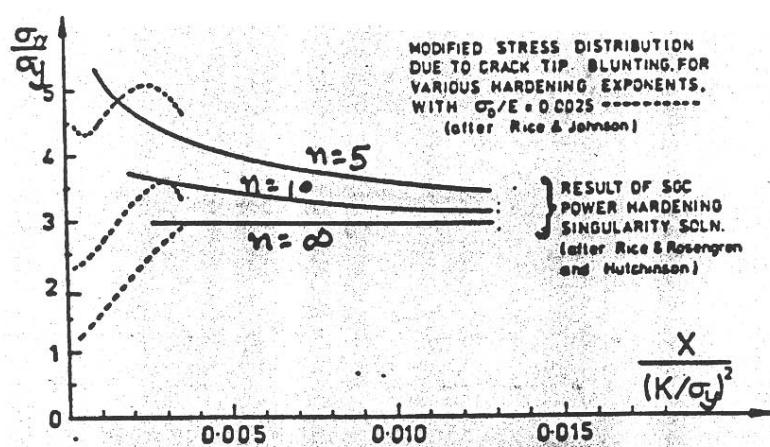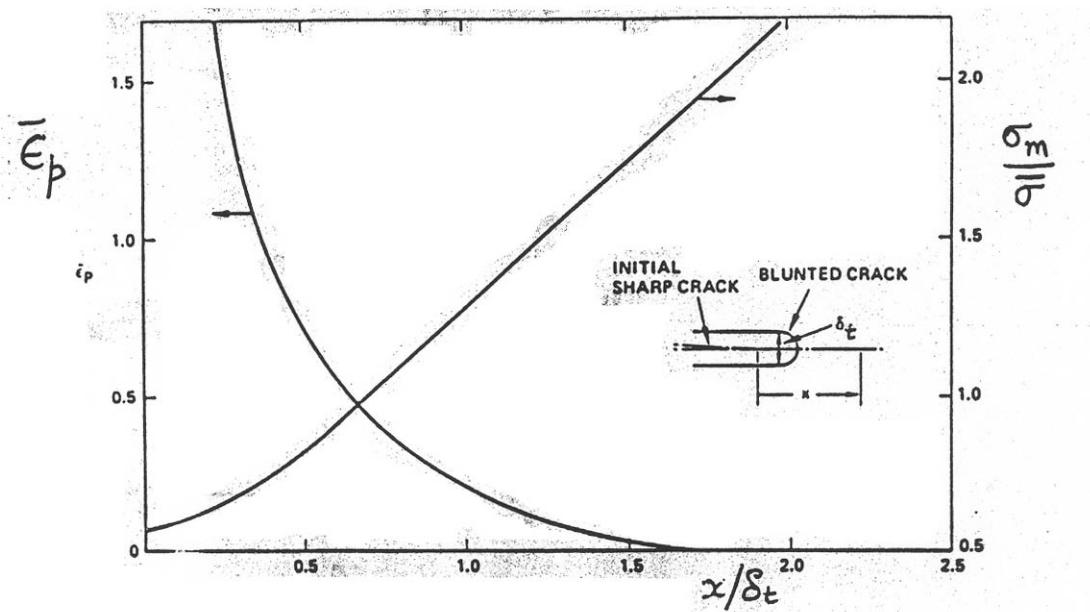
FIG. 1

FIGURE 2

MODIFIED STRESS DISTRIBUTION
DUE TO CRACK TIP BLUNTING FOR
VARIOUS HARDENING EXPONENTS.
WITH $\sigma_0/E = 0.0025$ ----------
(after Rice & Johnson)

RESULT OF SGC
POWER HARDENING
SINGULARITY SOLN.
(after Rice & Rosengren
and Hutchinson)

$n=5$

$n=10$

$n=\infty$

FIGURE 3

Fig. 4