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During irradiation of thin foils of an austenitic stainless steel in a high voltage electron microscope, steadily growing voids have been observed to suddenly shrink and disappear at the irradiation temperature of 650°C; the phenomenon has been observed in specimens both with and without implanted helium. Possible mechanisms for void shrinkage during irradiation are considered. It is suggested that the dislocation-pipe-diffusion of vacancies from or of self-interstitial atoms to the voids can explain the shrinkage behaviour of voids observed during our experiments.

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VOID SHRINKAGE IN STAINLESS STEEL DURI..G HIGH ENERGY ELECTRON IRRADIATION

by

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Abstract

During irradiation of thin foils of an austenitic stainless steel in a high voltage electron microscope, steadily growing voids have been observed to suddenly shrink and disappear at the irradiation temperature of 650°C; the phenomenon has been observed in specimens both with and without implanted helium. Possible mechanisms for void shrinkage during irradiation are considered. It is suggested that the dislocation-pipe-diffusion of vacancies from or of self-interstitial atoms to the voids can explain the shrinkage behaviour of voids observed during our experiments.

CONTENTS

1. INTRODUCTION

During thermal annealing, a void can shrink only through a loss of vacancies; the loss can occur due to thermal evaporation of vacancies which can then diffuse through the lattice (Volin and Balluffi 1968, Bowden and Balluffi 1969, Johnston, Dobson and Smallman 1969, and Smallman and Westmacott 1971) to internal sinks or surface(s). Even faster loss of vacancies can occur when voids are connected to dislocations which act as diffusion pipes (Volin, Lie and Balluffi 1971). During irradiation where (a) self-interstitials are produced continuously and (b) there exists a supersaturation of vacancies, a void can shrink not only by losing vacancies but also by gaining a net flux of self-interstitial atoms.

The void shrinkage by thermal evaporation at normal irradiation temperature is not likely to be a very efficient process; the dislocation pipe diffusion, on the other hand, can make a void shrink during irradiation (Norris 1971). Recently Nelson (1975) has proposed a mechanism which, under some circumstances, can lead to void shrinkage during irradiation via vacancy loss due to the intersection of replacement collision sequences with the void surface.

The shrinkage of voids has also been considered in terms of a preferential attraction for interstitial atoms to the void surface (Foreman 1971, Makin 1971). Foreman (1971) has suggested that the stress field of an interstitial would give rise to preferential drift effects close to the void surface, leading to the shrinkage of the void.

Since there exists a supersaturation of vacancies during irradiation, voids generally grow and the shrinkage of voids is a very rare event. We have however recently observed that in two (out of 35) irradiation experiments at 650°C some relatively big ($\geq$ 500Å in dia.) voids have shrunk and disappeared very rapidly. Voids of this size contain about 10 million vacancies per void and a considerable transport of matter is required to make them vanish. In the present note we first describe some of our observations showing shrinkage and disappearance of voids during 1 MeV electron irradiation of an austenitic stainless steel at 650°C in the EM-7 high voltage electron microscope. This is followed by considerations of possible shrinkage mechanisms in terms of loss of vacancies from and gain of self-interstitial atoms to the voids.

Theoretically it is also possible that the presence of high compressive stresses around a void can make the void emit dislocation loops and thermal vacancies. Both processes could produce void shrinkage if the stresses were to become sufficiently large. It is very difficult to comprehend, on the other hand, (a) why such high stresses should arise in an extremely small volume of the crystal surrounding an individual void and (b) how these stresses would be maintained while the void shrinks because the shrinkage will lead to stress relaxation around the void. Thus, it seems very unlikely that this mechanism can explain a sudden onset of very fast shrinkage rate in the case of a large and individual void. Hence, this mechanism will not be considered in the present paper.

2. EXPERIMENTS AND RESULTS

A piece of hot rolled sheet of an austenitic stainless steel (see Singh 1974 for composition) was annealed at 800°C or 1150°C for two hours in a vacuum of 10^{-6} torr. Thin foils obtained from this material were irradiated with 1 MeV electrons at 650°C in the EM-7 high voltage electron microscope (HVEM) at Harwell. In one of the thin foils helium atoms were implanted at room temperature using a 100 kv Heavy Ion Accelerator. During irradiation a displacement damage rate of 5×10^{-3} dpa (displacement per atom) per second was used. Around the specimen under irradiation a vacuum of $2\text{-}3 \times 10^{-7}$ torr was maintained. The thickness of the foil containing voids and the location of individual voids in the foil were determined from stereo pairs of micrographs.

In specimens both with and without helium atoms, voids were readily formed during irradiation at 650°C. As expected, the void density is markedly lower in the specimen without implanted helium (fig. 1) than in the specimen containing 100 ppm of helium (fig. 2). Once formed, these voids generally grow fairly rapidly at this temperature. In two out of 35 experiments at this temperature a few voids have, however, been observed to shrink and disappear in specimens both with and without implanted helium; this is demonstrated in figs. 1 and 2. Fig. 1 clearly shows that some of the voids present at a displacement dose of 15 dpa (fig. 1a) have disappeared before a dose of 20.4 dpa is reached (fig. 1b); there is no change in the specimen orientation between fig. 1a and fig. 1b. The locations of some of the voids in fig. 1 are quoted in fig. 3. The surface denuded zone width has been taken to be the same as the grain boundary denuded zone width which is in this steel found to be $\sim$ 1200Å at 650°C.

In the specimen containing implanted helium, voids have also shrunk and disappeared (fig. 2); in this experiment some voids have been observed in the process of shrinking. The growth and shrinkage behaviour of some of these voids are shown in fig. 4. The dose dependance of the average size of 200 voids is also shown in fig. 4. The locations of voids in the void layer is shown in fig. 5. Results quoted in fig. 4 demonstrate two significant points: firstly that the voids which have shrunk and disappeared have been growing steadily and at a rate faster than the average rate for 200 voids, and secondly that the voids have started shrinking very suddenly and have shrunk at a very fast rate. The average shrinkage rates of some of the shrinking voids are quoted in table 2.

3. POSSIBLE SHRINKAGE MECHANISMS

While considering mechanisms for void shrinkage during irradiation, it is important to note that at normal irradiation temperature the vacancy population is in a state of supersaturation against thermal vacancies as well as self interstitial atoms. It is this supersaturation which maintains the necessary net flux of vacancies to the growing voids. These voids can therefore shrink if they loose vacancies faster than they receive them. They would of course also shrink if they were to receive a net flux of self-interstitial atoms instead of vacancies. We therefore consider the shrinkage mechanisms in terms of both a net loss of vacancies from and a net gain of self-interstitial atoms to a void.

3.1. Vacancy Loss Mechanisms

During thermal annealing as well as irradiation experiments at elevated temperatures voids do loose vacancies by thermal evaporation. It can be shown however that during irradiation the shrinkage rate of voids due to thermal emission of vacancies is very small at moderate irradiation temperatures and is more than compensated by the net flux of irradiation induced vacancies into these voids. It is thus clear that the thermal evaporation of vacancies cannot lead to a net void shrinkage during irradiation at moderate temperatures. We therefore consider, in the following, the possibility of vacancies diffusing out of a void through a favourably oriented dislocation pipe.

We shall assume that a dislocation segment of length L connects a void of radius r_v to the foil surface. Then the shrinkage rate due to loss of vacancies through this dislocation pipe is, according to Volin et al. (1971) given by

$$-\frac{dr_v}{dt} = \frac{1}{4\pi} \frac{\pi r_d^2 D_d 2\gamma \Omega}{kT} \cdot \frac{1}{L-r_v} \cdot \left(\frac{1}{r_v}\right)^3 \qquad (1)$$

where r_d is the dislocation core radius, D_d the pipe diffusion coefficient, γ the surface energy, Ω the atomic volume, and kT has the usual meaning. Since the contribution of the thermally evaporated vacancies to the void shrinkage rate is negligible, (i.e. of the order of 10^{-3} Å/sec) it has not been included in eqn. (1). The numerical values of the constants used in eqn. 1 are given in table 1. In view of the lack of accurate data for

Table 1

Constants for Stainless Steel

$r_d = \underline{b}$ (Burgers Vector) $= 2.52 \times 10^{-8}$ cm
$\Omega = 1.13 \times 10^{-23}$ cm^3
$\gamma = 2000$ ergs/cm^2
$D_d = D_d^o \exp(-Q_d/kT)$
where $D_d^o = D_1^o = 0.6$ cm^2/sec
$Q_d/Q_1 = 0.4$ to 0.7
and $Q_1 = 2.9$ eV

D_d^o and the suggestion made by Balluffi (1970) that D_d^o values are, in general, somewhat smaller than those for lattice diffusion, we have assumed $d_d^o \simeq D_l^o$ (table 1). Since the experimental values of Q_d are found to be in the range 0.4 to 0.7 of the corresponding lattice diffusion energies, Q_l, (Balluffi 1970), we have calculated the shrinkage rate for some of the voids which have been observed to shrink during our irradiation experiments using Q_d/Q_l = 0.4, 0.5, 0.6 and 0.7; the results are quoted in table 2.

Table 2

Calculated and Observed Void Shrinkage Rate

Voids	d_v (Å)	L (Å)	Δt_{max}[a] (min.)	$-(dr_v/dt)$ in (Å/min.) Experimental average	Calculated (Q_d/Q_l = 0.4-0.7) 0.4[b]	0.5[c]	0.6[d,e]	0.7[f]
5	500	2062	8	-	336.4	8.8	0.2	0.0
6	710	1702	8	-	158.1	4.1	0.1	0.0
9	775	3298	-	7.5	56.2	1.5	0.0	0.0
9	625	3298	-		104.5	2.7	0.1	0.0
10	532	3088	-	4.1	179.3	4.7	0.1	0.0
10	450	3088	-		292.1	7.6	0.2	0.0
11	587	2949	10	-	141.9	3.7	0.1	0.0
12	760	2249	10	-	92.9	2.4	0.1	0.0
13	525	2055	-	9.1	293.8	7.7	0.2	0.0
13	380	2055	10		744.6	19.5	0.5	0.0

(a) Δt_{max} refers to the time within which voids are observed to have disappeared.

(b) Edge dislocations ($\underline{b}$ = a<100>) in a low angle tilt boundary (possibly close pairs of non-dissociated edge dislocations, $\underline{b}$ = a/2<110> + a/2<1$\bar{1}$0>), Upthegrove and Sinnott (1958).

(c) Edge dislocations ($\underline{b}$ = a/2<110>), Wuttig and Birnbaum (1966).

(d) Slip dislocations ($\underline{b}$ = a/2<110>) of varying edge-screw character, averaged over a wide range of edge-screw compositions, Canon and Stark (1969).

(e) Edge dislocations ($\underline{b}$ = a/2<110>) in a low angle (10°) tilt boundary, Canon and Stark (1969).

(f) Screw dislocations ($\underline{b}$ = a/2<110>) in a low angle (10°) twist boundary, Canon and Stark (1969).

It should be pointed out here that the calculated shrinkage rate quoted in table 2 is the minimum shrinkage rate because the possibility of more than one dislocation getting connected between a void and the foil surfaces does exist.

For comparison, the experimentally observed average shrinkage rate of voids 9, 10, and 13 is also quoted in table 2. Because the shrinkage rate is strongly dependent on the void radius (eqn. 1), the average shrinkage rate can be compared only qualitatively with the calculated shrinkage rate for a given void radius. While considering the comparison, it is also relevant to note that the calculated results (table 2) refer to nickel which has much higher (almost an order of magnitude) stacking fault energy than the stainless steel used in our experiments. Since the presence of stacking fault ribbon slows down the pipe diffusion (Balluffi 1970), the shrinkage rate in stainless steel would be expected to be lower than the calculated results quoted in table 2. Table 2 shows, however, that the calculated shrinkage rate for $Q_d/Q_1 = 0.5$ (i.e. pipe diffusion along edge dislocations) are in a reasonable agreement with the observed shrinkage rates. The calculated shrinkage rates for dislocations of screw and edge-screw characters are clearly too low to be compatible with our observations.

Another mechanism which can lead to vacancy losses from a void surface has been proposed by Nelson (1975). This is thought to arise as a result of the intersection of replacement collision sequences with the void surfaces. Whether or not this mechanism is capable of producing, rather suddenly, a net shrinkage in the diameter of a void which has been growing previously is examined in the following. According to Nelson's eqns. (2) and (7), the rate of change in void radius is given by

$$\frac{dr_v}{dt} = \frac{K\ \rho_d(Z_i - Z_v)}{r_v(\rho_d + 4\pi r_v C_v)^2} - \frac{\alpha\ \rho_d\ K\ \Omega^{1/3}}{(\rho_d + 4\pi r_v C_v)} \qquad (2)$$

where K is the damage rate, ρ_d is the dislocation density, C_v is the void density, $(Z_i - Z_v)$ is the dislocation preference, and α is a fitting parameter and has a value of between 0.5 and 5 depending on the efficiency of replacement sequences in the metal under consideration. The first term on the right hand side of eqn. (2) represents the void growth and the second term is the shrinkage rate due to dynamic vacancy injection at the void surfaces. In the derivation of eqn. (2) it has been assumed by Nelson that the injected vacancies are distributed uniformly throughout the solid and hence can easily get trapped at dislocations or/and other voids; vacancies lost at dislocations contribute to the shrinkage term.

In order to maximize the rate of shrinkage and also to make eqn. (2) applicable to rare events of net void shrinkage, as observed in our experiment, let us assume that during irradiation a void accumulates a large number of dislocations around i . Thus the void is isolated from the rest of the voids in the system. The dislocations around it absorb all the vacancies coming out of the isolated void as well as those coming towards it (i.e. the growth term in eqn. (2) becomes zero). Then the maximum shrinkage rate from eqn. (2) is given by

$$\frac{dr_v}{dt} = -\alpha K \Omega^{1/3} \qquad (3)$$

It is important to note here that the shrinkage rate, according to eqn. (3), is independent of void size and irradiation temperature. Our limited experimental results, on the other hand, indicate that the shrinkage rate increases with decreasing void size. Besides, the shrinkage rate according to eqn. (3) is too slow to explain our observation. For example, taking α=1 (as suggested by Nelson (1975) for stainless steel), $K = 5 \times 10^{-3}$ dpa/sec (in HVEM) and $\Omega = 10^{-23}$ cm^3, the shrinkage rate according to eqn. (3) is 0.6 Å/min. It will take about 435 minutes of irradiation before a void of 262Å (void 13 in table 2) in radius can shrink to r_v=0; in our experiment this particular void shrinks and disappears in less than 18 minutes. Furthermore our experiments suggest that not only do the voids shrink fast but also that the onset of shrinkage is rather sudden. This would require, according to the mechanism under discussion, a sudden and large increase in dislocation density around the voids which begin to shrink. It is, however, difficult to see what mechanism can lead to such a change.

Fig. 6 shows two micrographs taken during an irradiation experiment at 650^{o}C. It is quite clear that the dislocation density around void A is much higher than in its surroundings and around other voids. The void A however maintains a healthy growth and does not shrink at all. This is, of course, not conclusive evidence, but can be taken to indicate that the loss of vacancies through dynamic injection mechanism is not sufficient to cause a net shrinkage in void radius even in the situation when the mechanism should be most efficient.

3.2. Self-Interstitial Gain Mechanism

A void can gain an excess number of self-interstitial atoms required for its shrinkage if (a) it was possible for self-interstitial atoms to diffuse through a dislocation pipe to the void or (b) the void aquired a preference for self-interstitial atoms. In the following we first consider the dislocation pipe diffusion of self-interstitial atoms; the other possibility is dealt with in the later part of this section.

As in the previous section, we consider a void to be connected to a "hollow" dislocation of length L, but in this case the dislocation need not be connected to the foil surface. We assume that this particular dislocation segment acts as a very efficient pipe for the diffusion of all the excess self-interstitial atoms preferentially trapped at it. It should be noted here that this process could be operative only during irradiation. We then estimate the length of the dislocation pipe required to fill up the void at the observed rate; the estimate is based on our experimental data on the rate of swelling and void shrinkage. The observed swelling rate, in our experiment is found to be 1/3% per dpa. Thus the rate at which excess interstitial atoms are being deposited on a unit length of dislocation line in the material is $1/(300\ \Omega\rho_d)$ per dpa. The rate of shrinkage of a void due to the special length L of dislocation that is piping these excess interstitials back into the void is then

$$\frac{dr_v}{d(Kt)} = -\frac{1}{300}\,\frac{L}{4\pi r_v^2\rho_d} \tag{4}$$

We have assumed that (a) the 'hollow' dislocation has the normal preference for self-interstitial atoms, (b) the interstitial atoms can move along the dislocation pipe very easily, and (c) the interstitials cannot escape from the far end of the pipe. The average diametral shrinkage rate for void no. 13 (see table 2 and fig. 4) is 60.7 Å per dpa and its average diameter is 452.5 Å.

However the void would normally have been growing at a rate of 21.2 Å per dpa, so that the total diametral shrinkage effect must be 81.9 Å per dpa. Inserting this value into eqn. (4) and using $\rho_d = 10^{10}$ cm^{-2} we find L to be 7,902 Å. This length of dislocation line is comparable with the foil thickness (6,677 Å) but could easily be accomodated within the foil, especially if the dislocation were to be branched or convoluted.

A second mechanism for void shrinkage by interstitial gain has been proposed (Foreman 1971, Makin 1971). The voids may possess a bias towards trapping interstitial atoms, due to the short range attractive elastic interaction that might be expected to exist between an interstitial atom and a free surface. Calculations of void shrinkage by this mechanism have been made by Foreman (1971) on the basis of the cellular model. We shall here use the equations of rate theory, which gives similar results but in a simpler form. Let us suppose that due to the elastic interaction any interstitial that approaches within a distance Δr from the void surface is trapped. Then the sink strength of the voids for capturing vacancies is $4\pi r_v C_v$ and for interstitial atoms is $4\pi(r_v+\Delta r)C_v$. Thus the void has a bias equal to $\Delta r/r_v$ for capturing interstitial atoms. The precise strength of this effect is not known but it would be surprising if Δr were more than a few atom spacings, so that a 500 Å diameter void could have a bias of $\sim$1%.

The rate theory equations are

$$K = v\, D_v(4\pi r_v C_v + \rho_d) \qquad (5)$$

$$K - \Delta K = i\, D_i\left(4\pi(r_v+\Delta r)C_v + (1+p)\rho_d\right) \qquad (6)$$

where $p = Z_i - Z_v$ is the dislocation preference ($\sim$5%), v and i are the steady state concentrations of vacancies and interstitials respectively, D_v and D_i are their respective diffusion coefficients, and we have neglected the direct recombination of point defects. The term

$$\Delta K = K\, 4\pi\, r_v^2\, \Delta r\, C_v \qquad (7)$$

is a correction for the interstitials that are produced within a distance Δr from the void surfaces, since by definition these interstitials are irrevocably trapped. The growth rate of a void may then be shown to be

$$\frac{dr_v}{dt} = \frac{K}{r_v}\frac{\rho_d\, p}{(\rho_d+4\pi r_v C_v)^2} - \frac{K}{r_v^2}\frac{\rho_d\, \Delta r}{(\rho_d+4\pi r_v C_v)^2} - \frac{K\,\rho_d\,\Delta r}{(\rho_d+4\pi r_v C_v)} \qquad (8)$$

if $p \ll 1$ and $\Delta r/r_v \ll 1$. The first term in eqn. (8) is the normal void growth due to the vacancy supersaturation produced by the dislocation preference p. The second term is the shrinkage rate arising from the bias of the voids for trapping interstitials. The third term in eqn. (8) is the void shrinkage that arises because interstitials produced within a distance Δr from a void surface are irrevocably trapped.

We note that the principal void shrinkage term becomes progressively less important as the voids grow larger, due to the $1/r_v^2$ factor. Thus it is difficult to see how this term could account for the sudden and rapid shrinkage of a large void, even if the local dislocation density were to suddenly escalate. Only a sharp reduction in the dislocation preference p could produce such an effect. However a sudden increase in ρ_d might allow the second shrinkage term in eqn. (8) to dominate, giving

$$\frac{dr_v}{d(Kt)} = -\Delta r \qquad (9)$$

but this effect is rather small if Δr is only a few atom spacings. The observed shrinkage rate of 60.7Å per dpa would require Δr = 30.3Å, which is an incredibly long range for the elastic interaction between an interstitial atom and a void surface. Furthermore it would make the first shrinkage term in eqn. (8) very strong for normal dislocation densities, so that normal void growth would become difficult to explain. Thus we must conclude that the present mechanism cannot account for the observed rate of void shrinkage.

4. DISCUSSION AND SUMMARY

The analysis of section 3 would suggest that although all four mechanisms considered can, in some circumstances, lead to void shrinkage yet not all of them are capable of explaining the observed shrinkage behaviour which is typified by (i) a sudden onset and (ii) a very fast rate of shrinkage. It seems that only two mechanisms, based on diffusion of vacancies from or self--interstitial atoms to a void through a dislocation pipe, can appropriately apply to and be found to be reasonably consistent with the experimental results. Since dislocations during irradiation at 650^{o}C are highly mobile, it is reasonable to assume that a dislocation suitably orientated to act as an efficient diffusion pipe gets suddenly connected between a void and the foil surface, giving a sudden onset of the shrinkage. As far as the shrinkage rate is concerned, the calculated shrinkage rates, (based on eqn. 1) are in reasonable agreement with the experimentally observed ones.

The question why the shrinkage of voids as observed in our experiments is so rare an event cannot, at present, be answered with any certainty. We suggest the following which may at least be a part of the explanation. It may be that the pipe diffusion is sensitive to the character (see table 2) and orientation of the dislocation pipe (Volin et al., 1971, Canon and Stark 1969). The segregation of impurity atoms to dislocations, which is bound to occur during irradiation, can also affect the pipe diffusion. Furthermore, the presence of surface denuded zones during thin foil irradiation experiment may affect the possibility of dislocations getting connected between voids and foil surface(s). The occurrence of a hollow dislocation could also be a rare phenomenon, especially if it needs to have a larger than normal Burgers vector. The net effect of these factors will be that the pipe diffusivity is likely to vary fairly widely. This would suggest that the effect of pipe diffusion becomes noticeable in those rare cases when dislocation pipes are able to operate under ideal conditions. In other cases, although dislocation pipe diffusion might be operating but not efficiently enough to cause a net shrinkage in void radius at a fast rate. Finally, it should also be mentioned that edge dislocations with a<100> Burgers vectors would be rare and should give rise to rapid shrinkage. These might be causing the effect, especially if impurity atoms are inhibiting pipe diffusion along normal dislocations in stainless steel.

The voids should tend to remove more of impurities out of solution as they grow larger, giving a clearer matrix and more rapid pipe diffusion. This might explain why only large voids are seen to shrink.

In summary, then, we have observed very fast shrinkage and disappearance of some voids during 1 MeV electron irradiation at 650°C. This kind of chrinkage event is rather rare and should not affect the overall growth behaviour of voids in HVEM experiments. We have examined four possible shrinkage mechanisms and have found that the mechanism based on dislocation pipe diffusion of vacancies from or self-interstitial atoms to voids is consistent with the experimental observation.

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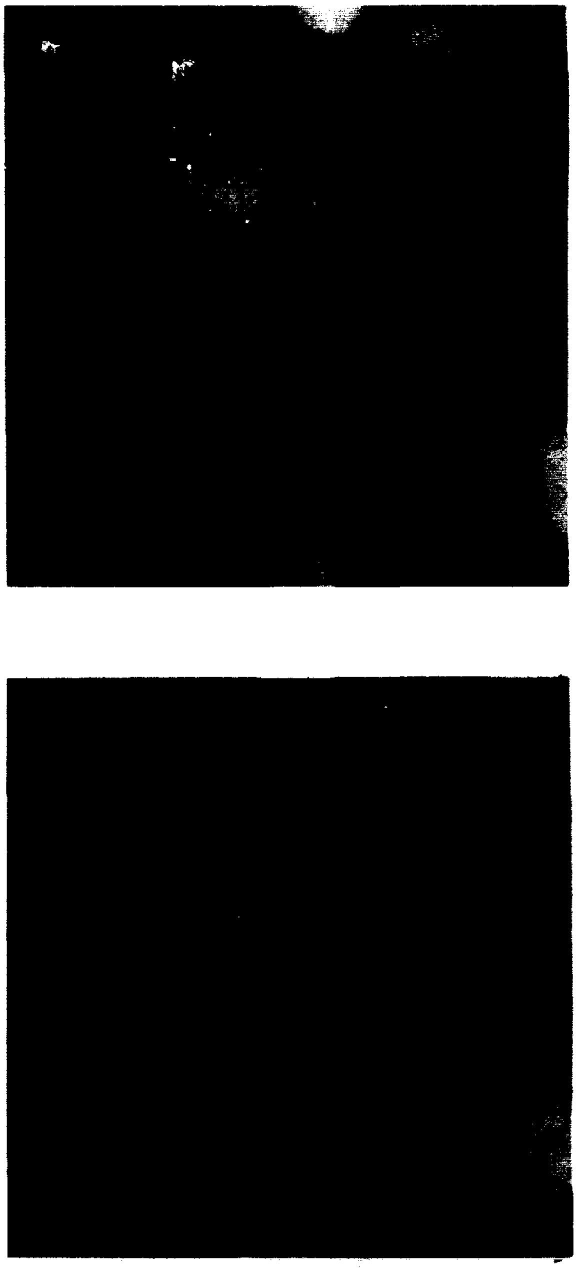
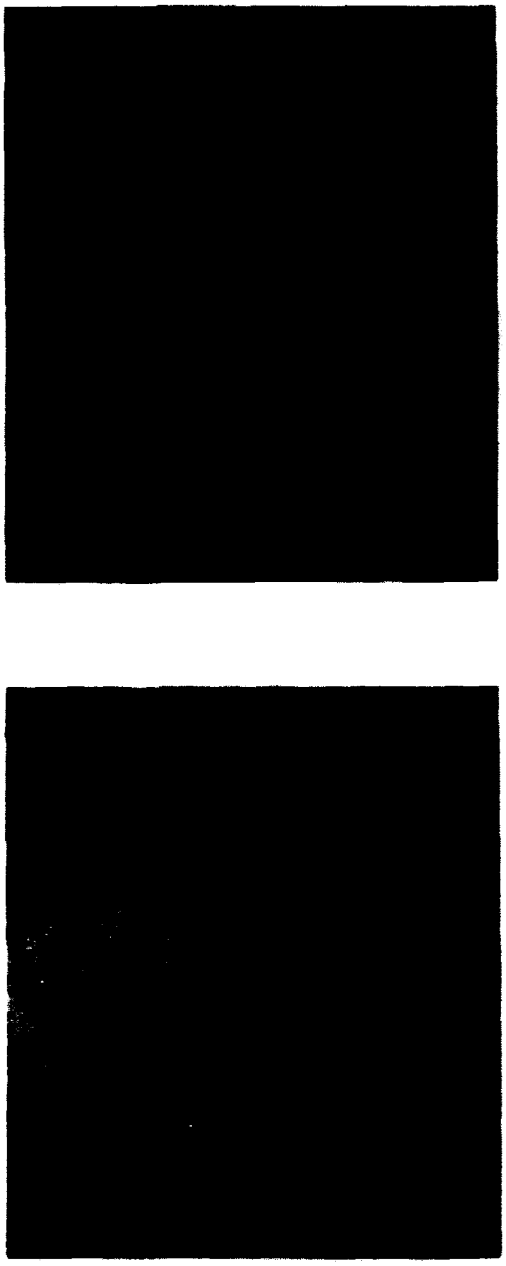


(a)
x 85,000

(b)
x 85,000

Fig. 1. 1 MeV electron micrographs of an austenitic stainless steel specimen (without implanted helium) irradiated at 650°C for (a) 50 min. (i.e. 15.5 dpa) and (b) 68 min (i.e. 21.1 dpa). Note that voids 3, 5, 6, 7 and 8 have shrunk and disappeared during 18 minutes of irradiation.



(a)
x 85,000

(b)
x 85,000

Fig. 2. 1 MeV electron micrographs of an austenitic stainless steel specimen (with 100 ppm of implanted helium) irradiated at 650°C for (a) 65 min (i.e. 20.1 dpa) and (b) 75 min (i.e. 23.2 dpa). Note the shrinkage of voids 9 and 10 and disappearance of voids 11, 12 and 13.

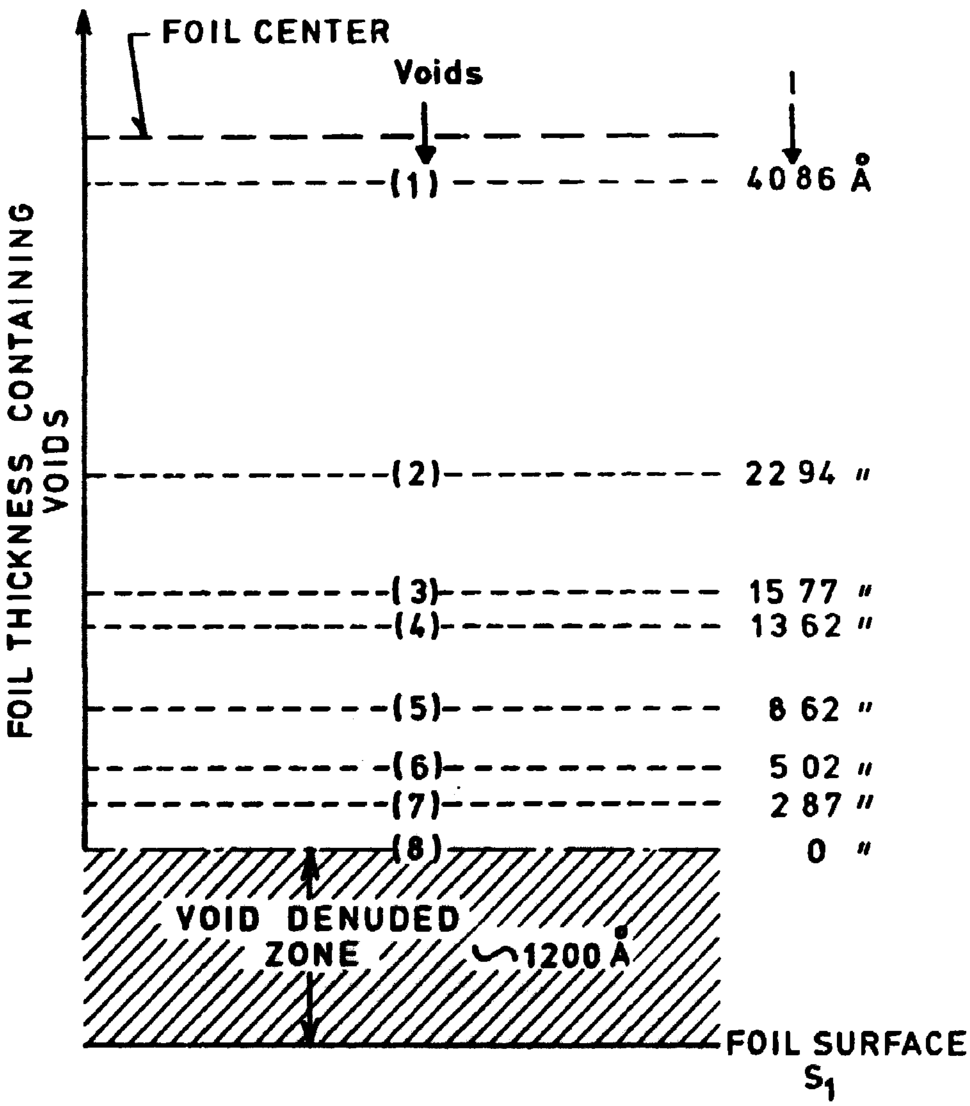


Fig. 3. Schematic representation of the location of voids (shown in fig. 1) in the void layer of the irradiated thin foil. l refers to the distance of a void from the inner end of the void denuded zone along the foil surface S_1.

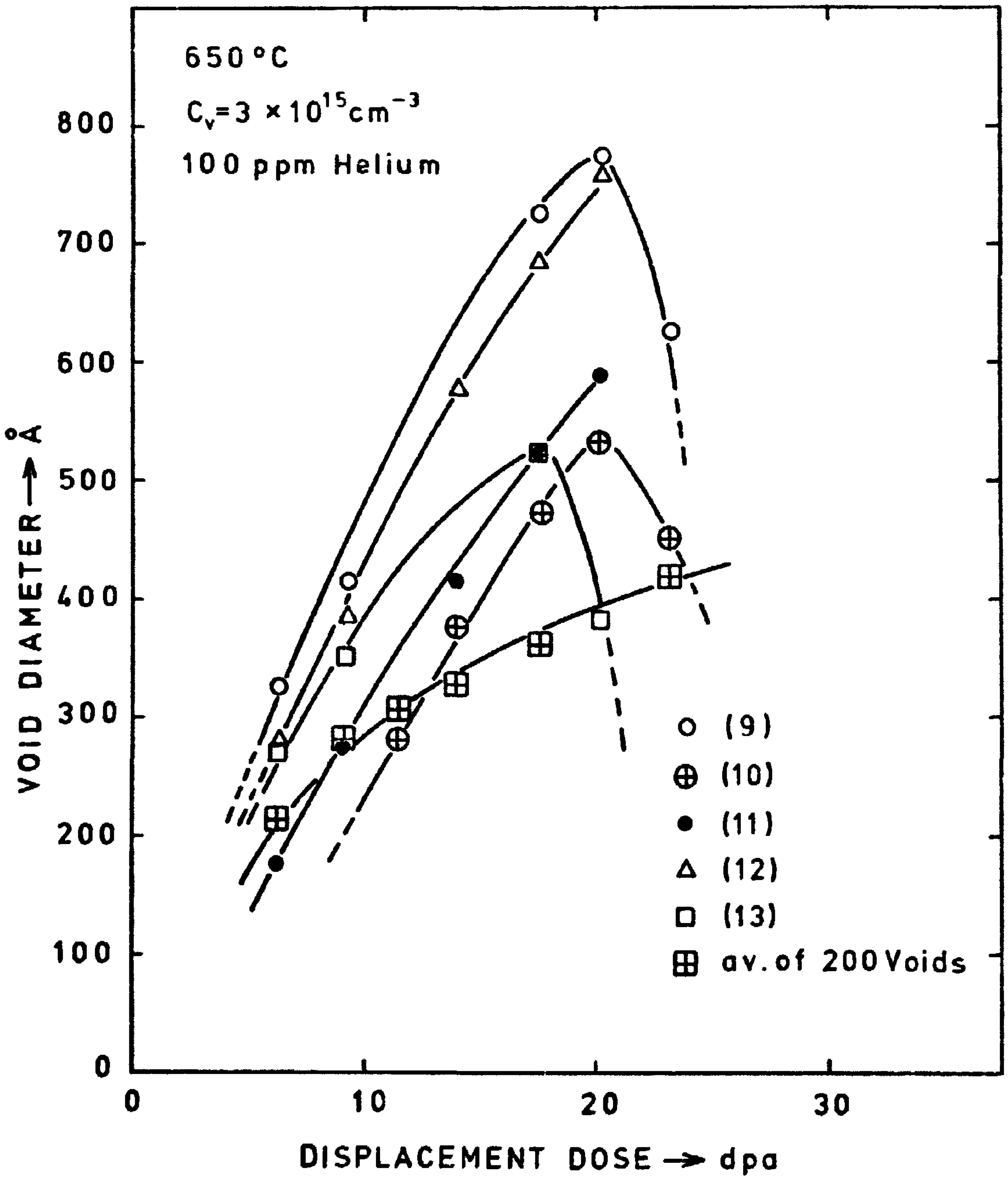


Fig. 4. Growth and shrinkage behaviour of voids (shown in fig. 2) during irradiation at 650°C. Voids 9 and 10 have shrunk and 11, 12 and 13 have shrunk and disappeared in ≤ 10 min.

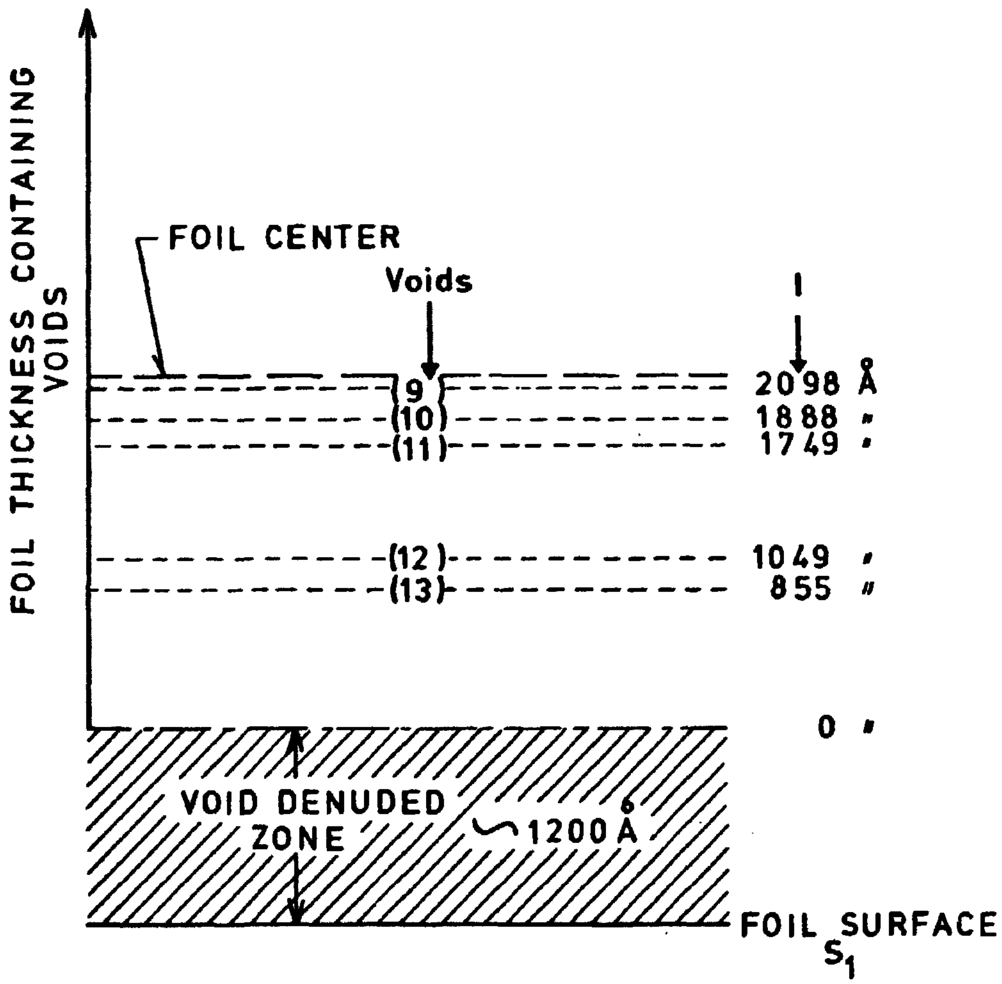
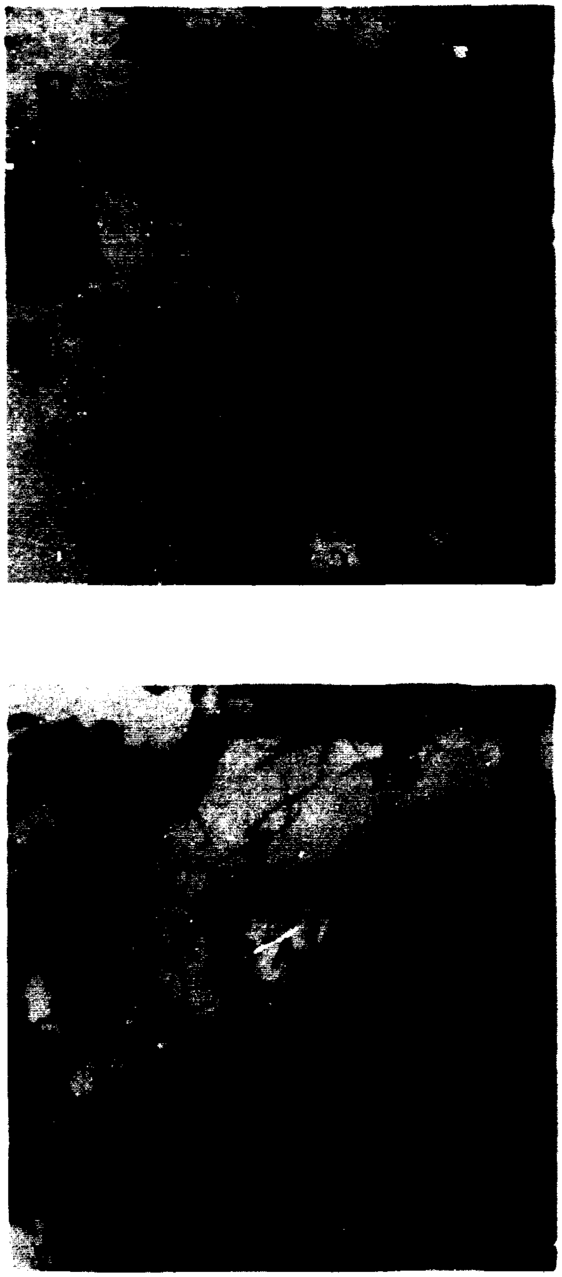


Fig. 5. Schematic representation of the location of voids (shown in fig. 2) in the void layer of the irradiated thin foil; I has the same meaning as in fig. 3.



(a)
x 85,000

(b)
x 85,000

Fig. 6. 1 MeV electron micrographs of an austenitic stainless steel irradiated at 650°C for (a) 29 min (i.e. 9 dpa) and (b) 65 min (i.e. 20.1 dpa).