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INTRODUCTION

Voids in neutron irradiated stainless steels were first discovered by Cawthorne and Fulton in 1967 while examining samples of fuel pin cladding from Dounreay Fast Reactor.[1] Because the cladding of the liquid-metal fast breeder reactor (LMFBR) will be operating in the temperature range where swelling might be of a significant magnitude, this phenomena could have a major impact on LMFBR design. For this reason experimental and theoretical programs, largely in the United States and United Kingdom, were initiated to study the void problem. There have also been several important conferences that have dealt with the void problem: IAEA Symposium in Vienna (1969),[2] ASTM Conference at Niagara Falls (1970),[3] BNES Conference in Reading (1971),[4] "Radiation-Induced Voids in Metals" at Albany (1971),[5] and ASTM Conference in Los Angeles (1972).[6]

The mechanism of void nucleation is still uncertain. There have been a few theoretical attempts at describing homogeneous nucleation.[7-10] Other nucleation mechanisms have been proposed, based on experimental observations, but no attempt has been made to treat these theoretically.[11-25] In contrast, the phenomona of void growth is understood much more clearly.[13-28,51,55] The effects of parameters such as temperature,[14,15,18-21,30-32,35,38,48] fluence,[14,15,18-23,30,31,34,35] flux,[15,30,32,51] microstructure,[11,14,20-23,32,34-37,41,42,46,48] and composition[11,12,43-45,48] are being studied in more detail as data at higher fluences is being obtained.

The above topics, as they relate to the swelling behavior of austenitic stainless steels and in particular 316 stainless steel, are discussed briefly in this paper. It is also apparent that swelling may become a serious problem, in controlled thermonuclear reactors. The present design proposed by the University of Wisconsin fusion reactor design group (hereafter referred to as the UWCTR design) incorporates 316 stainless steel as the first wall and main structural material in the blanket. A discussion of the swelling problems that might be encountered in the UWCTR design is also included in this report.

NUCLEATION OF VOIDS

The mechanism by which voids are formed in a metal is not known. Several theories have been proposed. These include homogeneous nucleation, heterogeneous nucleation on existing microstructural sites or second phase particles, nucleation on gas bubbles, and nucleation in displacement spikes that may or may not be stabilized by gas atoms. Several authors have suggested that probably no one single mechanism is adequate to describe void nucleation over the entire range of temperature, fluence, and microstructural conditions possible.[14,16,22] The theories proposed do agree that there must be a supersaturation of vacancies in the structure in order for void formation to occur. Katz and Wiedersich have calculated that a reduction in the vacancy supersaturation level by a factor of two can reduce the nucleation rate of voids by several orders of magnitude.[7] This is illustrated in Figure 1 where S represents the vacancy supersaturation level.

Irradiation of a metal results in the production of excess vacancies and interstitials in the matrix. The concentration of these defects is determined by the difference between the generation rate and the loss mechanisms. The generation rate is proportional to the number of defects that escape a displacement spike without recombination. This number is temperature dependent. Harkness and Li have estimated that for temperatures above 400°C the number of vacancies and interstitials that escape a spike is roughly constant.[15] For use in their model, used to analyze EBR-II irradiation data, it is assumed that 30 free vacancies and 30 free interstitials escape from each displacement spike. Based on data for 304 stainless steel, they predict that below 371°C, very few vacancies will escape the displacement spike due to the low vacancy mobility. In order to correctly predict the loss rate, the migration rate to all point defect sinks must be examined. These sinks include dislocations, loops, voids, grain boundaries, and precipitates. Each of these mechanisms has been treated by Harkness and Li.[15]

If there is no preferential attraction of a given point defect to a certain type of sink, there will be no void growth. Bullough and Perrin have stated that there is a substantial preferential attraction between dislocations and free interstitials in the lattice, while the attraction of vacancies to dislocations is small.[26] Harkness and Li,[15] and Brager et. al.[32] explain the higher sink efficiency of dislocation loops for interstitials as an attraction between the stress field of a dislocation and the large "misfit" strain of the interstitial, which is greater than the lattice strain of the vacancy. When the higher diffusivity of the interstitials is considered, the result is a much larger interaction between interstitials and dislocations than with vacancies.[11] Thus a point defect imbalance is created. The excess vacancy concentration can initiate void nucleation and add to the growth of existing voids.

The effect of the increase in vacancy flux to the voids relative to the interstitial flux to the voids can be seen in Figure 1 computed by Katz and Wiedersich.[7] The figure shows the void nucleation rate as a function of the ratio of the arrival rate of interstitials (β_i) to the arrival rate of vacancies (β_v) to a given void. If there were no preferential attraction of interstitials to dislocations, then $\beta_i=\beta_v$, and the nucleation rate would be zero as shown. However, due to the interaction mentioned above, there is a net increase in the flux of vacancies to a void embryo relative to the interstitial flux, thus nucleation occurs.

The interstitials add to the existing dislocation structure of which the most prominent feature is the Frank interstitial loop. In performing an overall evaluation of nucleation and growth, it is important that the dislocation loop generation rate be known because of their preferential attraction for interstitials.[15,23] The fluences and temperatures over which Frank loops are found in 316 stainless steel is mapped in Figure 2. Harkness has concluded that the loop generation rate is a controlling factor in both the magnitude and temperature dependence of void formation.[15,23]

The initial work by Harkness and Li[15] indicates the lack of understanding with regard to the nucleation of voids. Uncertain of how to treat the several possible processes, they developed an empirical relation for the density of critical nuclei, ρ_c.

$$\rho_c = \kappa_1 \exp \left[- \frac{\Delta F_c}{RT} \right] \qquad (1)$$

ΔF_c is the free energy of formation of a critical nucleus. A critical nucleus is defined as the size of vacancy cluster such that the cluster has an equal probability of growing into a void or decomposing. Since this factor is unknown, an expression derived on the assumption of homogeneous nucleation was used. The use of the free energy of formation for a critical nucleus was meant only to reflect the importance of a vacancy supersaturation in determining the temperature dependence of the nucleation rate.[15] The constant κ_1 is independent of temperature and depends only on flux and fluence. It can be used to signify the role played by displacement spikes or helium atoms in nucleation. If the nucleation mechanism is purely homogeneous, κ_1 becomes the steady state vacancy concentration. For heterogeneous nucleation, the constant can represent the density of nucleation sites.

The rate at which the critical nuclei increase beyond their critical size, β_c, is given by

$$\beta_c = \beta_v - \beta_i \Big|_{r_c} \qquad (2)$$

where β_v and β_i are the vacancy and interstitial fluxes respectively evaluated at the edge of the critical nucleaus, r_c. The addition of a vacancy to the cluster will result in an increase in the void size above the critical value, while the addition of an interstitial will make the cluster subcritical.

The nucleation rate, J, as developed from classical nucleation theory is given by [8]

$$J = \rho_c \beta_c \; z \; \exp \; (- t/\tau) \qquad (3)$$

where z is a correction factor to account for the fact that some supercritical nuclei decompose. It is a function of temperature and the free energy of formation of the critically sized cluster. The exponential term accounts for transient effects with τ being the incubation time or a measure of the time delay to reach steady state. Knowing the quantitites ρ_c and β_c computed from equations (1) and (2), the steady state nucleation rate is given by

$$J = \rho_c \beta_c z \qquad (4)$$

HOMOGENEOUS NUCLEATION

Homogeneous nucleation results from the agglomeration of vacancies into small clusters. The first attempts to describe homogeneous nucleation were made by Claudson et.al.[19] and Harkness and Li[18] who applied classical nucleation theory to void formation. It was recognized that these earlier attempts did not include the presence of a supersaturation of interstitials in their treatment of nucleation.[7,8,10] Because of this omission, classical nucleation theory may not be applicable to void formation. In void nucleation there is not a conservation of defect matter (vacancies and interstitials). For example in classical nucleation theory, a precipitate can only grow by the addition of a molecule, and can only shrink by the emission of one. A void nucleus, however, has two possible growth mechanisms: the addition of a vacancy and the emission of an interstitial. It also has two possible shrinking mechanisms: the addition of an interstitial and the emission of a vacancy.[7]

Since the vacancy and interstitial concentrations are independent, they can not be in thermodynamic equilibrium simutaneously with any one void distribution. Katz and Wiedersich have determined the void size distribution based on the kinetics of the vacancy and interstitial fluxes to the void embryos.[7] The rate of nucleation is then derived along the lines of classical nucleation theory.

Russell[8] also noted the problems in the original work of Harkness and Li and as such derived expressions for the nucleation of voids in the presence of a supersaturation of interstitials as well as vacancies. He also developed an expression analogous to the classical nucleation rate equation presented earlier. The nucleation rate is expressed as the flux of vacancy clusters between adjacent size classes in a phase space of vacancy cluster size. The equation to be solved is

$$J_n = \beta_v(n)\,\rho(n) - \alpha_v(n+1)\rho(n+1) - \beta_i(n+1)\rho(n+1) \qquad (5)$$

J_n is the flux of clusters between any two sizes containing n and n+1 vacancies. $\rho(n)$ and $\rho(n+1)$ are the concentrations of vacancy clusters containing n and n+1 vacancies respectively. $\beta_v(n)$ is the rate of vacancy capture by a cluster of n vacancies. $\alpha_v(n+1)$ is the rate of vacancy emission by a cluster of n+1 vacancies. $\beta_i(n+1)$ is the rate of interstitial capture by a cluster of n+1 vacancies. From analyzing each term on the right side of the equation, it can easily be seen that the first term represents the growth of a vacancy cluster from n to n+1 vacancies by vacancy capture. The last two term are the shrinkage rates of embryoes containing n+1 vacancies, by vacancy emission and interstitial absorption respectively. The probability of interstitial loss by a vacancy cluster is considered negligible.

Equation (5) is then solved except that the vacancy cluster concentrations, ρ, are not equilibrium values due to the supersaturation of interstitials present.[8] By defining a new function $\rho'(n)$, the equation is solved and the solution is analagous to classical nucleation theory.

$$J_c = \rho'_c \beta_c Z' \qquad (6)$$

The values of Z' and ρ_c' are now determined kinetically and the presence of interstitials is included.[8]

As a vacancy cluster increases in size, the tendency is for it to collapse into a loop or redissolve in the matrix. However, after the nucleus surpasses the critical size, it will be more energetically favorable for the cluster to expand and hence become a void. This critical size was defined earlier as the radius at which the cluster has an equal probability of growing into a void or decomposing. Russell has developed the following expression for the critical radius[8]

$$r_c = \frac{2\gamma\Omega}{kT \ln\left[\dfrac{\beta_v(n) - \beta_i(n+1)}{\beta_v^e(n)}\right]} \qquad (7)$$

γ is the surface energy and Ω is the atomic volume. $\beta_v^e(n)$ is the rate at which thermal equilibrium vacancies impinge upon the surface of an n vacancy cluster. kT has the usual meaning. From equation (7) it can be seen that the critical radius is proportional to the surface energy and varies inversely with temperature. The importance of interstitials is evident in the argument of the logarithm term. The presence of interstitials results in a larger critical radius. As the interstitial flux approaches the vacancy flux to the cavity, the critical radius approaches infinity. This can be seen graphically in Figure 3. This figure and equation (7) emphasize again that if there is no preferential attraction of interstitials to dislocations, there will be no void nucleation. If the material does not contain a sufficient number of defect sinks to lower the interstitial concentration, it is expected that Frank loop formation and other interstitial precipitates will precede the formation of voids.[7]

A final conclusion reached by Katz and Wiedersich[7] and Russell[8] is that the delay time to reach steady state following transients (incubation time τ) is much longer when interstitials are considered. The incubation time also increases as the interstitial concentration increases.

As mentioned previously, any attempt to predict void nucleation and growth in an irradiated metal depends critically upon the loop nucleation rate. The models developed by Burton,[10] Russell,[8] and Katz and Wiedersich[7] to treat homogeneous void nucleation can also be applied to loop nucleation. Loh[40] has taken the Katz and Wiedersich model and adjusted the parameters to form equations describing the nucleation of interstitial loops.

The original Harkness and Li model discussed earlier has been modified and now calculates the homogeneous nucleation rate from equations similar to those developed by Russell, and Katz and Wiedersich.[23]

HETEROGENEOUS NUCLEATION

Several reports have indicated that heterogeneous nucleation on incoherent precipitates may be of major importance in determining the swelling behavior of solution annealed 316 stainless steel.[12,14,35,45] Normally when an interstitial encounters a precipitate it will adhere to the particle since the matrix-precipitate interfere represents a lower energy state for the interstitial.[12] This forms a compressive stress field which will tend to repel other interstitials and attract vacancies. As a result the overall point defect concentrations are lowered. In 316 samples irradiated at temperatures from 468°C to 625°C to fluences of 7.32 x 10^{22} n/cm^2 (E>0.1 MeV), Appleby and Wolff found that intragranular $M_{23}C_6$ and M_6C precipitates did not act in this manner.[12] Instead they served as nucleation sites for voids. Appleby and Wolff believe that this occurred because the spacing between precipitates was too large. Therefore, interstitials were able to agglomerate and form loops between the particles. This prevented the interstitials from reaching the precipitates and subsequently being annihilated by vacancies. As a result of this, a high vacancy supersaturation was formed. The precipitate particles became nucleation sites either by providing a lower free energy surface for void formation or by attracting dissolved gaseous impurities that would stabilize vacancy clusters.

Appleby and Wolff have also suggested that most of the precipitation nucleation occurs prior to void nucleation.[12] Thus the particles can serve as nucleation sites for voids. They also conclude that even though precipitates influence void formation, they do not appear to affect the overall vacancy concentration.

Brager and Straalsund have attempted to characterize the swelling of 316 stainless steel with respect to various combinations of temperature and fluence that may exist during irradiation.[14] This is presented in Figure 4. It can be noted that above ∿ 425°C nucleation appears to be heterogeneous. In region II of Figure 4 most of the voids are associated with rod-shaped precipitates (RSP) which appear to be homogeneously nucleated throughout the matrix. For temperature range from ∿425°C to ∿485°C at the low fluence levels (region III in Figure 4), voids are also observed on RSP, however here the RSP are not distributed uniformly throughout the matrix, suggesting that they too are heterogeneously nucleated. At higher fluences (> ∿ 3 x $10^{22}n/cm^2$, E >0.1MeV) in the temperature range from ~450°C to ~550°C, some voids are observed on $M_{23}C_6$ precipitates as well as RSP. Trends indicate that as the temperature approaches 500°C, more voids become associated with $M_{23}C_6$ particles at the higher fluences. At higher irradiation temperatures from ∿625°C to 740°C (region V in Figure 4), a low density of very large voids attached to RSP is observed. As a contrast, below 400°C void formation is characterized by a high density of small voids which are probably homogeneously nucleated. The authors do note that it is not possible to rule out heterogeneous nucleation at low temperatures. Due to the damaged state of the matrix, it may be possible that nucleation could occur at impurities or very small precipitates.

Brager and Straalsund,[14] Appleby and Wolff,[12] and Bloom and Stiegler[39] have all examined the electron diffraction patterns of RSP and have concluded that it is not any of the presently identifiable phases. Brager and Straalsund have predicted that the structure is tetragonal with a lattice parameter between that of FeCr/σ and FeCr Mo/σ.[14] From annealing studies, they also found the upper temperature at which voids are observed during irradiation (~700°C) is the same as the upper temperature limit for the formation of RSP.

It should be noted that coherent precipitates are good point defect sinks and as such no nucleation occurs on them. The low interfacial energy characteristic of coherent precipitates results in the point defects being trapped at the interface.[24] They then serve as recombination centers for defects of the opposite type. In other words, coherent precipitates have only a finite capacity for point defects, while defects arriving at incoherent precipitates essentially lose their identity giving incoherent precipitates an infinite capacity for point defects.[24] It has been suggested that a fine dispersion of coherent precipitates might be a means by which swelling is reduced.[12,32,52] Researchers in the United Kingdom have shown that the γ' precipitates, $Ni_3(Ti,Al)$, formed in Nimonic PE16 largely reduce void nucleation and growth by providing recombination centers, and by pinning dislocations so they become immobile.[24,46,52]

EFFECTS OF GAS ON NUCLEATION

Cawthorne and Fulton were the first to recognize that gas may be important in void formation.[1] Gas atoms are thought to be a means by which subcritical vacancy clusters are stabilized against collapse. Bullough and Perrin have stated that the presence of gas in vacancy clusters is essential to ensure a three dimensional morphology.[25,26] In the absence of gas atoms, the vacancy cluster will collapse into a two-dimensional platelet which would have a preferential attraction for interstitials and thus no void formation would result. Although certain that insoluble gases will enhance nucleation, Katz and Wiedersich feel that gases are probably not an absolute prerequisite for void formation.[7]

The two most common gases produced in irradiated stainless steel are hydrogen and helium. These are formed from (n,p) and (n,α) reactions, respectively, with various components in the steel. Hydrogen does not appear to be of importance in void formation because of its high mobility and its solubility in the matrix.[43] Helium atoms, however, are insoluble, and can influence nucleation by exerting an internal pressure which stabilizes a vacancy cluster against collapse. The atoms themselves may also serve as a nucleation site.[15]

In their early work, Harkness and Li postulated that void formation in materials where the void is not the defect with the lowest formation energy may become a question of whether there are enough gas atoms in the neighborhood of a cavity to prevent its collapse into a stacking fault tetrahedron.[18] Frank loops and stacking fault tetrahedra both have lower formation energies than voids in stainless steels. This may imply that a relatively high gas concentration is needed to stabilize the clusters of vacancies until they become greater than critical size. For this reason, Harkness and Li state that perhaps the reason fluence thresholds for void formation are observed is that it takes time for a significant amount of helium to accumulate.

Much of the work on the effects of helium on swelling has been performed with ion bombardment experiments. Nelson and Mazey have shown helium to be a very effective nucleating agent and increase swelling substantially.[53] Their work showed that the presence of helium can increase the void volume by an order of magnitude. A recent report by McDonald and Taylor[47] on 316 stainless steel preinjected with helium and irradiated with 4-MeV Ni^+ ions showed no effect due to helium. However the high displacement rates (5×10^{-3}dpa/sec) may have resulted in homogeneous nucleation. Samples preinjected with helium have also been irradiated in EBR-II in order to eliminate some of the problems inherent in ion bombardment experiments, such as surface effects, differences in recoil spectra, positional dependence of damage, and the observed shift in the temperature dependence of swelling with the increased displacement rate.[54] These experiments showed that helium did influence swelling but not to the extent that was reported in Nelson and Mazey's ion bombardments. In similar irradiations in EBR-II, Harkness _et.al._[56] have found that helium may play a more important role in void nucleation than previously thought. Their results show that helium can apparently increase dislocation loop nucleation significantly. They also found evidence that substantiates the possibility of small helium bubbles as nucleation sites.

Nelson and Mazey feel that if more insight into the effects of helium on void formation are to be obtained, it must come from ion bombardment experiments, because it is impossible to isolate the individual effects of helium and radiation damage during a reactor irradiation.[53]

In the SCIM code developed by Harkness and Li, an empirical equation has been added to incorporate the effect of gas atoms on nucleation.[23] According to Norris,[20] no adequate theoretical treatment of nucleation with gas atoms has yet been presented. However, Russell has recently modified his nucleation theory in an attempt to calculate the coprecipitation rate of vacancies, interstitials, and gas atoms.[9] As mentioned earlier, he treats the void embryoes as moving in phase space. In this case the vacancy clusters are characterized by both the number of vacancies, n, and the number of gas atoms, x, contained in the embryo. This extends the phase space to two dimensions. Thus there are now flux equations to consider.

$$J_n = \beta_v(n,x)\rho(n,x)-\alpha_v(n+1,x)\rho(n+1,x)-\beta_i(n+1,x)\rho(n+1,x)$$
$$+\Theta\beta_c(n,x)\rho(n,x)-\Theta\alpha_c(n+\Theta,x+1)\rho(n+\Theta,x+1) \quad (8)$$

$$J_x = \beta_c(n,x)\rho(n,x)-\alpha_c(n+\Theta,x+1)\rho(n+\Theta,x+1) \quad (9)$$

The symbols are the same as defined previously except that the variable x now indicates the number of gas atoms present in the cluster. The new terms are: $\alpha_c(n+\Theta,x+1)$-the rate of vacancy-gas atom complex emission from an embryo of $n+\Theta$ vacancies and $x+1$ gas atoms, Θ - the number of vacancies associated with each gas atom, $\beta_c(n,x)$ - the rate of vacancy-gas atom complex impingement on an embryo of n vacancies and x gas atoms.

The treatment of these equations is more complex than in the case where interstitials and vacancies only are considered. In that case a "pseudo-equilibrium" distribution $\rho'(n)$ was defined in order to solve for J_n. When gas atoms are treated, if a similar function $\rho'(n,x)$ is postulated, there are very restrictive relationships between it and the equilibrium distribution of (n,x) - size clusters in the absence of interstitials.[9] If the gas atoms are substitutional in nature, as is generally thought, then it is impossible to find the function $\rho'(n,x)$ needed to compute J_n and J_x. However, if the gas atoms are treated as interstitials, a less probable situation, it may be possible to find such a function, by making some assumptions. Russell also presents a difference equation which in principle can be solved for the nucleation rate.[9]

$$\Delta_t \, \rho(n,x) + \Delta_n \, J_n + \Delta_x J_x = 0 \qquad (10)$$

This equation is obtained from converting equations (8) and (9) to time dependent equations by use of the difference operators Δ_t in time, Δ_n in number of vacancies, and Δ_x in gas atom content. Russell believes that the solution to this equation will result in a theoretical understanding of void formation with gas atoms in the matrix.

SPIKE NUCLEATION

Early work by Claudson et.al.[19] on the possible mechanisms of nucleation indicated that swelling behavior is best described assuming nucleation in a displacement spike. When a neutron interacts with the lattice, the atoms in the region of the interaction are displaced outwards forming a region rich in interstitials surrounding a volume of vacancies. These regions are unstable and some recombination occurs immediately after the event. If some of the interstitials are trapped at an impurity or a dislocation, a vacancy cluster may result. If the cluster is supercritical, it will probably grow into a void.[15,19,34] Subcritical size clusters will most likely shrink and redissolve into the matrix unless some driving force for growth, possibly helium atoms, is present.[15,19]

Bloom has stated that for nucleation to occur in displacement spikes, vacancy supersaturation would have to be high enough to stabilize such a cluster.[30] Thus nucleation rate would tend to be highest at low fluences. As the fluence level rises, the nucleation rate would decrease since the vacancy supersaturation decreases with increasing fluence (See Figure 5). This effect is not in agreement with experimental data.[30] Also, since the equilibrium vacancy concentrations are established in a relatively short time, the incubation period before swelling is observed that increases with temperature is difficult to rationalize.

Straalsund and Guthrie have stated that spike nucleation may be of major importance, especially in cladding irradiated at high temperatures and subjected to a hoop stress.[57] They postulated that spike nucleation may explain the discrepancies noted by Katz and Wiedersich[7]; that homogeneous nucleation could not account for all the nucleation observed experimentally at high irradiation temperatures.

Finally, Norris recently stated that spike nucleation is probably not of dominating importance since voids have been formed in metals by electron radiation where there were no pre-existing vacancy clusters.[20]

VOID GROWTH

The theory describing void growth is for the most part better understood than nucleation theory. A number of theoretical treatments of void growth have been presented.[18,19,22,24-28,51,55] A summary of the model developed by Bullough and Perrin is presented below.[25-27]

In order to describe growth, Bullough and Perrin began with a matrix containing a uniform density of spherical voids ρ_v each having a radius r_v which varies as a function of time during irradiation. Each void is associated with its own spherical cell of radius R. As r_v increases with dose, R also becomes larger. The coupled, steady-state diffusion equations which determine the fractional vacancy and interstitial concentrations, $C_v(r)$ and $C_i(r)$ respectively, are

$$D_v\left(\frac{d^2C_v}{dr^2} + \frac{2}{r}\frac{dC_v}{dr}\right) + K - \alpha_R C_v C_i - D_v Z_v \rho_d C_v = 0 \qquad (11)$$

$$D_i\left(\frac{d^2C_i}{dr^2} + \frac{2}{r}\frac{dC_i}{dr}\right) + K - \alpha_R C_v C_i - D_i Z_i \rho_d C_i = 0 \qquad (12)$$

for $r_v < r < R$.

D_i and D_v are the interstitial and vacancy diffusion coefficients. α_R is the recombination coefficient, K is the production rate of interstitials and vacancies. ρ_d is the dislocation density. Z_v and Z_i represent the vacancy and interstitial capture volumes per unit length of dislocation line. Due to the larger interaction between interstitials and dislocations than with vacancies and dislocations, $Z_i > Z_v$.

Since the outer boundary, r=R, represents the interface midway between each void, zero flux boundary conditions are imposed.

$$\frac{dC_v}{dr} = 0 \text{ at } r=R \qquad (13)$$

$$\frac{dC_i}{dr} = 0 \text{ at } r=R \qquad (14)$$

Two other boundary conditions are obtained by demanding continuity of the interstitial and vacancy flux at the void surface. For vacancies, it is required that "the diffusion controlled net flux of vacancies entering the void must be equal to the kinetically defined flux of vacancies" to the void.[26] A similar equation defines the interstitial flux continuity at the void surface.

$$D_v \frac{dC_v}{dr} = K_v [C_v - C_v^e \exp \{F_m b^3/kT\}] \qquad (15)$$

$$D_i \frac{dC_i}{dr} = K_i C_i \qquad (16)$$

$$\text{at } r=r_v$$

K_v and K_i are the velocities of transfer of vacancies and interstitials, respectively, across the void-matrix interface.

C_v^e is the thermal equilibrium concentration of vacancies. F_m is the mechanical force tending to shrink the void. b is "a distance on the order of the atomic lattice spacing."[26] Note that the probability of interstitial emission from the cluster is neglected since it is very small.

The rate of change of void radius is given by subtracting equation (16) from equation (15).

$$\frac{dr_v}{dt} = k_v C_v(r_v) - k_i C_i(r_v) - k_v C_v^e \exp [F_m b^3/kT] \quad (17)$$

If it is assumed that the activation energy for the defect jump across the void-matrix interface is the same as in the undisturbed matrix, then Bullough has shown that

$$k_v = \frac{D_v}{b} \text{ and } k_i = \frac{D_i}{b} . \quad (18)$$

Using the above, equation (17) takes the form

$$\frac{dr_v}{dt} = \frac{1}{b} [D_v C_v(r_v) - D_i C_i(r_v) - D_v C_v^e \exp (F_m b^3/kT)] \quad (19)$$

The procedure is then to solve the first pair of equations, (11) and (12), for C_i and C_v based on some initial value of r_v and R subject to the boundary conditions. The above equation is then used to find the increase in r_v and the cycle is repeated with new values for r_v, R, and F_m. Both R and F_m depend on r_v and change with increasing void size. An analytic solution to the problem is possible providing the nonlinear recombination terms in equation (11) and (12) are dropped. Bullough and Perrin indicate that annihilation will not be of major importance at high temperatures where swelling is most severe. Straaslund and Guthrie have also found that in general recombination efforts are significant only at low temperatures and low dislocation densities.[49]

Bullough and Perrin have also analytically described other factors influencing void growth such as stress, precipitates, dislocation depletion, and void-interstitial interactions.[25] They found that void growth kinetics are very sensitive to dislocation density. This is illustrated in Figure 6. At low dislocation densities the swelling is approximately linear with dose. At higher dislocation densities, the swelling rises rapidly after an intitally reduced swelling rate. This could have some bearing on the effectiveness of cold work in minimizing swelling, and will be considered in a later section. Their model also predicts that the kinetics of void growth are independent of dose rate.

One factor that complicated the study of void growth is whether or not that a void acts as a perfect sink, i.e. it absorbs vacancies fast enough so that the point defect concentration remains in equilibrium in the vicinity of the sink.[17] If a void acts as a perfect sink, vacancy absorption is easy. Otherwise, the growth rate may become limited by the kinetic processes at the void surface. Balluffi and Seidman have concluded that vacancy absorption by the void will occur at "ledges" on the surface of the void.[17] In this case, voids would tend to become polyhedra with faceted faces. If a void is to grow, it must continue to nucleate ledges on its surface. These ledges are just platelets of absorbed vacancies. According to Balluffi and Seidman, if a void is to act as a perfect sink, ledge nucleation on its surfaces must be relatively rapid and a resonable number of surface ledges must be maintained on the void faces.[17] If this occurs, the void will remain in essentially a quasi-steady state with the surrounding matrix.

Wiedersich has formulated another method to describe void growth.[13]

It is based on the use of rate equations to calcualte the spatially averaged point defect concentrations. By incorporating the effects of various sinks, the equations are solved to find the net accumulation of vacancies at voids. The advantage of the rate theory approach is that the equations are much simpler to handle than are the diffusion equations and accompanying boundary conditions of the "cellular model" proposed by Bullough and Perrin.[13,24] The model involves setting up rate equations for recombination and sink annihilation. These are then solved at steady state for the defect concentrations, recombination rate, and sink annihilation rate. The void growth rate is then derived from the sink annihilation rate.

The rate equations take the form of

$$\frac{dC_v}{dt} = K - (\nu_i a_i + \nu_v a_v)\ (C_v C_i - C_v^e\ C_i^e) - \nu_v p_v (C_v - C_v^e). \qquad (20)$$

$$\frac{dC_i}{dt} = K - (\nu_i a_i + \nu_v a_v)\ (C_v C_i - C_v^e\ C_i^e) - \nu_i p_i (C_i - C_i^e). \qquad (21)$$

Where C_v and C_i are the atomic fractions of vacancies and interstitials respectively. K is the radiation induced defect production rate (the same for interstitials and vacancies). ν_v and ν_i are the vacancy and interstitial jump frequencies respectively. a_v is the number of lattice sites around a stationary interstitial, where if a vacancy jumps into any one of these locations, it will immediately recombine with the interstitial. a_i is the number of lattice sites around a stationary vacancy where, if an interstitial should jump into one of these locations, it will immediately recombine with the vacancy. C_v^e and C_i^e are the thermal equilibrium concentrations of vacancies and interstitials, respectively. p_v and p_i are the probabilities that any single jump of a vacancy or interstitial, respectively, will result in an annihilation at a given sink.

The middle term in each equation represents the loss of defects due to recombination. The latter terms are the sink annihilation loss rates. The steady state form of these equations are solved based on different forms of the sink annihilation probabilities for the various sinks.

Recently, Brailsford and Bullough have reexamined the rate theory, giving special attention to the roles of each type of defect sink.[24] Sinks are classified into various categories and treated separately. The first type is neutral sinks, having no preference for either vacancies or interstitials. These include voids and incoherent precipitates. Variable bias sinks are those with only a finite capacity for defects of either type. Any sink with only a finite capacity for vacancies must acquire a bias in order to negate this accumulation. Examples of these are coherent precipitates and dislocations themselves, provided they are not able to climb. As was mentioned earlier, these sinks play an important role in swelling by serving as recombination centers. The final category is the fixed bias sink. These include dislocation loops formed during irradiation and mobile network dislocations. The origin of this fixed bias is the strong interaction between interstitials and the dislocation.

Brailsford and Bullough have developed an expression for the swelling rate and temperature dependence of swelling based on rate theory with attention given to the various types of sinks.[24] In order to obtain better results from rate theory, their work indicates that a better knowledge of the variation in dislocation density under irradiation is needed.

Harkness, et. al.[23] have treated growth similar to the Bullough and Perrin model. They have concluded that the maximum possible swelling will occur when the efficiency of a dislocation as a vacancy sink, $Q_{v,dis}$, equals the efficiency of a void as a vacancy sink, $Q_{v,void}$. During irradiation the efficiencies of both dislocations and voids as point defect sinks increases. If they increase at the same rate, a linear increase in swelling with time would be expected. Figure 7 shows the ratio of the two efficiencies as a function of irradiation temperature for 304 SS. It can be seen that the highest swelling should occur around 530° where the ratio $Q_{v,dis}/Q_{v,void}$ is closest to 1.0. This figure also illustrates the necessity for a more accurate description of the dislocation loop nucleation rate, since $Q_{v,dis}$ depends on the loop structure in the matrix.[23]

EFFECTS OF IRRADIATION PARAMETERS ON SWELLING

Temperature

Voids are observed in stainless steels over a temperature range extending from about 0.35 T_M to 0.55 T_M where T_M is the absolute melting point.[30] The limited amount of data available for 316 stainless steel makes the temperature extrema for 316 somewhat difficult to define. In 1970, Barton and Higgins proposed that the upper temperature limit for void formation on 316 stainless steel was between 560° and 570°C.[38] Later Brager et. al.[32] predicted an upper limit between 650° and 700°C. More recently the limit has been revised up to 700°C by Brager and Straalsund.[14] Figure 8 shows a prediction of the temperature dependence of swelling calculated by Harkness and Li[15] for 304 stainless steel. The bell-shaped distribution can be seen to increase with fluence. For temperatures below about 350°C, vacancy mobility is very low and thus a higher instantaneous vacancy concentration from irradiation is present. Because of the much higher mobility of interstitials at this temperature, there is a large recombination rate and therefore, migration of vacancies to sinks and void nucleation is suppressed. The upper limit of the curve in Figure 8 is due to the increased concentration of thermally produced vacancies. At high temperatures (~ 650 to ~700°C), there is no longer a sufficient supersaturation of irradiation produced vacancies required for void formation. This is illustrated in Figure 5 which shows the concentration of irradiation produced vacancies and the thermal equilibrium concentration of vacancies as a function of temperature. When the temperature approaches the upper limit for swelling, there is an increase in the thermal "evaporation" of vacancies from a void. Above the upper temperature limit there is a greater probability that a void will lose a vacancy rather than gain one.[38] There can still be considerable swelling at these high temperatures just below the maximum temperature for swelling. Appreciable supersaturation exists within a few degrees of the upper limit. When this is coupled with the high growth rate present at these temperatures, considerable swelling can result.[18] Figure 9 shows the temperature dependence of nucleation and growth rates as predicted by Harkness and Li from classical nucleation theory.[18] These curves were developed before the work of Katz and Wiedersich[7], and Russell[8] discussed in the previous sections.

Investigations have repeatedly shown that for a constant fluence, void number density decreases with increasing irradiation temperature while the average void size increases with increasing temperature.[12,14,19,32,34,38] Figure 10 shows the decreasing in void number density with temperature for 304 stainless steel. Figure 11 illustrates the increase in void size with temperature.

Interstitial loops follow the same pattern of temperature dependency as voids.[14,32,34] Figure 12 shows that for increasing temperature, the loop number density decreases for 316 stainless steel. The increase in loop diameter with temperature is illustrated in Figure 13.

Temperature has also been shown to influence the shape of voids.[19,30] For low temperatures, the voids are smaller and spherical in shape. The few large voids are polyhedral or irregularly shaped. At high temperatures, Claudson et. al.[19] and Keefer et. al.[33] have observed the voids to be octahedral with (111) planes as faces truncated by (100) surfaces.

<u>Fluence</u>

The swelling behavior of stainless steel with respect to fluence can be divided into three periods. Initially during irradiation, very little swelling occurs. After a time, the fractional volume increase of the steel begins to increase with fluence in roughly a linear fashion. Finally, at high dose levels, a saturation in swelling may occur. This last effect is a point upon which there is some disagreement. There have been indications that swelling may not saturate. These will be considered later in this section.

Bloom has stated that a threshold level of fluence exists for void formation.[30] He observes that this threshold will increase for higher irradiation temperatures. Harkness and Li,[18] and Norris[21] have reported no swelling below a fluence of 10^{22} n/cm^2 (E > 0.1 MeV). Brager et. al.[32] have detected voids at 0.4 X 10^{22}n/cm^2 (E > 0.1 MeV) in 304 stainless steel, so the threshold value is not well defined. There are two possible explanations for the occurence of such a threshold. First, the delay in void formation may be due to the time required for a sufficient amount of helium to build up so that void nuclei can be stabilized against collapse.[30] Incubation times are longer at higher temperatures because more helium is needed to stabilize a void embryo due to the lower levels of vacancy supersaturation present. As mentioned earlier, the preferential attraction of interstitials to Frank loops is a determining factor in the nucleation rate. Thus void formation could be delayed until a sufficiently large network of dislocations and loops has formed.

Above $\sim 10^{22}$ n/cm^2 ($E > 0.1$ MeV), the fractional void volume change increases with fluence by a power law relationship.

$$\frac{\Delta V}{V} \propto (\phi\ t)^n \qquad (22)$$

The exponent n of the fluence ϕt varies with temperature but is generally close to 1.0. Bates and Straalsund[50] have fitted the 316 stainless steel data to an engineering design equation similar to the above form. The data and plots of this equation for several temperature ranges is given in Figure 14.

There are differing opinions as to when and if saturation occurs. A qualitative explanation of the phenomena has been offered by Sandusky et. al.[58] Once a void is formed, it acts as a sink for vacancies. As the vacancies diffuse toward the void, a concentration gradient is set up. The region around the void containing this gradient can be referred to as the "sphere of influence" of the void. The size of this sphere is proportional to the vacancy mobility so it is larger at higher temperatures. As irradiation continues, more voids nucleate and existing voids increase in size. Eventually, the spheres of influence will begin to impinge upon one another. At this point, void concentration is said to be saturated. Nucleation has essentially stopped since all the vacancies produced have a higher probability of contributing to the growth of existing voids.

The work of Sandusky et. al.[58] and Bloom[30,59] indicate that at higher irradiation temperatures (above 500°C), saturation in void concentration may occur for fluences as low as 3 to 4 X 10^{22} n/cm^2 in 304 stainless steel. Harkness[15] also predicts a rapid saturation in void number density with fluence at the higher irradiation temperatures due to the reduction in vacancy supersaturation caused by the increase in sink population (voids and loops). At low temperatures (370-380°C), Bloom et. al.[59] found that void density appeared to saturate above 10^{23} n/cm^2 but that the overall swelling does not saturate.

Claudson has explained swelling saturation in terms of void-loop interactions that will become more important at higher fluences.[19] When impingement between a void and a loop occurs, the interstitial atoms in the loop may migrate by pipe diffusion to the void, thus reducing its size and eliminating the loop. This process is complex and not well understood. There are two effects of such an interaction. First, there is the obvious reduction in void number density. Also, the removal of the loop results in the loss of a sink biased toward interstitials. This in turn reduces the void growth rate. Harkness has incorporated this effect in the SCIM code and has predicted the swelling behavior of 304 stainless steel. His results are shown in

Figure 15. Except for the very low temperature plot (372°C), swelling has saturated for all temperatures before 1.5 X 10^{23} n/cm^2 (E > 0.1 MeV).

The highest fluence data presently reported shows no swelling saturation at 10% swelling for 304 stainless steel irradiated in EBR-II to a fluence of 11.7 X 10^{22} n/cm^2.[60] The maximum temperature of the specimens was 470°C.

Brager and Straalsund[14] do not predict any saturation in the void number density with fluence in 316 stainless steel at the fluences noted by Sandusky[58] in 304. Their description of heterogeneous nucleation on precipitates results in a void growth mechanism which predicts a rapid growth rate for small voids and a slower growth rate for the larger voids.[14] This prediction is based on the assumption that all voids are attached to precipitates. The migration of vacancies to precipitates will occur regardless of void size. However, as a void grows larger, it should emit vacancies at a higher rate due to its increasing surface area. The growth rate would decrease until a quasi-equilibrium was reached where vacancy absorption is balanced by vacancy emission.

Since there is presently no high fluence data for 316 stainless steel, most of the experimental data for examining saturation in swelling must come from ion bombardment experiments. In the United Kingdom, Mazey and Nelson[46,52,53] have irradiated 316 stainless steel at 525°C with 20 MeV carbon ions. They reported a trend toward saturation at 40 dpa. Data from these experiments is plotted in Figure 16 along with results for other metals. Data from experiments in the United States does not show any saturation effects. 1 MeV proton irradiations conducted by Keefer et. al.[33,48] show no saturation at 50 dpa at 500°C. Figure 17 presents their data for various temperatures.[67] 4 MeV nickel ion bombardments by Taylor and McDonald[47] also showed no trend toward saturation at 525°C for doses from 20-250 dpa.

Neutron Flux

Although most treatments of void formation in stainless steel consider only fluence, Harkness and Li have stated that it is necessary to separate the flux and time dependence.[15] Their work shows that for the same dose, void size and number density will be different for varying flux levels. This effect is most apparent at high temperatures. Figure 18 shows the void volume as a function of fluence for 304 stainless steel at 400°C and 600°C as predicted by the Harkness and Li model.[15] Calculations were made at each temperature using fluxes of 2×10^{15} n/cm^2 sec and 2×10^{14} n/cm^{2}sec. It can be seen that at 600°C for a fluence of 10^{23} n/cm^2, the higher flux increases the void volume by a factor of 100. Harkness and Li have also predicted the effect of flux on void number density and void size by assuming constant exposure times at various fluxes. These are shown in Figures 19 and 20. Their results show that the void density increases in a somewhat linear fashion with increasing flux. The void size is unaffected by the flux level below 500°C. The large drop in void radius with increasing flux at 600°C in Figure 20 is not explained by Harkness and Li.[15]

The results of investigations by Bloom[30] show that a flux dependency is much less than that predicted by Harkness and Li, if one exists at all. Brager et. al.[32] concluded that the defect structure is independent of flux, depending only on fluence, However, he does mention that a significant flux effect could be concealed within the scatter of the data.

Shively[51], in studying nucleation rate, has stated that the nucleation rate depends on the production rate of displacement spikes, if spike nucleation is assumed to be of major importance. Thus, he concludes that since spike production is proportional to the flux above the PKA threshold, that the nucleation rate is proportional to the instantaneous flux.

Neutron Energy Spectrum

The energy distribution of the neutron flux is important in the development of models to simulate displacement damage. In applying the irradiation data from EBR-II and other reactors to practical useage in component design, the spectral differences in flux must be accounted for.[62] In addition, since helium production has been shown to be of importance in void formation, the energy distribution of the neutrons will affect the helium production rate.

Stress

The effect of stress on swelling was first questioned in the initial discovery of voids in stainless steel by Cawthorne and Fulton.[1] Straalsund and Guthrie[49] listed the possible effects of stress as (1) changing the dislocation structure, (2) altering the precipitation process, and (3) changing the mechanism of nucleation. Appleby[11] has stated that an external stress can increase the intragranular precipitation at 500°C. Due to the association of voids with precipitates in 316 stainless steel mentioned earlier, this effect could result in an increased void concentration and swelling. Garner et. al.[63] have presented ways that stress might affect nucleation and growth processes. The first is the formation and growth of grain boundary creep cavities under tensile forces. Helium bubbles located at grain boundaries may also grow. Appleby[11] predicts this effect to be minor. In addition, the volume change associated with the hydrostatic component of irradiation creep may result in a substantial swelling increase.[63] Finally, stress motivated point defect currents could result in an increase in the void growth kinetics.[63] Garner et. al. state that the only effect of stress is to alter the free energies of formation and migration of the point defects. Void growth under a tensile stress will then increase due to the enhanced vacancy diffusivity and the alteration of the equilibrium interstitial concentrations near the void surface.

Straalsund and Guthrie[49] have incorporated the effects of stress into the Bullough and Perrin growth model. They found that the results are very sensitive to the microstructure present. The predictions also depend heavily on values of the surface energy and the vacancy formation energy which are not well defined. Figures 21 and 22 show what effects variations in these parameters can have on the swelling behavior for both the stressed and unstressed case. Their results showed little or no effect of hydrostatic tensile stress at temperatures below 450°C. However, at temperatures above which maximum swelling occurs, void growth rate increased linearly with hydrostatic stress. Other theoretical treatments of the effects of stress have been performed by Li and Nolti[29] and Loh[65,66].

Analysis of irradiation experiments by Appleby[11] at temperatures from 500 to 550°C show no evidence of stress induced void growth. He did observe some strain assisted void formation in the samples.

EFFECTS OF COLD WORK ON SWELLING

Of the methods postulated to minimize swelling in stainless steels, cold work has been studied the most intensively. Cold working introduces a high density of dislocations and a smaller grain size into the matrix. The high sink density results in a reduction of the overall point defect supersaturation, because the dislocations are now more efficient sinks for both interstitials and vacancies. In the solution treated alloy, the large capture volume around a dislocation for interstitials is thought to be dominant. However, at higher dislocation densities this effect is reduced.[41] A high dislocation density may also trap helium atoms by themselves or in groups too small to stabilize void nucleation, under the reduced vacancy supersaturation caused by cold work.[30,41]

It is generally agreed that swelling of cold worked stainless steel is reduced when compared to the swelling of solution annealed steel for low fluence irradiations. There is some question however that at higher fluences the swelling in cold worked 316 stainless steel may eventually exceed that of the solution annealed metal.[23,26] Harkness has analyzed the effects of cold work using the SCIM code.[23]

For 304 stainless steel at 372°C, he found no swelling occurs until the fluence level reaches ~7 x 10^{22} n/cm^2 (E > 0.1 MeV). At this point, helium concentration has built up sufficiently to allow nucleation to occur. According to the model, all voids are nucleated on gas atoms. The dose dependency of the swelling is now much higher than for the solution annealed metal. This is because the sink efficiency ratio, described previously, (see Figure 7) drops rapidly. When the fluence level reaches 2 x 10^{23} n/cm^2, the sink efficiency ratio is actually less for the cold worked material than for the solution treated steel. Thus a higher swelling rate for cold worked material cound eventually prevail. Stiegler and Bloom[41] have stated that the helium content in cold worked stainless steel may build up to an effective level for nucleation at fluences above only 2.5 x 10^{22} n/cm^2. Also the larger number of trapping sites for helium available in the cold worked matrix could eventually increase the concentration of voids to a higher value than in annealed materials. The Bullough and Perrin growth model[25-27] presented earlier predicts the same effect. Initially, cold work will substantially reduce swelling, but at higher fluences, the large dislocation density results in higher growth kinetics so that the swelling eventually may exceed that of a comparably irradiated solution annealed sample. This effect was noted earlier in Figure 6. Ion bombardment experiments by Nelson et. al.[46] also show the same trend.

Presently, there is some disagreement over how cold working affects the nucleation and growth process. Experimental evidence presented by several investigators in the United States[11,30,31,36,41,61] is that cold working reduces the number of density voids, but does not influence the growth of voids to any appreciable extent. This effect is illustrated in

Figure 23. Stiegler and Bloom have shown that cold work reduces the void number density by a factor of 10 in EBR-II irradiations at a fluence of 2.5 x 10^{22} n/cm^2 (E > 0.1 MeV) and a temperature of 455°C, while the void size is only slightly decreased. Thus, the lower void concentration is the reason for the initial decrease in swelling in cold worked material relative to the solution treated metal.

Analyses from the United Kingdom indicate the opposite effect.[35] Void number density is not affected appreciably by cold work but the the void size is reduced. Swelling is thus reduced because of the smaller void size observed in cold worked 316 stainless steel. Appleby[11] has stated that the differences may be due to the manner in which the irradiation tests were conducted.

Straalsund et. al.[36] have noted that the thermal stability and effectiveness of the cold worked dislocation structure in preventing void formation depend strongly on minor variations in alloy chemistry. Compositional changes can lead to recrystallization followed by void growth during irradiation.

The degree of cold work can also be an important variable. Brager and Straalsund [64] have characterized the effect of various degrees of cold work on swelling as shown in Figure 24. The curve labeled

$$y = e^{-0.115x}$$

is based on improved TEM data (solid line). The other curve (dashed line) was formulated when only bulk density data was available. Originally, it was thought that cold working beyond 20% would do little to increase swelling resistance especially since recrystallization was detected at 732°C in a 75% cold worked sample. However, based on the new TEM data, Brager and Straalsund have now concluded that degrees of cold work above 20% merit some investigation.[64] It should be noted that these results are only for low fluence (2.8×10^{22} n/cm^2, E > 0.1 MeV) and low temperature (427°C) conditions.

A recent report by Brager and Straalsund[42] examines the change in dislocation structure in cold worked 316 stainless steel during irradiation. They found that for 20% cold worked stainless steel irradiated at ~450°C, irradiation induces the replacement of nearly the entire dislocation structure with Frank interstitial loops. At the low fluence of 0.6×10^{22} n/cm^2 the dislocation density decreased by an order of magnitude, and a high density of Frank loops (~3×10^{15} loops/cm^3) was formed. Solution treated 316 irradiated to the same fluence resulted in only one-third of this Frank loop density. The study has shown that increasing dislocation density suppresses void formation but not Frank loop nucleation.

The temperature dependence of swelling in cold worked austenitic stainless steels varies considerably from the "bell-shaped" curve of the solution treated alloy.[36,61] There are two maxima in the cold worked swelling curve. One occurs at low temperatures and the other at high temperatures. Figure 25 exhibits such a curve for 304 stainless steel. Brager and Straalsund have noted that the temperature dependence of cold worked 316 is similar to that shown in Figure 25 but it is not quite so pronounced. This is because in the temperature region for the second maximum, in Figure 25, significant recrystallization occured in the 50% cold worked 304 stainless steel. Voids later formed in these zones. No recrystallization effects were observed in the 316 samples. At low temperatures (in the region of the first maximum), voids and Frank faulted loops are produced in the same manner as described for the solution annealed alloy.[61] However, this occurs at a much slower rate since the cold work induced dislocations are efficient sinks and reduce the vacancy and interstitial supersaturation levels. For temperatures immediately above this first maximum, the increase in vacancy mobility and thermally produced vacancy concentration results in a lowering of the nucleation rate. This is shown in Figure 26 for 316 stainless steel. As the neutron fluence level increases, the intermediate temperature range and fluence values over which void formation is reduced decreases. Brager and Straalsund have attributed this to: (1) the gradual buildup of helium with fluence which would tend to enhance nucleation; (2) the development of a dislocation structure more characteristic of solution annealed alloys (discussed above); or (3) time dependent thermal recovery. At 600°C (see Figure 26), recovery of the cold worked induced dislocation structure occurs. Voids are observed to form in the areas where this recovery has occurred. At temperatures approaching 700°C, the thermal equilibrium concentration of vacancies is high and there is no longer a significant supersaturation of radiation induced defects, hence nucleation ceases. This study empahsizes that when determining the temperature dependence of swelling cold worked 316 stainless steel that the temperature dependence of the recovery of cold worked induced dislocations must be included. Brager and Straalsund's work[61] is conveniently summarized in Figure 27 for temperatures from about 400°C to over 800°C and fluences up to 5.5×10^{22} n/cm^2 (E > 0.1 MeV). It can be seen that at high fluences, voids will be formed at all temperatures below 700°C, thus the reduced swelling effect around 500°C is not permanent.

EFFECTS OF SWELLING ON THE UWCTR DESIGN

The structure of the blanket and shield region for the UWCTR design is illustrated in Figures 28-30: Figure 28 shows the cutaway view of the blanket module; Figure 29 is an enlarged view of the blanket; and Figure 30 illustrates the design of a single heat removal cell. The design of the blanket and heat removal system is described in detail elsewhere.[68] 316 stainless steel has been chosen as the structural and first wall material for the blanket. Since the mechanical properties of the steel are degraded at temperatures above about 650°C, the maximum operating temperature must be kept below this level. The corrosion rate of the stainless steel by the lithium coolant places even more restrictive limits on the system, requiring that the maximum temperature of the steel be no larger than 500°C. The direction of flow of the lithium is indicated in Figures 29 and 30. The temperatures of the lithium coolant in each heat removal cell have been calculated[68] and are listed in Table I. The letter under the maximum wall temperatures refer to the location in Figure 29 where these temperatures occur.

TABLE I

Lithium Temperatures in Heat Removal Cell

	First U-Bend	Second U-Bend	Third U-Bend	Fourth U-Bend
T_{inlet}	275	325	375	425°C
T_{outlet}	325	375	425	475°C
T max, first wall	350	400	450	500°C
(Location on Figure 29)	(A)	(B)	(C)	(D)

TABLE II

DISPLACEMENT RATES AT VARIOUS BLANKET LOCATIONS

	First Wall	Blanket/Shield Interface	Inside Surface of B_4C	Exterior Surface of Outer Shield
dpa/sec	4.37×10^{-7}	3.46×10^{-9}	3.13×10^{-10}	7.26×10^{-15}
dpa/year	13.72	0.108	0.0098	2.28×10^{-7}
dpa after 20 years	274.4	2.16	0.196	4.56×10^{-6}

Kulcinski has calculated the displacement rates at several positions throughout the blanket for a first wall loading of 0.50 MW/m^2.[69] These values have been modified to conform to the present design wall loading of 0.53 MW/m^2 and are listed in Table II. In order to compare these displacement rates to irradiation data, they must be converted to fluence levels. Doran[70] has calculated that dpa values can be converted to fission neutron fluences for the core center of EBR-II by the following:

$$7.03 \text{ dpa} \approx 10^{22} \text{ n/cm}^2 \ (E > 0.1 \text{ MeV}).$$

Using these values, the fluence levels at various stages of the UWCTR life have been calculated for the first wall.

TABLE III

dpa and Fluence Levels At Various Irradiation Times In UWCTR First Wall

UWCTR Irradiation Time	dpa	Equivalent Fast Reactor Fluence (n/cm^2, E > 0.1 MeV)
1 year	13.72	1.95×10^{22}
2 years	22.74	3.90×10^{22}
5 years	68.60	9.76×10^{22}
10 years	137.20	19.52×10^{22}
20 years	274.40	39.03×10^{22}

As can be seen from Table II, the displacement rates in the outer shield areas are so small that no swelling will result even after 20 years of operation. The same is true of the poloidal header region above the graphite reflector. If any swelling is present after 20 years of operation it will be less than 1% and not of importance in design considerations. This leaves only the first wall and structural components of the heat removal cells as areas where swelling is apt to influence the design.

It was noted earlier that the swelling behavior of stainless steel as a function of temperature is a "bell-shaped" distribution having a maximum at approximately 500°C. Below approximately 350°C, there is essentially no swelling due to void formation. The range of first wall temperatures in the UWCTR design extends from 300°C to 500°C. Thus it is evident that the swelling will be non-uniform over the first wall. The regions where the first wall temperature is around 300-350°C will experience little or no swelling over the entire 20 year reactor life. However at 500°C, swelling will probably be greater than 10% and possibly much higher than this value.

Solution Annealed 316 Stainless Steel

All of the neutron irradiation data presently available for solution annealed 316 stainless steel is for fluences less than 8×10^{22} n/cm^2 (E > 0.1 MeV). This corresponds to about 4 years of the UWCTR lifetime. Thus predictions of the swelling behavior of the first wall at 20 years must be made by extrapolation of the current results, or by the use of ion bombardment data.

Brager and Straalsund[14] have compiled the following equation for the swelling as a function of fluence and temperature based on EBR-II data. They anticipate that the equation will yield good results for fluences at least up to 10^{23}n/cm^2.

$$\frac{\Delta V}{V}(\%) = \frac{\pi}{6} \cdot 10^{-9}\ K(T)\ (\phi t)^{D(T)}\ \exp\left(E(T)\right) \qquad (23)$$

Where

$$K(T) = 0.48 + 9.2 \times 10^{-4}T \qquad \text{for } T < 873^{\circ}K \qquad (24)$$

$$K(T) = 1.28 \qquad \text{for } T > 873^{\circ}K$$

$$D(T) = 0.72 + \frac{1.7}{1 + \exp\ 0.04\ (700-T)} \qquad (25)$$

$$E(T) = 49.46 - 0.0243T - \frac{9440}{T} \qquad (26)$$

The temperature is ^{o}K and the fluence in 10^{22} n/cm^2 (E > 0.1 MeV).

Using these relations,values for the per cent volume change were calculated for the UWCTR first wall at 500^{o}C after 1, 2, 5, 10, and 20 year exposures to a 0.53 MW/m^2 wall loading. The fluences used were the values mentioned earlier. The following results were obtained.

Irradiation Time (years)	Equivalent Fast Reactor Fluence to First Wall ($\times 10^{22}$ n/cm^2, E > 0.1MeV)	Swelling % $\frac{\Delta V}{V}$
1	1.95	0.31%
2	3.90	1.61%
5	9.76	13.10%
10	19.52	67.
20	39.03	331.

If these equations provide good estimates up to 10^{23} n/cm^2, then the swelling predictions up to 5 years are reliable. The highest fluence data for solution treated 316 stainless steel has been provided by Appleby and Wolff.[12] They found 7.76% swelling in a sample irradiated at 508^{o}C to 7.32 x 10^{22} n/cm^2 (E > 0.1 MeV).*

Higher fluence data has been obtained from ion bombardment experiments, however care must be taken in using these results to predict neutron irradiations. The higher displacement rates result in a shift of the swelling versus temperature curve to higher temperatures. Bullough and Perrin[26] have developed an equation to describe this increase in temperature, which Kulcinski et. al.[71] presents

*Swelling determined from immersion density measurements. TEM analysis yielded 7.1% swelling.

in the following form

$$\Delta T = T_i^2 \frac{k}{E_v} \frac{1}{1 + \frac{kT_i}{E_v} \ln \frac{K_i}{K_n}} \ln \frac{K_i}{K_n} \qquad (27)$$

ΔT is the temperature shift; k is Boltzman's constant; T_i is the ion irradiation temperature; E_v is the energy of diffusion for vacancies; K_i is the defect production rate during ion bombardment; and K_n is the neutron bombardment defect production rate. The theoretical work of Bullough and Perrin[25] has also shown that the maximum swelling value does not change significantly for the higher dose rates.

Nelson and Mazey[46] have irradiated solution annealed 316 stainless steel with 20 MeV C^{++} ions at 525°C. The damage rate was approximately 10^{-3}dpa. As noted before they observed saturation around 40 dpa (see Figure 16). They calculated, using equation (27) that the temperature shift was approximately 100°C. Thus for neutron irradiations this data corresponds to approximately 425°C. From Figure 16, the following swelling levels would be observed.

UWCTR Irradiation Time	dpa	Swelling
1 year	13.72	0.8%
2 years	27.44	2.6%
5 years	68.60	9.0%
10 years	137.20	10.0%
20 years	274.40	10.2%

Empirical relations describing the swelling as a function of dpa have been computed by Keefer et. al.[67] based on data from their proton irradiation experiments. Most of these irradiations were at about the same displacement rates as the C^{++} experiments of Nelson et. al.[46]. The maximum dpa values of the data reported are less than 50 dpa. The samples also were preinjected with about 5 appm helium. The equations derived are listed below.

Temperature	Equation	
400°C	$\frac{\Delta V}{V}(\%) = 0.19\ (dpa)^{0.72}$	(28)
500°C	$\frac{\Delta V}{V}(\%) = 0.14\ (dpa)^{1.3}$	(29)
600°C	$\frac{\Delta V}{V}(\%) = 0.0013\ (dpa)^{2.9}$	(30)

If it is assumed that the temperature dependence has shifted by about 100°C, estimates of the swelling can be made for low damage levels using these equations. Assuming that equation (29) will correspond to about 400°C for neutron irradiation, and equation (30) roughly approximates a 500°C neutron irradiation, the following values can be calculated.

UWCTR Irradiation Time	dpa	Swelling 400°C	Swelling 500°C
1 year	13.72	4.21%	2.58%
2 years	27.44	10.38%	19.29%

For higher fluences, the swelling predictions become unrealistic. These numbers illustrate that the nucleation of voids at higher temperatures is delayed, but that the growth rate is much greater.

4MeV Ni^+ bombardments by McDonald and Taylor[47] have yielded smaller volume increases than those of Keefer et.al.[67] mentioned above. For a displacement rate of 5 x 10^{-3} dpa/sec, solution annealed 316 swells to about 20% at 100 dpa at 600°C. The authors calculate an increase in the peak swelling temperature of 125°C, thus this would correspond to approximately 475°C in an EBR-II irradiation. Data for irradiations at 525°C (400°C in a neutron irradiation) predict less than 10% swelling at the reactor lifetime damage level of 274.4 dpa. Their results also show that helium preinjection does not affect the void number density for ion irradiation temperatures below 600°C. As mentioned previously, McDonald and Taylor observed no saturation effects.

Cold Worked 316 Stainless Steel

Even less data is available for cold worked 316 stainless steel than for the solution annealed 316. As mentioned earlier, cold working reduces the volume increase relative to a solution annealed sample irradiated under identical conditions. The C^{++} bombardment experiments of Nelson et. al.[46] have shown that cold work initially reduces swelling, but at 200 dpa the swelling is the same as in solution treated metal. However, when ion bombardments are used to irradiate cold worked samples, the temperature shift mentioned previously presents problems in analyzing the data. While the swelling curve may shift to higher temperatures with higher displacement rates, the temperature dependence of recovery of the cold work induced dislocation structure remains the same. As an example, suppose a cold worked sample underwent an ion bombardment at a displacement rate of about 10^{-3} dpa/sec at $600^{o}C$. The swelling behavior should be characteristic of a neutron irradiation at roughly $500^{o}C$ to the same fluence. However, at $600^{o}C$ significant recovery of the dislocation structure can occur resulting in higher swelling. At $500^{o}C$ the thermal recovery is only slight and thus the swelling in a neutron irradiation would be less than that predicted by the ion experiment. Thus, the recovery that can occur in ion bombardments is not indicative of the corresponding neutron irradiated sample. This effect should be kept in mind when analyzing ion bombardment data for cold worked samples.

Figure 27 shows that at $500^{o}C$ (the maximum operating temperature of the first wall), that cold work initially eliminates swelling. According to Figure 27, no voids are formed until a fluence of 4.0×10^{22} n/cm^2 (E > 0.1 MeV) is reached. This corresponds to just over 2 years exposure in the UWCTR. After this time swelling will be observed at all temperatures.

The Ni^{+} irradiations of McDonald and Taylor[47] indicate that cold work may be very effective. At 120 dpa in a 20% cold worked sample irradiated at $525^{o}C$ very few voids were observed. At $600^{o}C$ voids were formed in the localized regions where recovery had occurred, but not elsewhere in the sample.

CONCLUSIONS

As stated earlier, only the first wall and structural components of the heat removal cells are subject to high enough fluences that swelling will be of concern. The first wall can be taken as the limiting case since it receives the highest dose. If the first wall can withstand the radiation-induced swelling, it is probable that the remainder of the structure, spread out over 50 cm from the first wall to the graphite, can also tolerate the effects.

It appears that cold worked 316 stainless steel will exhibit less swelling than solution annealed, at least over most of the reactor lifetime. Whether the swelling of solution annealed 316 will be maintained below an acceptable level depends on how much, if any, saturation occurs. If there is not saturation, extrapolation of neutron data for 500°C, predicts swelling values for solution treated 316 stainless steel in excess of 100% will be attained before 20 years in the UWCTR. This level of swelling is clearly unacceptable. At least for the lower doses, it appears that cold worked 316 stainless steel offers a better resistance to swelling.

It was mentioned previously that helium is observed to enhance void nucleation. This fact may be of even more significance in CTR's than in LMFBR's. Although accurate cross sections for 316 stainless steel have not yet been compiled, it is known that the (n, α) cross sections for several of the components of stainless steel are higher at neutron energies around the 14 MeV energy range than at the fission energies present in EBR-II. Thus swelling may be greater

UWCTR than levels currently projected for LMFBR's at the same fluence because of the increased helium production. The helium production rates and (n,α) cross sections are to be discussed in more detail by Lott.[72]

The effect of stress on swelling in the first wall can not be determined because at present the effects of stress on void nucleation and growth are not understood. Based on the preliminary work it is probable that any stress effects will only be of concern at the higher first wall temperatures. This again brings up the nonuniform behavior of swelling in the first wall.

Because of the 200^{o}C variation between the minimum and maximum first wall temperatures, a very large swelling gradient will exist. There will also be a differential volume change due to varying degrees of thermal expansion in the first wall. However, this will be at most 0.1% and when compared to swelling volume changes is negligible. It is possible that such a high differential swelling rate would produce high stresses resulting in bowing and deformation of the first wall and heat removal cell structure. The Appendix to this paper shows estimates of the swelling expected at 5 years at various locations throughout the heat removal cell. As can be seen, swelling varies widely over this region. The nature of the stresses that will result are complex and should be investigated.

Swelling may affect the cooling capabilities of the heat removal cell in two ways. First of all void formation will reduce the coolant flow area. For the present UWCTR design this effect will be negligible since the channels are relatively large (7.5 cm tube size is used[68]). Even uniform swelling approaching 100% will not reduce the flow area more than a few per cent. However, should bowing result, significant reductions in flow area might be created and "hot spot" effects could be encountered. The other effect of swelling would be to decrease the thermal conductivity of the 316 stainless steel which would result in decreased heat transfer performance and increase thermal stresses in the first wall.

In summary, the true impact of void formation on the performance of the blanket structure and first wall can not be confidently predicted until higher fluence neutron and ion bombardment data has been gathered and analyzed. The key point will be whether or not saturation in swelling occurs at the high neutron fluences expected during the UWCTR lifetime.

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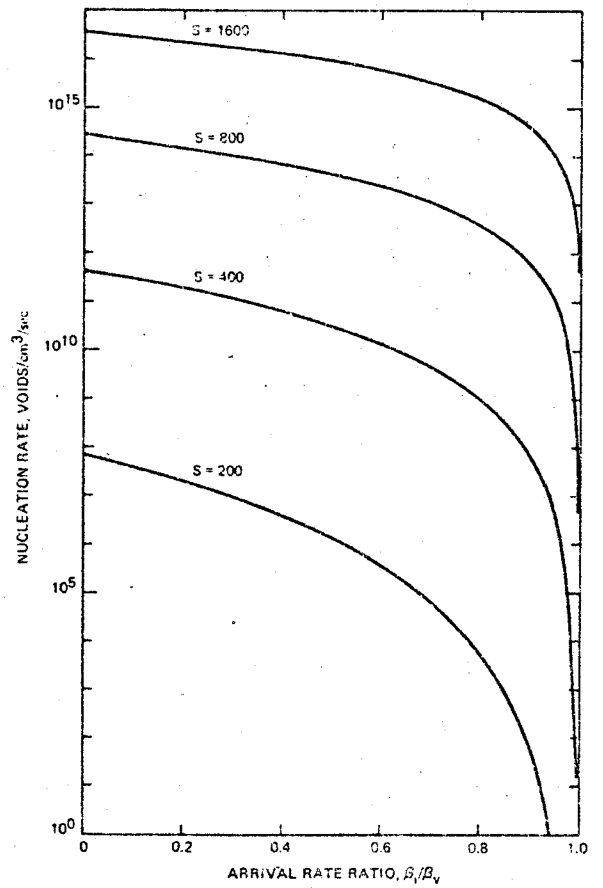
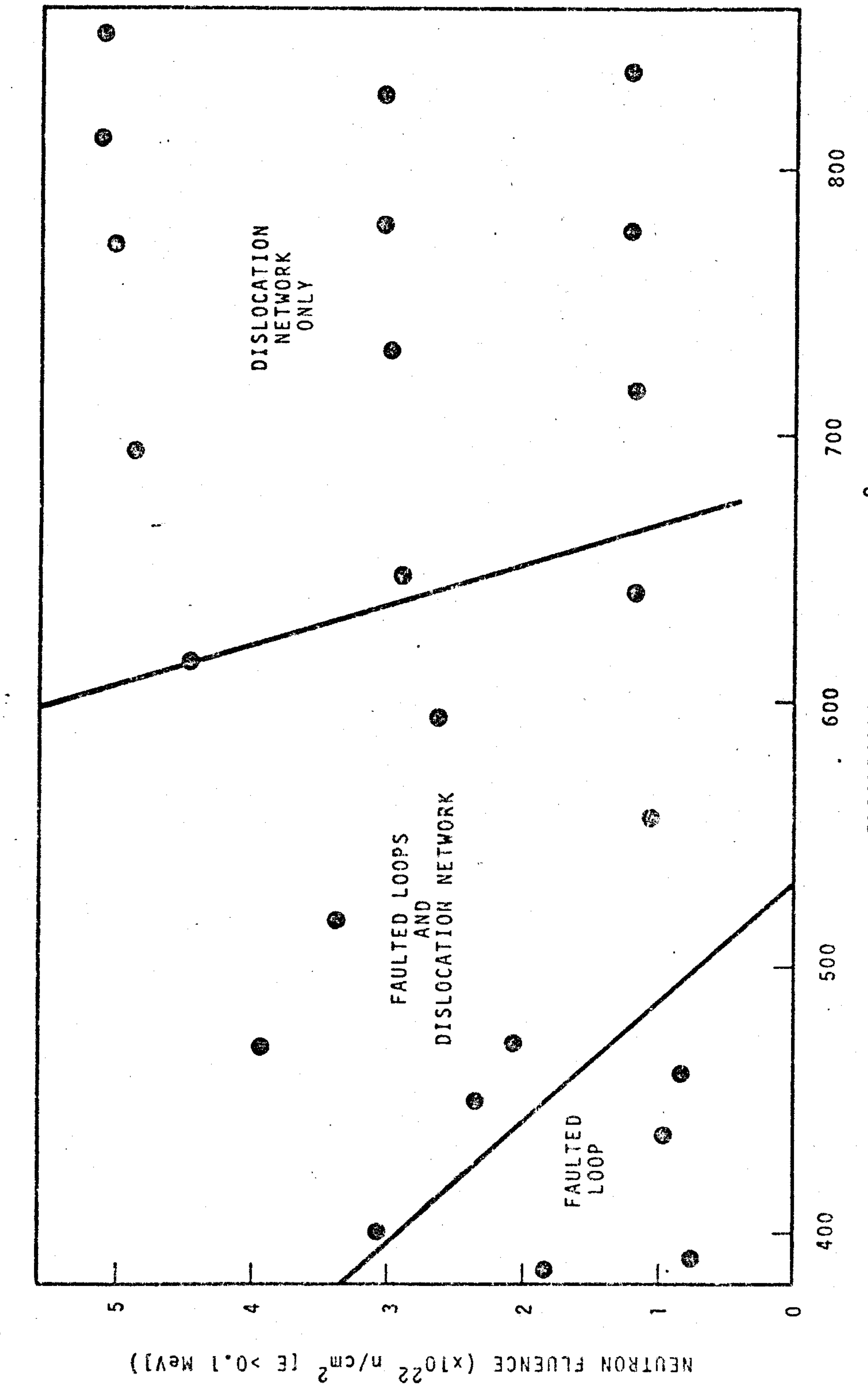


Void nucleation rate as function of the arrival rate ratio for several vacancy supersaturations S. $T=900°K$.

Figure 1

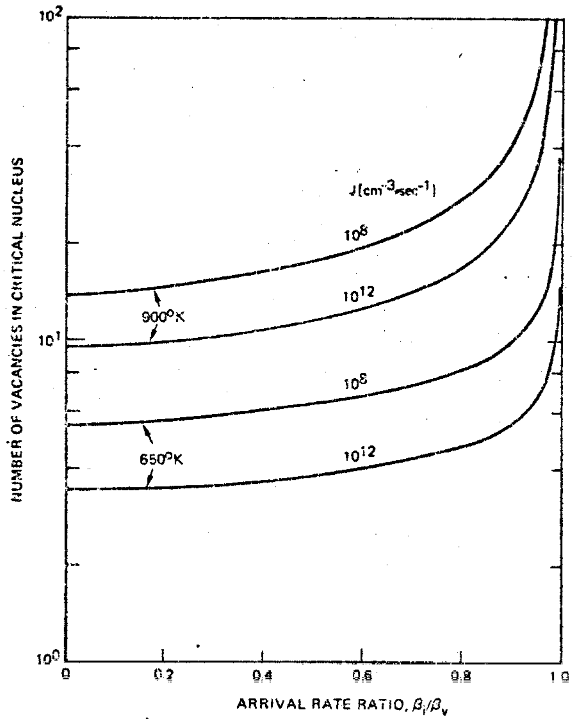
J. L. Katz and H. Wiedersich, "Nucleation of Voids in Materials Supersaturated with Vacancies and Interstitials," J. Chem. Phys. 55: 1414, (Aug. 1, 1971).



Fluence and Temperature-Dependence of Frank Loop and Dislocation Development in Fast Reactor Irradiated, Solution-Treated 316 SS.

Figure 2

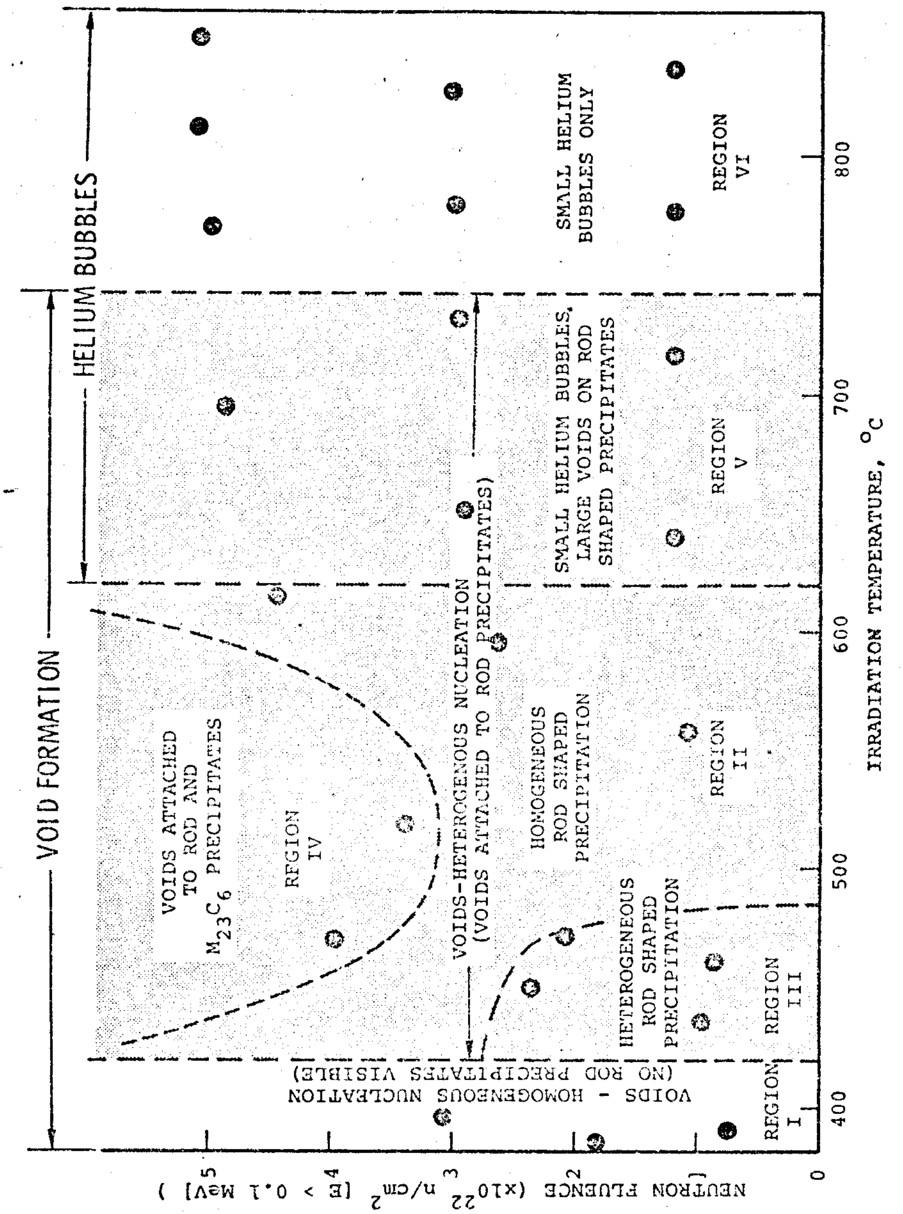
H. R. Brager and J. L. Straalsund, "Defect Development in Neutron-Irradiated Type 316 Stainless Steel," HEDL-TME 72-108, (1972).



Size of critical nuclei as function of the arrival rate ratio. The nucleation rates J and temperatures are indicated as parameters.

Figure 3

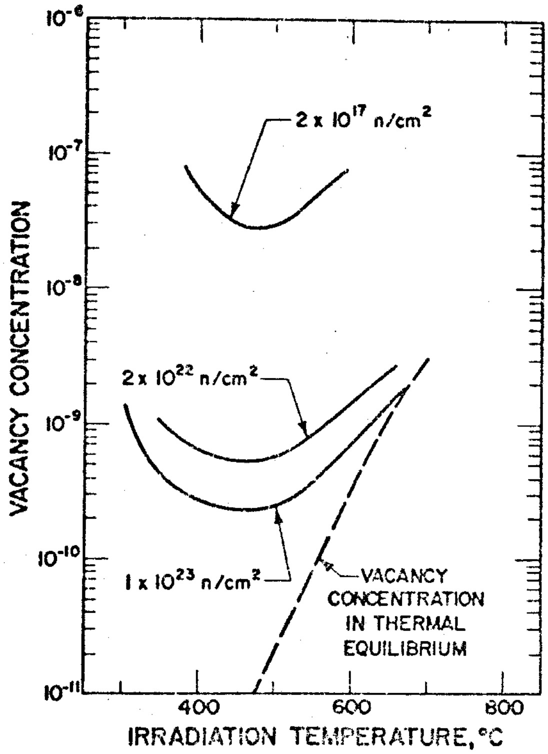
J. L. Katz and H. Wiedersich, "Nucleation of Voids in Materials Supersaturated with Vacancies and Interstitials," J. Chem. Phys. 55 : 1414, (Aug. 1, 1971).



Neutron Fluence and Irradiation Temperature Regions of Void-Helium Bubble Formation in Solution-Treated 316 SS

Figure 4

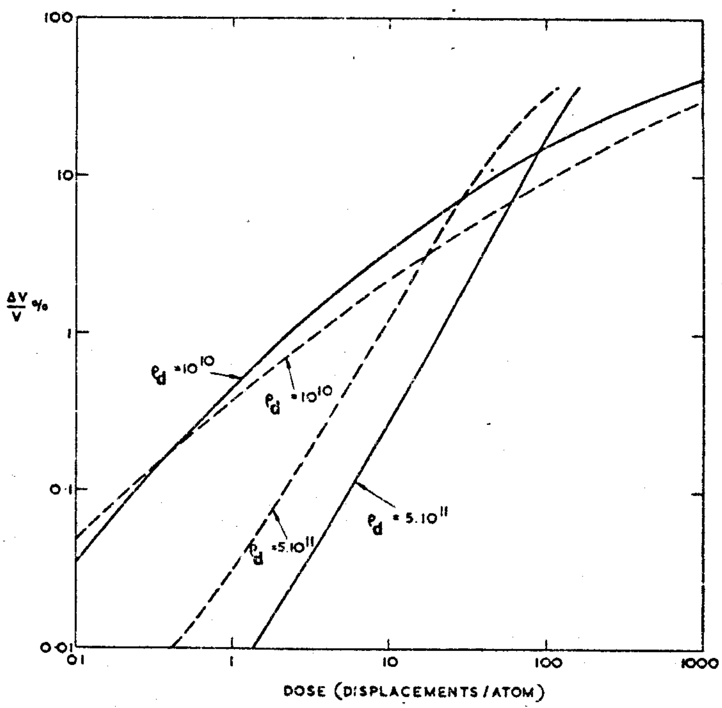
H. R. Brager and J. L. Straalsund, "Defect Development in Neutron-Irradiated Type 316 Stainless Steel," HEDL-TME 72-108, (1972).



·The predicted effect of irradiation temperature and neutron dose on the steady-state vacancy concentrations.

Figure 5

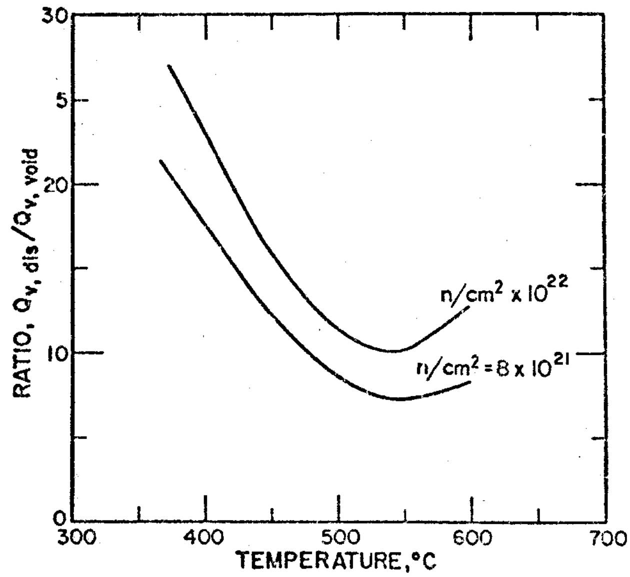
S. D. Harkness and Che-Yu Li, "A Study of Void Formation in Fast Neutron-Irradiated Metals," <u>Met. Trans. 2</u> : 1457, (May, 1971).



—*The kinetics of void swelling when $Z_v = 1.00$ and $Z_i = 1.02$ for various void and dislocation densities. The solid line denotes a void density $\rho_v = 10^{15}/cm^3$, the dashed line denotes a void density $\rho_v = 5 \times 10^{15}/cm^3$, and ρ_d is the dislocation density in lines/cm^2.*

Figure 6

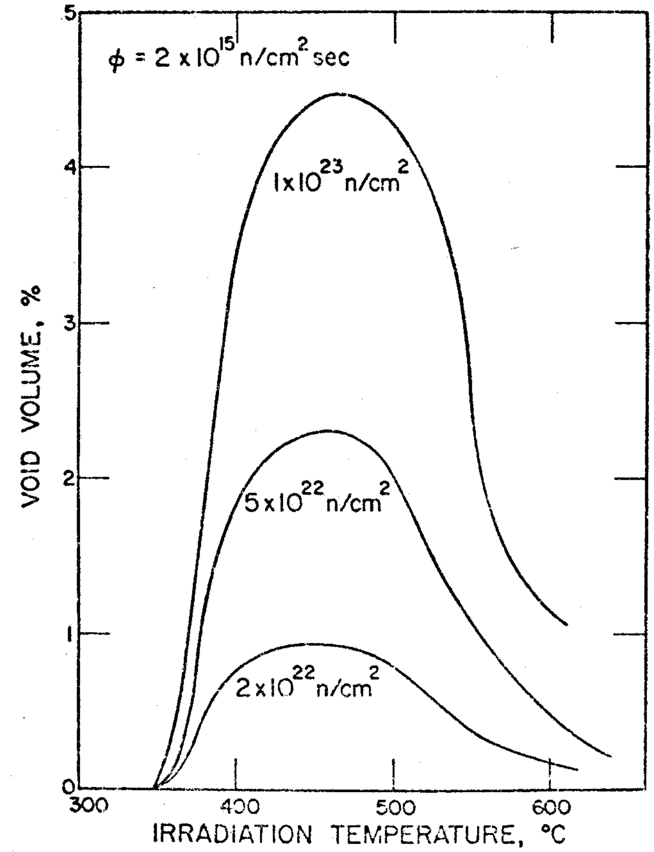
R. Bullough and R. C. Perrin, "The Mechanisms and Kinetics of Void Growth During Neutron Irradiation," Irradiation Effects on Structural Alloys for Nuclear Reactor Applications, ASTM STP No. 484, p. 317, (June 29-July 1, 1970).



Change in relative efficiencies of voids and dislocations as sinks as a function of irradiation temperature.

Figure 7

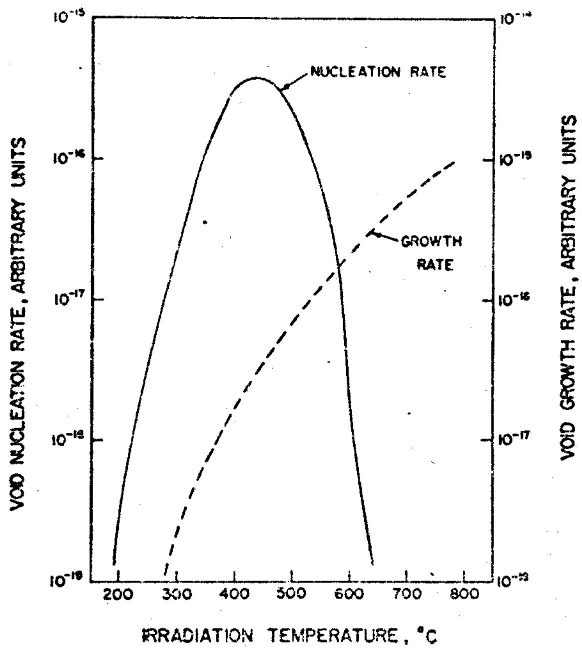
S. D. Harkness, R. Grappel, and S. G. McDonald, "SCIM - A Theory Based Computer Code for the Prediction of the In-Pile Behavior of Type 304 Stainless Steel," Nucl. Tech. 16 : 25, (Oct. 1972).



—The predicted effect of irradiation temperature on void volume.

Figure 8

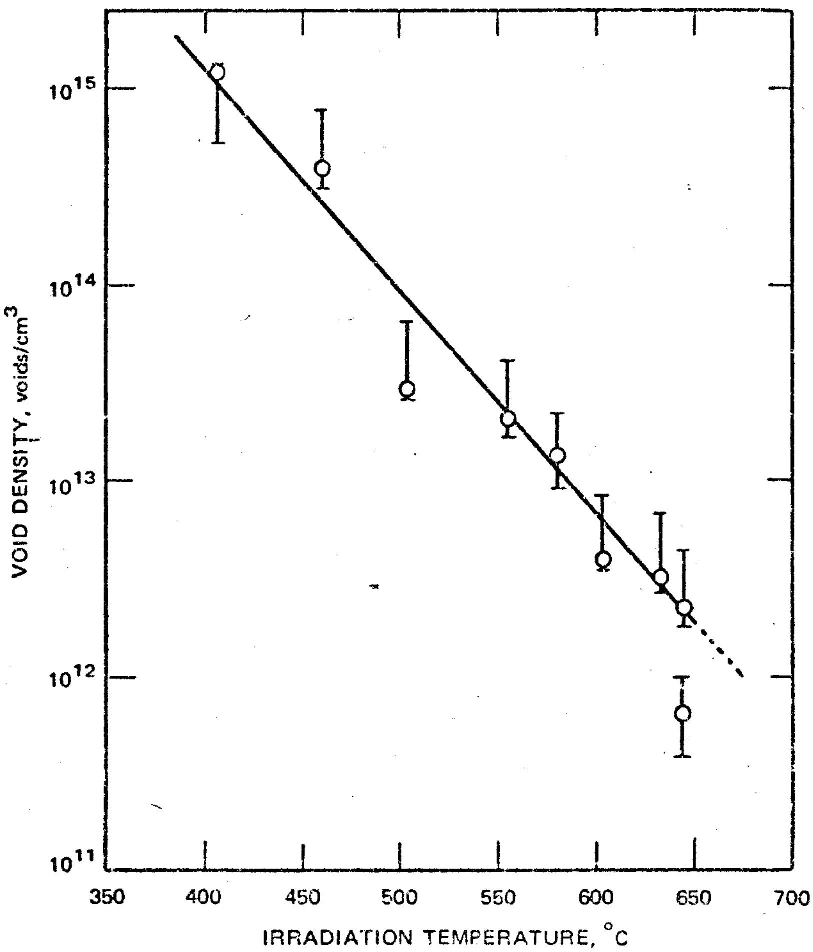
S. D. Harkness and Che-Yu Li, "A Study of Void Formation in Fast Neutron-Irradiated Metals," <u>Met. Trans. 2</u> : 1457, (May, 1971).



Temperature dependence of the rates of void nucleation and void growth in type 304 stainless steel irradiated in EBR-II.

Figure 9

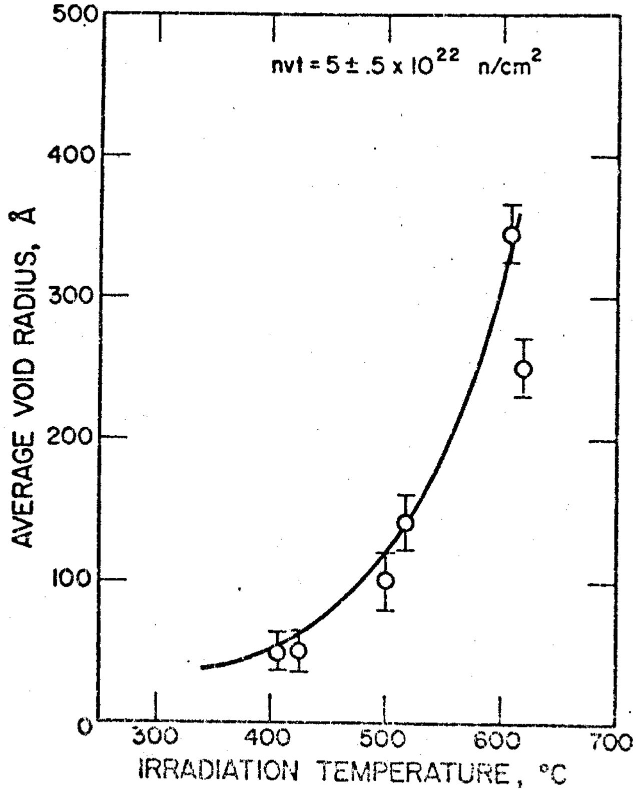
S. D. Harkness and Che-Yu Li, "A Model for Void Formation in Metals Irradiated in a Fast-Neutron Environment," Proceedings of the Symposium on Radiation Damage in Reactor Materials, Vol. II : 165, (June 2-6, 1969).



Void number density in 304 stainless steel vs. irradiation temperature.

Figure 10

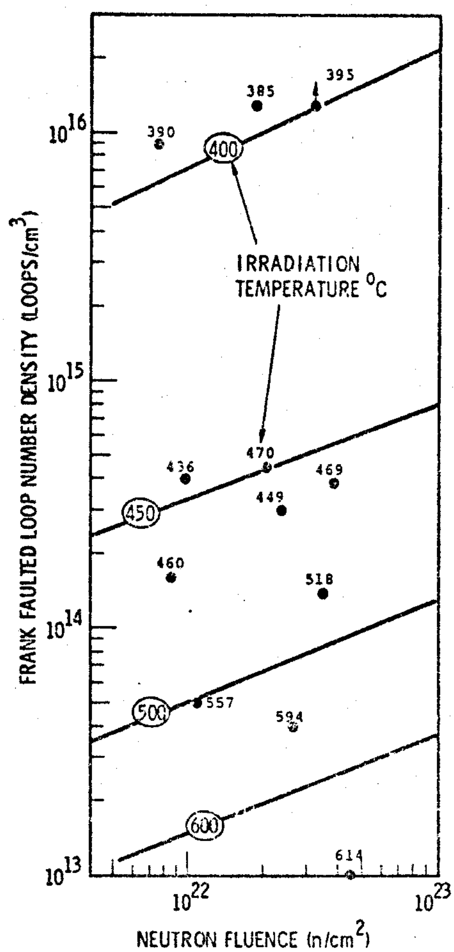
W. K. Appleby, D. W. Sandusky, and U. E. Wolff, "Fluence and Temperature Dependence of Void Formation in Highly Irradiated Stainless Steels," Radiation-Induced Voids in Metals, Conference Proceedings, p. 156, (June 9-11, 1971).



The effect of irradiation temperature on the average void size in Type 304 stainless steel. Solid line: prediction of model.

Figure 11

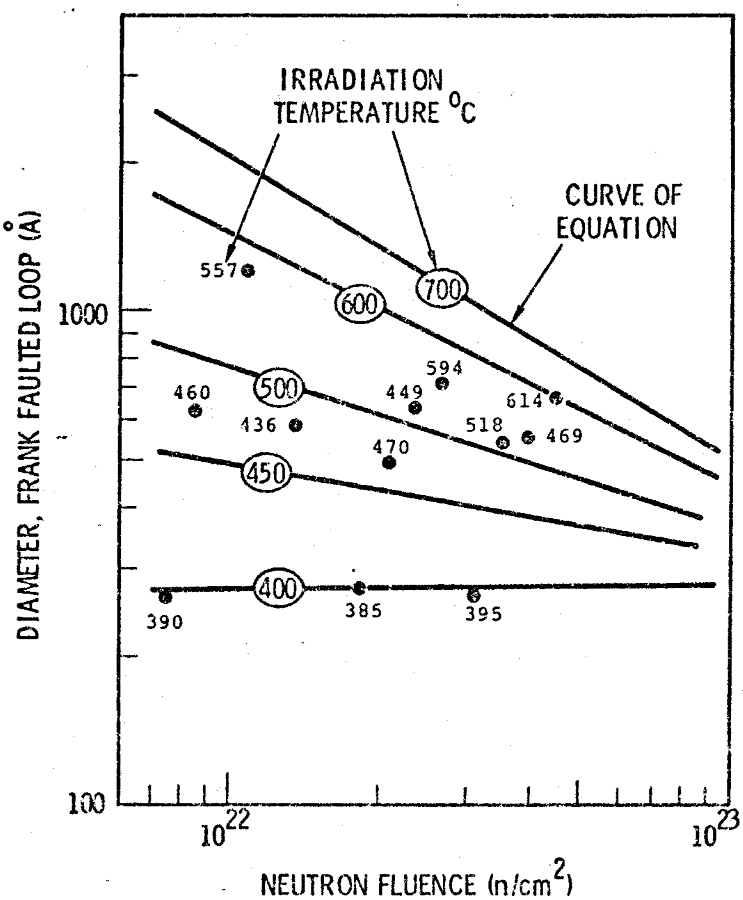
S. D. Harkness and Che-Yu Li, "A Study of Void Formation in Fast Neutron-Irradiated Metals," Met. Trans. 2 : 1457, (May 1971).



Comparison of Frank Faulted Loop Density Data as a Function of Neutron Fluence and Irradiation Temperature with Correlated Values of Equation.

Figure 12

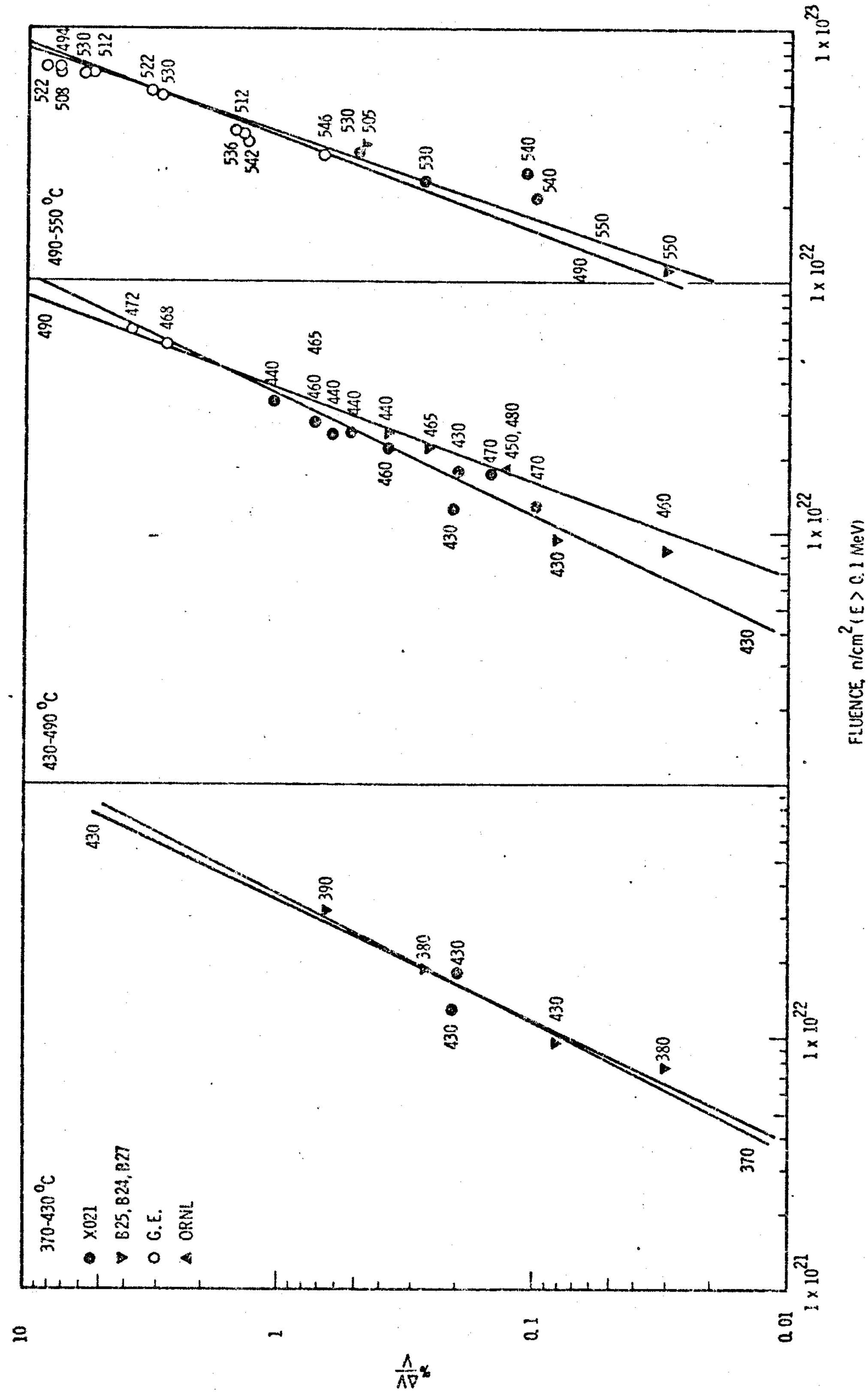
H. R. Brager and J. L. Straalsund, "Defect Development in Neutron-Irradiated Type 316 Stainless Steel," HEDL-TME 72-108, (1972).



Variation of the Frank Faulted Loop Mean Diameter as a Function of Neutron Fluence and Irradiation Temperature.

Figure 13

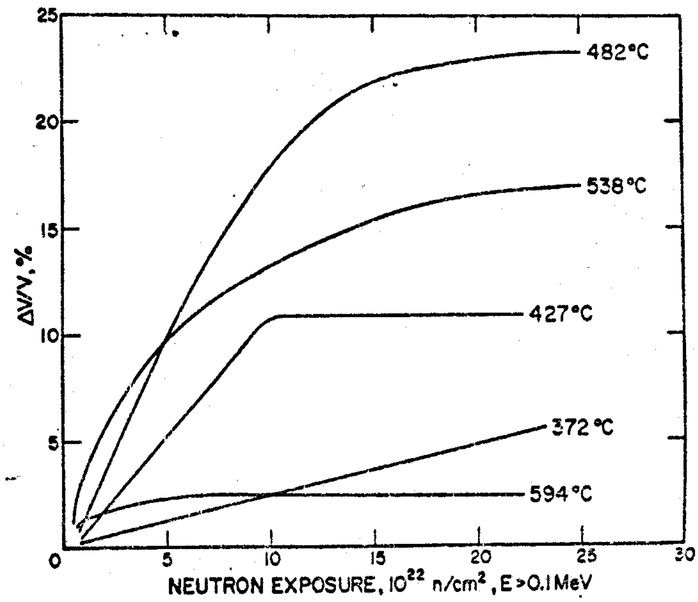
H. R. Brager and J. L. Straalsund, "Defect Development in Neutron-Irradiated Type 316 Stainless Steel," HEDL-TME 72-108, (1972).



Swelling in Solution Treated Type 316 Stainless Steel

Figure 14

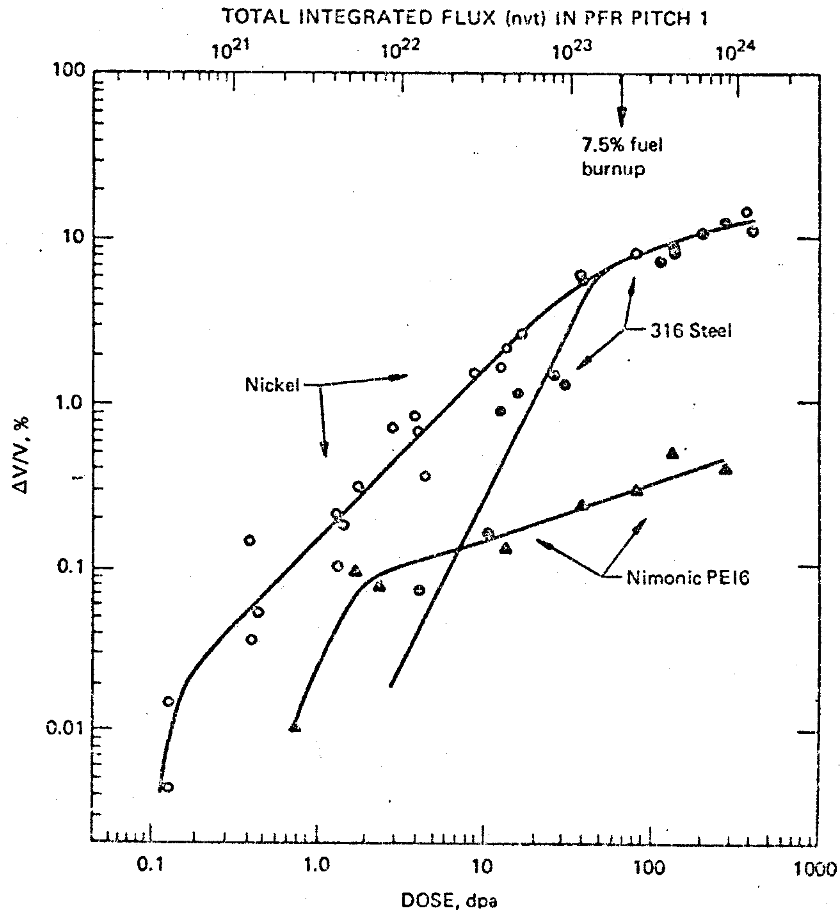
J. F. Bates and J. L. Straalsund, "A Compilation of Data and Empirical Representations of Irradiation-Induced Swelling of Solution Treated Types 304 and 316 Stainless Steel," HEDL-TME 71-139, (Sept. 1971).



*. Effect of Neutron Exposure and Irradiation Temperature on Void Volume in 304 SS. Note Predicted Saturation Behavior. MSD-56774.

Figure 15

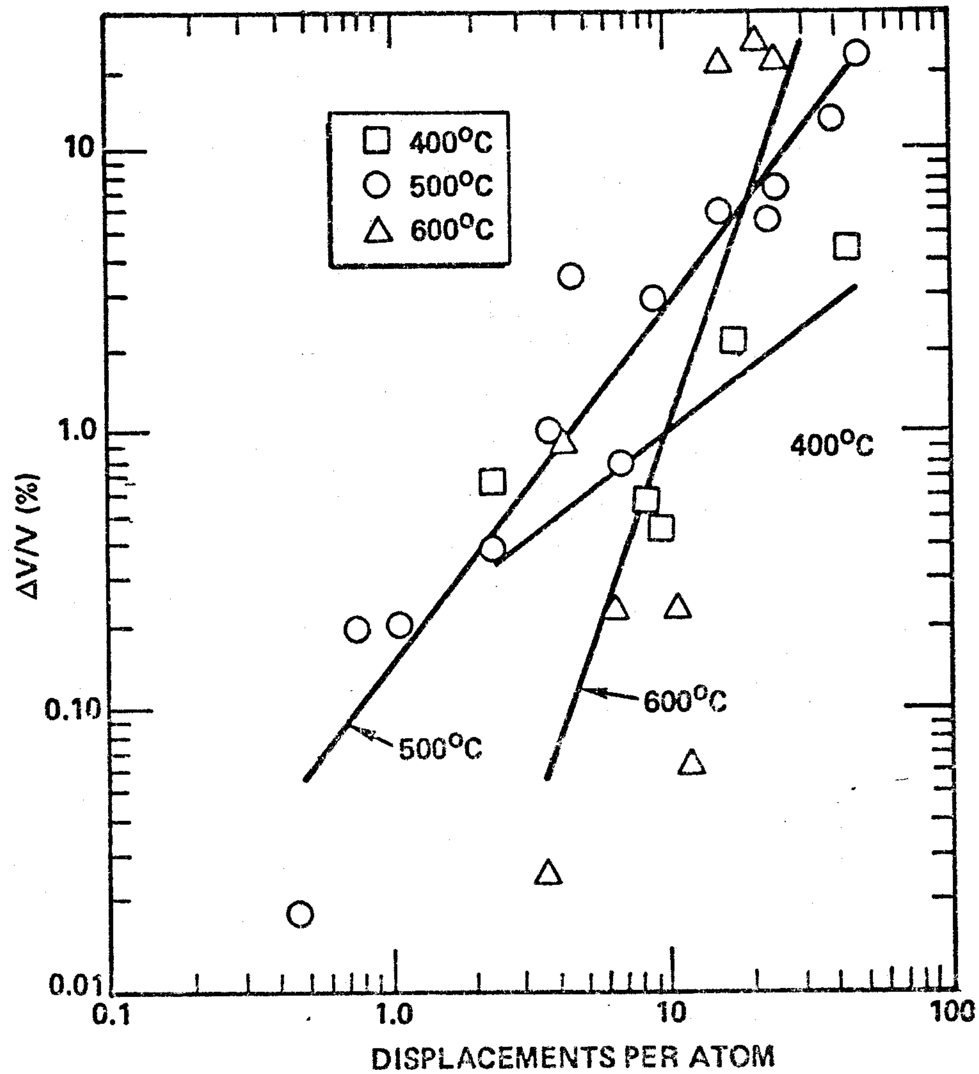
S. D. Harkness, R. Grappel, and S. G. McDonald, "Recent Swelling and Creep Results of the SCIM Code," <u>Quarterly Progress Report, Irradiation Effects on Reactor Structural Materials : February - April, 1972</u>, HEDL-TME 72-64, p. ANL-1.



Log swelling vs. log dose in nickel, 316 stainless steel and Nimonic PE16 at 525°C.

Figure 16

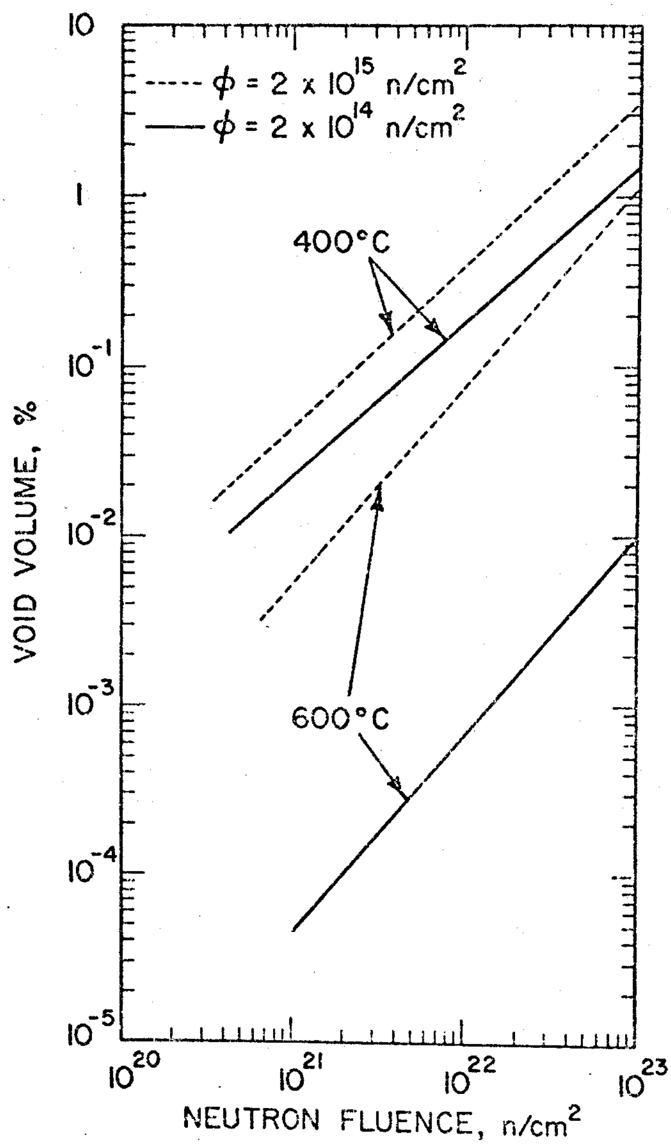
R. S. Nelson, et. al., "Void Formation in Metals During Ion Bombardment," Radiation-Induced Voids in Metals, Conference Proceedings, p. 430, (June 9-11, 1971).



Swelling as a function of damage at 400, 500 and 600°C for Type 316 stainless steel, solution annealed and aged.

Figure 17

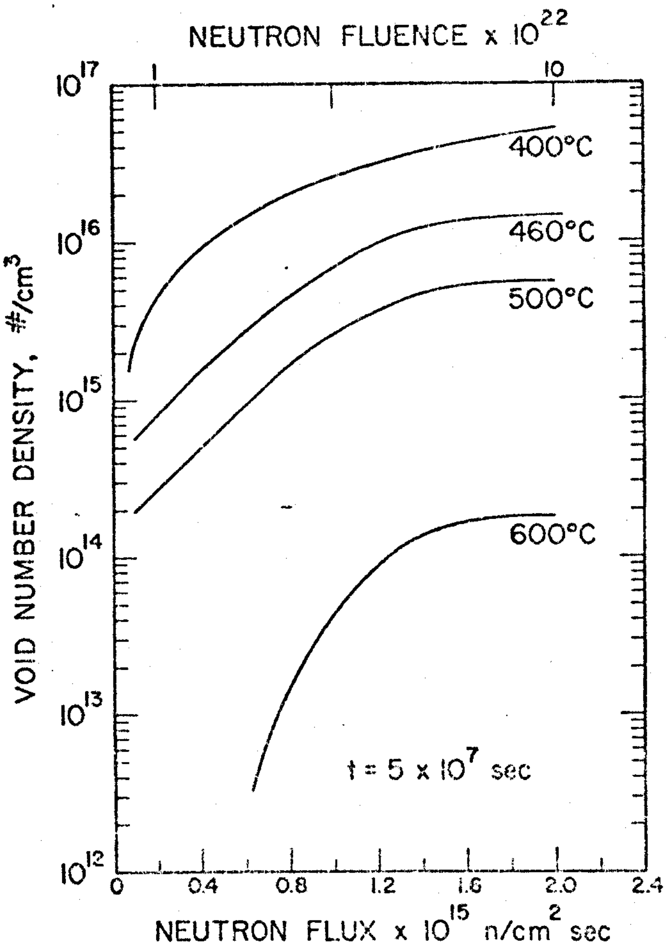
D. W. Keefer, et. al., "Proton Irradiations of Type 316 Stainless Steel," Quarterly Progress Report, Irradiation Effects on Reactor Structural Materials: May-June 1972, HEDL-TME 72-105, p. A1-1.

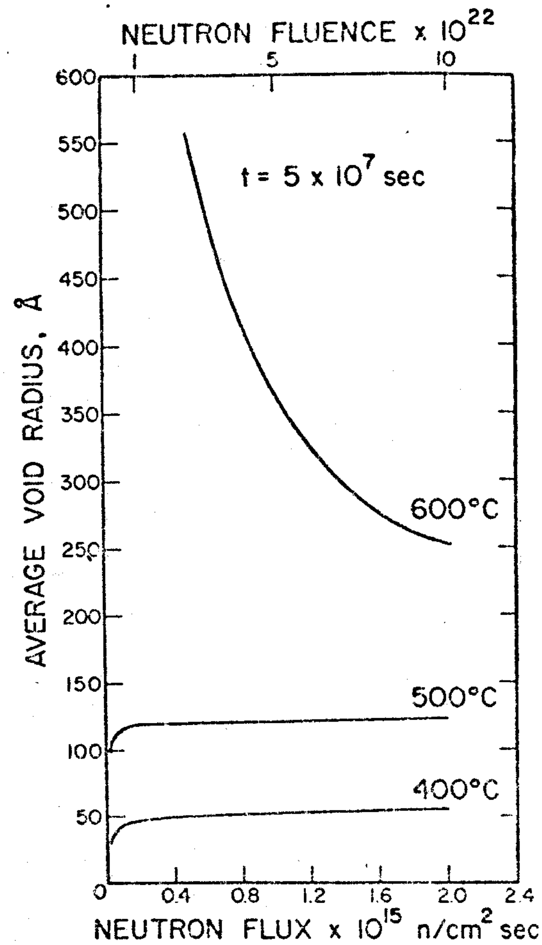


—The predicted effect of flux level on the overall void volume.

Figure 18

S. D. Harkness and Che-Yu Li, "A Study of Void Formation in Fast Neutron-Irradiated Metals," Met. Trans. 2 : 1457, (May 1971).

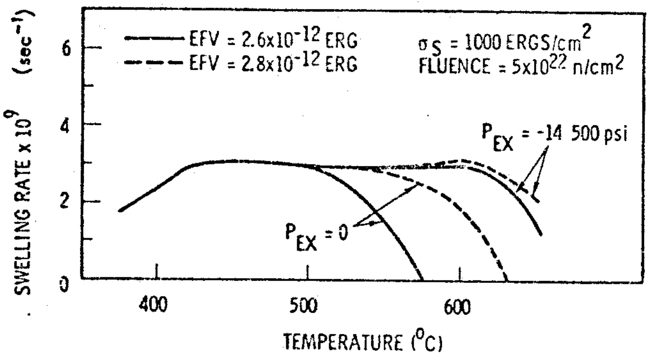




—(*a*) The predicted effect of irradiation at constant time over a range of fluxes on the void number density. (*b*) The predicted effect of irradiation at constant time over a range of fluxes on the average void size.

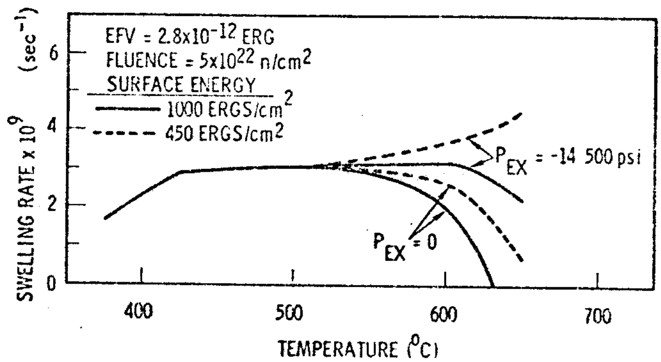
Figure 19 Figure 20

S. D. Harkness and Che-Yu Li, "A Study of Void Formation in Fast Neutron-Irradiated Metals," Met. Trans. 2 : 1457, (May 1971).



Swelling rate vs temperature, using observed microstructure—effect of vacancy formation energy (EFV).

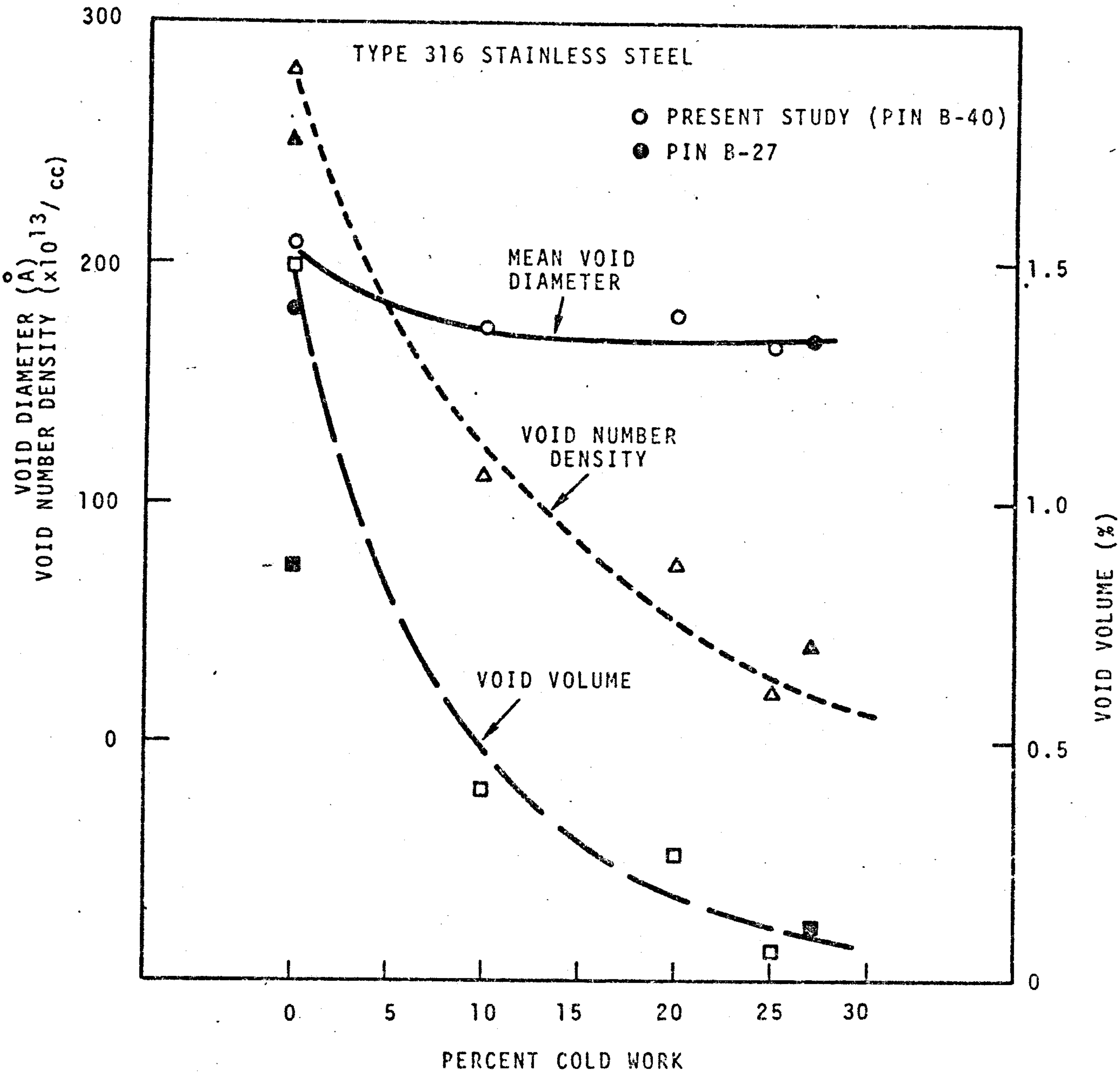
Figure 21



Swelling rate vs temperature, using observed microstructure—effect of surface energy.

Figure 22

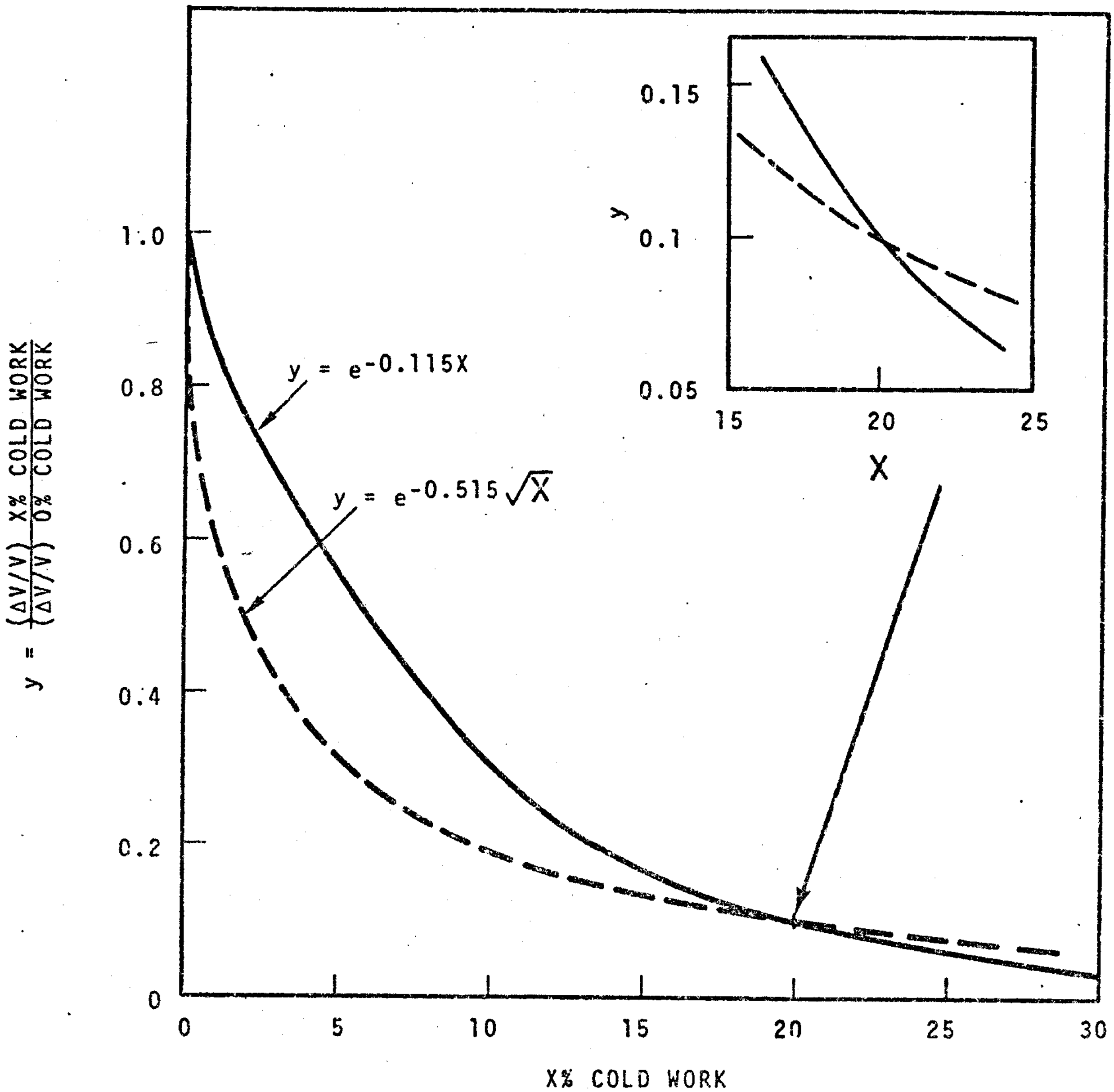
J. L. Straalsund and G. L. Guthrie, "An Analysis of the Effects of Hydrostatic Stress on Swelling," Nucl. Tech. 16 : 36, (Oct. 1972).



Effect of Cold-Work Level on Void Formation.

Figure 23

H. R. Brager and J. L. Straalsund, "Effect of Cold Work Level on Void Formation in Type 316 Stainless Steel," Quarterly Progress Report, Irradiation Effects on Reactor Structural Materials : February - April, 1972, HEDL-TME 72-64, p. HEDL-52.

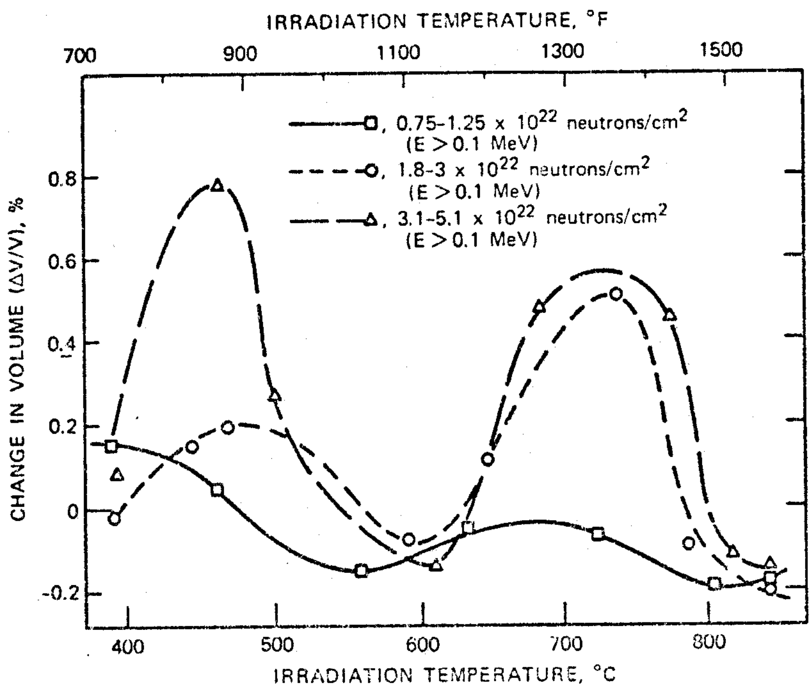


Graphic Relationship between Swelling and Cold-Work Level.

Figure 24

H. R. Brager and J. L. Straalsund, "Effect of Cold Work Level on Void Formation in Type 316 Stainless Steel," Quarterly Progress Report, Irradiation Effects on Reactor Structural Materials : February - April, 1972, HEDL -TME 72-64, p. HEDL-52.

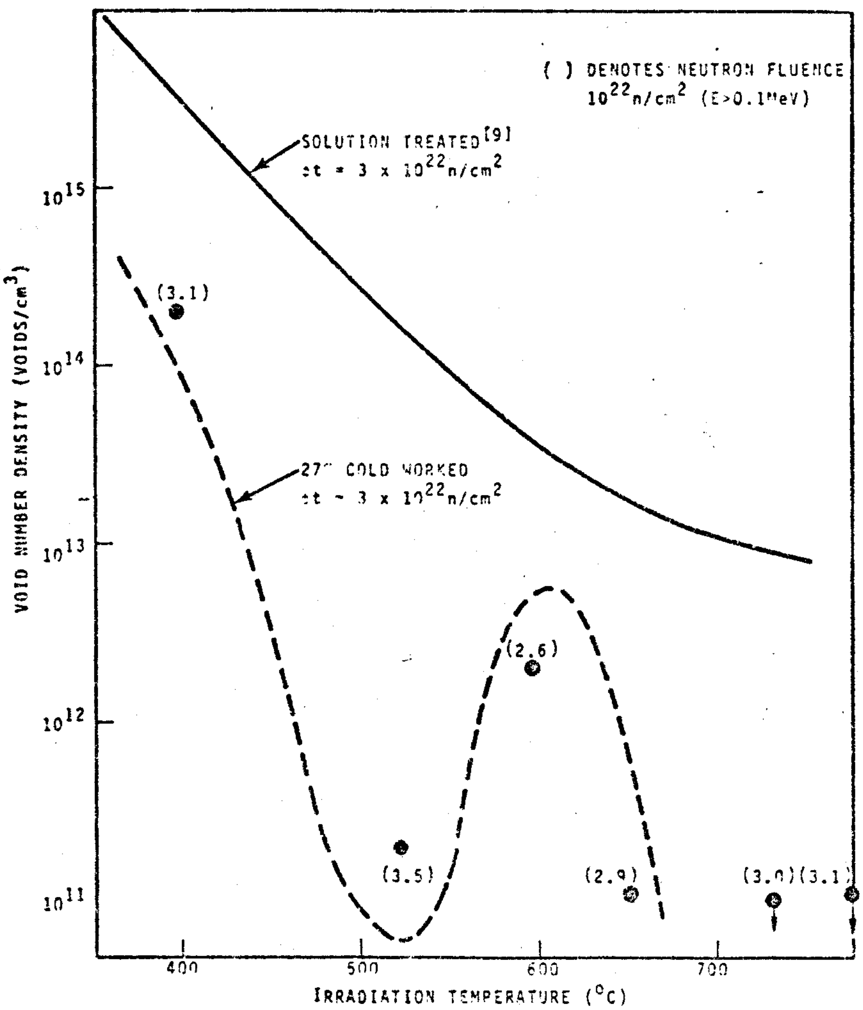
EFFECTS OF COLD WORK ON VOID FORMATION IN STEEL



Temperature dependence of swelling in 50% cold-worked 304 stainless steel.

Figure 25

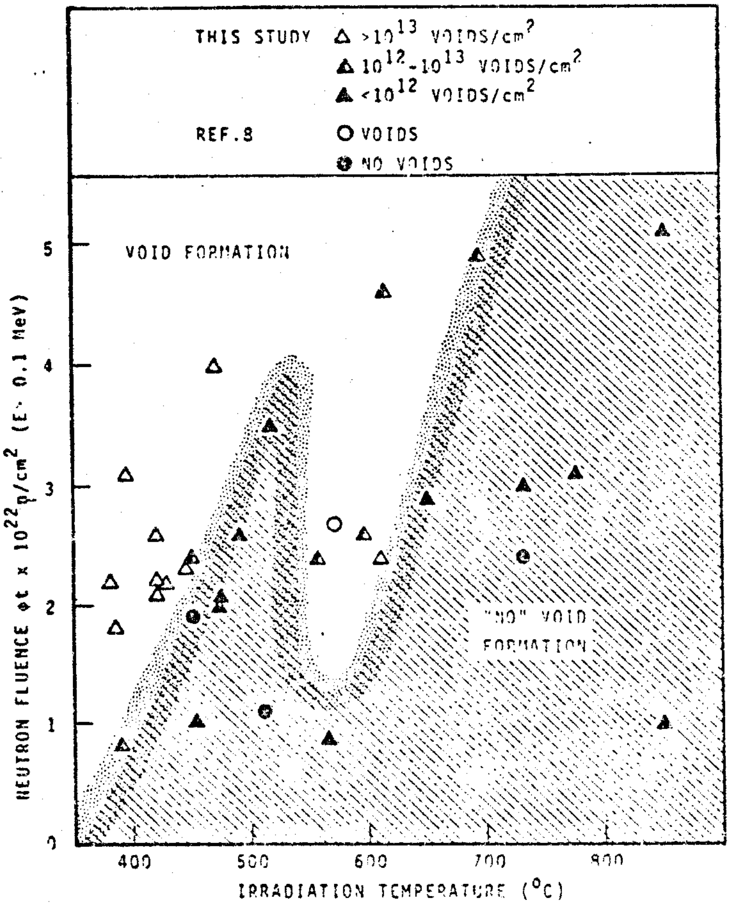
J. L. Straalsund, H. R. Brager, and J. J. Holmes, "Effects of Cold Work on Void Formation in Austenitic Stainless Steel," Radiation-Induced Voids in Metals, Conference Proceedings, p. 142, (June 9-11, 1971).



Void Number Density Data of 27% Cold Worked 316 SS Irradiated to a Nominal Neutron Fluence of 3×10^{22} n/cm^2, Shown as a Function of Irradiation Temperature. The dashed curve, which was fit to this cold worked data, exhibits a markedly different temperature dependence than that for the same steel irradiated in the solution treated condition as shown by the solid line.

Figure 26

H. R. Brager and J. L. Straalsund, "Irradiation Temperature Dependence of Void Formation in Cold Worked Type 316 Stainless Steel," HEDL-72-121, (Oct. 1972).



Irradiation Conditions for Void Formation in 316 SS Cold Worked 20% to 27% by Swaging.

Figure 27

H. R. Brager and J. L. Straalsund, "Irradiation Temperature Dependence of Void Formation in Cold Worked Type 316 Stainless Steel," HEDL-72-121, (Oct. 1972).

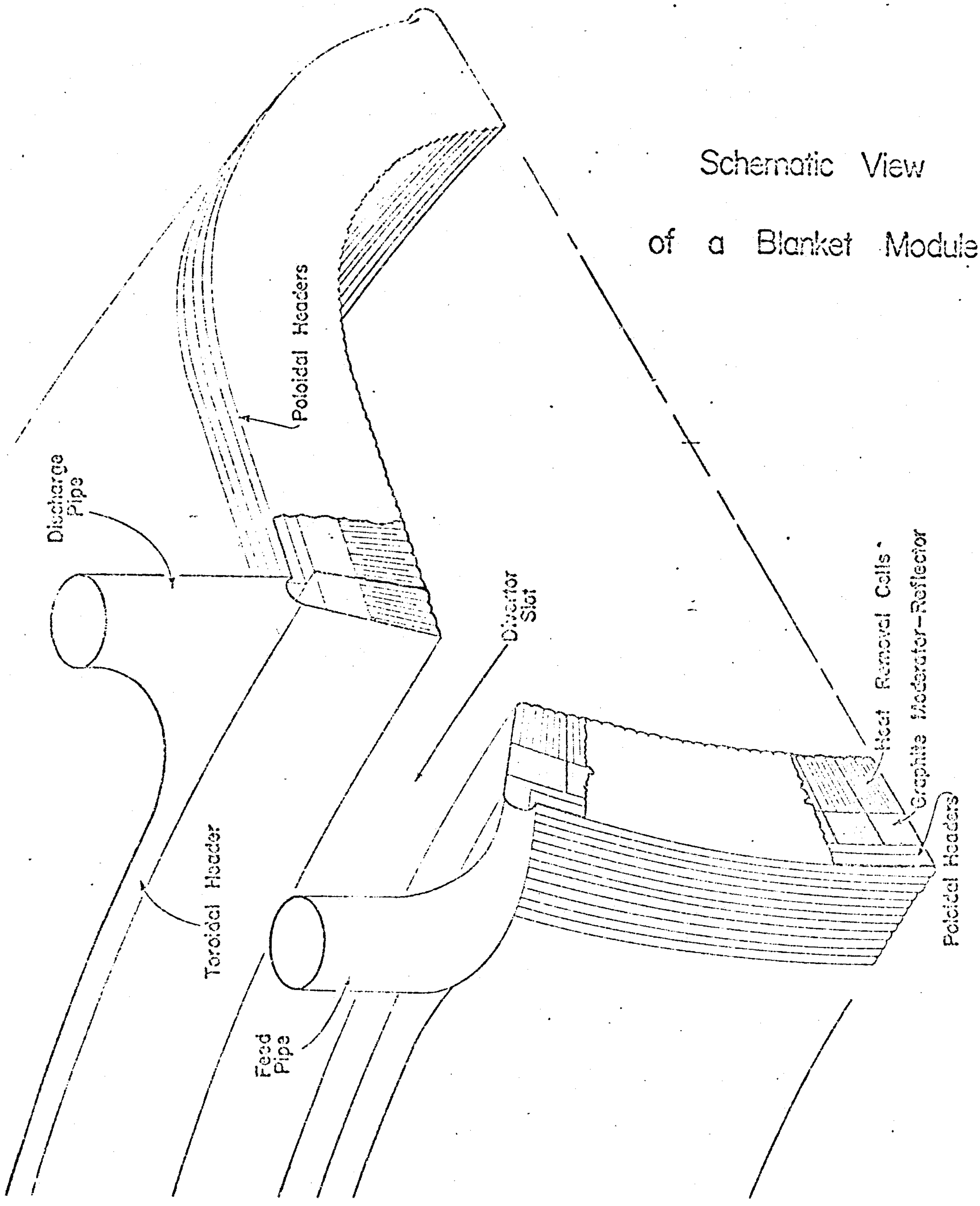
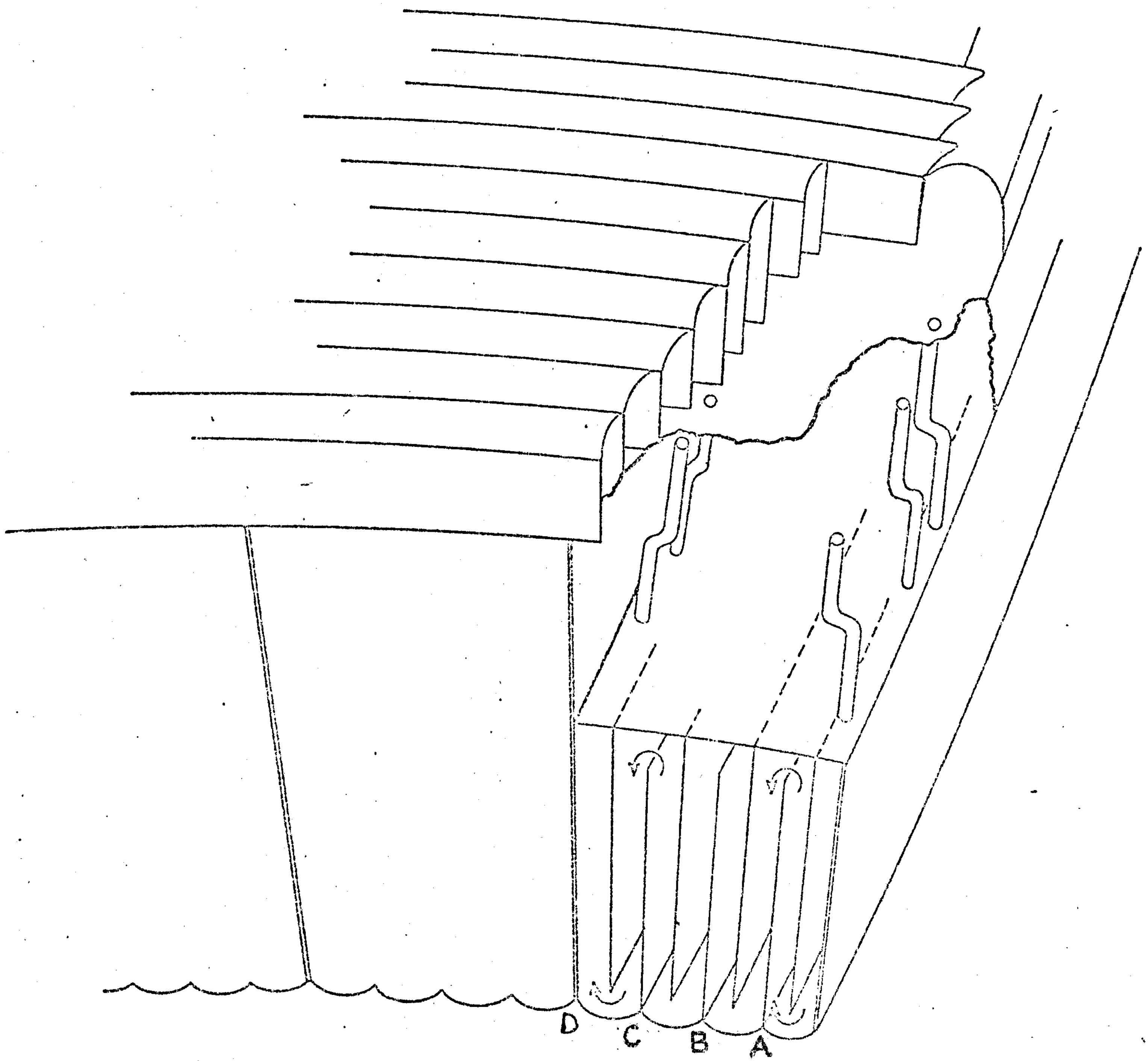


Figure 28

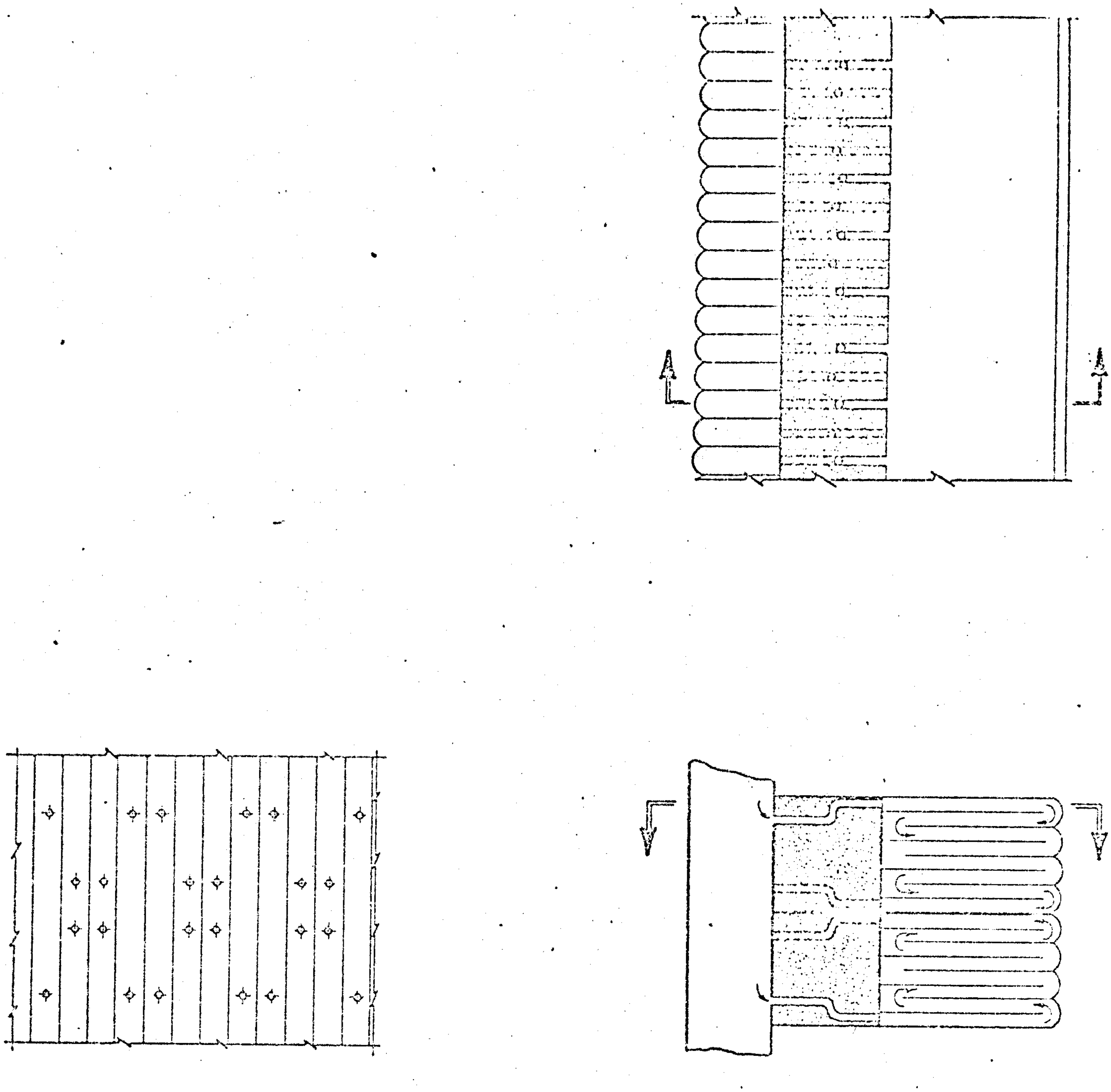
D. K. Sze and W. E. Stewart, "Lithium Cooling for a Low – β Tokamak Reactor," to be published in proceedings of Texas Symposium on the Technology of Controlled Thermonuclear Fusion Experiments and the Engineering Aspects of Fusion Reactors, FDM-32, (Nov. 1972).



Cutaway View of Blanket

Figure 29

D. K. Sze and W. E. Stewart, "Lithium Cooling for a Low - β Tokamak Reactor," to be published in proceedings of Texas Symposium on the Technology of Controlled Thermonuclear Fusion Experiments and the Engineering Aspects of Fusion Reactors, FDM-32, (Nov. 1972).



Section Views of Blanket

Figure 30

D. K. Sze and W. E. Stewart, "Lithium Cooling for a Low $-\beta$ Tokamak Reactor," to be published in proceedings of Texas Symposium on the Technology of Controlled Thermonuclear Fusion Experiments and the Engineering Aspects of Fusion Reactors, FDM-32, (Nov. 1972).