Irradiation Induced Strengthening in the UWCTR Blanket

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FUSION TECHNOLOGY INSTITUTE

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INTRODUCTION

The effects of neutron irradiation on the tensile properties of various reactor structural materials has been studied quite extensively. The changes in the tensile behavior that occur during irradiation at various temperatures and fluences are of importance in contemporary light-water reactors, and will be even more influential on the designs of the fast breeder and fusion reactors.

In reactor design it is important that strength of the structural components be known accurately. Due to the large amounts of structural material that are required for a reactor, especially for the larger fusion reactor systems proposed, the cost of this material will influence the economics of the plant. In analyzing reactor accidents it is advantageous to know the strength of the structure so an adequate, but not excessive, safety margin can be added in order that the system can withstand the loads incurred. If the maximum strength required by the design can be satisfied by a smaller amount of material, a savings will result. Also, the less structural material there is in the reactor, the lower will be the parasitic capture in the structure. This is important in both fast breeders and fusion reactors where it is essential to breed new fuel. Neutron capture by the structure also leads to after-heat problems and radioactivity considerations.

As will be discussed in this paper, irradiation results in a strengthening of solution annealed metals. This would seem to be advantageous. However, accompanying this strengthening is a decrease in the ductility of the metal. It is not desirable to have an increase in strength if there is a large drop in the ductility. Embrittlement is a very serious problem, but it will not be considered in this presentation.

A brief discussion of the general characteristics of strength in metals will first be presented. Then the effects of irradiation on strength will be considered, specifically with regard to austenitic stainless steels. Finally, the effects irradiation-induced strengthening might have on the UWCTR design will be discussed.

GENERAL CHARACTERISTICS OF STRENGTH IN METALS

Before the effects of radiation on strength are considered, a brief summary of yielding in unirradiated metals under applied stress will be presented. Consider a metal sample undergoing a tensile test. A typical stress-strain curve for such a sample is given in Figure 1. For low stresses the deformation is entirely recoverable when the load is removed. The material is said to behave elastically. There is a linear relationship between stress and strain given by Hooke's Law,

$$E = \frac{\sigma}{\varepsilon}$$

E is Young's Modulus; σ is the stress; and ε is the resultant strain.

However, if the applied stress is high enough, above the value indicated YS in Figure 1, the specimen will not return to its original shape once the load is removed. The sample has deformed plastically. The stress at which there is a transition from purely elastic behavior to the occurrence of some permanent deformation is defined as the yield or flow stress. As the stress is increased above the yield stress plastic deformation increases. If for example the sample is stressed to point P, once the load is removed the elastic strain will be recovered (as indicated by the dashed line from P to B in Figure 1). The permanent strain is indicated by point B on the strain axis. Once a metal has yielded, elastic strain continues to increase with increasing stress, but this is negligible when compared to the plastic strain.

There are two general types of stress-strain curves. Figure 1 shows the curve for a metal which exhibits a sharp yield point. This curve is typical of iron and low carbon steels. Two yield points may be distinguished. This type of behavior illustrated in Figure 1 is often termed discontinuous yielding. When the stress reaches the value YS in Figure 1, the material begins to yield plastically. This point is termed the upper yield point (UYP). As the material yields, there is a drop in the stress required for continued deformation. This lower stress is called the lower yield point (LYP) and is characterized by considerable plastic strain at an almost constant stress. Discontinuous yielding is caused by foreign atoms in the matrix, such as carbon. They exhibit a tendency to find stable locations in the lattice, such as the spaces just beneath the extra row of atoms of an edge dislocation. These anchor the dislocation against movement. Before slip can occur these dislocations must be moved. Once the dislocation moves away from its original site, it requires a lower stress to continue its motion since it is no longer anchored by the foreign atom. Figure 2 illustrates a metal that does not exhibit a distinct yield point. In this case, the yield stress is usually defined as the stress required to give a 0.2% plastic strain (indicated in Figure 2). This value is often referred to as the offset yield strength.

Another parameter of importance is ultimate tensile strength, labeled UTS on Figure 1. This is the maximum stress that the metal can withstand. It is not the true stress in the sample, however since it is calculated based on the original cross sectional area of the sample, and not the reduced area that is present under such high tensile loads. Beyond the maximum in the stress-strain curve, necking occurs and the metal eventually fractures (indicated by an x in Figures 1 and 2).

Figures 1 and 2 are examples of the stress-strain behavior of ductile metals. The stress-strain curves typical of brittle materials can be seen in Figure 3. A perfectly brittle material exhibits no plastic deformation prior to fracture. Its stress-strain curve is entirely elastic (Figure 4(a)). The shape of the stress-strain curve and the value of the yield stress are dependent on many factors. Among these are the metal or alloy, temperature, grain size, microstructural features, and the testing equipment and procedure. Figure 4 shows that increasing the temperature decreases the yield stress. The effects of some of the other variables will be discussed later.

Plastic deformation results from the movement of dislocations through the matrix as a result of the applied stress. Dislocations can originate from several types of sources inside the material. It is also believed that they can multiply while traveling through the matrix.[1] One source of dislocation generation might be an impurity or precipitate particle. Dislocations attached to precipitates can act as anchors for a Frank-Read source. The Frank-Read source generates dislocation loops under shear stress. Figure 5 shows the various stages in the development of a loop from a Frank-Read source. The dislocation anchored between x and y generates the loops. The applied stress causes the dislocation to bulge outward as shown by the profiles a, b, and c. At point m annihilation occurs and the result is a loop and the original dislocation between x and y regenerated (indicated by d in Figure 5). A more complete description of the Frank-Read source can be found elsewhere.[2-4] Grain boundaries and twin boundaries also act as sources for dislocations. Other sources are vacancy platelets and the original "grown in" dislocation network.[1]

The yield stress can be represented by the summation of the effect of the various forces required to move dislocations. These will be mentioned briefly. More detailed discussions can be found elsewhere.[1,5] One of the forces to be considered is the Peierls-Nabarro force.[1,5] This is the force required for a dislocation to surmount a "potential hill" between two equilibrium positions. This force is highly sensitive to the atomic structure and difficult to calculate. The contribution of the Peierls force to the yield strength is uncertain.

Another factor influencing dislocation motion is the stress field that exists around all dislocations. These stresses can both aid and hinder dislocation motion. Dislocations that become "piled up" at a strong barrier can create a sizeable internal stress field as well.

Even in well annealed metals, little plastic deformation can occur without the occurrence of a large number of dislocation intersections. The interactions between intersecting dislocations may be attractive or repulsive in nature. If an attraction occurs, the intersecting dislocations will form a juncture of maximum length in which case a large amount of energy is released (hundreds of eV). For the dislocation to keep moving, this juncture must be broken, in which case this energy must be supplied by the applied stress. This results in a high resistance to flow. The strength of repulsive intersections is hard to estimate, but is much less than that for the attractive mechanism.

Jogs are another factor that must be considered. Jogs are created when dislocations intersect. They also create resistance to flow. In addition, when a dislocation containing a jog moves it may result in the increase or decrease in a row of point defects (Figure 6). Since energy is needed to produce these point defects, the jog exerts a force on the dislocation which tends to restrain its motion.

Grain boundaries also contribute to the yield strength. When a dislocation encounters a grain boundary, its slip plane ends and it is difficult for it to proceed onward, since slip planes of adjacent grains are not usually aligned. Therefore in order for the dislocation to continue on into the next grain, it must by a zig-zag motion into a slip plane in the adjacent grain. This results in a disturbance at the grain boundary. In general the smaller the grain size, the higher the yield stress, since there are more grain boundaries in the sample to inhibit dislocation movement.

Alloying elements and precipitates exert a very strong influence on the yield strength. The effect of precipitate particles is illustrated schematically in Figure 7. Under an applied stress dislocations move forward in their slip plane. Assuming the particles are "strong" (i.e. The dislocation does not enter the particle itself and attempt to shear the precipitate), the dislocation will bow between the precipitates. In order for slip to continue Orowan[13] postulated that the dislocation must bend to a radius of curvature of one half the particle separation (x in Figure 7). The segment can then expand and move beyond the particles leaving a dislocation loop around the precipitate. The stress, τ , required to accomplish this procedure is approximately the stress required to bend a dislocation into a semicircle of radius $x/2$[12]. This is given by

$$\tau \approx \frac{2T}{bx}$$

Where T is the line tension of the dislocation and b the magnitude of the Burgers vector. Strictly speaking the space between the particles through which the dislocation bulges should be used rather than the interparticle spacing, however the error involved is usually negligible. Using the formula for the line energy per unit length of a dislocation, T becomes

$$T = \frac{G_m b^2}{2}$$

where G_m is the shear modulus of the matrix. τ now becomes

$$\tau \simeq \frac{G_m b}{x}$$

The above formula indicates that dislocations can move relatively easy between well-separated precipitates. As the particle spacing decreases, the stress needed to accomplish the above process increases, thus the yield stress is increased to a higher value.

There are other factors that also influence the yield strength in a metal with a dispersion of precipitates. Even when no external stress is applied, there will be "self-stresses" in the matrix associated with the particles. The causes of this stress can be the coherency between the particle and the lattice, a difference in the coefficient of thermal expansion between the precipitate and the matrix, a phase change in the particle or matrix, or the condensation of point defects onto the particle.[12] When the metal is stressed to some value less than the yield point, a local stress pattern occurs in the matrix as a result of the different elastic properties of the matrix and the particle. This effect is local in nature and does not extend out into the matrix away from the particles. The effect needs only to be considered when examining the slip process on a microscopic scale.

IRRADIATION EFFECTS ON STRENGTH

High energy neutron irradiation of a metal displaces atoms in the matrix which are referred to as primary knock-on atoms or PKA's. These PKA's travel throught the lattice themselves interacting to produce vacancies and interstitials. The point defects may then agglomerate and form such structures as dislocation loops, voids or stacking fault tetrahedra. As the fluence level increases, the concentration of these structures increases, until possible saturation occurs. From the discussion in the previous sections, the strength of a given metal is dependent on the stress required to move dislocations through the matrix. In view of the barriers to dislocation motion already discussed, it follows that irradiation produced defect aggregates will increase the resistance to dislocation movement through the matrix for a given stress.

Bement[11] has categorized the various barriers as to the force required to move a dislocation past them. Vacancies and solute atoms with symmetrical strain fields provide the weakest barriers. Faulted dislocation loops and voids are classified as relatively strong barriers.

It has been shown experimentally that voids act as hard barriers.[9] It is possible that voids can be cut by a dislocation and examples are shown in Figure 8. However the forces postulated are quite high and the entire process has not been examined in great depth.[11]

The increase in yield stress due to voids and dislocation loops has been treated in some detail.[35-38] Garr et. al.[14] have utilized these treatments in predicting yield stress increases. The basic equations used are outlined below. Coulomb[35] calculates the increase in shear stress due to a void, $\Delta\tau_v$, as

$$\Delta\tau_v = \frac{Gb}{2\beta L}$$

where β is a constant whose value is approximately one. L is the spacing between voids which Westmacott et. al.[28] have computed to be

$$L = 0.5(nd)^{-1/2}$$

where n is the void number density and d is the average void diameter.

Fleischer [37] calculates the increase in shear stress due to dislocation loops, $\Delta\tau_\ell$, as

$$\Delta\tau_\ell = \frac{Gb}{2.5L}$$

In this case L is the average spacing between loops in the slip plane. It is given by the same formula as L for voids, however n is now the loop density and d is the average loop diameter.

The total increase in shear stress is given by[38]

$$\Delta\tau_{shear} = (\Delta\tau_v^2 + \Delta\tau_\ell^2)^{1/2}$$

The increase in yield stress is given by the expression

$$\Delta\sigma_{tensile} = 2\Delta\tau_{shear}$$

Table I shows some results obtained by Garr et. al.[14] using these equations, and a comparison of these numbers with the results of tensile tests. These numbers indicate that loop hardening is more important than void hardening. Olsen and Beck[22] used a similar procedure and found void hardening to be dominant. The relative effectiveness of voids and loops in strengthening is strong function of irradiation temperature and fluence. This effect will be seen later when strengthening in the UWCTR design is considered.

Figure 9 is an example of how irradiation affects the stress-strain relationship. In Figure 9, irradiation results in a distinct yield point where formerly the metal had none. The opposite effect may also occur. A sharp yield point in an unirradiated sample may disappear with irradiation.[7] In general the parameters such as the proportional elastic limit, offset yield strength, yield point, and ultimate tensile strength are all increased as the irradiation dose increases for solution treated metals.

Irradiation may also result in a decrease in the ability of a metal to strain harden. Strain hardening is the increase in strength resulting from plastic deformation. Once a metal has been deformed plastically, it requires an increasingly higher stress in order to increase the level of permanent damage. Figure 9 illustrates the decrease in the strain hardening ability of a material. The slope of the stress-strain curve once the yield point has been exceeded for the irradiated curve than for the unirradiated plot. The smaller slope indicates a lower strain hardening effect.

IRRADIATION STRENGTHENING OF SOLUTION ANNEALED AUSTENITIC STAINLESS STEELS

All experimental evidence with solution annealed 316 stainless steel confirms the theory outlined previously: The yield strength and ultimate tensile strength increase as a result of irradiation. The various parameters that affect the yield strength of austenitic stainless steels will now be discussed.

Temperature

Temperature, both during irradiation and during testing is an important variable in the tensile behavior of all metals. Figure 10 shows how the yield stress and ultimate tensile stress vary with irradiation temperature for solution treated 316.[30] The irradiation temperature and the testing temperature were approximately equal in the samples tested. The yield stress increase after irradiation is highest at the lower temperatures. The change in yield stress due to irradiation decreases as the temperature increases. At about 800^{o}C the irradiated and unirradiated samples have approximately equal yield strengths. Figure 10 also shows that the ultimate tensile strength increases very slightly at the lower temperatures following irradiation. At higher temperatures, irradiation causes the ultimate tensile strength to drop slightly compared to the control value. Figure 10 also shows the strong temperature dependence of the ultimate tensile strength for unirradiated 316, while the yield stress shows only a slight decrease with increasing temperature in the unirradiated metal.

Fish et. al.[26] have irradiated solution treated 316 to 7x10^{22} n/cm^{2} (E > 0.1 MeV) at 760^{o}C and found no change in yield stress.

Early, low fluence data indicated that above irradiation temperatures of $1/2T_m$ (T_m - absolute melting point), significant hardening would not occur.[9,18] This was thought to be caused by the high rate at which defects would anneal out at these temperatures. However, more recent and higher fluence data such as that just cited indicate that this is not the case.[9,26,29,30] Bloom and Fahr[29] have found that above 600°C precipitation of carbides and the sigma phase is the controlling factor. Yield strengths of samples aged for 4000 hours at 650°C show an increase in yield strength the same as a solution annealed sample undergoing irradiation at the same temperature to a fluence of 1.5 to 2.3 x 10^{22} n/cm^2 ($E > 0.1$ MeV). The data of Bloom and Fahr is plotted in Figure 11.

Holmes et. al.[8] **conclude** that below 650°C strengthening is due to voids and loops. At 650°C Frank loops disappear rapidly. Any strengthening effects at temperatures above this are due to voids and the dislocation network. At 750°C voids anneal out and little or no increase in yield strength is predicted. This effect is confirmed by the results of Fish et. al.[26] mentioned above. Holmes et. al.[8] conclude that elevated temperature strengthening is caused by the formation of large point defect aggregates. They also found that the irradiation and test temperatures at which radiation hardening is no longer significant increases with higher fluences.

The effect of the testing temperature on the yield stress is illustrated in Figure 12 for solution annealed 304 stainless steel irradiated to a fluence level of 1.4 x 10^{22} n/cm^2 ($E > 0.18$ MeV) in EBR-II.[9] The irradiation temperature for these tests was calculated to be 538°C ± 48°C (0.49 ± 0.03 T_m, where T_m = 1672°K). The data was corrected for any temperature effects on the shear modulus by multiplying the yield stress by the ratio of the shear modulus at 21^{o}C to the shear modulus at each test temperature. Figure 12 shows that the yield stress in unirradiated 304 stainless steel decreases very slightly with increasing temperatures over the range from 100 to 871^{o}C. The irradiated samples show a large increase in strength at low temperatures which decreases in magnitude as the testing temperature is raised. At 816^{o}C (0.65 T_m) the yield stress increase is insignificant.

Fluence

Generally speaking, as the exposure increases, the yield strength increases. This can be seen in Figure 13.[26] The offset yield strength is plotted versus fluence for several irradiation temperatures. The increase in yield stress is the highest for the low testing and irradiation temperatures, as has been mentioned earlier. It can also be seen from Figure 13 that the threshold fluence, above which an increase in yield stress due to irradiation is observed, increases with increasing irradiation temperature. At 427^{o}C (800^{o}F) the threshold fluence is predicted to be approximately 5 x 10^{20} n/cm^2 (E > 0.1 MeV). For an irradiation temperature of 704^{o}C (1300^{o}F), an increase in yield stress is not observed until approximately 2 x 10^{21} n/cm^2 (E >0.1 MeV). This behavior can be more clearly understood if the Coulomb relation [35] presented earlier is examined.

$$\Delta\tau_v = \frac{Gb}{2\beta L}$$

Substituting the value of L,

$$\Delta\tau_v = \frac{Gb}{\beta} (nd)^{1/2}$$

Although this equation is for voids, the same general remarks will hold for dislocation loops.

In general, as the irradiation temperature increases, the void number density decreases and the average void size increases for a constant fluence. However the void number density decreases much faster than the void size increases, thus the product nd will be less for an equivalent fluence at a higher temperature. This explains why the increase in yield strength is less for higher temperatures. The product nd increases with increasing fluence at any given temperature. At high temperatures where nd is less it takes a larger fluence to produce a noticeable change in yield strength, thus the higher threshold dose at higher irradiation temperatures.

There may be a saturation in the yield strength at high fluence values. This effect is indicated in Figures 14 and 15 which shows tensile tests on 304 stainless steel EBR-II safety rod thimbles.[31] The samples were tested and irradiated at roughly the same temperature. The two figures show that for irradiation temperatures of 371°C (700°F), 427°C (800°F), and 482°C (900°F), the yield strength increases rapidly for fluences up to 1 to 2 x 10^{22} n/cm^2 (E > 0.1 MeV). Above this value the fluence dependence of yield strength diminishes. Fish et. al.[31] state that above 7 x 10^{22} n/cm^2 (E > 0.1 MeV) the yield stress and ultimate tensile stress are constant. One data point in Figure 14 seems to indicate that the yield strength may actually decrease at high fluences. However, due to the insufficient number

number of tests and the possible scatter of the data, the authors drew no conclusions based on this point.[31]

Examining the Coulomb formula again, an explanation of saturation can be obtained. For both voids and loaps, as the fluence level increases at a constant temperature, the defect size increases very slightly while the number density increases rapidly at first and later tends to saturate. As a result the product nd levels off somewhat at higher fluences, and the increase in yield stress changes very little at higher fluences.

Flux

Little experimental work has been done to examine the effect of different flux spectra on the strength of austenitic stainless steels. Kangilaski and Bauer[27] have compared 347 stainless steel tensile test results from irradiations in ETR and EBR-II to equivalent fast fluence levels (3×10^{22} n/cm^2, E > 1 MeV). The former reactor has a mixed thermal and fast flux while EBR-II has predominantly a fast flux. The results shown in Figures 16 and 17 indicate a slightly larger increase in strength for the ETR irradiation at low testing temperatures. At higher irradiation temperatures the EBR-II spectrum results in the highest values of yield and ultimate tensile strength. At 750°C all the irradiation produced increase in yield strength is annealed out during testing.

IRRADIATION STRENGTHENING OF COLD WORKED AUSTENITIC STAINLESS STEELS

Cold worked samples show a different tensile behavior following irradiation. Reports have shown that the yield strength and ultimate tensile strength decrease upon irradiation.[14,30] This is opposite to the behavior of solution treated steels discussed in the previous section. Figure 18 shows the effect of irradiation on the yield stress and ultimate tensile stress for 20% cold worked 316 stainless steel irradiated from 1.5 to 2.7 x 10^{22} n/cm^2 (E ≥ 0.1MeV).[30]

The large increase in yield strength due to cold work alone can be seen by comparing the unirradiated yield stress values at 450^{o}C for 20% cold worked 316 (Figure 18) with the 450^{o}C yield stress for unirradiated solution annealed 316 (Figure 10). Yield stress is increased by a factor of six (~20 ksi to 130~ksi) by cold working. Above 450^{o}C, the yield stress in unirradiated cold worked 316 drops rapidly and eventually equals the solution treated yield strength at about 850^{o}C. Figure 18 shows that for all irradiation temperatures, irradiation of 20% cold worked 316 results in a slight decrease in the yield and ultimate tensile strength. It has been postulated that above 450^{o}C irradiation may enhance recovery thus contributing to the lower values observed in the irradiated samples.

Cold work has been shown to inhibit void formation.[16] As stated earlier voids are a major barrier to dislocation motion. Thus if no voids are present, the large increase in yield strength observed in solution annealed 316 will not be present in the cold worked sample. The Frank dislocation loops that form in cold worked samples during irradiation are not present in large enough quantities to influence the yield stress when compared to the large increases in strength induced by the cold work.[16]

IRRADIATION STRENGTHENING IN UWCTR DESIGN

As mentioned in the introduction to this report, an accurate knowledge of strength is important to fusion reactor design. From an economic standpoint, it is imperative that the minimum amount of structural material be incorporated in the design, provided that design requirements and safety factors are satisfied. The economic impact is much greater in a fusion reactor than a fission reactor simply because the CTR is so much larger, thus requiring a greater quantity of structural material.

Another problem of special concern to the UWCTR is the thermal stress present in the first wall. The internal heating in the wall by the photons and charged particles results in a substantial temperature gradient across the wall. This gives rise to a thermal stress which increases as the wall becomes thicker. The stresses are much higher in 316 stainless steel than for refractory metals, which have also been considered for the first wall.[39] This is caused by the relatively lower thermal conductivity and higher thermal expansion coefficient for the steel. Thermal stresses provide an upper limit for the wall thickness. It is thus important to satisfy the necessary strength requirements in a wall thickness that does not exceed a value where thermal stresses become excessive.

Solution Treated 316 Stainless Steel

Presently there is no tensile test data for solution treated 316 stainless steel that even approaches the high fluence values predicted for the first wall of the UWCTR. The data reported thus far seems to indicate that there will be a saturation in the increase in yield strength (see Figures 14 and 15).[15,31] Assuming that saturation does occur, estimates of the increase in yield stress have been made based on an extrapolation of the most recent data of Fish et. al.[15] His tests were performed at temperatures between 800 and 900°F (427-482°C) at fluence levels less than 3.5 x 10^{22} n/cm^2 (E > 0.1 meV). Two testing temperatures were used, 430°C and 480°C, depending on the calculation of the irradiation temperature for the given sample. The yield stresses for the unirradiated control samples were 19.0 ksi for the 430°C test temperature and 18.5 ksi for the 480°C tests.

In order to calculate the yield stress increases, the fluence at various locations throughout the blanket must be known. Kulcinski has calculated the displacement rates for stainless steel at two locations in the UWCTR blanket.[41] These are located at the first wall and the blanket-shield interface. These two locations are indicated in Figure 19. These displacement rates have been corrected for the present wall loading of 0.53 MW/m^2 and are as follows.

First wall	13.72 dpa/year
Blanket-Shield Interface	0.108 dpa/year

Using these two points, a first approximation of the dpa rate throughout the blanket is shown in Figure 20. This type of approximation is not far from being correct. It tends to underestimate the dpa rate in front of the graphite due to the reflection of neutrons by the graphite. On the outside of the graphite, the displacement rate may be overestimated since, the spectrum has been thermalized significantly and thus is less effective in displacing atoms.

Increases in yield strength were computed for irradiation times of 5 years and 20 years (end of life). Fluence levels were calculated using Figure 20 and Doran's approximation[42] to convert dpa to fluence. Doran has calculated that 7.03 dpa is approximately equivalent to a fluence of 10^{22} n/cm^2 (E > 0.1 MeV) at the center of the EBR-II core. The data on the yield strength increases is listed in Tables II and III, and plotted in Figures 21 and 22.

From examining these figures it is evident that very early in life there will be a substantial increase in the strength of the first wall. After the first year there will be a 66 ksi increase in yield strength in the first wall. By two years the saturation value should be reached at the first wall. Increases in yield stress beyond the reflector occur at a much slower rate but are quite significant by the end of life.

By extrapolating Figure 20 to the stainless steel separating the lead and the B_4C (see Figure 19), the 20 year dpa value in this section of stainless steel is only 0.56 dpa which results in a negligible amount of hardening. Thus for purposes of examining strengthening, only the blanket region needs to be considered.

It should be remembered that these results are only estimates for the temperature range from 427 to 482°C. This omits the colder portion of the heat removal cell and the headers through which the "cold" lithium flows. Due to the previous discussion, it would seem that the increase in yield stress for these regions would be only slightly larger (see Figure 10). Only a very small portion of the blanket is above 482°C, and again, from previous discussions, the increase in yield stress is well approximated by the values presented.

The critical assumption is that the yield stress saturates. Although there is no 316 data to substantiate this assumption, 304 tensile tests for samples exposed to fluences greater than 10^{23} n/cm^2 (E > 0.1 MeV) show a leveling off of yield strength at high fluences (See Figures 14 and 15).[31]

Cold Worked 316 Stainless Steel

There has been very little data presented on the tensile properties of cold worked 316 stainless steel after irradiation. The latest data reported by Fish et. al.[16] shows little effect of irradiation on the tensile properties of 20% cold worked 316 below $540^{\circ}C$. The maximum fluence level was only 1×10^{22} n/cm^2 (E > 0.1 MeV) so the data is far below that needed to estimate the tensile properties above 10^{23} n/cm^2. Slightly higher fluence data (up to 2.7×10^{22} n/cm^2, E > 0.1 MeV) has been reported[30] for 20% cold worked 316. This is illustrated in Fiugre 18. A slight drop in the yield stress was observed with irradiation.

Since void nucleation is decreased by cold work, irradiation induced changes in yield strenghth, at lower fluences, will be due primarily to loop formation and any recovery that might occur. Fish et. al.[16] believe that the loop concentration is not significant when compared to the cold work induced dislocation structure as far as strengthening is concerned. However, Brager and Straalsund[40] have recently reported that for 20% cold worked 316 stainless steel irradiated at ~$450^{\circ}C$, that irradiation induced the replacement of nearly the entire dislocation structure by Frank loops. At a low fluence of 0.6×10^{22} n/cm^2 (E > 0.1 MeV), they report reduction in the dislocation density by an order of magnitude and a high density of Frank loops ($\sim 3 \times 10^{15}$ loops/cm^3). This process may result in opposite effects. The recovery and disappearance of the cold work induced dislocation network will result in a lowering of the yield stress. However, the high Frank loop concentration will tend to increase yield stress. Once the fluence level becomes large enough void formation will add to the increase in yield stress.

Relative Effects of Voids and Loops on Strengthening

Since both voids and loops contribute to strengthening, it may be of interest to know what areas of the UWCTR in which the strengthening is due mainly to voids and what areas that Frank loops dominate the hardening. A single heat removal cell unit as illustrated in Figure 23 was chosen for analysis.

In order to examine the effects of voids and loops, it is first necessary to estimate the concentrations and sizes of these defect structures. In doing this, the latest, and highest fluence data, for solution treated 316 stainless steel presented by Brager and Straalsund[43] has been used. The authors estimate that their empirical relations will give good results up to a fluence of 10^{23} n/cm^2 (E > 0.1 MeV). On this basis a 5 year irradiation time in UWCTR was chosen, since the maximum first wall fluence after 5 years is 9.77×10^{22} n/cm^2 (E > 0.1 MeV).

The relations used will now be listed. The mean void diamater is given by

$$d_v = (\phi t)^m \quad \exp[C\ (T)]$$

where:

d_v is in Å

ϕt is the fluence (10^{22} n/cm^2, (E > 0.1 MeV))

m =0.24

$$C(T) = 37.27 - 0.0173T - \frac{14110}{T} \qquad (T \text{ in } ^{o}K)$$

The void number density is given by

$$n_v = (\phi t)^{A(T)} \quad \exp[B(T)]$$

where:

n_v is the void number density (10^{13} voids/cm^3)

$$A(T) = \frac{1.7}{1 + \exp[0.04\ (700-T)]}$$

$$B(T) = -62.34 + 0.0275T + \frac{32880}{T}$$

The mean Frank loop diameter is

$$d_\ell + (\phi t)^{H(T)} \quad \exp\ [J(T)]$$

where:

d_ℓ is the mean loop diameter (Å)

$$H(T) = -6.313 + 0.00262T + \frac{3064}{T}$$

$$J(T) = 23.84 - 0.0071T - \frac{9045}{T}$$

The Frank loop density is given by

$$n_\ell = (\ t)^{0.532} \quad \exp[L(T)]$$

where:

n_ℓ is the loop density (10^{15} loops/cm^3)

$$L(T) = -203.46 + 0.1155T + \frac{85920}{T}$$

Using these empirical relations, values for d_v, n_v, d_ℓ, and n_ℓ were calculated for several points in the heat removal cell. These are labeled by the letters A through K in Figure 23. The values calculated are listed in Tables IV through VII.

The effect of voids and Frank loops on yield strength was examined using the Coulomb[35] and Fleischer[37] relations as presented by Garr et. al.[14] The increase in shear stress due to voids is

$$\Delta\tau_v = \frac{Gb}{2\beta L}$$

The symbols in the above equation have been defined earlier. The increase in shear stress due to loops is

$$\Delta\tau_\ell = \frac{Gb}{2.5L}$$

For voids, L is given by

$$L = 0.5\ (n_v d_v)^{-1/2}$$

where n_v and d_v are the void concentration and mean void diameter respectively.

For loops, L is given by

$$L = 0.5\ (n_\ell d_\ell)^{-1/2}$$

where n_ℓ and d_ℓ are the loop concentration and mean loop diameter respectively.

The increase in yield stress can be expressed as

$$\Delta\sigma_{\text{tensile}} = 2\Delta\tau_{\text{shear}} = 2(\Delta\tau_v^{\ 2} + \Delta\tau_\ell^{\ 2})^{1/2}$$

However for purposes of examining the relative effectiveness of loops and voids only $\Delta\tau_\ell$ and $\Delta\tau_v$ are needed. The following ratio can be formed:

$$R = \frac{\Delta\tau_v}{\Delta\tau_\ell}$$

Letting $\beta = 1.0$, R becomes

$$R = \frac{2.5L_{\ell}}{2.0L_{v}}$$

where the subscripts ℓ and v on L refer to L calculated for loops and voids respectively. Using the definitions of L,

$$R = \frac{1.25\ (n_v d_v)^{1/2}}{1.00\ (n_{\ell} d_{\ell})^{1/2}}$$

When $R > 1, \Delta\tau_v > \Delta\tau_{\ell}$ thus voids will have a greater influence on the strengthening. For $R < 1$, Frank loops dominate the strengthening process.

Values of R were calculated using the data in Tables IV through VII. These values are tabulated in Table VIII. Based on these results, the regions over which voids and loops dominate hardening are estimated in Figure 24. The figure shows that after 5 years irradiation, most of the irradiation induced strengthening in the heat removal cell will be dominated by the formation of Frank dislocation loops.

This situation might be expected to change as void formation and growth increases with fluence. However since there is no neutron irradiation data for the high first wall fluences that will be attained, no reasonable estimate of higher fluence effects will be predicted.

CONCLUSIONS

If solution treated 316 stainless steel is chosen as the chief blanket structural component, it is evident that irradiation induced strengthening will become appreciable throughout the blanket by the end of life. The most drastic increases in yield strength will occur in the heat removal cell regions because the fluences will be much higher. Beyond the lead shield, strengthening effects will be negligible. Tensile test data for solution annealed 316 stainless steel is needed at the high fluence levels (above 5×10^{22} n/cm^2, E $>$0.1 MeV) so more accurate predictions of irradiation induced hardening can be performed.

Irradiation effects on cold worked 316 are not known because again the data does not exist for the higher fluences that will be experienced by UWCTR.

In closing it is important to reiterate an important point. Although an increase in yield strength is a desirable property, the accompanying loss in ductility during irradiation may lead to severe limitations on the UWCTR design. This topic will be considered in a later report.

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TABLE I

TABLE 6
COMPARISON OF CALCULATED AND OBSERVED INCREASE IN YIELD STRENGTH OF IRRADIATED TYPE 316 STAINLESS STEEL

Treatment	Temperature (°C)	Calculated (kN/cm^2)			Observed (kN/cm^2)
		$\Delta\sigma_v$	$\Delta\sigma_\ell$	$\Delta\sigma_{tensile}$	
A	room	7.7	15.9	35.3	31.2
	500	6.4	13.2	29.2	32.4
B	room	7.8	10.0	25.4	33.4
	500	6.4	8.3	21.0	33.0

Pre-irradiation Treatments

A	One hour at 980°C
B	One hour at 980°C and eight hours at 760°C

K. R. Garr, C. G. Rhodes, and D. Kramer, "Effects of Microstructure on Swelling and Tensile Properties of Neutron Irradiated Types 316 and 405 Stainless Steels," AI-AEC-13031, (June 15, 1972).

TABLE II

INCREASE IN YIELD STRESS AS A FUNCTION OF DISTANCE IN THE UWCTR BLANKET FOR SOLUTION TREATED 316 STAINLESS STEEL STEEL AFTER 5 YEARS

Blanket[1] Location cm	dpa/year	5 Year dpa	5 Year Fluence $\times 10^{22}$ n/cm^2 (E > 0.1 MeV)	Yield Stress ksi	Increase in Yield Stress ksi
0	13.7	68.6	9.77	110	91
10	8.5	42.5	6.05	110	91
20	5.2	26.0	3.70	105	86
30	3.5	17.5	2.49	94	75
40	2.0	10.0	1.42	74	55
50	1.2	6.00	0.85	58	39
70	0.47	2.35	0.33	36	17
80	0.29	1.45	0.21	30	11
90	0.18	0.90	0.13	24	5
100	0.11	0.55	0.078	21	2

[1]Distance from first wall.

TABLE III

INCREASE IN YIELD STRESS AS A FUNCTION OF DISTANCE IN THE

UWCTR BLANKET FOR SOLUTION TREATED 316

STAINLESS STEEL AFTER 20 YEARS

Blanket[1] Location (cm)	dpa/year	20 Year dpa	20 Year Fluence $x10^{22} n/cm^2$ (E > 0.1 MeV)	Yield Stress ksi	Increase in Yield Stress ksi
0	13.7	274.	39.0	110	91
10	8.5	170.	24.2	110	91
20	5.2	104	14.8	110	91
30	3.5	70.	9.95	110	91
40	2.0	40.	5.70	110	91
50	1.2	24.	3.42	103	84
70	0.47	9.4	1.34	71	52
80	0.29	5.8	0.82	57	38
90	0.18	3.6	0.51	46	27
100	0.11	2.2	0.31	35	16

[1] Distance from first wall

TABLE IV

MEAN VOID DIAMETER AT VARIOUS LOCATIONS

IN THE HEAT REMOVAL CELL AFTER 5 YEARS

Location cm from First Wall	Temperature ^{o}C	5 Year Fluence $\times 10^{22}$ n/cm^2 E > 0.1 MeV	Mean Void Dia. Å
First Wall			
A*	500	9.77	491
B	450	9.77	329
C	425	9.77	251
D	400	9.77	184
E	350	9.77	81
F	300	9.77	**
10 cm			
G	445	6.12	280
25 cm.			
H	487.5	2.28	316
50 cm			
I	475	0.85	228
J	425	0.85	140
K	375	0.85	70

*letter refers to location in Figure 23.

** empirical equation not applicable at 300°C since voids are not observed in 316 stainless steel at this temperature.

TABLE V

VOID NUMBER DENSITY AT VARIOUS LOCATIONS

IN THE HEAT REMOVAL CELL AFTER 5 YEARS

Location cm from First Wall	Temperature ^{o}C	5 Year Fluence $\times 10^{22}$ n/cm^2 E > 0.1 MeV	Void No. Density $\times 10^{15}$ voids/cm^3
First Wall			
A*	500	9.77	1.71
B	450	9.77	3.44
C	425	9.77	3.54
D	400	9.77	4.24
E	350	9.77	24.16
10 cm.			
G	445	6.12	2.03
25 cm.			
H	487.5	2.28	0.23
50 cm.			
I	475	0.85	0.074
J	425	0.85	0.48
K	375	0.85	6.85

* Letter refers to location in Figure 23

TABLE VI

MEAN FRANK LOOP DIAMETER AT VARIOUS LOCATIONS

IN THE HEAT REMOVAL CELL AFTER 5 YEARS

Location cm from First Wall	Temperature ^{o}C	5 Year Fluence $x\ 10^{22}\ n/cm^2$ $E > 0.1$ MeV	Mean Loop Dia. Å
First Wall			
A*	500	9.77	365
B	450	9.77	325
C	425	9.77	307
D	400	9.77	274
E	350	9.77	230
F	300	9.77	184
10 cm.			
G	445	6.12	344
25 cm.			
H	487.5	2.28	522
50 cm.			
I	475	0.85	652
J	425	0.85	386
K	375	0.85	192

*letter refers to location in Figure 23.

TABLE VII

FRANK LOOP NUMBER DENSITY AT VARIOUS LOCATIONS

IN THE HEAT REMOVAL CELL AFTER 5 YEARS

Location cm from First Wall	Temperature ^oC	5 Year Fluence $\times 10^{22}$ n/cm^2 $E > 0.1$ MeV	Loop No. Density $\times 10^{15}$ $loops/cm^3$
FIRST WALL			
A*	500	9.77	0.16
B	450	9.77	1.05
C	425	9.77	4.31
D	400	9.77	24.1
E	350	9.77	2022
F	300	9.77	**
10 cm.			
G	445	6.12	1.10
25 cm.			
H	487.5	2.28	0.11
50 cm.			
I	475	0.85	0.095
J	425	0.85	1.17
K	375	0.85	49.1

* Letter refers to location in Figure 23.
**empirical equation not applicable at this temperature.

TABLE VIII

RATIO OF THE INCREASE IN SHEAR STRESS DUE TO VOIDS TO THE INCREASE IN SHEAR STRESS DUE TO FRANK LOOPS AT VARIOUS LOCATIONS IN UWCTR AFTER 5 YEARS

Location cm from First Wall	Temperature ^oC	5 Year Fluence $\times 10^{22}$ n/cm^2 E > 0.1 MeV	$R = \frac{\Delta\tau_v}{\Delta\tau_\ell}$
First Wall			
A*	500	9.77	4.75
B	450	9.77	2.27
C	425	9.77	1.02
D	400	9.77	0.43
10 cm			
G	445	6.12	1.53
25 cm.			
H	487.5	2.28	1.39
50 cm.			
I	475	0.85	0.78
J	425	0.85	0.48

*Letter refers to location in Figure 23.

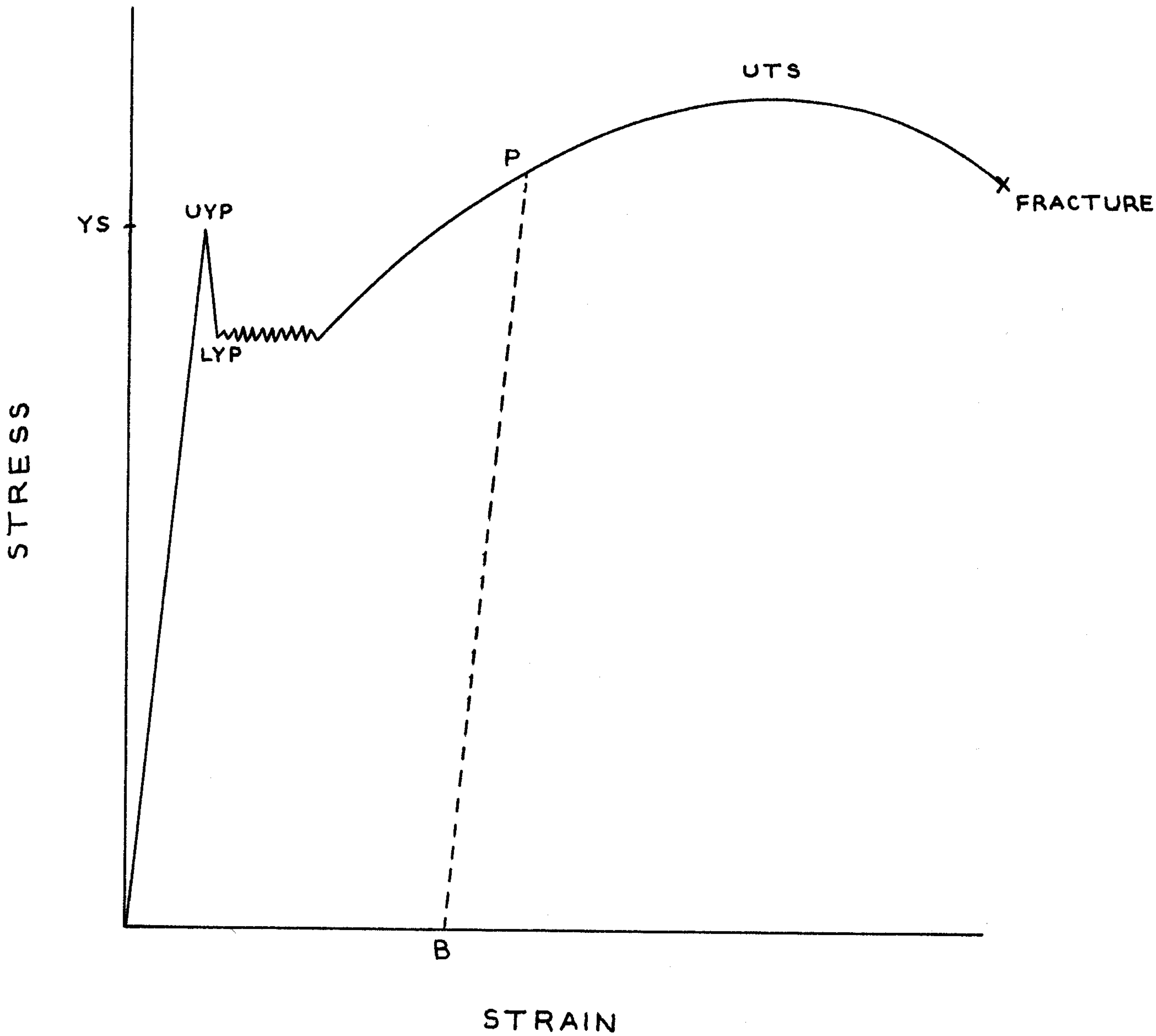


Figure 1

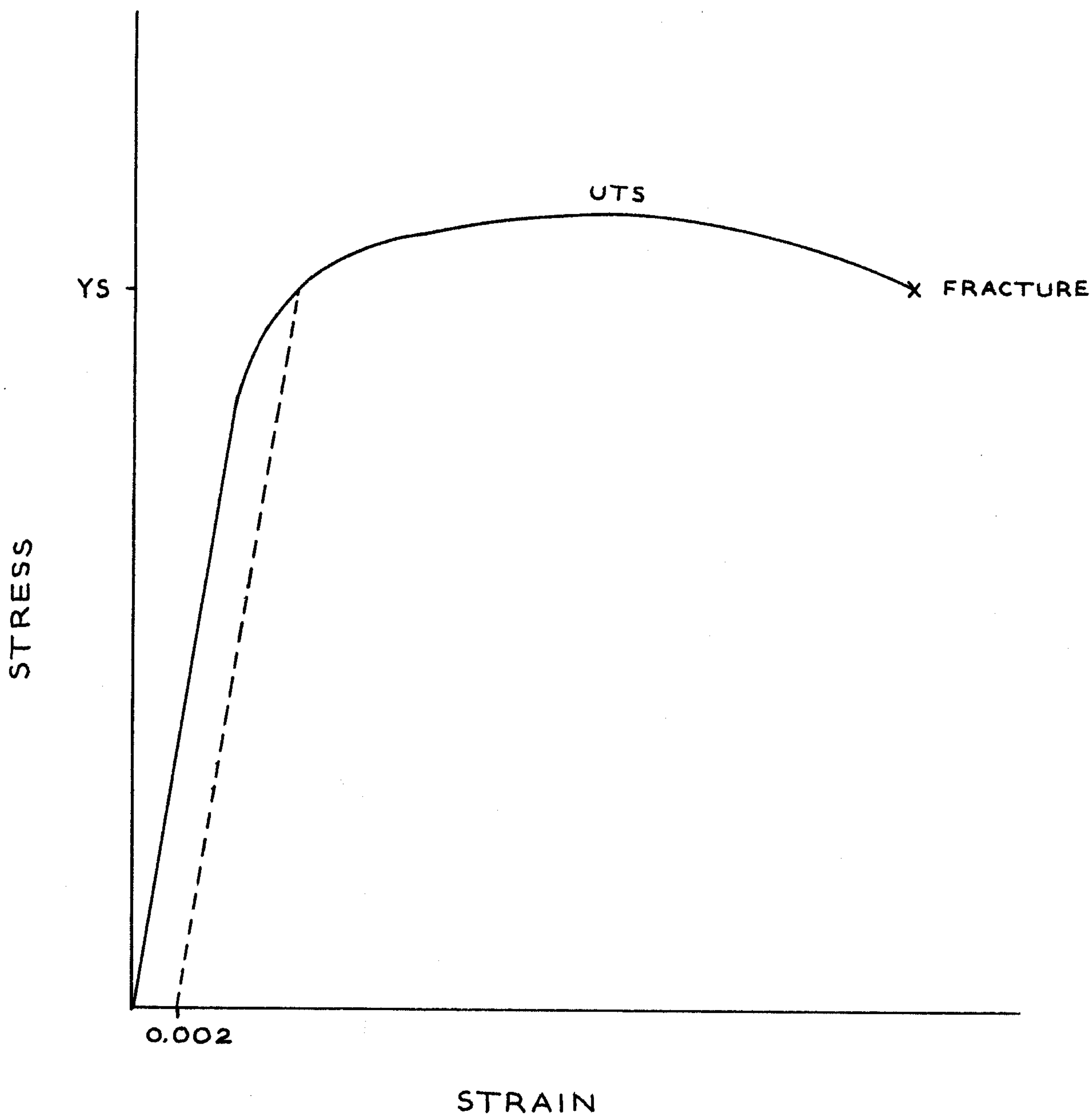
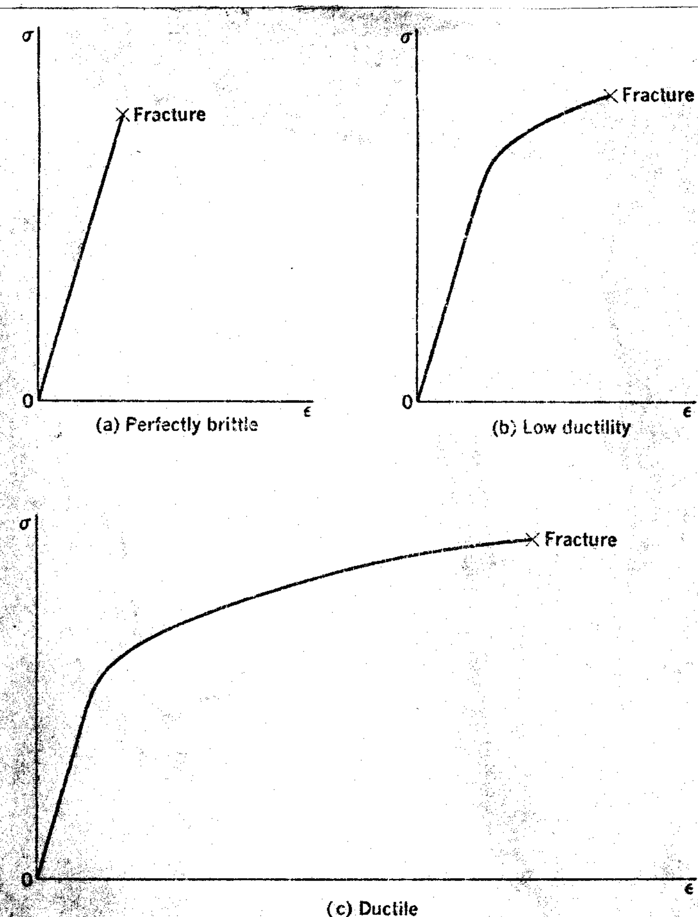


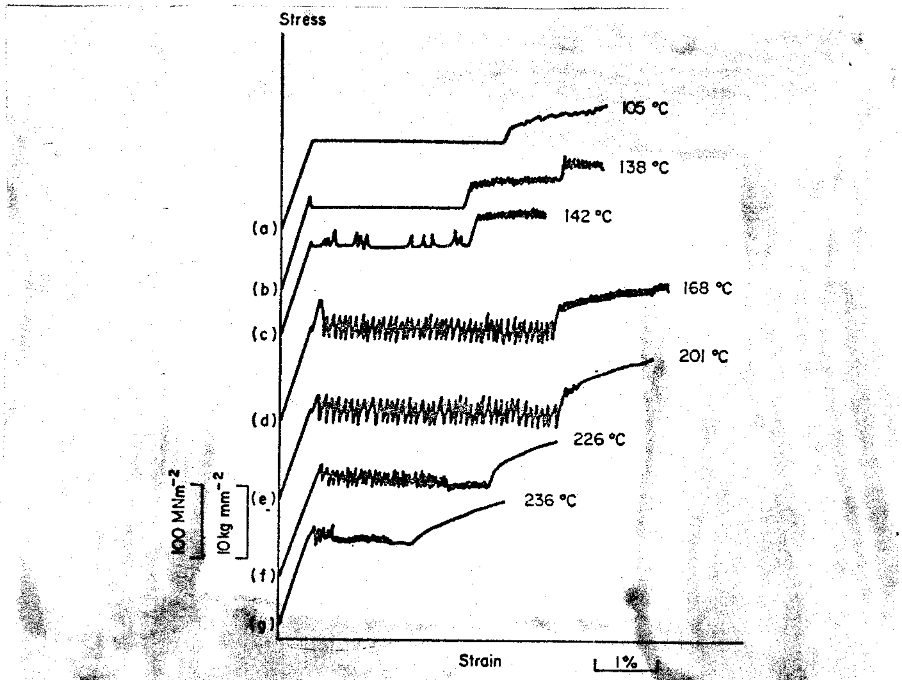
Figure 2



Typical stress-strain diagrams for three classes of materials. A material such as (b) is frequently classed as brittle, *since its ductility is so much less than that of material (c).*

Figure 3

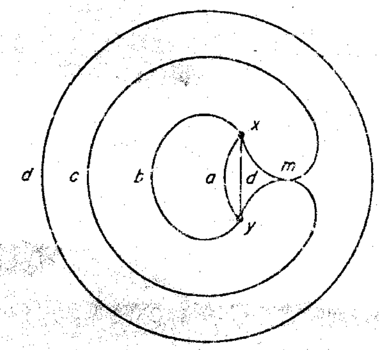
C. W. Richards, Engineering Materials Science, (Wadsworth Publishing Co., Belmont, Cal.), 1967.



Stress–strain curves for polycrystalline mild steel at elevated temperature (Blakemore and Hall, 1966)

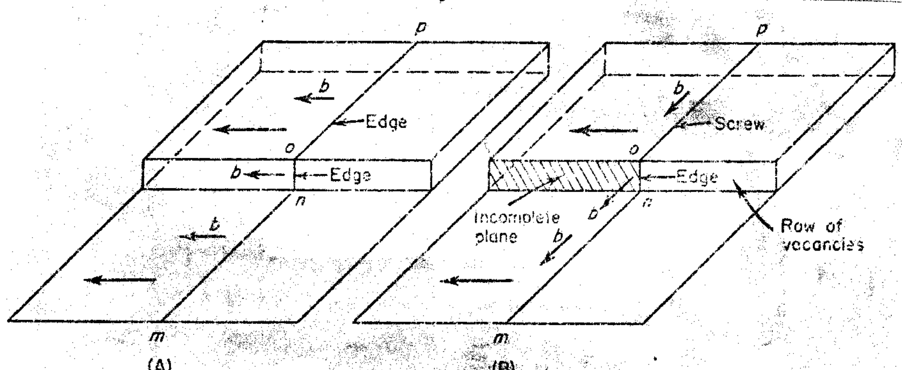
Figure 4

E. O. Hall, <u>Yield Point Phenomena in Metals and Alloys</u>, (Plenum Press, New York), 1970.



Various stages in the generation of a dislocation loop at a Frank-Read source

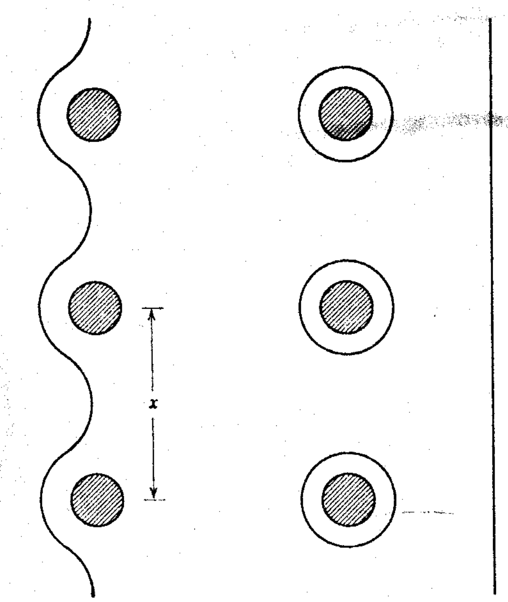
Figure 5



Dislocations with jogs normal to their slip planes

Figure 6

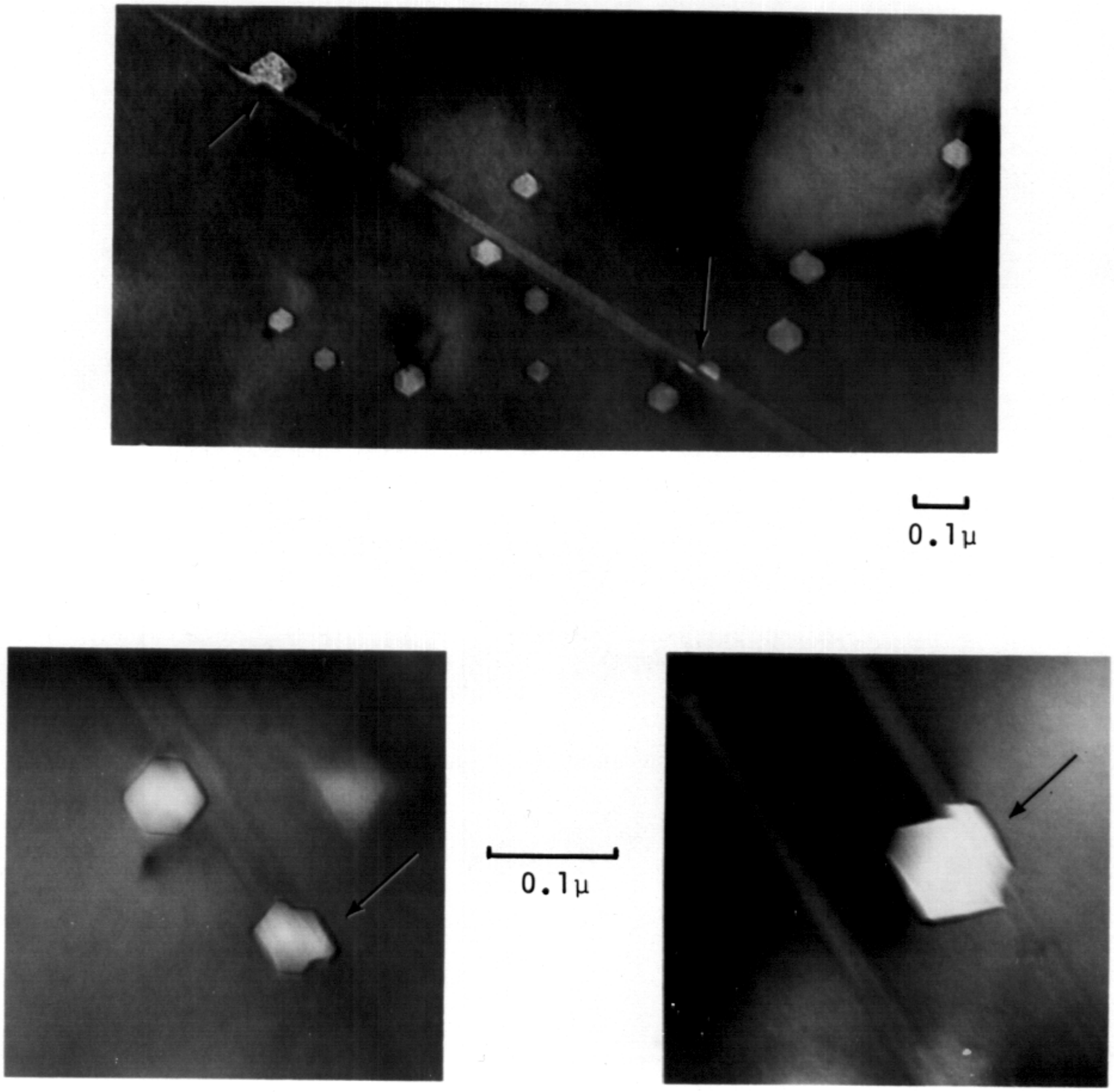
R. E. Reed-Hill, Physical Metallurgy Principles, (Van Nostrand Rheinhold Co., New York), 1964.



A dislocation bulges between particles (*a*). If the particles are widely enough spaced, the bulges spread around the particles and join up, leaving a ring of dislocation around each particle, and the dislocation moves on (*b*).

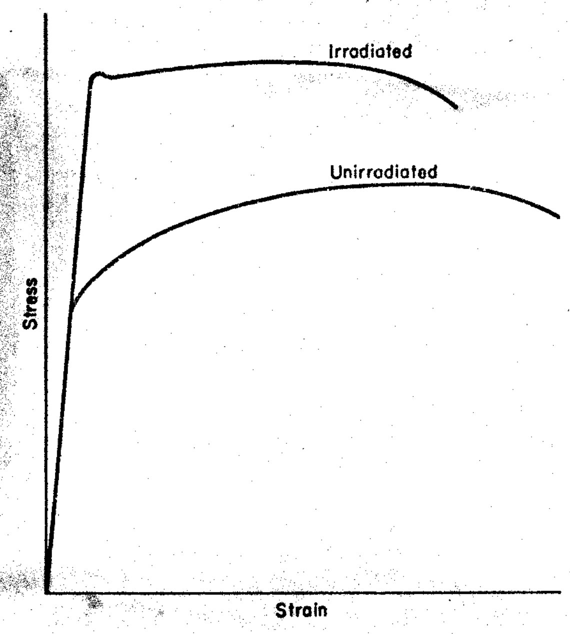
Figure 7

D. McLean, Mechanical Properties of Metals, (John Wiley and Sons, New York), 1962.



Examples of Voids in Irradiated Nickel Which Have Been Intersected By Slip Traces

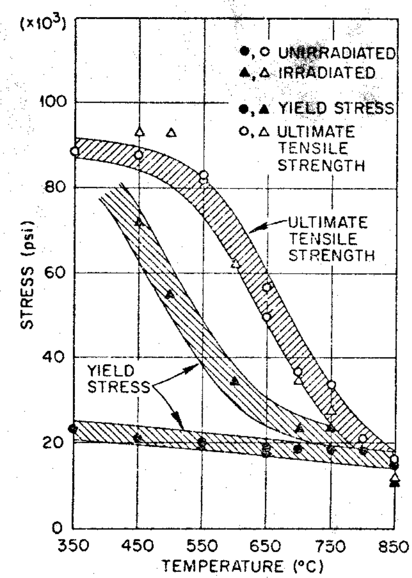
Figure 8



Representative engineering stress-strain curves for a metal before and after irradiation.

Figure 9

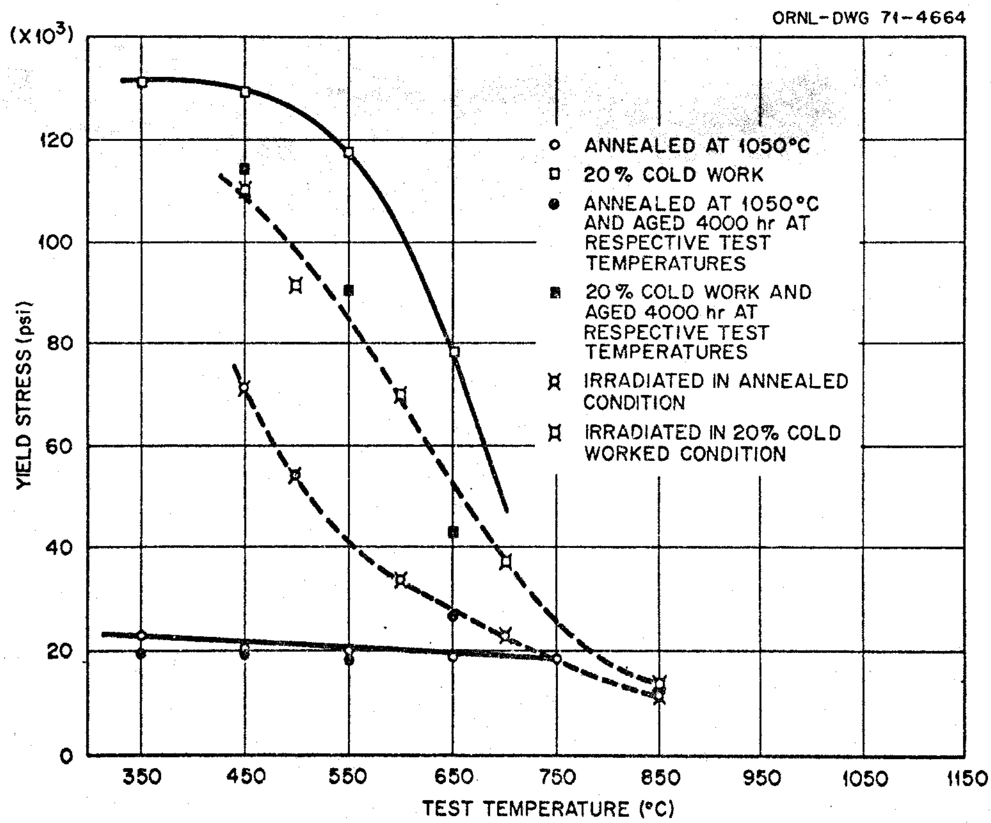
S. H. Bush, Irradiation Effects in Cladding and Structural Materials, (Rowman and Littlefield Inc., New York), 1965.



Effect of Irradiation to Fluences in the Range 1.5-2.7 X 10^{22} n/cm^2 (>0.1 MeV) on the Tensile Properties of Annealed 316 SS. Specimens were Irradiated and Tested at Approximately the Same Temperature.

Figure 10

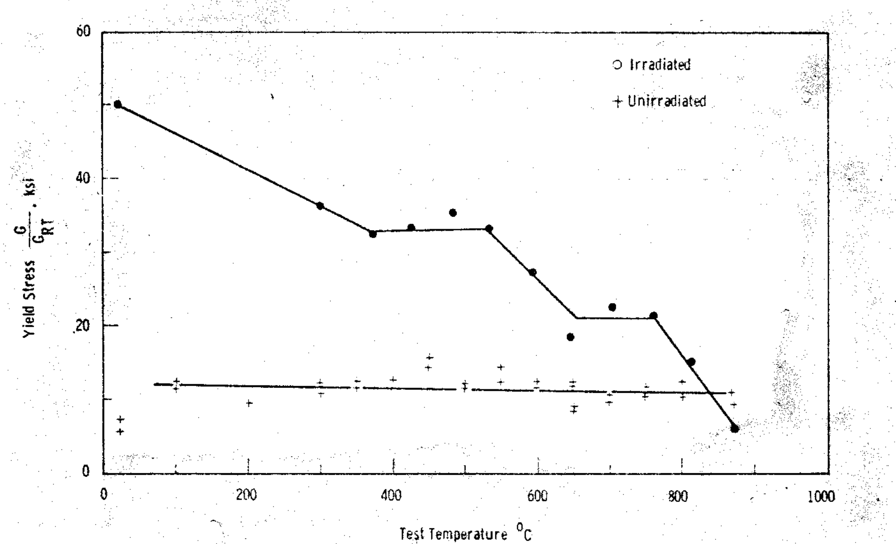
"Postirradiation Mechanical Properties of Types 316 and Titanium Modified 316 SS," <u>Quarterly Progress Report, Irradiation Effects on Reactor Structural Materials,</u> Aug.-Oct., 1971, HEDL-TME 71-161, p. ORNL-9.



Effect of Irradiation and Thermal Aging on the Yield Strength of Annealed and 20% Cold Worked Type 316 Stainless Steel. Specimens were irradiated at temperatures near the test temperature to fluences in the range 1.2 to 2.3 x 10^{22} neutrons/cm^2 (>0.1 MeV)

Figure 11

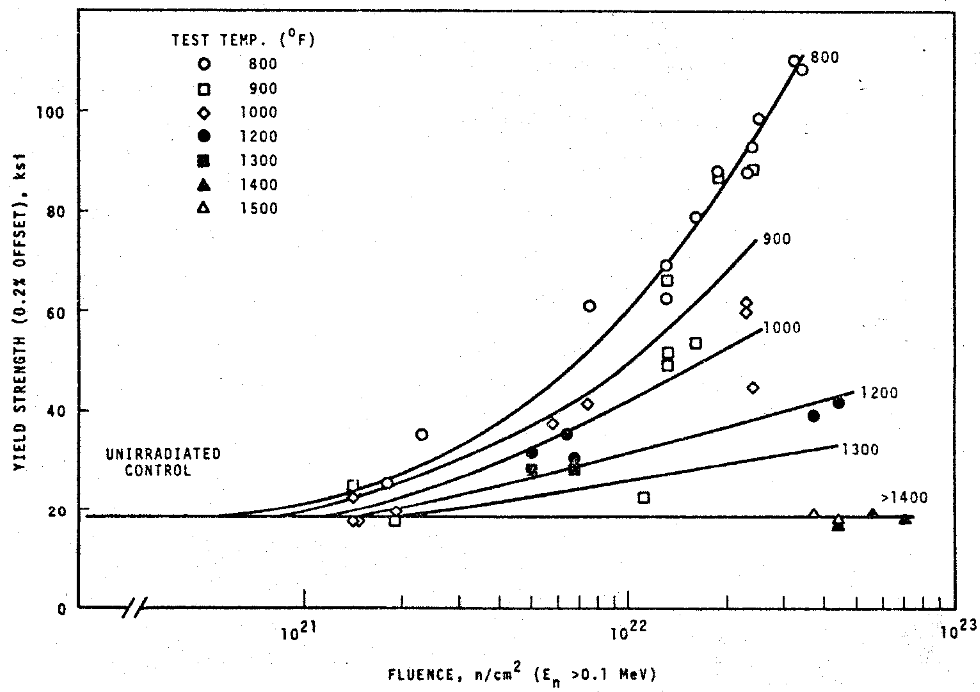
E. E. Bloom and D. Fahr, "Mechanical Properties of Standard and Titanium-Modified Types 316 Stainless Steel," Quarterly Progress Report, Irradiation Effects on Reactor Structural Materials, Feb.-March, 1971, HEDL-TME 71-66, p. ORNL-11.



Yield strength (proportional elastic limit) of AISI 304 stainless steel after irradiation in the EBR-II to $1{\cdot}7 \times 10^{22}$ n/cm² at $0{\cdot}49\ T_m$.

Figure 12

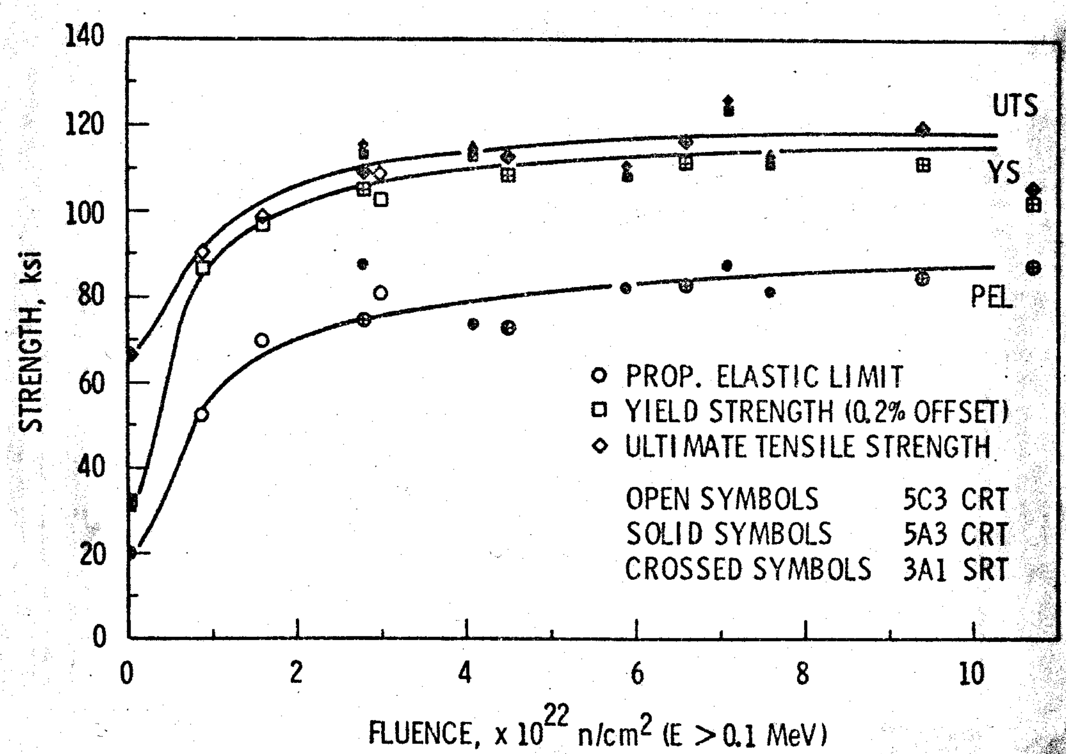
J. J. Holmes, et. al., "Elevated Temperature Irradiation Hardening in Austenitic Stainless Steel," Acta. Met. 16: 955, (July, 1968).



0.2% Offset Yield Strength in Annealed 316 SS After EBR-II Irradiation. Test and Irradiation Temperatures are nearly Identical in the Above Presentation.

Figure 13

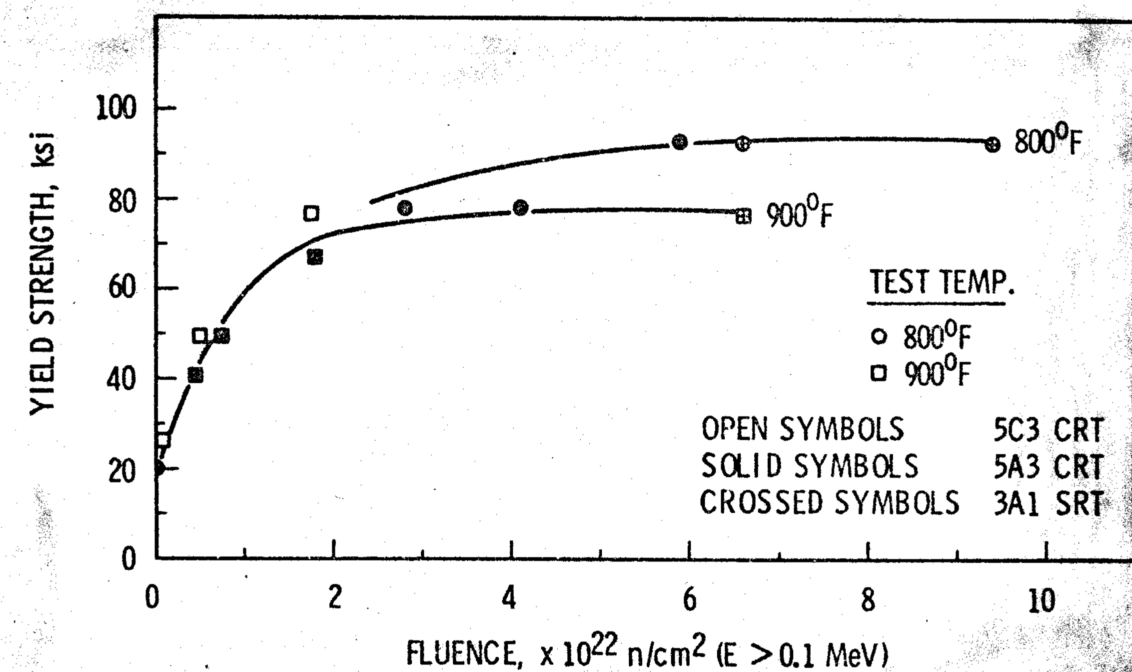
R. L. Fish, C. W. Hunter, and J. J. Holmes, "Tensile Properties of 316 and 304 Stainless Steel After Irradiation," Quarterly Progress Report, Irradiation Effects on Reactor Structural Materials, May-July, 1971, HEDL-TME 71-116, p. HEDL-85.



Effect of Fluence on the 700°F Strength of EBR-II Type 304 SS Thimbles.

Figure 14

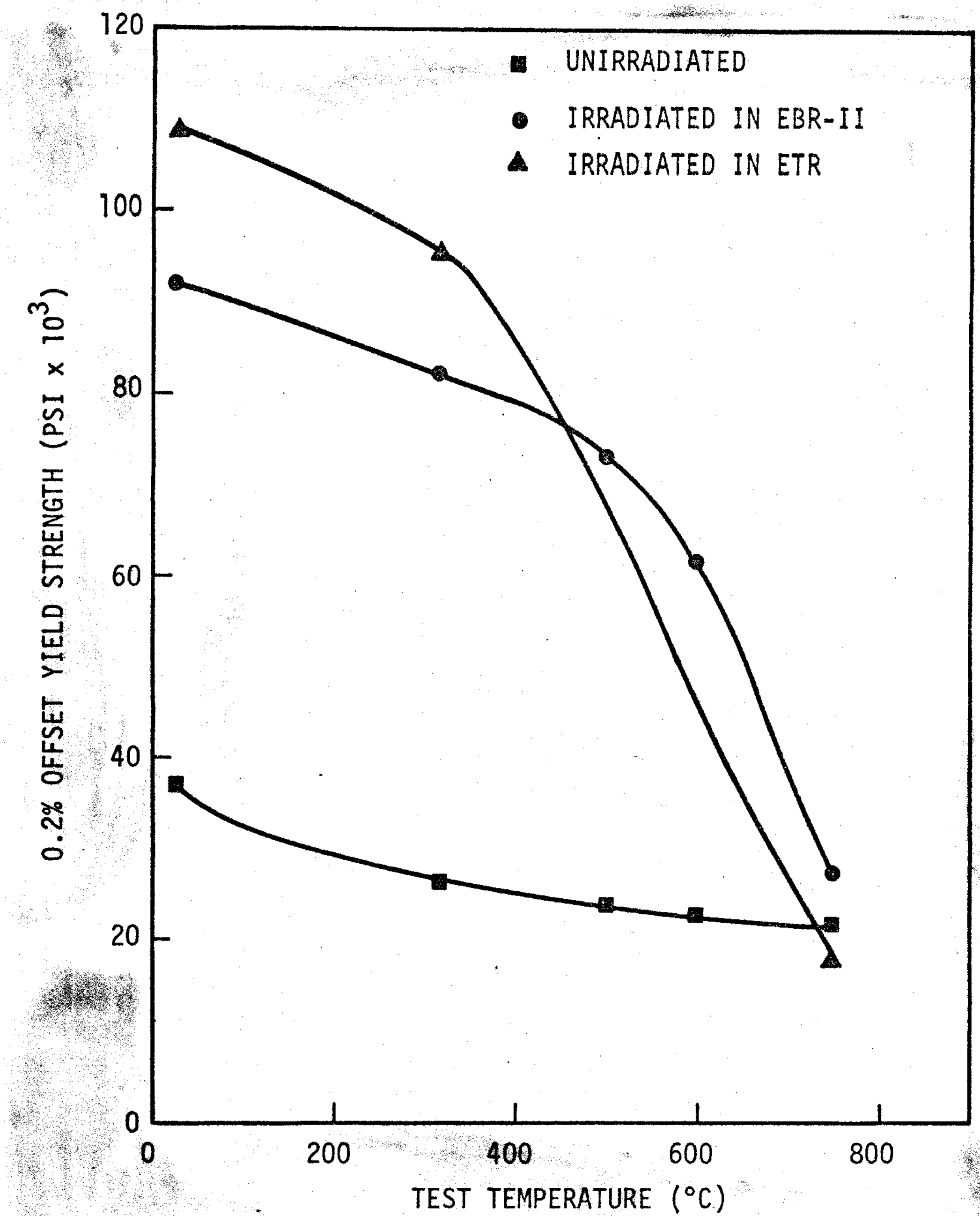
R. L. Fish, C. W. Hunter, and J. J. Holmes, "Tensile Properties at 700 - 900°F of High Fluence EBR-II Thimbles," Quarterly Progress Report, Irradiation Effects on Reactor Structural Materials, Feb.-March, 1972, HEDL-TME 72-64, p. HEDL-17.



Effect of Fluence on the 800 and 900°F Yield Strength of EBR-II 304 SS Thimbles.

Figure 15

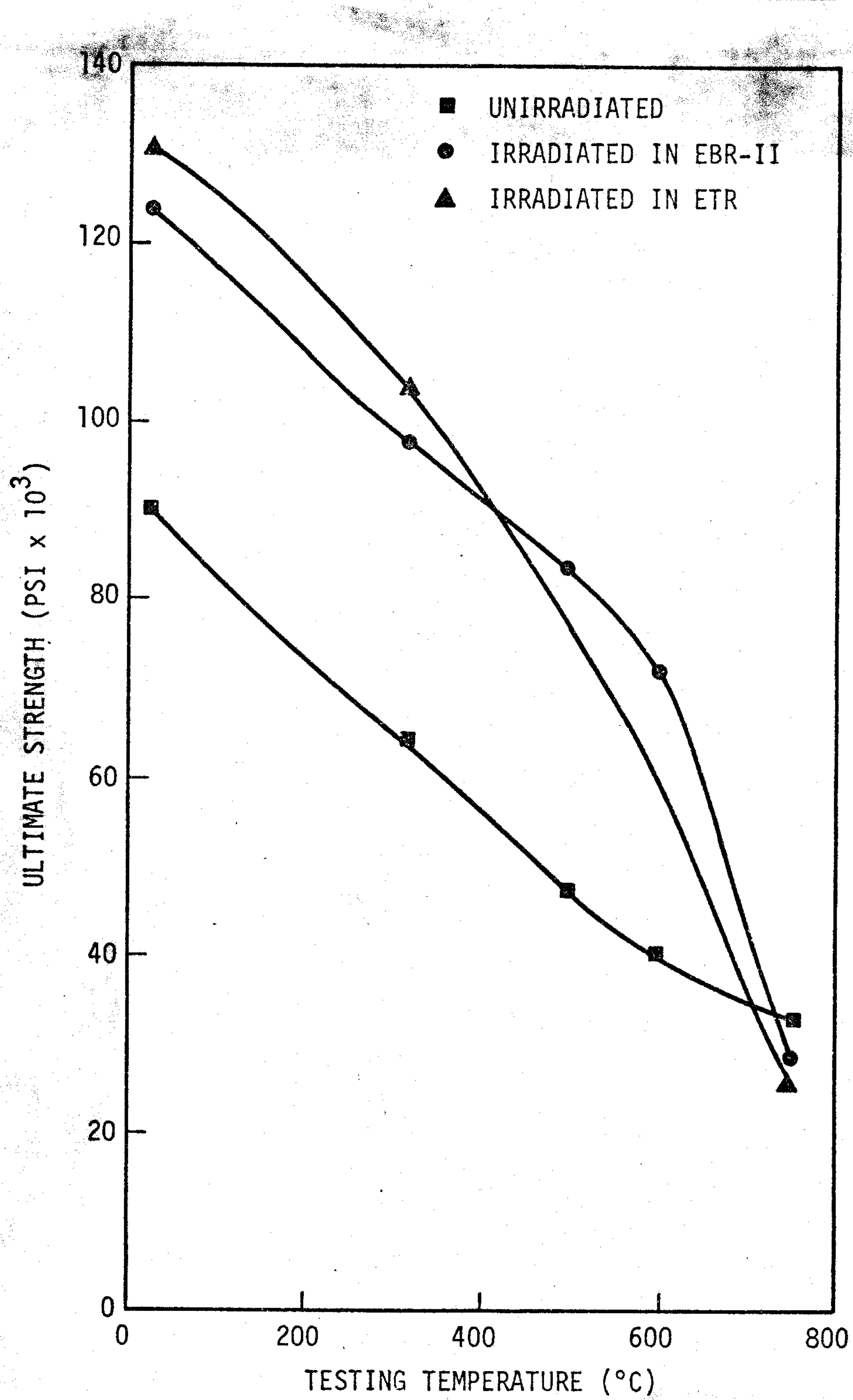
R. L. Fish, C. W. Hunter, and J. J. Holmes, "Tensile Properties at 700 - 900°F of High Fluence EBR-II Thimbles," Quarterly Progress Report, Irradiation Effects on Reactor Structural Materials, Feb.-March, 1972, HEDL-TME 72-64, p. HEDL-17.



Yield Strength of Type 347 Stainless Steel Irradiated at 400°C.

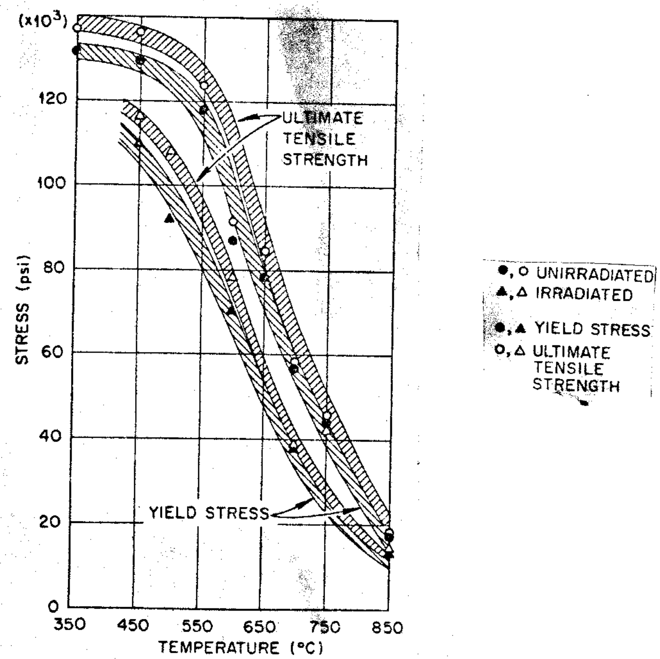
Figure 16

M. Kangilaski and A. A. Bauer, "Correlation of Radiation Effects on Mechanical Properties as Caused by a Predominantly Fast Flux and a Mixed Fast and Thermal Flux," Quarterly Progress Report, Irradiation Effects on Reactor Structural Materials, Feb.-March, 1971, HEDL-TME 71-66, p. BMI-1.



Ultimate Strength of Type 347 Stainless Steel Irradiated at 400°C.

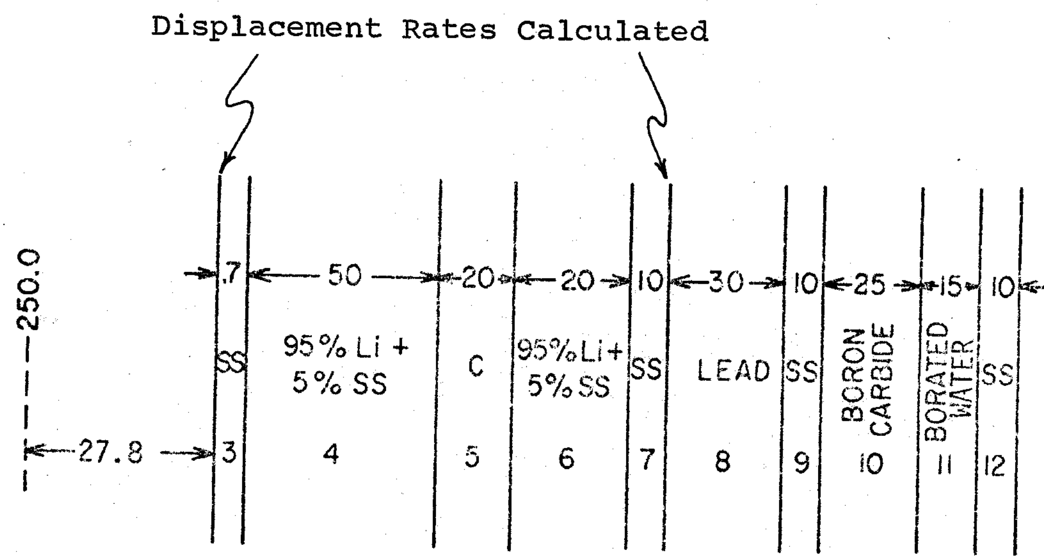
M. Kangilaski and A. A. Bauer, "Correlation of Radiation Effects on Mechanical Properties as Caused by a Predominantly Fast Flux and a Mixed Fast and Thermal Flux," Quarterly Progress Report, Irradiation Effects on Reactor Structural Materials, Feb.-April, 1971, HEDL-TME 71-66, p. BMI-1.



Effect of Irradiation to Fluences in the Range 1.5-2.7 X 10^{22} n/cm^2 (>0.1 MeV) on the Tensile Properties of 20% CW 316 SS. Specimens were Irradiated and Tested at Approximately the Same Temperature.

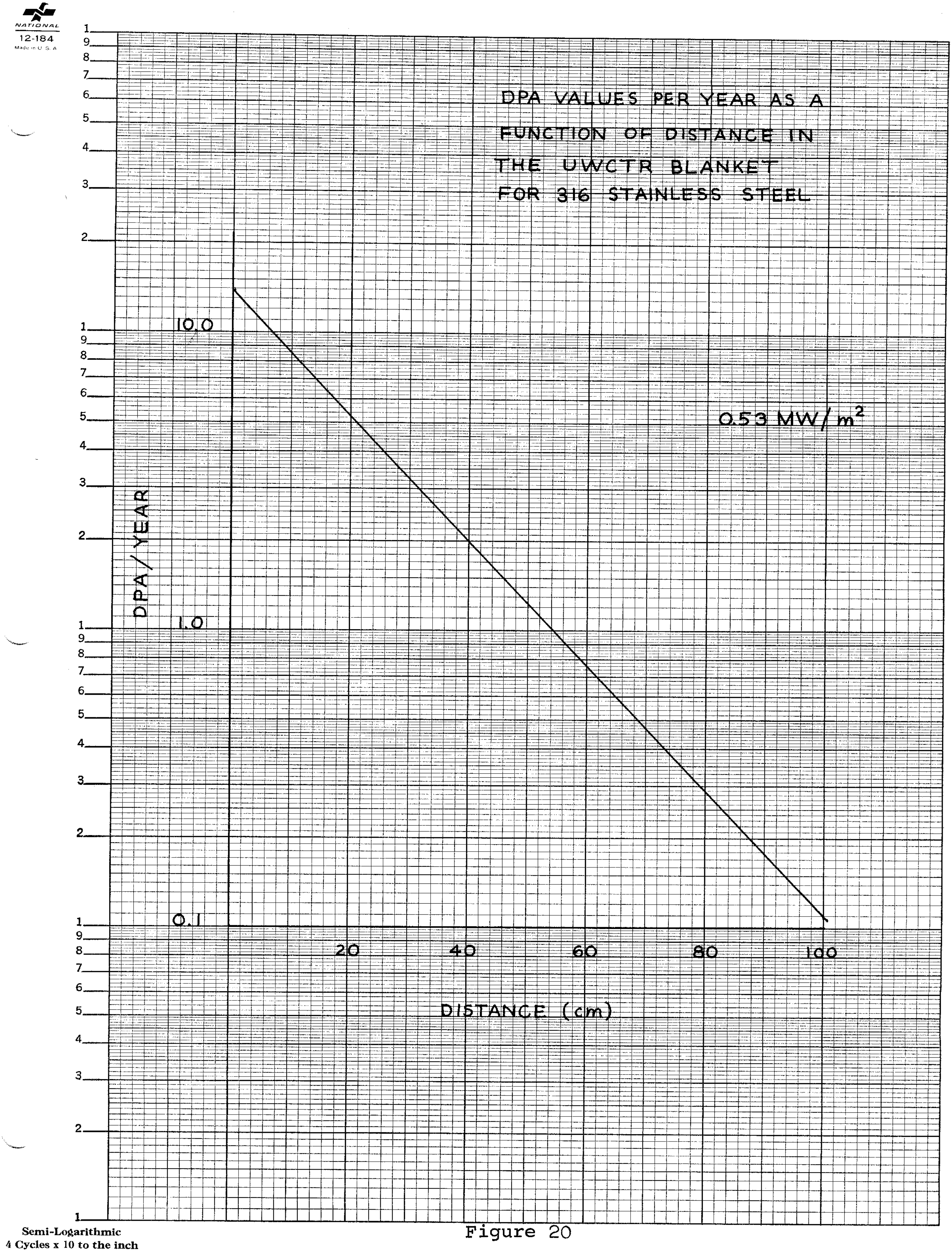
Figure 18

"Postirradiation Mechanical Properties of Types 316 and Titanium-Modified 316 SS," Quarterly Progress Report, Irradiation Effects on Reactor Structural Materials, Aug.-Oct., 1971, HEDL-TME 71-161, p.ORNL-9.



SCHEMATIC OF BLANKET AND SHIELD

Figure 19



DPA VALUES PER YEAR AS A
FUNCTION OF DISTANCE IN
THE UWCTR BLANKET
FOR 316 STAINLESS STEEL
0.53 MW/m²
DPA/YEAR
10.0
1.0
0.1
20
40
60
80
100
DISTANCE (cm)
NATIONAL
12-184
Semi-Logarithmic
4 Cycles x 10 to the inch

Figure 20

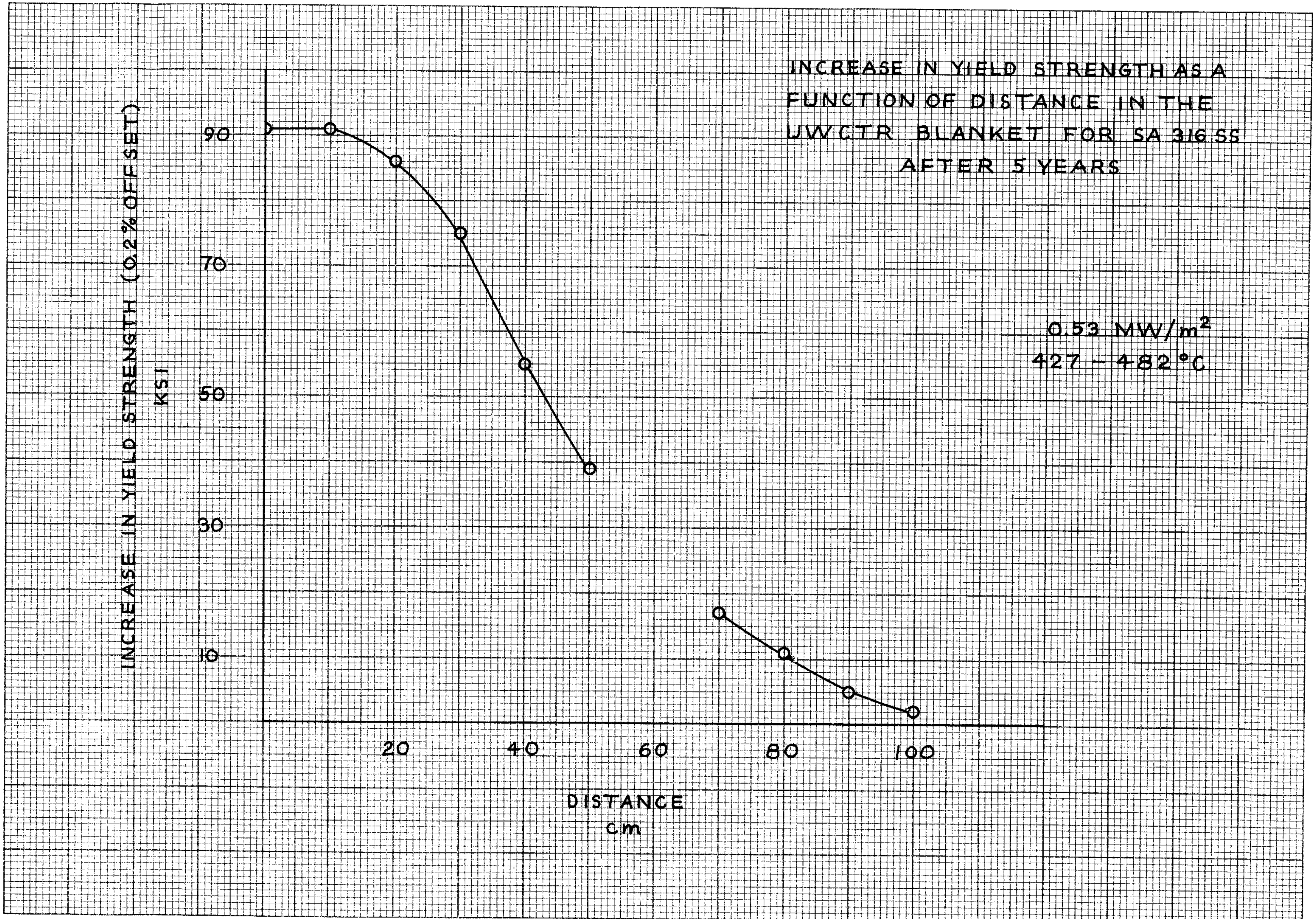

INCREASE IN YIELD STRENGTH AS A
FUNCTION OF DISTANCE IN THE
UWCTR BLANKET FOR SA 316 SS
AFTER 5 YEARS
0.53 MW/m²
427 – 482 °C
INCREASE IN YIELD STRENGTH (0.2% OFFSET)
KSI
90
70
50
30
10
20
40
60
80
100
DISTANCE
cm

Figure 21

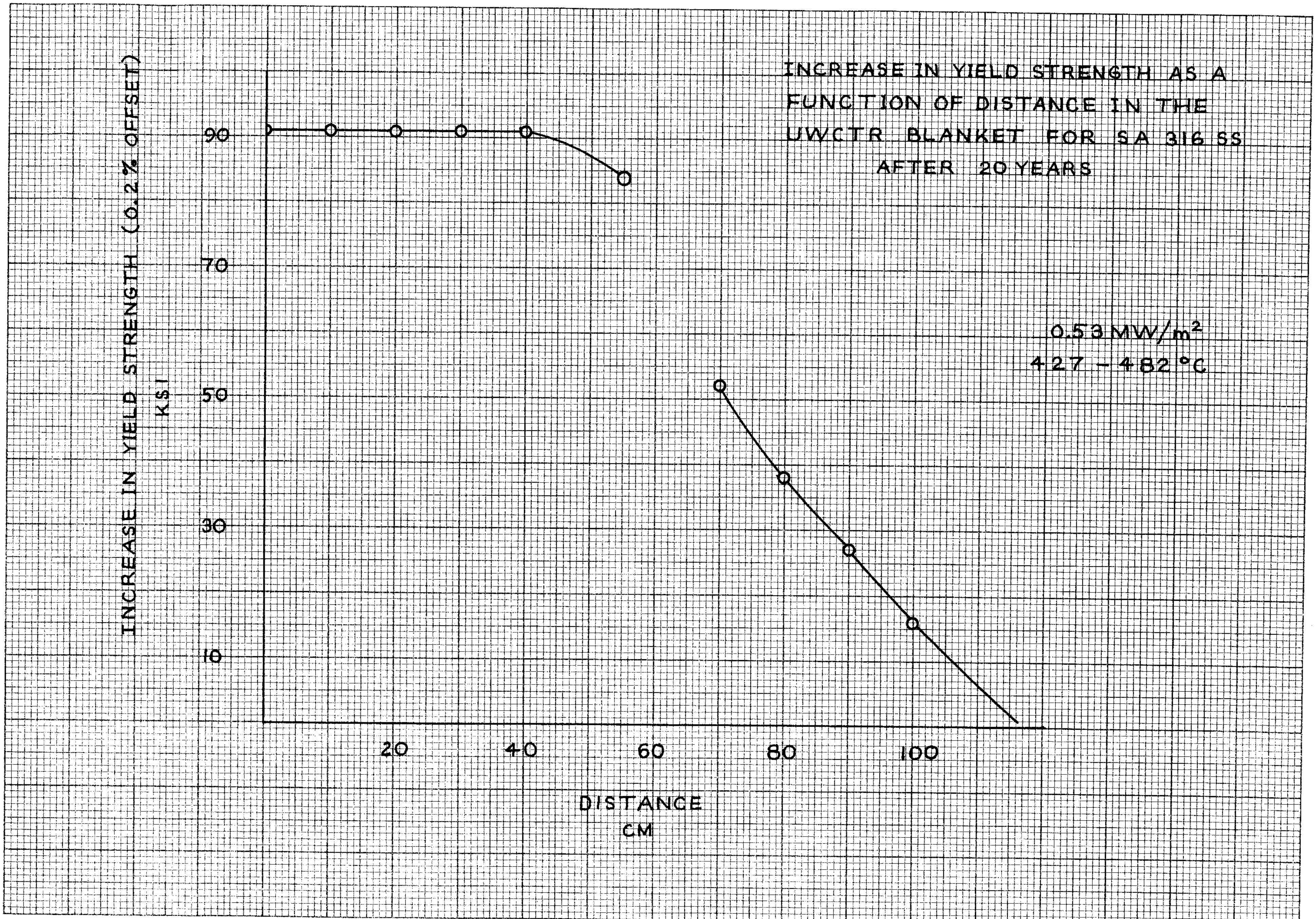

INCREASE IN YIELD STRENGTH AS A
FUNCTION OF DISTANCE IN THE
UWCTR BLANKET FOR SA 316 SS
AFTER 20 YEARS
0.53 MW/m²
427 - 482 °C
INCREASE IN YIELD STRENGTH (0.2% OFFSET)
KSI
90
70
50
30
10
20
40
60
80
100
DISTANCE
CM

Figure 22

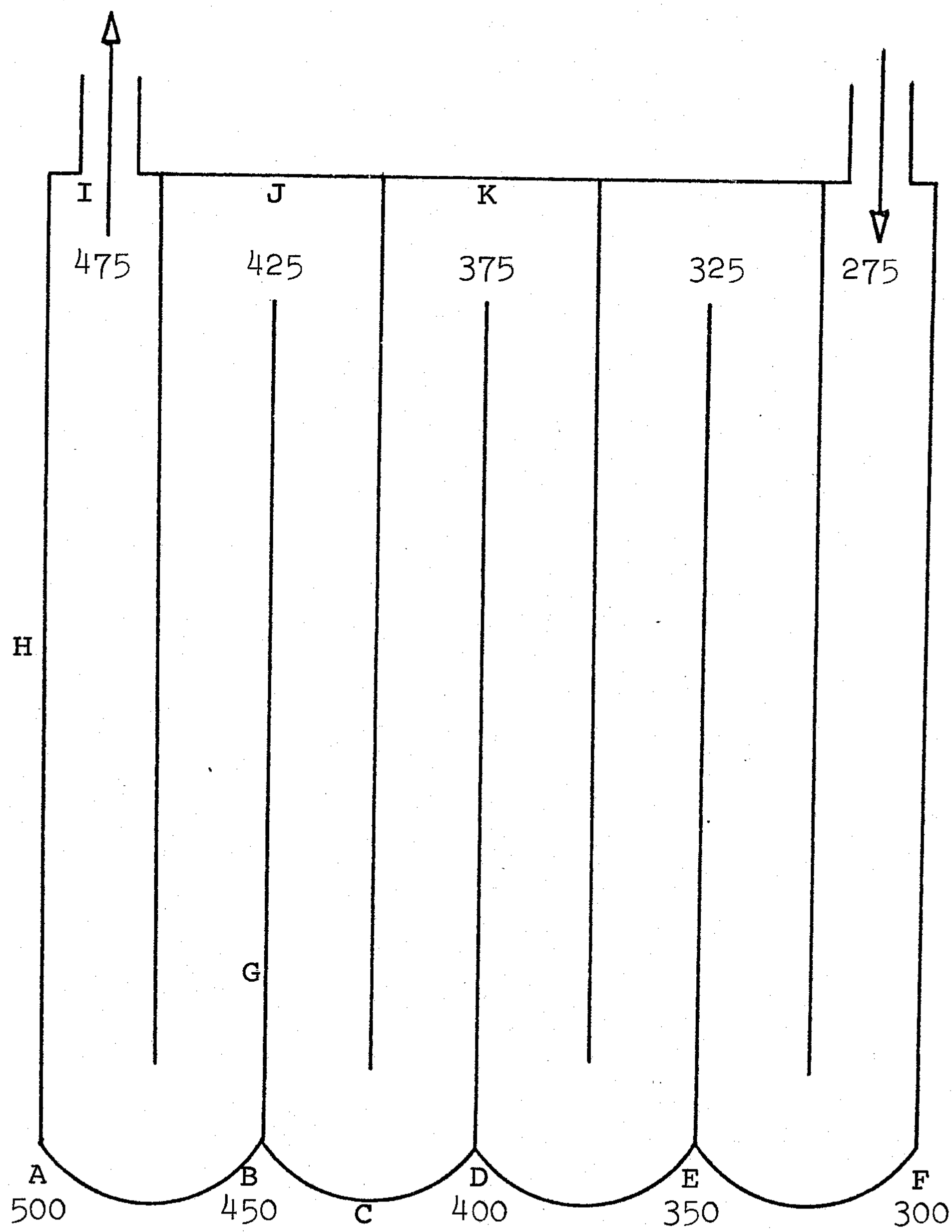
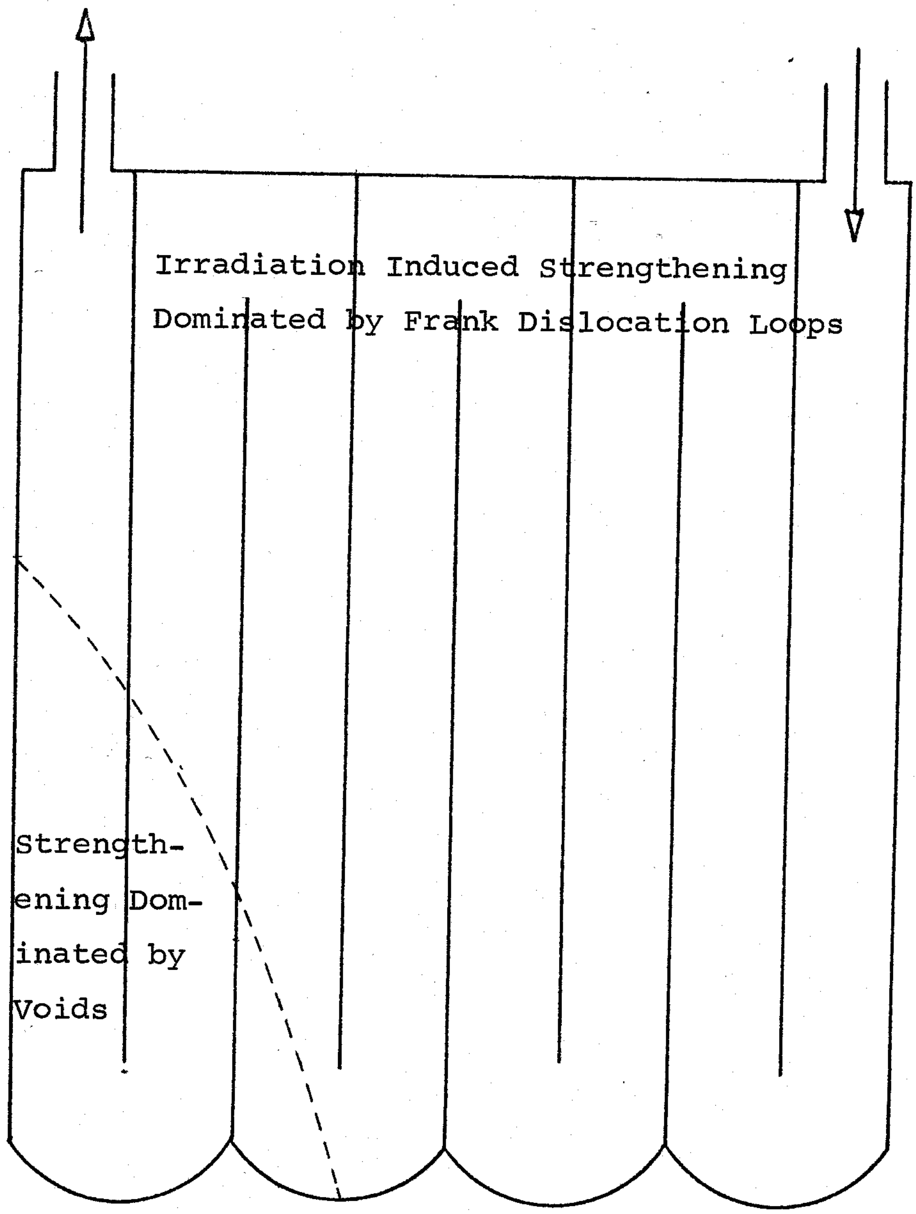


Figure 23

Heat Removal Cell of UWCTR. Temperatures indicated are in °C.



The Relative Effectiveness of Voids and Frank Loops on Irradiation Induced Strengthening After 5 Years Irradiation in the UWCTR Heat Removal Cell. (Solution treated 316 SS)

Figure 24