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PROCEEDINGS OF SECOND TOPICAL MEETING ON FUSION REACTOR MATERIALS

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THE MICROSTRUCTURAL ORIGINS OF YIELD STRENGTH CHANGES IN AISI 316 DURING FISSION OR FUSION IRRADIATION

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The changes in yield strength of AISI 316 irradiated in breeder reactors have been successfully modeled in terms of concurrent changes in microstructural components. Two new insights involving the strength contributions of voids and Frank loops have been incorporated into the hardening models. Both the radiation-induced microstructure and the yield strength exhibit transients which are then followed by saturation at a level dependent on the irradiation temperature. Extrapolation to anticipated fusion behavior based on microstructural comparisons leads to the conclusion that the primary influence of transmutational differences is only to alter the transient behavior and not the saturation level of yield strength.

1. INTRODUCTION

AISI 316 stainless steel is the major structural alloy employed in the fast reactor programs of the United States, Britain and France. It may also be chosen for service in first generation fusion devices. In the absence of data developed in fusion neutron spectra, the design of such devices requires that the data developed in fission reactors be extrapolated to the spectral and operational environment projected to be characteristic of fusion devices.

If the consequences of differences in environmental variables such as neutron flux and stress state are discounted there are still substantial differences in the displacive and transmutational characteristics of the neutron spectra of fusion devices and those of various fission reactors. To a first approximation it appears that the differences in atomic displacement characteristics of neutrons can be adequately factored into a fission-fusion correlation by expressing the exposure dose in damage energy units (eV/atom). This was recently demonstrated for AISI 316 irradiated at room temperature with T(d,n), Be (d,n) and fission (thermal reactor) neutrons.(1) The changes in yield strength and total elongation, as well as in the density of radiation-induced defect clusters that cause these changes, were all found to correlate with damage energy.

In reactor environments operating at realistic power-generating temperatures, however, the microstructural alterations are quite different and much more extensive than that observed at room temperature. There is also a substantial radiation-induced and temperature-sensitive elemental redistribution that occurs. There is concern that in the two neutron spectra the differences in both the identity and generation rates of gaseous (2) and solid (3) transmutants will lead to differences in the microstructural (4) and microchemical (5) evolution. This possibility can be partially addressed by comparing the microscopic response of the steel to irradiation in two fission reactors with different transmutational characteristics. A recent effort of this type concerned the microstructural and microchemical evolution of this steel in EBR-II (Experimental Breeder Reactor II) and HFIR (High Flux Isotope Reactor).(6)

In this paper the microstructural origins of yield strength changes are examined and projections of anticipated behavior in fusion spectra are made on the basis of conclusions drawn from dual ion irradiations and EBR-II/HFIR comparisons.

2. MODELING OF YIELD STRENGTH CHANGES

The modeling effort requires knowledge of the relevant microstructural components, their densities and sizes, and their individual action with respect to determination of the yield strength. Previous attempts on austenitic alloys have been made to determine the contribution of each microstructural component to the hardening or softening observed for a given set of irradiation conditions.(7-10) These efforts were hampered by incomplete microstructural descriptions and some ambiguity concerning the nature of the hardness model for each component.

Recent developments now allow a potentially more successful microstructural description of yield strength. Breeder reactor programs have yielded not only detailed strength data as a function of neutron exposure, temperature and starting condition,(11-16) but have also provided insight on the nature of the microstructural/microchemical evolution of this alloy. The central insight is that all microstructural components evolve toward saturation densities which are only functions of temperature and displacement rate. It has been shown that the yield strength also saturates at a level which is dependent on irradiation temperature (Figure 1) but not cold-work level (Figure 2). Although there is some variation in the strength of various heats of unirradiated AISI 316, Blackburn and coworkers (17) have shown that the radiation-induced strength changes reported for breeder-irradiated heats of annealed AISI 316 are independent of heat identity and depend only on temperature.

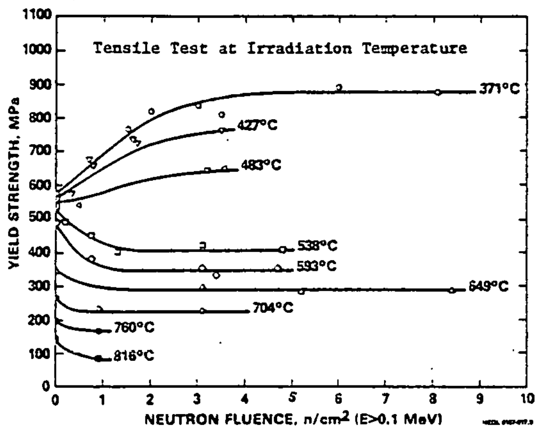

FIGURE 1. Yield Strength of 20% Cold Worked AISI 316 After Irradiation in EBR-II.

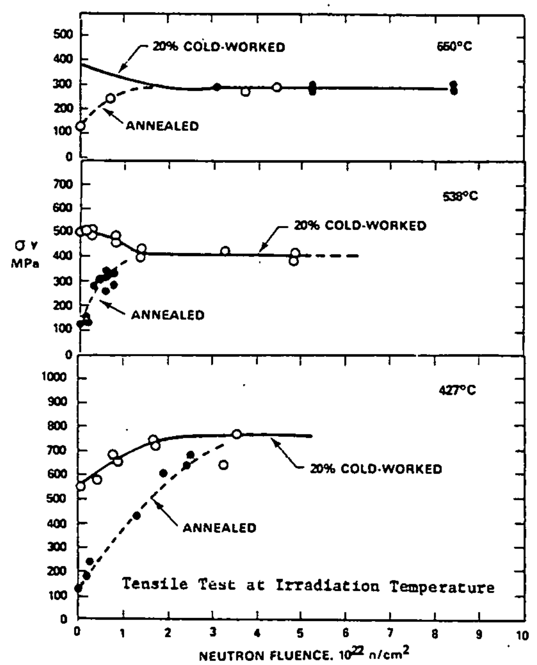

FIGURE 2. Yield Strength of Irradiated AISI 316 at Temperatures of 427, 538 and 650°C.

It is traditional to describe irradiation hardening with models that invoke the interaction of various defects with moving dislocations.(18) Barriers which resist the motion of dislocations have been classified as either long range (LR) or short range (SR). Long range forces are due to the interaction of moving dislocations with the dislocation network of the material. Obstacles lying in the slip plane of the moving dislocation produce short range forces when the dislocation is in close proximity to the obstacles. The increase in the shear stress, $\Delta\tau$, is given by:

$$\Delta\tau = \Delta\tau_{LR} + \Delta\tau_{SR} \tag{1}$$

The long range term is given as

$$\Delta\tau_{LR} = \alpha G b \sqrt{\rho_d}, \tag{2}$$

where ρ_d is the dislocation density, G the shear modulus, b the Burgers vector and α ranges from 0.15 to 0.30.(19) The short range contribution of an obstacle is

$$\Delta\tau_i = \frac{Gb\sqrt{N_i d_i}}{\beta_i} \tag{3}$$

and

$$\Delta\tau_{SR} = [\Sigma\Delta\tau_i^{\,2}]^{1/2}, \tag{4}$$

where N_i is the number of defects of a given type and diameter d_i per unit volume and β_i is a constant for each type of defect. The β values for loops range from 2 to 4,(18-20) while those for voids (21) and precipitates (22) are about 1.0.

Finally, in calculating the flow or yield stress, it is necessary to convert from shear stress to uniaxial stress, namely $\Delta\sigma_y = 3\sqrt{\Delta\tau}$ based on the Von Mises criterion. Thus, the strength of the irradiated steel, σ_i is

$$\sigma_i = \sigma_o + \Delta\sigma_{cw} + \Delta\sigma_i, \tag{5}$$

where $\sigma_o + \Delta\sigma_{cw}$ is the intrinsic strength plus work-hardening of the unirradiated steel. σ_o is assumed not to change during irradiation.

The microstructural data required to generate yield stress predictions using equations 1-5 have been presented earlier.(11) With only minor modifications, these same data were employed using the values $\beta_{FL} = 3$, $\beta_v = 1$ and $\beta_p = 1$. The value of $\alpha = 0.2$ was determined from the room temperature strength data (Figure 3) and the knowledge that the dislocation density of this heat of steel in the 20% cold-worked condition is approximately 3×10^{11} cm/cm^3.(4) Since there are no discernible precipitates in this steel at room temperature and none that develop on the short time frame of tensile tests at elevated temperatures, it is reasonable to assume that the softening that occurs is due to the temperature dependence of the shear modulus and the relaxation of dislocation densities. Therefore, the data of Figure 3 were also used to generate the initial dislocation density ρ_d^{cw} of the cold-worked steel after relaxation at temperature and prior to significant irradiation.

$$\rho_d^{cw} = \left[\frac{\tau_{cw} - \tau_{sa}}{0.2Gb}\right]^2 \tag{6}$$

The preirradiation dislocation density of the solution-annealed (sa) steel was assumed to be 10^8 cm/cm^3 at all temperatures.

3. RESULTS OF CALCULATIONS

In a previous report, it was shown that the major features of the evolution in strength of AISI 316 could be described using the above models.(11) However, the microstructurally-based correlation

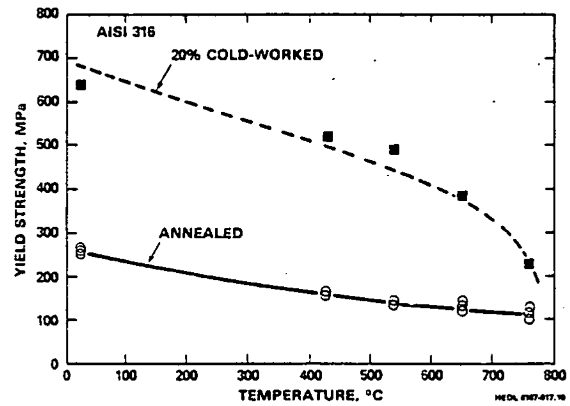

FIGURE 3. Yield Strength of Unirradiated AISI 316.

tended to underpredict the observed behavior at most temperatures. In reviewing the assumptions employed in the treatment of Fleischer (20) for Frank loops it was determined that at 300-600°C, Frank loops cannot be considered as short range obstacles. The original model was normalized to data derived from low temperature irradiation, leading to low dislocation densities and very high densities of very small loops. At temperatures relevant to fast reactors, Frank loops are two-dimensional in nature and very large, with diameters on the order of the dislocation spacing; they also contain a large fraction of the total dislocation line length.

Assuming for the moment that equation 3 is still valid and β_{FL} = 3.0, the Frank loop contribution to hardening was treated as a long range contribution; the loop contribution was included in equation 2 and not 4. As shown in Figure 4, the microstructurally-based predictions exhibit remarkable agreement with the data.

4. DISCUSSION

It can be seen in Figure 4 that the relaxation of dislocation density of cold-worked steels during irradiation leads to a softening contribution to yield strength, particularly evident at temperatures above ~550°C. The presence of voids, Frank loops and radiation-stable precipitates leads to a hardening contribution, particularly at lower temperatures where the densities of these defects are largest. Since these components are sensitive to displacement rate, it is expected that the saturation strength will vary with neutron flux and spectra. (11)

If Frank loops are not short range obstacles, why was the short range formulation of equation 3 employed for loops in this study? Actually, the apparent success of the "wrong" model was fortuitous since the total line length of Frank loops is $\rho_{FL} = N_{FL}\,(\pi d_{FL})$. Therefore,

$$\frac{\sqrt{N_{FL}\ d_{FL}}}{\beta_{FL}} = \frac{\sqrt{\pi\ N_{FL}\ d_{FL}}}{3\sqrt{\pi}} = 0.19\sqrt{\rho_{FL}} \qquad (7)$$

and the short and long range formulations are identical; the value of 0.19 agrees with the value of α = 0.2 determined earlier. Since the predictions agree with the data at high temperatures where dislocations provide the major hardening contribution, it appears that the assumption is valid that the intrinsic yield stress σ_o is largely unaffected by radiation-induced segregation. The major effect of the microchemical evolution on yield strength thus lies in the formation of radiation-stable precipitates.

The data of Figure 1 are from steel that had not developed large void swelling at the fluences attained. Although the strength saturates at all temperatures, it is expected that a late-term softening will occur as the voids change the shear modulus of the steel, particularly at low temperatures. Several researchers (23-24) have shown that the modulus of irradiated stainless steels is decreased according to the relationship $G' = G(1-2\ \Delta V/V_o)$. Decreases in modulus of 20% have been observed for 10% swelling. (24)

5. PROJECTION TO FUSION ENVIRONMENTS

In principle, it should be possible to predict strength changes in fusion environments by studying the response of fission-induced microstructure to anticipated changes in transmutants. The available neutron data fall into two classes: direct comparisons of irradiation response in two reactors with different spectra and studies in one reactor involving elemental variations of expected transmutants. In a recent comparative study of AISI 316 it was shown that the microchemical and microstructural evolution at 500-620°C was remarkably insensitive at 40-70 dpa to more than two orders of magnitude difference in helium generation rates found in EBR-II and HFIR reactors. (6) This study was complicated by the burn-out of manganese and the formation of ~0.8% vanadium at relatively moderate doses in HFIR. (3) Neither of these elements suffer measurable changes at comparable doses in EBR-II. Breeder reactor studies have also shown a sensitivity of void swelling to manganese content (3) and also to carbide-forming elements similar to vanadium. (5) Both of these elements are projected to increase slowly in fusion irradiation of AISI 316, with possible late-term consequences in phase stability and yield strength.

In another study it has been shown that large helium generation rates in AISI 316 irradiated in HFIR lead to earlier void nucleation at all temperatures in the range 300-700°C, but not to void densities which are substantially different at saturation than those that develop in breeder reactors. (25) Therefore, the effect of helium is concentrated in the transient regime.

Dual ion irradiation experiments to date have concentrated only on the gaseous transmutants. The development of Frank loops and their subsequent conversion to dislocations have been shown to be sensitive to the helium level. (26-29) At high fluences, however, the dislocation density

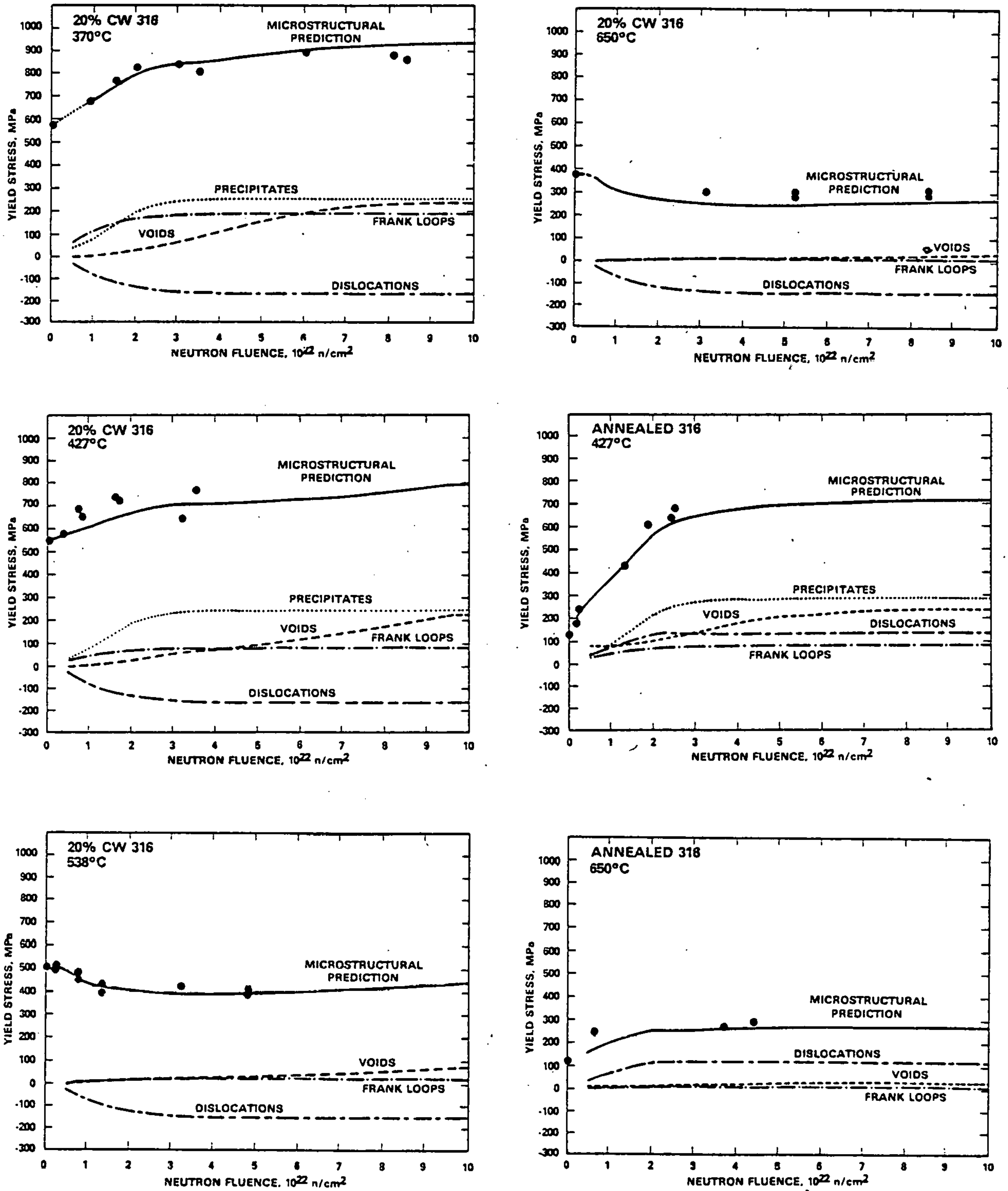

FIGURE 4. Comparison of Microstructurally-Based Predictions With Yield Stress Measurements of AISI 316. The separate strengthening contributions of each microstructural component are shown.

has been found to be remarkably insensitive to helium concentration or the mode of implantation.(28-29) It also appears that large levels of helium can alter the early phase evolution, particularly in accelerated simulation experiments.(26-27) There are no supporting neutron data.

Since the steady-state swelling rate is insensitive to helium content (6) and the saturation dislocation density is also unaffected it appears that the primary effect of helium is in the transient regime, particularly the contributions of voids and Frank loops. Unless solid transmutants change the precipitation behavior or matrix strength, no effect is anticipated in the saturation regime. This conclusion may not be valid for pulsed reactor systems due to the flux sensitivity of voids, Frank loops and radiation-stable precipitates.(5)

6. CONCLUSIONS

Yield strength changes in AISI 316 during breeder reactor irradiation can be described in terms of microstructurally-based models. Microchemical changes are of importance only in that they lead to extensive precipitation. The saturation observed in yield strength is a reflection of the saturation of microstructural densities, but it is expected that a second-order softening will eventually occur as accumulated voidage decreases the shear modulus. While voids and precipitates can be considered as short range obstacles, large Frank loops cannot and therefore should be treated as additional dislocation line length.

The projection of these insights to the anticipated response of AISI 316 in fusion devices yields the conclusion that under steady-state irradiation, the strength behavior at a given displacement rate will differ primarily in the transient and not the saturation regime.

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