SCIM, A THEORY-BASED COMPUTER CODE FOR THE PREDICTION
OF THE IN-PILE BEHAVIOR OF TYPE 304 STAINLESS STEEL

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ABSTRACT

A model for the behavior of Type 304 stainless steel during fast-reactor irradiation has been developed into the computer program SCIM (Swelling and Creep of Irradiated Metals). The model incorporates recent concepts on high-temperature radiation damage into an analytical tool for predicting in-pile behavior of Type 304 stainless steel.

Swelling rates are discussed in terms of the relative efficiencies of voids and dislocations as sinks. The calculation for the swelling rate shows it is a maximum when the sink efficiencies are equal. The dose dependence of swelling is found to be a function of the relative rates of void and dislocation loop formation. Saturation mechanisms are discussed with respect to their effect on the swelling rate. Saturation is favored by increased void number density and increased irradiation temperature. This results in a compromise between irradiation temperature and the expected void volume at saturation.

Cold work is expected to be increasingly effective with increasing irradiation temperature. At 372°C the dose dependence of swelling in cold-worked material is expected to be much higher than for solution-annealed material because of the rapidly changing relative effectiveness of voids and dislocations as point-defect sinks.

High number densities of incoherent precipitate should limit swelling at intermediate irradiation temperatures by forming a saturation microstructure at low void volumes.

A climb-controlled in-pile creep mechanism is discussed. The expression that results depend on the radiation-produced excess interstitial flux to glissile dislocations as the mechanism for enhanced in-pile creep. The glissile dislocations are created by the unfauling of irradiation-produced interstitial dislocation loops. The principal obstacle is taken as dislocations pinned by the void structure. The maximum in-pile creep rate is expected to occur at nearly the same temperature at which swelling is a maximum. The creep rate is expected to slowly decrease with dose.

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The modeling of void formation in Type 304 stainless steel irradiated under fast-reactor conditions has been the subject of several recent papers.¹⁻³⁾ It is the purpose of this paper to summarize the most recent results of the ANL radiation-damage modeling program. The results of the program have been developed into a computer routine called SCIM (Swelling and Creep of Irradiated Metals), which is included as a subroutine in the LIFE-II fuel-element modeling code.

The general nature of the SCIM computer routine has been described fully elsewhere.⁽¹⁾ The concentration of vacancies and interstitials is derived from the solutions of quasi-steady-state equations that balance the point-defect generation rate with the annihilation rate of point defects at sinks, such as voids, dislocation loops, grain boundaries and precipitate particles, and the annihilation rate of point defects by mutual recombination events. These equations are

and
$$G_v - Q_{v,sink} D_v C_v - R = 0 \quad (1a)$$

$$G_i - Q_{i,sink} D_i C_i - R = 0, \quad (1b)$$

where -

G_v, G_i = effective generation rate of vacancies;

interstitials into matrix,

$Q_{v,sink}, Q_{i,sink}$ = efficiency of sinks for vacancies and interstitials,

C_v, C_i = concentration of vacancies and interstitials

present during quasi-steady-state conditions,

and

D_v, D_i = vacancy and interstitial diffusion coefficients.

The evolution of the radiation damage is followed by assuming steady-state conditions for short periods. The void and dislocation nucleation rates are calculated in terms of the quasi-steady-state vacancy and interstitial concentrations for a given time interval.

Currently, the nucleation of voids is assumed to be the sum of a homogeneous term and an empirical, heterogeneous expression. The homogeneous expression was developed by Russell⁽²⁾ and by Katz and Wiedersich.⁽³⁾ The empirical expression used for nucleation on gas atoms follows a similar form and is written as

$$I_{He} = N_{He} (\beta_v - \beta_i)_{crit} \exp \left[\frac{16 \pi \gamma^3}{3 K T} \left(P + R T \ln \left(\frac{\beta_v - \beta_i}{\beta_{vo}} \right) \right)^2 \right],$$

where

N_{He} = the number of free helium atoms in cm^3 ,

β_v, β_{vo} = flux of vacancies to a void in terms of the radiation-induced concentration, and the thermal concentration,

and

β_i = the flux of the interstitials to a void.

The value of the surface energy γ was selected as 1125 ergs/cm^2 to give good agreement with ion-bombardment results, where all nucleation is assumed to be homogeneous, and with the 370°C neutron data.

P , the internal gas pressure in a critical size void, was used as a fitting parameter to gain agreement with the number densities observed in

higher temperature EBR-II experiments. A value of 7.5×10^{10} dynes/cm² gave the best fit.

The dislocation loop nucleation expressions were similarly derived as the sum of a homogeneous and a heterogeneous term. The form of the energy barrier used is presented elsewhere.⁽¹⁾ These expressions can only be treated, at this time, as empirical fits to observed dislocation density results. The growth of voids is computed in terms of the difference in the vacancy and interstitial flux to voids of a given size. The calculated excess of from 1 to 0.1 % of the total vacancy flux that arrives at a void is the direct result of including an expression for the preferential attraction of interstitials to dislocations through the interaction of the dislocation stress field with the misfit strain of the interstitial. As discussed later, the excess interstitial flux to dislocations may well serve as the basis for an enhanced dislocation climb velocity and thereby a mechanism for radiation-enhanced creep.

Swelling Rates

The accumulation rate of excess vacancies is a parameter of key importance. This may be written as

$$\frac{dv}{dt} = Q_{v,void} \left(D_v (C_v - C_{v,T}) - D_i C_i \right), \quad (2)$$

where

$Q_{v,void}$ = the efficiency of voids as sinks for
vacancies or interstitials,

$C_{v,T}$ = thermal vacancy concentration in equilibrium
with a void, $C_{v,T} = C_{v,o} \exp\left(-\frac{2\gamma\Omega}{rKT}\right)$,

$C_{v,o}$ = thermal equilibrium vacancy concentration,

γ = surface energy,

r = void radius,

and

Ω = atomic volume.

Assuming recombination is not important and solving Eqs. 1(a) and 1(b) for the case where voids and dislocations are the only important sinks in the system, one obtains

$$C_v = \frac{G_v}{D_v(Q_{v,void} + Q_{v,dis})},$$

and

$$C_i = \frac{G_i}{D_i(Q_{i,void} + Q_{i,dis})}.$$

By realizing that since the void is a nonpreferential sink

$$Q_{v,void} = Q_{i,void} = 4\pi N_v R_v,$$

where N_v is the number of voids/cm³, and R_v is the average void radius,

and at higher temperatures

$$G_v = G_i = G,$$

and that the efficiency of dislocations for capturing interstitials is only slightly greater than for capturing vacancies so that

$$Q_{i,dis} = (1 + \epsilon) Q_{v,dis} = (1 + \epsilon) \frac{2\pi\rho}{\ln(S_s/R_c)},$$

where

ρ = dislocation density,

S_s = sink spacing,

R_c = dislocation core radius,

and

$$\epsilon \text{ is } \ll 1.$$

Substitution of these results in Eq. (2) yields

$$\frac{dv}{dt} = \frac{Q_{v,void} G \epsilon Q_{v,dis}}{Q_{v,void} + Q_{v,dis}} + \left[F(Q_{v,void}, Q_{v,dis}, T) - D_v C_{v,T} Q_{v,void} \right]$$

where

$$F = \frac{Q_{v,void} (Q_{v,void} C_{v,T} D_v + Q_{v,dis} D_v C_{v,dis})}{Q_{v,void} + Q_{v,dis}}$$

An estimate of the magnitudes of the two terms may be obtained by writing

$$Q_{v,dis} = M Q_{v,void}$$

where usually M lies between values of $1/2$ and 30 . Thus the term in the bracket has a value of

$$Q_{v,void} D_v \left(\frac{M C_{v,dis}}{1 + M} - \left(1 - \frac{1}{1+M} \right) C_{v,T} \right)$$

For $vM = 1$

$$= \frac{Q_{v,void} D_v}{2} (C_{v,dis} - C_{v,T})$$

For $M = 30$

$$F \approx Q_{v,void} D_v (C_{v,dis} - C_{v,T}).$$

In an unstressed solid

$$C_{v,dis} \approx C_{v,o}$$

while

$$C_{v,T} = 1.5 - 3 C_{v,o}.$$

Thus for purposes of discussion one can write the overall swelling rate as

$$\frac{dv}{dt} = \frac{M G \epsilon}{(1 + M)^2} - \alpha Q_{v,void} D_v C_{v,o}$$

and α typically lies between .5 and 2.

During most of the irradiation the term containing the equilibrium vacancy concentration will be small $\frac{M}{M+1}$ or so that the swelling rate will be a maximum when $Q_{v,void} = Q_{v,dis}$. During irradiation, the efficiencies of both dislocations and voids/will increase. If they increase at the same rate, the swelling rate will remain constant, and a linear relationship between swelling and time would be expected. If the relative efficiencies change, however, the swelling rate would be expected to change with time, and swelling would be expected to proceed as dose raised to some power. In practice, the dislocation and void structure usually evolve simultaneously such that dose exponents near unity are expected. The calculated ratio of the dislocation efficiency to the void efficiency, in terms of the computer model for Type 304 stainless steel at two doses as a function of temperature, is presented in Fig. 1. The value of the ratio nearest unity should correspond

to the highest swelling rates, whereas the largest relative change between ratio values at the two doses should indicate the highest dose dependence. For example, the results indicate the highest swelling rate at these doses should occur near 530°C. The dose dependence according to these calculations would be expected to decrease slightly at all temperatures within this dose range. This example of the sensitivity of swelling to the relative numbers of dislocations and voids present illustrates one of the difficulties encountered in both modeling and characterizing swelling results. The dislocation loop nucleation rate is one of the least understood parameters involved in the radiation-damage process. It is clear, however, from the previous discussion that the exact temperature, flux and dose dependence of the dislocation loop nucleation rate must be known to characterize fully the swelling rate. To illustrate this point, Fig. 2 shows the expected swelling behavior of Type 304 stainless steel assuming two dislocation loop nucleation rates as the result of changing the interstitial formation energies. The higher interstitial energy of formation results in a higher dislocation density and thus a higher ratio value. Judging from this result, a contributing factor to the differences in swelling behavior between alloys may be due to differences in the dislocation structures that are developed.

It is instructive to consider possible saturation in swelling in terms of Eq. (3). Swelling will cease when

$$\frac{M G \epsilon}{(1 + M)^2} = \alpha Q_{v, \text{void}} D_v C_{v, 0}$$

The left-hand expression can decrease either by changing the relative efficiencies of the void and dislocations as sinks so that they are less equal or by reducing the biasing of interstitials to dislocations, thereby reducing ϵ .

The first case could conceivably occur if the dislocation structure buildup at a faster rate than the void structure. A decreasing swelling rate would thus be expected, leading eventually to a saturation in the void volume. This process might be important at low irradiation temperatures where the dislocation structure cannot recover as easily through glide and annihilation processes. Of course, the magnitude of $D_v C_{v,0}$ will decrease sharply as the temperature decreases, which will tend to minimize the effect.

At higher temperatures where $D_v C_{v,0}$ values are larger, rapid dislocation buildup seems unlikely because at higher temperatures fewer dislocation loops are nucleated, which results in individual loops growing larger. As the loops become larger, the probability they will unfault to form a glissile prismatic dislocation loop increases. Recovery processes can now occur among the glissile dislocations thereby limiting, to some extent, the buildup of the dislocation density. In the SCIM code, recovery is permitted through an empirical expression that limits the dislocation density as a function of temperature.

Glissile dislocations will also tend to intersect and be pinned by the void structure. As discussed in detail elsewhere⁽⁴⁾ the intersection of the void structure by dislocations allows a mechanism that effectively reduces the number of biased point defects and thus, in terms of Eq. (2), ϵ .

This occurs by the pipe-diffusion movement of vacancies from the void down the dislocation core to recombine with the biased interstitials arriving at the dislocation. A central theme to this argument⁽³⁾ is that the dislocation configuration between two voids is essentially what it would be at thermal equilibrium because during irradiation the vacancy flux from the voids can be much larger than the interstitial flux from the grain matrix to the dislocation. Thus, as excess interstitials arrive at the dislocation, which cause it to climb out of the thermal equilibrium configuration, vacancies will be emitted from the void to recombine with the excess interstitial and readjust the dislocation position to thermal equilibrium. A vacancy moving along a dislocation core has a finite probability of reentering the matrix of the material. This means that the average distance between voids must be less than some finite amount. Calculations of this distance are not precise, but typical values range between 1000 and 5000 Å. For the purposes of modeling, the bias was reduced by the fraction $\lambda/(1/2 S_v)$, where λ is assumed to be the longest distance a vacancy can move along a dislocation core, taken as 3000 Å. S_v is the average mean free spacing between voids and is defined as

$$S_v = \frac{1}{4\pi N_v (R_v)^2},$$

where N_v is the number of voids/cm³, and R_v is the average void radius.

The incorporation of this saturation expression in the SCIM code leads to the swelling profiles for Type 304 stainless steel presented in Fig. 3. The saturation mechanism is only allowed to operate in the code when the radius of the average dislocation loop is greater than 500 Å.

It is assumed that all loops less than this size are sessile and thus cannot glide to interact with the void structure. The 372°C results reflect this assumption because no saturation is indicated since loop number densities are high and average sizes small at this temperature. The overall swelling rate is low due to the high value for the ratio of dislocation-to-void sink efficiencies, as presented in Fig. 1. At 427°C, saturation is predicted at about 11% swelling as compared with higher values at 482 and 538°C. This is due to the much higher number density of voids formed at 427°C than at 482 or 538°C which results in smaller void spacings S_v and thus earlier saturation. The fact that saturation void volume is lower at 427 and 538°C than at 482°C, where the void number density is higher, is due to the increased value of the $D_v C_{v,T}$ term in Eq. (2) at the higher temperatures. In agreement with Fig. 1, the initial swelling rate is highest at 538°C until saturation mechanisms become effective. Of course, the exact level of saturation is a function of the imprecisely known critical spacing. For example, reducing the critical spacing from 300 to 1500 Å increases the void volume from 11 to 28% at 427°C, from 23 to 44% at 482°C, and from 7 to 20% at 538°C. Exact values of the spacing must await experiment.

Predictions on the Effects of Cold Work and Precipitate

The effect of cold work was incorporated into the model by assuming that

1. No homogeneous nucleation could occur within 50 Å of the dislocation core because of the low vacancy supersaturation in this region.

2. Helium atoms within 50 Å of a dislocation are trapped and not available for nucleation.
3. Dislocation density is related to cold-work amounts up to 20% by the expression

$$\rho = 10^8 \exp(0.45 * \text{PERCW}),$$

where PERCW is the percentage of cold work. This expression is a rough fit of Stickler's⁽⁵⁾ results of dislocation densities in cold-worked Type stainless steel.

4. The dislocation structure is uniform throughout the material.

The effect of a high initial dislocation density is to reduce void nucleation, thereby leading to the overwhelming predominance of dislocations as sinks for both types of point defects. The model predictions for the effect of cold work at 372°C are particularly interesting to analyze (Fig. 4). The values for void volume, void size, void number density, and sink effectiveness ratio are identical for solution-annealed, 5 and 10% cold-worked material. The 20% material, however, is predicted to show no swelling until neutron doses in excess of 7×10^{22} n/cm² are attained. At this point, the helium concentration has accumulated sufficiently to allow void nucleation. In contrast to the solution-annealed case, all voids formed in the 20% cold-worked material are gas-atom rather than homogeneously nucleated. The dose dependence of the swelling in the 20% material is much higher than linear because the sink efficiency ratio decreases rapidly after void nucleation occurs.

At the highest dose modeled (2×10^{23} n/cm², $E > 0.1$ MeV), the sink efficiency ratio is less than for the solution-annealed material, which indicates the possibility that the cold-worked material could eventually swell more than the solution annealed material.

The predicted swelling results at 482°C (Fig. 5) are somewhat less interesting. No effect on swelling is predicted for cold-work percentages of 10 or less; however, the 20% cold-worked material never exhibits swelling.

Figure 6 presents the effect of cold work at two fluences as a function of irradiation temperature. The shift of the peak swelling temperature from 530 to 480°C in going from 3×10^{22} to 2.2×10^{23} n/cm², $E > 0.1$ MeV is due to the onset of saturation mechanisms at the higher temperature.

The effect of a high density of incoherent precipitate on void formation is somewhat complicated. Combining incoherent precipitates in a microstructure with dislocations results in an excess flux of vacancies to the precipitate. Thus, if a small vacancy supersaturation exists in the system, voids will form on the precipitate. Unlike other types of sinks (grain boundaries or dislocations), increasing the sink density does not necessarily restrict void nucleation. In fact, precipitates should extend void formation to higher temperatures by allowing voids to form at lower supersaturations than would otherwise be possible. The principal means by which incoherent precipitate structure could limit swelling is by formation of a saturation microstructure at low void volume. That

is, if the dislocation structure threads among precipitate particles, excess vacancies would be expected to recombine with biased interstitials on the dislocation in the same manner as described in the section on saturation. For this reason, the SCIM program predicts that the swelling will be sharply reduced between about 435 and 590°C when $10^{15}/\text{cm}^3$ of 200 Å precipitate particles are added. At lower temperatures, the dislocation loop structure is not expected to unfault to form glissile dislocations that can interact with the precipitate particles. At higher temperatures, the swelling exceeds that of solution-annealed material because of the formation of voids on precipitates at temperatures where voids otherwise could not form.

Comparison of Model Predictions with EBR-II Results

A comparison with experimental results is the only true test for any model. In general, such comparisons first consider functional dependencies and later quantitative agreements. The ability of the SCIM codes to predict the correct temperature dependencies for swelling, decreasing number density and increasing size with increasing temperatures has been well established. The dose dependencies predicted by the code are nearly linear, whereas current design equations use a relationship in which swelling is proportional to dose raised to the 1.5 power. However, recent work⁽⁷⁾ has indicated much of the current swelling data can easily be plotted linearly, if a small initial incubation period is allowed. Other general trends, such as the increasing effectiveness of cold work in limiting swelling with increasing irradiation temperature, have also been substantiated. Any deviations from this trend

could reasonably be attributed to local recovery of the dislocation structure. The saturation behavior has not been demonstrated experimentally. Samples are only beginning to attain doses high enough to observe saturation.

Quantitative comparisons of SCIM code predictions with EBR-II results are difficult because of the combination of poorly known fluxes and temperatures during the irradiation history of any particular experiment. Furthermore, most of the high dose data have come from structural components that have experienced quite a variety of reactor conditions. An example of how different irradiation histories might affect the void microstructure is presented in Table 1. Each power history, as defined in Table 2, yields the same average values for flux and temperature. The differences allowed in both flux and temperature are quite small, but substantial differences in the expected irradiated microstructure result. Much larger discrepancies must be expected when trying to model actual reactor experiments. A comparison between measured and predicted void sizes, number densities, and volumes is presented in Table 3. In general, the agreement seems quite good.

In-Pile Mechanical Properties

The SCIM code computes the yield strength and minimum creep rate expected in-pile. These values are calculated in terms of the microstructure developed during the irradiation. The yield stress is calculated from an expression discussed previously⁽⁸⁾

$$\Delta\tau_{ys} = (\Delta\tau_{void}^2 + \Delta\tau_{loop}^2)^{1/2},$$

where

$\Delta\tau_{\text{void}}$ = increment in critical resolved shear
stress due to voids. (see Ref. 8).

$\Delta\tau_{\text{loop}}$ = increment due to dislocation loops
(see Ref. 8).

The in-pile creep properties are a function of both the irradiated microstructure and the excess point-defect concentrations generated by the irradiation. The physical picture modeled is the climb of a glissile dislocation over a dislocation pinned by the voids. Only strain accumulated by glide is considered in this model.

Following Cottrell's treatment,⁽⁹⁾ the distance between the leading dislocation and the second dislocation in a pile-up may be written for an edge dislocation as:

$$d = 1.84 \mu b / (2\pi (1-\nu) n \tau) \quad (4)$$

where

μ = shear modulus,

ν = Poisson's ratio,

and

τ = resolved shear stress on the glide
plane.

The term n is the total number of piled-up dislocations that exert a stress on the leading dislocation. For the case of a microstructure with dispersed dislocation pinning points, such as voids, the appropriate value of n is the number of piled-up dislocations that occur within the average spacing between voids. This may be written as

$$n = \pi \tau (1-\nu) \lambda / \mu b,$$

which leads to a value of

$$d = \frac{0.92}{\lambda} \frac{\mu}{\pi \tau (1-\nu)} \quad 2$$

If pile-ups are not considered, as may be appropriate for high dislocation density structures with strongly overlapping stress fields, the stress dependency of the expected strain rate becomes strictly linear.

To bypass the pinned dislocation, the second dislocation must climb to a position where the force $\tau b L$, exerted by the applied stress is equal to the back stress of the pinned dislocation. L is defined as the effective free length of the dislocation and is taken as equal to the average dislocation loop time length. The bypass condition can be written as

$$\tau b L = \frac{\mu b^2}{2\pi(1-\nu)} \frac{x(x^2 - y^2)}{(x^2 + y^2)^2} \quad (5)$$

Thus, initially, the second dislocation is a distance d from the pinned dislocation. As it climbs, it can move forward according to Eq. (5).

Solution of Eq.(5) shows that y goes through a maximum with decreasing x . This maximum value of y ($y_{\max}$) defines the effective obstacle height of

the pinned dislocation. Higher stresses cause the climbing dislocation to start nearer the pinned dislocation (lower values of d), which results in lower climb distances to bypass the obstacle and thereby higher creep rates. Of course, in some circumstances the required climb distance is less than the void radius and the bypassing dislocation will intersect the void. Subsequent dislocations will then be forced to climb the calculated distance to overcome the pinned dislocations.

The rate of climb of the dislocation is a function of the difference in the interstitial and vacancy fluxes reaching the dislocation. An excess interstitial flux is expected due to the previously discussed preferential attraction of interstitials by dislocations. The climb rate R in cm/sec is thus

$$R = (J_{i,dis} - J_{v,dis}) b^2,$$

where $J_{i,dis}$ and $J_{v,dis}$ are the interstitial and vacancy fluxes, respectively arriving at a dislocation line per centimeter per second.

The values for the fluxes are calculated in the SCIM code as a function of irradiation history. The rate of overcoming the pinned dislocation thus becomes, on the average,

$$R/y_{max}.$$

The total strain associated with the climbing dislocation density, ρ_{mobile} , overcoming an obstacle is $\rho_{mobile} \lambda b$. The expression for the in-reactor creep rate thus becomes

$$\dot{\epsilon} = \rho_{mobile} (J_{i,dis} - J_{v,dis}) b^3 \lambda / y_{max}.$$

When comparing this result with experiment, the most difficult aspect was to define what fraction of the total dislocation density was mobile. The total dislocation density during irradiation increases because of the creation of interstitial dislocation loops. The assumption that one-half (average dislocation loop radius/500 Å) of the total dislocation density was mobile agreed with Walter's⁽⁹⁾ results and reflected the physical consideration that loops had to reach a size of about 500 Å to unfault and become glissile. Of course, when the average loop size exceeds 500 Å, the mobile dislocation density and the total are taken as identical.

A comparison of the present formulation with Walter's results⁽⁹⁾ is shown in Fig. 7 as a function of stress. At lower stresses, the formulation shows a near linear stress dependence whereas at high stresses it approaches a γ^2 dependence. An orientation factor of 3.1 was used to convert hoop stresses to resolved shear stresses.

Figure 8 presents the expected in-reactor creep rates as a function of dose and irradiation temperature. The rather large increase in creep rate obtained in going from 372 to 427°C is due to the increased fraction of the dislocation density, which is assumed mobile, and to an increased dislocation climb rate. The dislocation climb rate increases for the same reason the swelling rate increases in going between these two temperatures. The climb rate is a function of the excess interstitial flux that arrives at a dislocation. This term will be a maximum when the void and dislocation sink efficiencies are equal. Therefore, as shown in Fig. 1, the climb rate and swelling rates are both expected to reach a maximum near 530°C. At high doses, when saturation mechanisms become operative, both swelling

and enhanced creep must be expected to saturate under this model. The general reduction in creep rate with dose is a result of the increasing sink density with irradiation time. The odd behavior at 372°C is due to a balancing of the decreasing climb rate with an increasing fraction of dislocation loops that have grown large enough to become mobile.

TABLE 1. A Comparison of Predicted Swelling after Different Power Histories

Irradiation History	Average Temperature, °C	Average Neutron Flux, 10^{15} n/cm ² · sec ($\epsilon > 0.1$ MeV)	Fluence, 10^{22} n/cm ² > 0.1 MeV	Average Void Diameter, Å	Void Number Density, 10^{15} /cm ³	Void Volume, %
A	384	1.28	8.2	182.	10.8	3.5
B	384	1.28	8.2	190	8.7	3.3
C	384	1.28	8.2	152	22.3	4.2
A	417	1.32	8.4	262	4.7	7.0
D	417	1.32	8.4	260	5.0	11.7
C	417	1.32	8.4	264	5.2	7.4

TABLE 2. Definition of Power Histories

Irradiation History	Flux during Equal Time Intervals, $10^{15} \text{ n/cm}^2 (\epsilon > 0.1 \text{ MeV})$			Temperature during Equal Time Intervals, $^{\circ}\text{C}$		
	1	2	3	1	2	3
A	1.2	1.3	1.4	-10 $^{\circ}\text{C}$	Ave.	+10 $^{\circ}\text{C}$
B	1.2	1.3	1.4	Ave.	Ave.	Ave.
C	1.3	1.3	1.3	Ave.	Ave.	Ave.
D	1.2	1.3	1.4	+10	Ave.	-10 $^{\circ}\text{C}$

TABLE 3. A Comparison of Experimental and Predicted Void Size Distributions

Irradiation Temperature, °C	Fluence, 10^{22} n/cm ² (E > 0.1 MeV)	Neutron Flux, 10^{15} n/cm ² ·sec (E > 0.1 MeV)	Average Void Diameter, Å	Predicted Average Void Diameter, Å	Void Number Density, 10^{15} /cm ³	Predicted Void Number Density, 10^{15} /cm ³	Immersion Density Change, %	Predicted Void Volume Change	Ref.
370	0.4	0.11	105	110	1.5	4.4		.04	11
370	0.6	0.16	110	114	2.0	1.1		.07	11
370	0.9	0.34	125	112	0.7	2.0		.13	11
370	1.5	0.56	145	114	1.9	3.6		.25	11
371	5.4	0.85	94	124	20.0	13.5	1.2	1.4	12
371	4.2	0.66	94	128	20.0	8.3	1.3	0.9	12
372	8.8	1.38	103	118	20.0	35.6	2.0	3.2	12
374	6.8	1.07	134	128	10.0	20.5	2.4	2.3	12
375	5.6	1.52	145	120	5.0	23.9		2.3	12
384	8.2	1.28	135	152	24.0	22.3	4.1	4.2	12
385	3.4	1.27	150	144	6.5	9.9		1.5	11
386	10.5	1.65	175	152	9.0	36.0	3.8	6.8	12
390	3.4	1.56	210	150	2.3	10.1		1.8	11
400	5.9	1.60	200	195	6.0	9.7		3.9	11
401	8.5	1.33	203	214	13.0	10.5	6.5	5.7	12
405	11.0	1.73	198	222	15.0	13.9	8.0	8.8	12
417	8.4	1.32	237	264	9.2	5.2	7.4	7.4	12
420	2.4	0.90	200	202	2.0	2.6		1.6	12
420	10.00	1.20	215	280	16.0	5.9	10.6	10.3	12
425	2.2	0.82	210	214	1.7	2.1		1.6	11
430	3.5	0.95	190	254	4.8	2.2		3.4	11
435	6.6	1.03	217	320	8.2	2.2	5.3	7.3	11
460	0.5	0.18	245	300	0.3	.13			11

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Figure 3. Change in Relative Efficiencies of Voids and Dislocations as Sinks as a Function of Irradiation Temperature.
Fig. No. MSD-56771.

Figure 2. Effect of Changing Dislocation Loop Nucleation Rate on Swelling Rate of Type 304 Stainless Steel. Neg. No. MSD-56772.

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Figure 4. Effect of Cold Work on Void Formation at 372°C . (a) Relative

Figure 4. Effect of Cold Work on Void Formation at 372°C. (a) Relative Sink Efficiencies, (b) Void Volume, (c) Void Number Density, and (d) Void Size. Neg. Nos. MSD-56766, 56770, 56760, and 56768.

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Figure 8. Predicted Effect of Neutron Exposure and Irradiation Temperature on In-reactor Creep Rate of Type 304 Stainless Steel.
Neg. No. ANL-306-1405.

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