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Implications of Radiation-Induced Reductions in Ductility to the Design of Austenitic Stainless Steel Structures

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Implications of Radiation-Induced Reductions in Ductility to the Design of Austenitic Stainless Steel Structures

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Abstract

In the dose and temperature range anticipated for ITER, austenitic stainless steels exhibit significant hardening with a concomitant loss in work hardening and uniform elongation. However, significant post-necking ductility may still be retained. When uniform elongation (e_u) is well defined in terms of a plastic instability criterion, e_u is found to sustain reasonably high values out to about 7 dpa in the temperature range 250-350°C, beyond which it decreases to about 0.3% for 316LN. This loss of ductility has significant implications to fracture toughness and the onset of new failure modes associated with shear instability. However, the retention of a significant reduction in area at failure following irradiation indicates a less severe degradation of low-cycle fatigue life in agreement with a limited amount of data obtained to date. Suggestions are made for incorporating these results into design criteria and future testing programs.

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Introduction

Austenitic stainless steel has been selected as the reference structural material for the International Thermonuclear Experimental Reactor (ITER). While the anticipated operating temperature for the structure and the peak neutron dose are both relatively low, for planning purposes the properties of austenitic stainless steels in general, and 316 LN in particular, are being evaluated for temperatures in the range 200°C to 450°C and for neutron doses up to about 25 dpa. In this temperature and dose range, 316LN is expected to harden considerably with an attendant loss of ductility and fracture toughness. Since ductility parameters such as uniform elongation (e_u) and reduction of area (RA) are used in design criteria, the very low values of e_u reported for some austenitics after modest exposure to neutrons at temperatures near 300°C require some re-assessment of these data and an evaluation of their implications to design. The first step is to ascertain a consistent and correct definition of ductility parameters, particularly e_u; the second step is to understand the implications of reductions of ductility to fracture and failure modes; and the third step is to ensure that this understanding is correctly incorporated into the design criteria. It is the purpose of this paper to review the status of this assessment.

Review of Tensile Data

Tensile data generated at low temperatures (less than 400°C) in support of fusion machines like ITER have shown that neutron-induced changes in yield stress (σ_y) are considerably enhanced for irradiation and test temperatures in the vicinity of 300°C [e.g., Refs 1, 2]. Figure 1 is a composite plot of yield stress as a function of irradiation and test temperature for several variants of solution annealed 316 stainless steel irradiated in a variety of reactors under different experimental conditions; the data reveal a maximum in both the yield stress and the increase in yield stress ($\Delta\sigma_y$) around 300°C. In addition, as shown in Figure 2, the yield stress (YS) approaches the ultimate tensile strength (UTS) in the 200-330°C temperature range after modest doses [17].

This increase in strength is accompanied by a loss of ductility and strain hardening capacity in this temperature range. There are discrepancies in the way uniform elongation (e_u) is defined, as addressed below; however, as an illustration of the effect of irradiation on ductility, e_u for 316LN decreases from about 30% for unirradiated material to about 0.3% at 11 dpa, and the strain-hardening capability -- as measured by the average plastic modulus $E_p \approx (UTS-\sigma_y)/e_u$ -- decreases from about 1000 MPa for unirradiated material to about 0 at 5±2 dpa.[18]

This loss of ductility has raised concerns over using austenitics in ITER in this temperature range. For instance, engineering design codes use e_u to characterize austenitic steels as ductile ($e_u \geq 5\%$), semi-brittle ($1 < e_u < 5\%$), or brittle ($e_u \leq 1\%$). Moreover, irradiation can profoundly affect the nature of the stress-strain relationship, and the effects are sensitive to both irradiation (and test) temperature and dose. [3, 7-9, 11, 19-21]. For instance, at a dose of about 7 dpa: irradiation at 60°C results in the appearance of a yield drop in the tensile stress-strain curve followed by work hardening; at 200°C the work hardening may be insufficient to produce a UTS larger than the upper yield stress; at 330°C no strain hardening occurs after initial yield; and at 400°C there is no yield drop but some modest work hardening is exhibited.[10] Similarly, as illustrated in Figure 3 for 316LN irradiated and tested in the temperature range of 250-270°C, increasing dose gives rise to a range of behavior starting with the onset of a yield point and post yield work hardening at low dose, diminishing work hardening (and hence a flatter post-yield plateau) with increasing dose, and finally no work hardening after yield at doses above about 7 dpa. The manifestation of two maxima in the stress strain curves (yield point and UTS) has given rise to confusion about the definition of e_u as discussed below.

Nonetheless, a significant post-yield ductility is still retained, even in the absence of work hardening. For instance, the reduction in area (RA), which is used to determine the true strain at failure (ε_{tf}) by $\varepsilon_{tf} = \ln (1 - RA)^{-1}$, only decreases from an unirradiated value of about 70% to values of 61±4% at 11 dpa in 316LN in the 250-270°C temperature range [11]. Hence, the questions arise as to how to best define ductility under these conditions, what these ductility changes imply with respect to failure and fracture modes, how to incorporate this into design criteria for irradiated structures, and what additional testing is required to address these questions.

Defining Uniform Ductility

The uniform elongation is intended as a simple measure of the permanent plastic strain accumulated prior to necking and failure in a uniaxial tensile test. For unirradiated material it is determined from the engineering stress-strain curves as the permanent plastic strain corresponding to UTS or maximum load. However, as noted above, load drops exhibited by irradiated material can give rise to double maxima, and this in turn has resulted in a variation in the way e_u has been evaluated and reported. For example, values of e_u ranging from 0.3%-14% have been reported for 316LN irradiated to 5±2 dpa. Not only are load drops a manifestation of displacement-controlled testing, but when a load drop is followed by a work hardening or perfectly plastic plateau, it is inappropriate to use the displacement at the first peak load to define e_u. A more fundamentally correct criterion for determining the true uniform strain (ε_{tu}) at the onset of necking can be obtained by equating the rate of material work hardening to the rate of increase in stress required by geometrical softening (e.g., specimen thinning) to continue deformation [22]; this defines a point of plastic instability as

$$(1/\varepsilon_t)\,\partial(\ln \sigma_t)/\partial(\ln \varepsilon_t) = (1/\sigma_t)(\partial\sigma_t/\partial\varepsilon_t) = 1 \qquad (1)$$

where σ_t and ε_t are the true stress and strain, respectively. Hence, the uniform engineering strain can be obtained from $e_u = \exp(\varepsilon_{tu}) - 1$. For 316LN stainless steel, this definition of e_u results in a fairly consistent estimate of $e_u \sim 13\%$ at 5.1 dpa (versus the range of values in the literature from 0.3 to 14% noted above); and $e_u \sim 0.3\%$ at doses > 7 dpa (which also corresponds to the lower values reported in the literature).

The data from Refs 2, 3, 9-11, 18, 23-28 have been reanalyzed on the basis of Eq. (1) and are plotted in Figure 4. The data include the SUPERPHENIX heat 316L(N)-SPH and two Japanese steels, JPCA and J316. Irradiations were performed in HFIR where the He/dpa ratio was on the order 30, as well as HFR, R2 and ORR where the He/dpa ratios were smaller than about 10; and the data have been divided to reflect this. In addition the data have been divided into low ($200 \le T < 250$°C), intermediate ($250 \le T \le 350$°C), and high ($350 < T \le 400$°C) irradiation/test temperature regimes. The uniform ductility remains relatively high out to about 7 dpa for all conditions. It drops to relatively low values (<1%) at this dose for intermediate temperatures (with the exception of two data at 10 dpa), with an average of about 0.3% to doses out to 20 dpa. There does not appear to be an effect of He/dpa on this trend. At the lower temperatures, the uniform elongation remains high out to 10 dpa (one datum), and at the higher temperatures it appears that uniform elongation is beginning to decrease to less than 1% at about 10 dpa, at least in material irradiated with the higher He/dpa (3 data points).

Implications to Fracture and Failure Modes

The data for fracture toughness (K_{IC}) of austenitic stainless steels are rather limited. Most data have been generated at relatively high irradiation and test temperatures. Odette and Lucas [29] have shown that the K_{IC} data for austenitic stainless steels irradiated and tested at temperatures >400°C can be reasonably correlated by

$$K^{i}_{Ic}/K_{Ic}{}^{u} \sim \sqrt{(e_u{}^{i}/e_u{}^{u})(\sigma_o{}^{i}/\sigma_o{}^{u})} \tag{2}$$

where σ_o is the flow stress (average of σ_y and UTS) and the superscripts indicate unirradiated (u) and irradiated (i) material. Assuming that such a correlation also applies to lower irradiation/test temperatures and using nominal values of $\sigma_o{}^{i}$ = 650 MPa and $e_u{}^{i}$=10% in Eq (2) suggests a retention of 75% of the unirradiated fracture toughness for doses up to about 7 dpa in 316LN in the temperature range 250-400°C. This is consistent with the relatively high values of retained fracture toughness obtained for austenitic stainless steels irradiated to 3 dpa in this temperature range [30]. However, the severe loss of uniform elongation beyond 7 dpa shown in Figure 4 would suggest a much more dramatic loss of fracture toughness at higher doses. Again using nominal values of $\sigma_o{}^{i}$ = 800 MPa and $e_u{}^{i}$ = 0.3% in Eq. (2) results in values of K_{IC} < 50 MPa$\sqrt{\text{m}}$. This potentially severe loss of fracture toughness must be explored experimentally.

Odette and Lucas [31] have also recently argued that the loss of fracture toughness and severe degradation in tearing modulus observed in austenitics irradiated at higher temperatures are consistent with a change in fracture mode with increasing dose, transitioning from a ductile-dimple process to a shear decohesion process where severe dislocation channeling and channel fracture predominate. They modeled the decohesion zone ahead of a loaded crack as a continuous set of plastic ligaments bridging the face of a virtual crack over a length L_z The ligaments deform and fail according to a characteristic stress-displacement function $\sigma(\Delta)$ modeled as a triangular function of height σ_{zm} and base Δ_z. Their analysis suggests that fracture toughness scales with these parameters as

$$K_{Jc} = \sqrt{0.5\, E'\, \sigma_{zm}\, \Delta_z} \tag{3}$$

where $E' = E/(1-\nu^2)$ is the plane strain elastic modulus. Initial estimates of $\sigma_{zm} \sim 2\sigma_o$ and $\Delta_z \sim 9$ µm yield estimates of $K_{Jc} \sim 50$ MPa$\sqrt{m}$. Moreover, the model suggests that once initiated, crack growth would take place without matrix plasticity, and hence the tearing modulus would approach 0. Both of these predictions are in agreement with the lower bound on the high temperature data. Such a process could also be manifest in austenitics at the lower irradiation temperatures at high doses, since the loss of work hardening capacity would be consistent with the flow localization leading to such a fracture mode. However, key experiments are required to verify and develop this further. These experiments should include characterization of deformation and fracture zones ahead of cracks in austenitics irradiated to higher doses at low temperatures combined with quantitative fractography and strain mapping.

The onset of flow localization and the transition from ductile fracture to a shear decohesion mode also has implications on notch ductility and the possible appearance of alternative failure modes. For instance, in the presence of a notch, tensile fracture in a ductile material, particularly in thin sheet, is preceded by the development of a narrow zone of intense plasticity ahead of the notch, and fracture occurs by the initiation and growth of a crack within this zone.[32] Again, this process can also be modeled by treating the plastic zone as a hypothetical crack, bridged by the material in the zone.[33,34] The fracture energy Γ thus becomes the integral of the characteristic stress-displacement function of the material in the bridging zone [35], and the degree of notch sensitivity is governed by a characteristic bridging length $\Gamma E/\sigma_{ts}^2$ where σ_{ts} is the tensile strength in the absence of the notch. When Γ is large and/or σ_{ts} is small, this characteristic length is large relative to the notch size, and the tensile strength is realized in a notched specimen well before significant extension of the plastic strip. However, if Γ is reduced and/or σ_{ts} is increased (e.g., by irradiation), then the characteristic length becomes small relative to the notch size, and the (reduced) toughness is fully utilized in the plastic strip prior to catastrophic failure. Thus, for instance from Equation (3), the onset of a shear decohesion mode of failure would reduce Γ to $K_{Jc}^2/E' \sim 0.5\,\sigma_{zm}\Delta_z$; again using $\sigma_{zm} \sim 2\sigma_o$ and $\Delta_z \sim 9$ µm, the characteristic bridging length is reduced to only 3mm. This implies the induction of severe notch sensitivity. Moreover, the implication of loss of work hardening to localized plasticity suggests additional failure modes associated with shear decohesion between offset cracks, notches, and holes may become even more important at high doses. This must also be examined experimentally.

There are even fewer data on the effects of neutron irradiation on the fatigue properties of austenitic stainless steel.[36-39] In strain-controlled fatigue tests, irradiation to about 10 dpa at 550°C results in little change from the unirradiated behavior; at 430°C this results in about a factor of three to ten reduction in fatigue life at stresses/strains above the endurance limit. The post-irradiation (10 dpa) fatigue data at lower temperatures, although limited to a few tests, show no reduction in fatigue life at irradiation/test temperatures of 227 and 427°C and about a 50% reduction in fatigue life at 327°C. There appears to be little effect of irradiation on the endurance limit. Changes in low cycle fatigue life are generally scaled with decreases in total ductility. Since this scales with RA, the small loss in fatigue life observed is consistent with the significant retention of RA in irradiated material.

Implications to Design Criteria

Hence, based on scaling laws, one would expect the low-cycle fatigue life to be relatively independent of dpa in the range of interest. In addition, the limitation on the peak local design stress (primary membrane + primary bending + secondary + notch effects) for monotonic loading should be relatively insensitive to dpa even in the intermediate temperature range for doses <7dpa because the ability of the material to blunt the notch effects through local plastic flow remains about the same. However, notch sensitivity and shear failure modes must be considered at higher doses. Moreover, because of strain localization resulting from the lack of strain hardening capability, additional constraints may need to be placed on the primary membrane + bending stress limit (K S_m, where K is a shape factor and S_m is the maximum allowable stress). Certainly the shape factor K (normally 1.5 for a rectangular cross-section) would have to be reduced to values approaching 1 as e_u decreases below 1% and E_p approaches 0.

Summary and Conclusions

Austenitic stainless steels undergo significant increases in strength with an attendant loss of ductility for irradiation and test temperatures in the range 250-400°C. As the dose increases the tensile stress-strain curves begin to exhibit a load drop followed by a deformation plateau that transitions from work hardening to perfectly plastic to work softening behavior. Under these conditions the appropriate definition for uniform ductility is obtained from a formal definition of the onset of plastic instability. When the data in the literature are corrected for this definition, the uniform elongation stays relatively high out to a dose of about 7 dpa for all temperatures of interest; but at temperatures in the range 250-350°C, doses beyond 7 dpa appear to result in severe ductility loss. At temperatures outside this range, the dose to severe ductility loss appears to be higher. Severe ductility loss has significant implications with respect to loss of fracture toughness and the onset of perhaps more severe failure modes associated with shear decohesion in the presence of notches, holes and offset cracks. This must be examined experimentally and design criteria must account for the competition between these failure modes. In addition to standard fracture toughness tests, specimen geometries should be explored to determine notch sensitivity and the possibility of shear decohesion failures. Strain mapping and quantitative fractography should be employed to greatly enhance the data obtained from these mechanical tests. On the other hand, the significant retention of RA following irradiation suggests that a less severe degradation of fatigue life might be expected. While limited data to date support this, a larger data base is required for confirmation.

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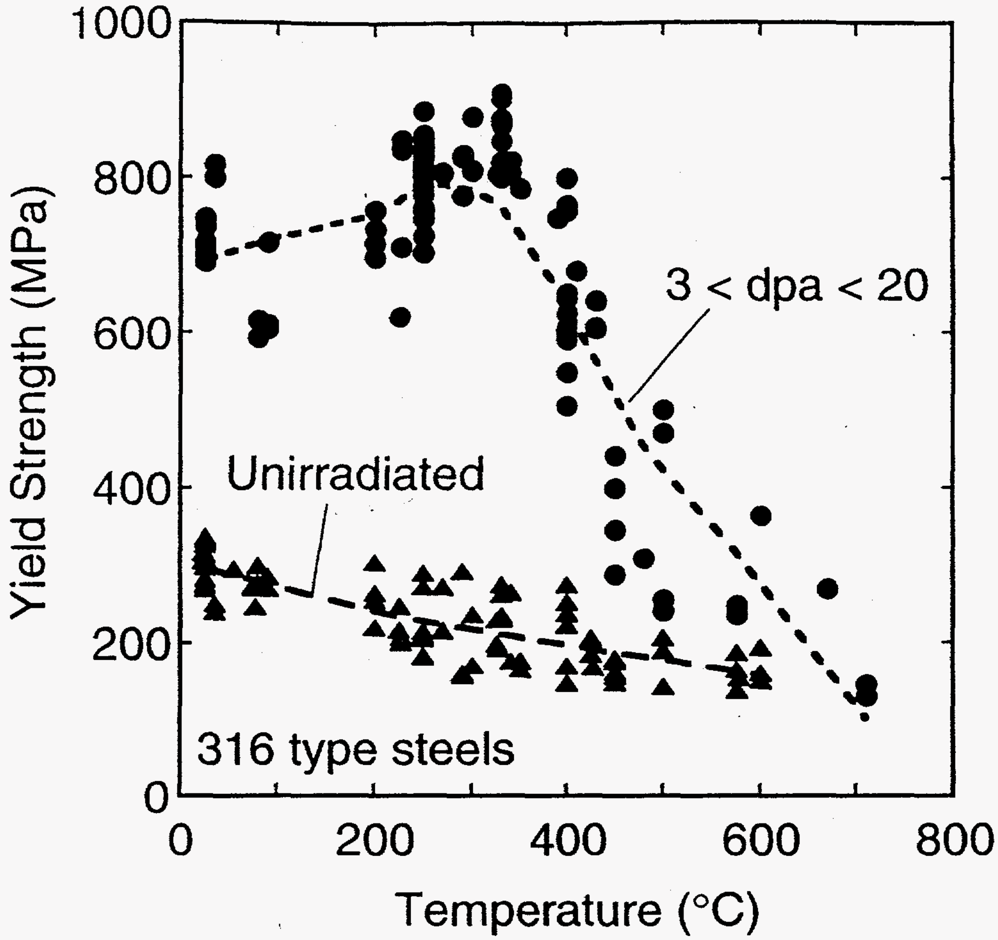
List of Figures



Yield Strength (MPa)
1000
800
600
400
200
0
0
200
400
600
800
Temperature (°C)
3 < dpa < 20
Unirradiated
316 type steels

Fig 1

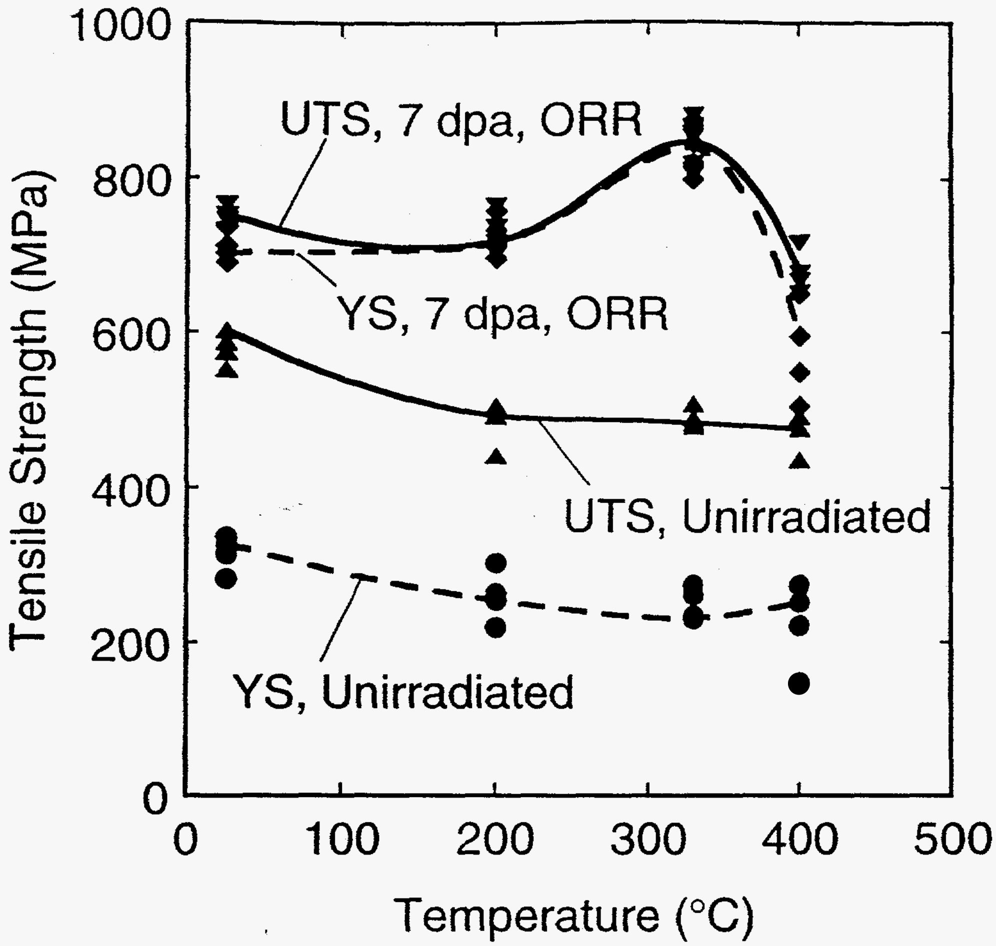

1000
800
600
400
200
0
0
100
200
300
400
500
Tensile Strength (MPa)
Temperature (°C)
UTS, 7 dpa, ORR
YS, 7 dpa, ORR
UTS, Unirradiated
YS, Unirradiated

Fig 2

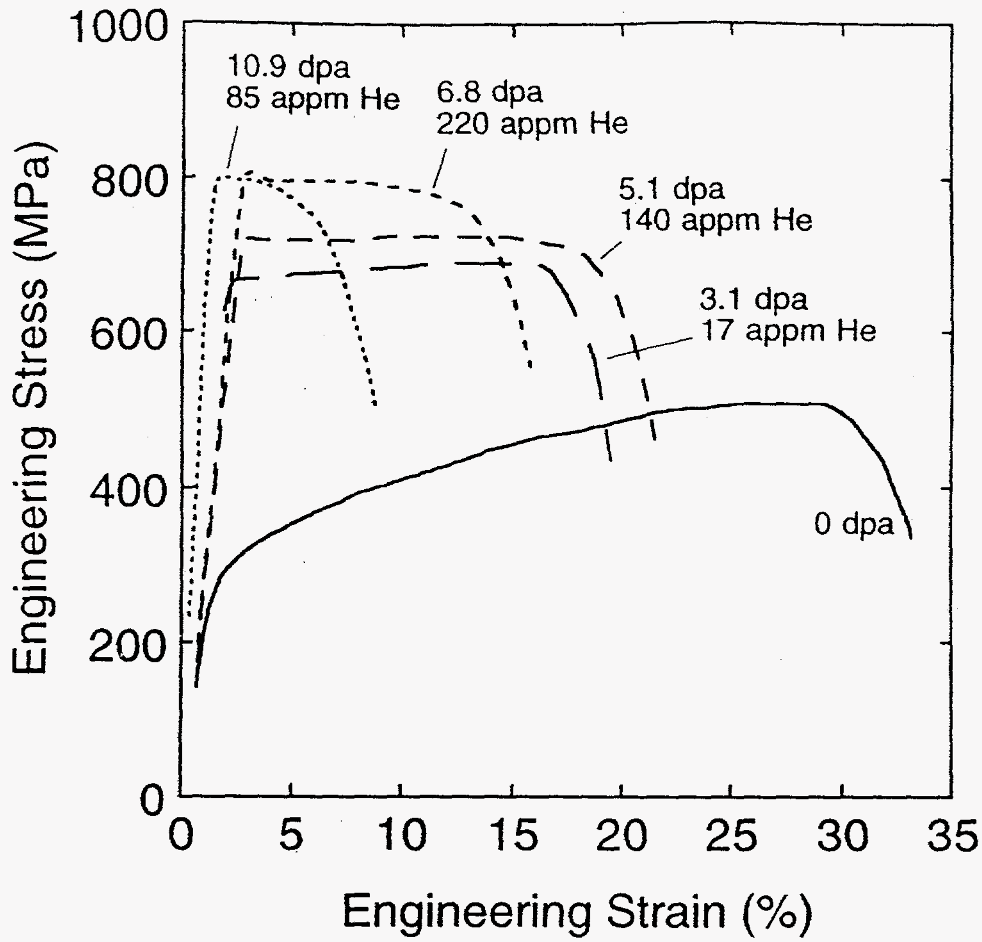
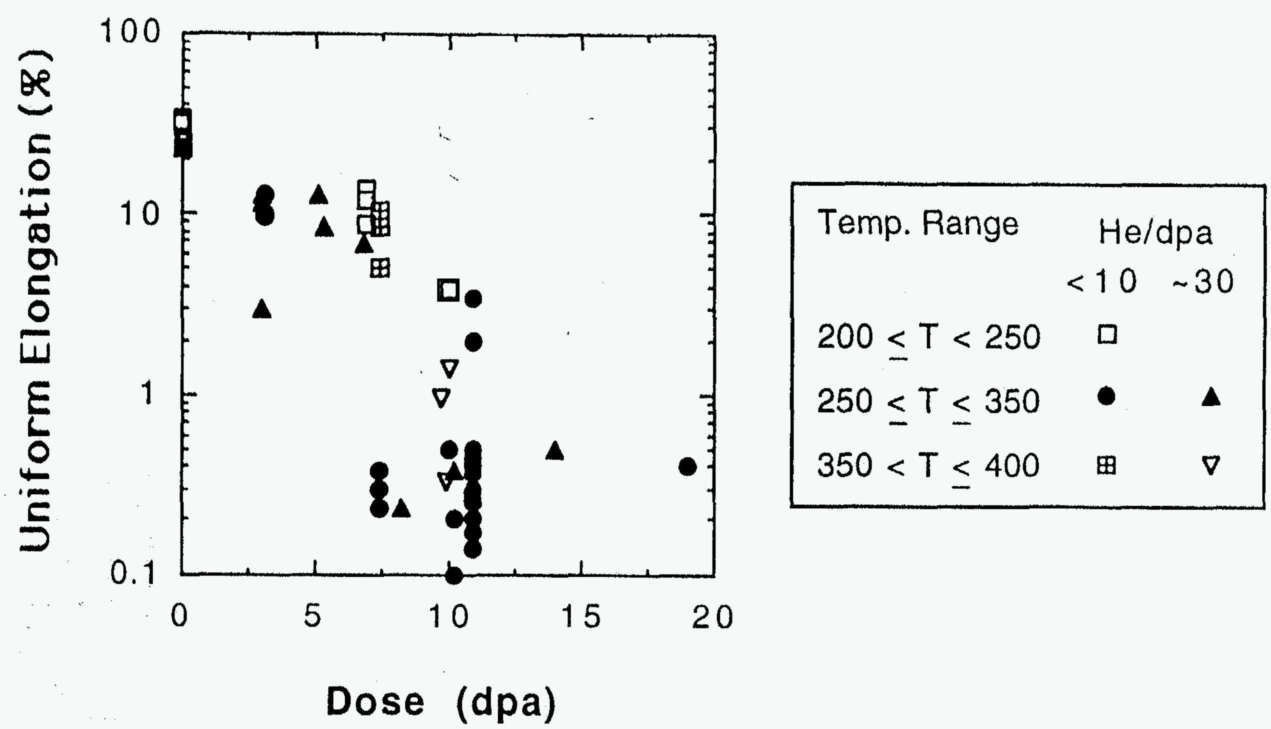

1000
800
600
400
200
0
0
5
10
15
20
25
30
35
Engineering Stress (MPa)
Engineering Strain (%)
10.9 dpa
85 appm He
6.8 dpa
220 appm He
5.1 dpa
140 appm He
3.1 dpa
17 appm He
0 dpa

Fig 3



Uniform Elongation (%)
100
10
1
0.1
0
5
10
15
20
Dose (dpa)
Temp. Range
He/dpa
< 10
~30
200 ≤ T < 250
250 ≤ T ≤ 350
350 < T ≤ 400

Fig 4