INVESTIGATIONS OF THE MICROSTRUCTURE
OF HEAVY ION BOMBARDED NICKEL

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JULY 1974

Prepared for the U.S. Atomic Energy
Commission under Contract AT(45-1):1830

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PACIFIC NORTHWEST LABORATORY
operated by
BATTELLE
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U.S. ATOMIC ENERGY COMMISSION
Under Contract AT(45-1)-1830

Printed in the United States of America
Available from
National Technical Information Service
U.S. Department of Commerce
5285 Port Royal Road
Springfield, Virginia 22151
Price: Printed Copy \$4.00; Microfiche \$1.45

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BNWL-1839
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ABSTRACT

The microstructure of pure nickel bombarded with Ni ions in the temperature range 180°C to 450°C has been studied. The defect structure consist of small dislocation loops and stacking fault tetrahedra (SFT) which are of vacancy type. Large dislocation loops and dislocation fragments comprise the interstitial type damage. The small loops and SFT formed a f.c.c. defect lattice similar to the void lattice. The preference of the vacancies to form dislocation type defects rather than voids can be related to the high displacement rate during ion bombardment. Several explanations are given for the defect lattice although no one explanation is completely adequate.

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INTRODUCTION

The study of void formation in metals as a result of heavy ion bombardment has received increased impetus in recent years. Heavy ion bombardment has been used because of the greatly increased atom displacement rates which can be obtained. However, because of the great differences in dose rates between ion bombardment and neutron irradiation, one must be cautious in equating the structures in neutrons irradiated with ion bombarded materials which have been irradiated at the same temperature. Specifically it has been proposed and shown in some cases that a temperature correction must be applied when comparing void microstructures due to irradiation at different dose rates.

There have been several ion bombardment studies of nickel at temperatures greater than 450 °C.⁽¹⁻⁵⁾ However, there has been little reported on ion bombardment studies at lower temperatures. Neutron irradiation of nickel at temperatures above approximately 250 °C will produce voids.⁽⁶⁾ However, due to the temperature shift, it was not known whether voids would be produced at these low temperatures during ion bombardment, and, if not, what type of vacancy defect structure would prevail.

To clarify these questions, the microstructure of ion bombarded nickel was studied as a function of temperature over the range 185° - 435 °C, and as a function of dose over the range 0.1 to 10 displacements per atom (dpa).

The results are analyzed in light of present knowledge of defect migration in nickel and the ease of nucleation of voids and dislocation type defects.

EXPERIMENTAL PROCEDURES

MRC nickel foil, approximately 0.75 mm thick, was used in the experiment. The chemical analysis is shown in Table 1. The foils

Table 1

Chemical Analysis of the Nickel

<u>Impurity</u>	<u>ppm wt.</u>
Mg	2
Al	4
Si	5
Ca	2
Fe	7
Cu	5
C	50
Oxygen	57
Nitrogen	<5
Hydrogen	6

were annealed at 750 °C for two hours in a vacuum of 1×10^{-6} torr. This treatment was followed by an electropolish prior to bombardment. There was no mechanical polishing of the surface.

The foils were bombarded in a Tandem Van de Graaff accelerator with 5 MeV Ni^{2+} ions. The diameter of the beam was approximately 3 mm. The ion current was periodically checked with a Faraday cup, inserted directly in front of the specimen. The specimen was clamped to a heated block to maintain temperature and the temperature was monitored with a thermocouple mounted on the heating block.

The ion dose was converted to displacements per atom using the approach of Kulcinski et al.⁽⁷⁾ and Manning and Mueller.⁽⁸⁾ The range of the nickel ion in nickel is of the order of 1 micron. The variation in displacement damage as a function of distance from the bombarded surface is shown in Figure 1. The bombardment times varied from 1 minute to 30 minutes and the dose rates from 6×10^{-3} to 2×10^{-2} dpa/sec. The most common dose rate was 1×10^{-2} dpa/sec.

The foils were electrolytically thinned such that a section representing approximately 0.5 micron from the bombarded surface was viewed in the electron microscope. The particular thinning technique has been described previously.^(1,9) The actual measurements of surface removal varied between 0.2 to 0.8 μm . In several instances, either by accident or design, the area viewed represented regions beyond the maximum in the displacement versus distance curve.

RESULTS

In the next section, the general characteristics of the microstructure are discussed as a function of bombardment temperature and ion dose. A more detailed analysis of the identity of the various defects and their relationship to temperature and dose are then presented.

EFFECT OF TEMPERATURE

The microstructure reveals a high density of black spot defects and a dislocation network, Figure 2. The dislocations appear to be the remains of large loops or are a result of the interaction of large loops. At the low temperature (185 °C), the black spots are distributed randomly, Figure 2a. From 280 °C up to 435 °C, the black spots become aligned into

preferred crystallographic directions, Figures 2b, c, and d. The rows are aligned parallel to the $\langle 100 \rangle$ direction of the parent atomic lattice for the $\{100\}$ foil orientation shown. The size of the spots increases and the density of spots decreases with temperature. At 370 - 435 °C, small dislocation loops can be clearly resolved within the rows. The spacing of the rows increases slightly with temperature. Only an occasional void was present after bombardment and only at the higher temperatures of 370 - 435 °C. The quantitative measurements of the various defect parameters are listed in Table 2 as a function of temperature.

Table 2

<u>Temp.</u>	<u>Size of Spots</u>	<u>Measured Density</u>	<u>Lattice Spacing</u>	<u>Density from Lattice Spacing</u>
185 °C	35 Å	$1 \times 10^{17}/\text{cm}^3$	Not measurable	--
280 °C	55 Å	$4 \times 10^{16}/\text{cm}^3$	700 Å	$1.2 \times 10^{16}/\text{cm}^3$
370 °C	70-80 Å	$3.3 \times 10^{16}/\text{cm}^3$	800 Å	$0.8 \times 10^{16}/\text{cm}^3$
435 °C	120 Å	$2 \times 10^{16}/\text{cm}^3$	900 Å	$0.55 \times 10^{16}/\text{cm}^3$

The crystallography of this "spot" lattice is consistent with a f.c.c. arrangement. Tilting the foil from the $\langle 001 \rangle$ orientation to near $\langle 011 \rangle$ produces the alignment shown in Figure 3. This microstructure is consistent with a face centered cubic arrangement of spots when viewed along a $\langle 011 \rangle$ axis. The foils have a very strong $[001]$ texture, and it is difficult to obtain a $\langle 110 \rangle$ exactly parallel to the beam. Figure 3 represents a tilt of approximately 37° from the $\langle 001 \rangle$, so it is not a perfect $\langle 110 \rangle$. However, the b.c.c. and simple cubic would show a different arrangement of spots for the $\langle 110 \rangle$ orientation. The directions in the f.c.c. spot lattice also coincide with the directions of the parent f.c.c. atomic lattice. In this sense, it is similar to the void lattice which has been reported in nickel and aluminum. (1,10)

The number density of spots and loops determined from the lattice spacing is also listed in Table 2. The counted number density is about a factor of four higher than that calculated from the lattice spacing. This means that more than one defect is occupying a lattice site.

EFFECT OF DOSE

The effect of dose is illustrated in Figure 4 which shows the microstructure after doses of 0.1 and 1 dpa at a constant bombardment temperature of 370 °C. The defect lattice is not obvious at the lowest dose; however, it is well formed after a dose of only 1 dpa. The absence of an observable defect lattice at the lowest dose is partly due to the lower defect density. Alignment of the defects becomes apparent if a sufficiently thick foil area is observed, as shown in Figure 5. Both the defect size and density increase with dose as listed in Table 3, however, resolvable dislocation loops are only present at the highest dose. The spot lattice spacing does not change with dose.

Table 3

<u>Dose</u>	<u>Size</u>	<u>Density</u>
0.1	50 Å ^o	1.2×10^{16}
1.5	60 Å ^o	2.5×10^{16}
3	60 Å ^o	2.5×10^{16}
10	70-80 Å ^o	3.3×10^{16}

The defect density appears to reach a maximum after 1 dpa. The principal change from 1 to 10 dpa is in the size of the defects.

There is some change in the background dislocation structure as a function of dose. At low doses very large loops could be resolved. At high doses, only a dislocation network or possibly loop fragments are present.

IDENTITY OF DEFECTS

At low doses and low temperatures, the small defects which comprise the lattice are principally Frank sessile loops with Burgers vector $\vec{b} = \frac{a}{3} \langle 111 \rangle$. Figure 6 shows the typical black-white strain contrast from the small defects when imaged in dark field. It has been shown that the direction of the black-white streak originating from the defects is parallel to its Burgers vector or its projection in the foil plane.⁽¹¹⁾ The analysis of the figure indicates that the predominant Burgers vector is $\langle 111 \rangle$. There are only a few spots which could be interpreted as having $\langle 110 \rangle$ type Burgers vector.

In addition to the loop defects, stacking fault tetrahedra are also present. SFT are reported not to give any strain contrast for low index reflection.⁽¹²⁾ However, they can be identified from their shape if they are sufficiently large, i.e., they will project as squares on the $\{001\}$ planes. Figure 7 shows some defects believed to be SFT. SFT have been identified in quenched nickel⁽¹³⁾ and neutron irradiated nickel.⁽¹⁴⁾ The SFT could not be identified at low doses or at temperatures below 370 °C, but this may be due to the very small size of the defects at the lower temperature.

Prismatic dislocation loops were only positively identified at temperatures of 370 and 435 °C and at the higher doses. Figure 8 is a high magnification print illustrating some prismatic loops. The shape of the loop and the direction of the line of no contrast indicates that the loops lie on $\{111\}$ planes. If the loops planes were $\{110\}$, the line of no contrast would lie along the $\langle 001 \rangle$ direction. The loops show no stacking

fault so the Burgers vector must be $a/2 \langle 110 \rangle$. Although prismatic loops are present, a large number of Frank sessile loops are also still present at these higher temperatures and doses.

For dislocation loops large enough to be distinguished as loops, they were always prismatic in nature. It appears that loops originally form as Frank loops at low doses or temperatures, but convert to prismatic loops as they grow larger. The SFT never become very large, and it is not possible to determine if they formed by glide process from an existing Frank loop or if they nucleated directly as tetrahedra.

There was no direct experimental confirmation of the nature of the defects, i.e., vacancy or interstitial; however, there is strong circumstantial evidence that the small, aligned defects originate from vacancies, and the large dislocation loops and network originate from interstitials. The mobility of vacancies is much less than that of interstitials and the vacancy type loops are expected to be much smaller than interstitial type for identical irradiation temperatures. In studies of void formation in irradiated metals, a high density of small voids is normally superimposed upon a low density of interstitial dislocation loops or a dislocation network. Figure 9 shows nickel after neutron irradiation at 260 °C and voids are present after this type of irradiation. The background dislocation structure is very similar to that observed in the nickel ion bombarded at nearly the same temperature. It is reasonable to conclude that the aligned defect structure is directly analogous to the void microstructure in the neutron irradiated nickel. During ion bombardment the vacancies choose to precipitate as a dislocation type defect, whereas during neutron irradiation, the vacancies precipitate as voids.

Neutron irradiation of nickel at a temperature below which voids form, however, does not necessarily produce alignment of the defects. Figure 10 is a micrograph of nickel neutron irradiated at 50 °C and shows a very dense, random distribution of black spot defects. It has not been possible to separate the interstitial from the vacancy type defects in such a structure. However, there has been no neutron experiments in the range 100 - 250 °C in which some alignment of the dislocation-type defects could conceivably occur. *

DISCUSSION

In analyzing the results, two questions must be answered. Why do voids form in neutron irradiated nickel in the temperature range 260 - 440 °C, but do not form in ion bombarded nickel in the same temperature range, and, secondly, why do the dislocation loops form a crystallographic array in the ion bombarded nickel? Each question will be discussed in turn.

A. Type of Defects

The nucleation and growth of voids in neutron irradiated nickel, irradiated over the temperature range 260 - 650 °C has been well documented. (14-17) The formation of voids in nickel by electron irradiation in the high voltage microscope has been found at temperatures of >300 °C. (18,19) However, voids form in ion bombarded nickel only at temperatures greater than 450 °C. (1-5)

It has been proposed that the growth of voids at a particular temperature should be related to the displacement rate. A high displacement rate will result in a much greater recombination rate of vacancies and interstitials. This produces the same effect as irradiating at a lower temperature.

* Stiegler and Farrell (ORNL) have recently observed both voids and the loop lattice in nickel irradiated with neutrons at 300°C but only voids after irradiation at 350°C. (to be published in Scripta Metallurgica)

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Bullough and Perrin have given a simple expression for calculating the temperature shift at the temperature of maximum swelling which is of the form: (20)

$$\frac{K_i}{D_i} = \frac{K_n}{D_n}$$

When K_i and K_n are the displacement rates for ion bombardment and neutron irradiation, respectively, and D_i and D_n are the diffusion coefficient at the respective bombarding temperatures. The temperatures are then calculated from the expression $D = D_0 e^{-Q/KT}$.

The "temperature shift" is applied when comparing the differences in the void microstructure, particularly the swelling, of neutron irradiated and ion bombarded material. It does not necessarily predict that the critical temperature for void nucleation will be greater when irradiating at higher dose rates.

The lower temperature limit for void formation is, in fact, dependent on the dose rate and this conclusion is supported by the following experimental observations. Several specimens were prepared which represented regions beyond the peak of the displacement versus distance curve shown in Figure 1. Figure 11 shows a sample bombarded at 370 °C in which approximately $1.2 \mu \pm .2 \mu$ was removed from the surface. Very small voids are observed and there is no indication of aligned spot defects. This foil could represent an area where the number of displacements is two orders of magnitude less than the maximum. Consequently, the displacement rate is reduced by the same amount since the time is constant. We, therefore, have a situation in which voids form near the end of the damage region where the dose rate is low, but not near the beginning of range where the dose rate

is high. At sufficiently high temperatures, >500 °C, voids form throughout the damage zone at least for the currently available dose rates.

These experimental results still do not answer the basic question as to why voids nucleate in preference to loops other than the fact that the rate at which the defects are generated plays an important role.

It has been predicted theoretically that gas atoms are necessary for the nucleation of voids.⁽²⁰⁾ Experiments also show that gas plays a critical role in void nucleation.⁽²¹⁾ Based on this criterion, it can be shown through straightforward reasoning that the critical temperature for void formation is dependent on the dose rate.

We make the basic assumption that vacancies migrate and combine with gas atoms before combining with other vacancies in order to form voids. Combining only with other vacancies will, in general, produce only dislocation loops. We further assume that the gas atoms, e.g., oxygen, are immobile in the temperature range of interest, approximately 200 to 400 °C. The average spacing of the gas atoms in the matrix, $\bar{X}_g$, is approximately given by:

$$\bar{X}_g = \left(\frac{1}{\rho_g} \right)^{1/3} \quad (1)$$

Where ρ_g is the gas concentration.

The time required for a vacancy to migrate this distance is

$$t_g = \frac{\bar{X}_g^2}{D_v}$$

or

$$t_g = \frac{1}{D_v} (1/\rho_g)^{2/3} \quad (2)$$

Where D_v is the vacancy diffusion coefficient and equal to $D_0 e^{-Q_m/KT}$

Now the concentration of vacancies created by irradiation in time t is given as $\dot{n}t$, where $\dot{n}$ is the net generation rate. The thermal equilibrium concentration can be neglected at the low temperature. The average spacing of vacancies at time t is given by

$$X_v = \left(\frac{1}{\dot{n}t} \right)^{1/3}$$

rearranging gives

$$t_v = \frac{1}{\dot{n}} \left(\frac{1}{X_v} \right)^3 \quad (3)$$

A critical concentration will be reached when the average spacing of the vacancies is equal to the average spacing of the gas atoms;

$$t_v = \frac{1}{\dot{n}} (\rho_g) \quad (4)$$

In order for voids to nucleate, the time required for migration to gas atoms must be less than or equal to the time required for vacancies to reach this critical concentration. Therefore, we equate t_v with t_g

$$\frac{1}{D_v} \left(\frac{1}{\rho_g} \right)^{2/3} \leq \frac{1}{\dot{n}} \rho_g$$

or

$$\dot{n} \leq D_v (\rho_g)^{5/3} \quad (5)$$

This is the same form as the Bullough-Perrin expression except that the vacancy diffusion coefficient is used rather than the self-diffusion coefficient.

However, Braislford and Bullough⁽²²⁾ have developed a more complete model of swelling and show that the vacancy diffusion coefficient should be used when analyzing the low temperature end of the swelling region so it is not inconsistent. When the temperature is low (D_v is small) or the dose rate is high, vacancies tend to combine with other vacancies without the presence of gas atoms and, hence, form loops. Conversely, when the temperature is high or dose rate low, vacancies combine with gas atoms before combining with other vacancies and, hence, form voids.

In the MRC Ni, the concentration of oxygen and nitrogen is about 50 ppm by weight or 180 appm. Using an ion bombardment dose rate of 10^{-2} dpa/sec, D_v is calculated from Equation (5) to be $9 \times 10^{-12}/\text{cm}^2\text{-sec}$. This corresponds to a critical temperature of 500 °C if the migration energy for vacancies is 1.4 e.v.⁽²³⁾ This is consistent with the observation that voids are observed in ion bombarded nickel at 525 °C but not at 435 °C. A corresponding critical temperature of 225 °C is calculated for a displacement rate of $10^{-7}/\text{sec}$ which is typical of neutron irradiation. This is consistent with the observation that voids are observed at temperatures >250 °C in neutron irradiated nickel.⁽¹⁴⁾ Although this is an admittedly simplified approach, the experimental evidence supports the assumption of vacancy migration and combination with gas to nucleate voids.

B. Defect Lattice

The formation of a void lattice has been reported in a number of ion bombarded and neutron irradiated metals, and the subject has recently been reviewed.⁽²⁴⁾ The observation of a lattice of small dislocation loops in nickel was reported in the above reference.⁽²⁴⁾ Sprague and Smidt have also recently made very similar observations in ion bombarded nickel.⁽²⁵⁾

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No one has reported such a lattice in electron irradiated nickel, however. This would indicate that it is not solely a result of the high dose rate used in bombardment studies.

There appears to be two possible mechanisms which could explain this alignment of defects into an f.c.c. type structure. The small loops can migrate by either self-climb or glide into a position where the interaction force between loops is zero, at which point they will come to rest. The position of the loops at this point may closely represent an f.c.c. arrangement. The other mechanism is similar to that proposed by Foreman to explain the void lattice.⁽²⁶⁾ In this mechanism, the directional migration of interstitial crowdions annihilates the vacancy defects in a preferential manner, allowing the defects to align along close packed directions.

The first mechanism is similar to that discussed by Hudson to explain the alignment of loops in irradiated uranium.⁽²⁷⁾ In this model, the loops glide along the slip cylinder and come to rest at points of zero zlip force. This produces rows of loops in which the crystallography, in uranium at least, corresponded to that predicted.

In nickel, the Frank sessile loops cannot glide, yet there is evidence that the Frank loops are positioned in the spot lattice as well as the glissile prismatic loops. Therefore, if this mechanism is valid for nickel, the loops must position themselves by self-climb rather than by prismatic glide. The mechanism is the same except that the loops migrate by self-climb into a position where the climb force is zero.

We can estimate the interaction distances between loops to determine how effective this mechanism is. The interaction energy between two prismatic loops lying on parallel (111) plane with identical $\frac{a}{2} \langle 110 \rangle$ Burgers vector

is shown in Figure 12 where the lines delineate a surface of constant interaction energy.⁽²⁸⁾ The expression for the velocity of a loop, migrating by self-climb is taken from the work of Eyre and Maher and is given by:⁽²⁹⁾

$$V_L = \frac{B r_o^2 a^2 v_o \exp(-Q_c/kT) (\sin^4 \theta) g(\theta)}{2kT x^4}$$

where: V_L = velocity of the loop

$$B = \frac{\mu b^2}{4\pi(1-\sigma)}$$

$$\mu = \text{shear modulus} = 7 \times 10^{11} \text{ dynes/cm}^2$$

$$b = \text{Burgers vector} = 2.5 \times 10^{-8} \text{ cm}$$

$$\sigma = \text{Poissons ratio} = 0.3$$

$$r_o = \text{radius of stationary loop}$$

$$r_i = \text{radius of moving loop} \cong r_o$$

$$a = \text{lattice parameter} = 3.5 \times 10^{-8} \text{ cm}$$

$$v_o = 1 \times 10^{13} \text{ sec}^{-1}$$

$$T = \text{temperature}$$

$$\theta \text{ and } x \text{ from Figure 12}$$

$$Q_c = \text{activation energy for self-climb} = Q_{S.D.}/2 = 1.5 \text{ e.v.}^{(23)}$$

$$g(\theta) = 3 \sin \theta \cos^2 \phi (1 + 10 \cos^7 \theta - 35 \cos^4 \theta) + 3 \cos \theta \sin 2\phi (4 - 35 \sin^2 \theta \cos^2 \theta) + \sin \theta \sin^2 \phi (1 + \sigma + 10 \cos^2 \theta - 35 \sin^2 \theta \cos^2 \theta)$$

The region of maximum climb motion occurs between $\theta = 75^\circ$ and $\theta = 130^\circ$. A maximum value of $\sin^4 \theta g(\theta)$ was taken as 4. The velocity of the loops for various values of X are shown in Table IV. Since the maximum bombardment times are of the order of thousands of seconds, loops must be closer than

Table IV

$\bar{x}$ (Å)	Velocity (Å/sec) T = 280 °C	Velocity (Å/sec) T = 370 °C
50	4.5×10^{-3}	0.29
100	2.8×10^{-4}	1.8×10^{-2}
200	1.75×10^{-5}	1.1×10^{-3}
300	1.1×10^{-6}	7×10^{-5}

100 Å for significant migration to occur. Therefore, it appears that self-climb is an insignificant contribution to the formation of the lattice.

The self-climb mechanism would be possible if Q_c were much less than 1.5 e.v. The value of one-half the self-diffusion is based on estimation made by other investigations and is related to the diffusion along dislocation lines.⁽²⁹⁾ However, only prismatic loops have been considered in the analysis. It is possible that the faulted Frank loops could self-climb with a lower activation energy. There is no firm data to support this, so this must remain speculation at this point.

The various reasons for and against this mechanism of self-climb can be summarized as follows:

Reasons supporting this mechanism:

- The spacing and perfection of the loop lattice show a temperature dependence. The diffusion energy is also temperature dependent.
- There are several similar observations in irradiated uranium which have been explained by a similar mechanism.

Reasons against this mechanism:

- The calculated interaction distances are much smaller than the observed spacing. A much smaller energy for self-climb would have to be assumed to make the calculation agree with the observations.
- It is difficult to explain the presence of more than one defect per lattice site by this mechanism.

Foreman has proposed a mechanism for the formation of a void lattice in an irradiated metal which is based on the non-random motion of the dynamic crowdion.⁽²⁶⁾ The crowdions move along close pack directions, i.e., $\langle 110 \rangle$ in f.c.c. Due to the nonuniform flux of these interstitials around the void, the void shifts during the irradiation. The voids eventually move into the lattice configuration as a result of shadowing one another from the incoming crowdions. A similar mechanism may explain the loop lattice where the vacancy loops will undergo shifting due to being eroded by the interstitial crowdions. As the loops move into a position in which they can shield one another, they will naturally arrange themselves in an f.c.c. structure. There are several reasons for supporting this particular mechanism:

- 1) It explains the formation of a face centered cubic type lattice,
- 2) It is compatible with more than one defect per lattice site if several defects could be acting together as a shield or sink for the crowdions,
- 3) It does not require any self-climb or glide of the loops, and
- 4) It is compatible with either faulted or unfaulted loops.

However, it is difficult to visualize the loops shielding one another as voids do since the loops are essentially a line sink, whereas the void is a surface sink.

One other possible mechanism relates to impurities. The loop lattice may be nucleated on an impurity lattice. There are some suggestions that the void lattice is purity dependent. However, the lattice is observed in nickel of various purities.⁽²⁵⁾ Figure 13 shows a lattice in Ni-270 which is considerably less pure than the MRC nickel. The lattice does appear to be less perfect in the Ni-270.

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In summary, there is no completely satisfactory explanation for the formation of this loop lattice. At the present time, channeling, crowdion motion, or some type of directional migration along close packed direction appears to be an important factor. This is at least compatible with most of the experimental observations. It is also tempting to speculate that this loop lattice and the void lattice have the same origin.

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FIGURE 1. Displacement Damage (dpa) as a Function of Distance from Bombarded Surface for Nickel Bombarded with 5 meV Ni^{++} Ions to an Ion Dose of 3×10^{16} Ions/cm²

(a) 185°C

(b) 280°C

(c) 370°C

(d) 430°C

FIGURE 2. Microstructure of Nickel After Bombardment with 5 meV Ni^{++} Ions to 10 dpa at the Temperatures Indicated. Marker = 0.1 μ . Foil Normal is [001]

FIGURE 3. Nickel Bombarded with 5 meV Ni^{++} Ions to 10 dpa at 370°C. Same Foil as Figure 2c Tilted to $\sim[110]$ Orientation. Marker = 0.1 μ

FIGURE 4. Nickel Ion Bombarded at 370°C to the Doses Indicated.
Marker = 0.1 μ

FIGURE 5. Nickel Ion Bombarded at 370°C to 0.1 dpa. Same Foil as
Figure 4a Except Much Thicker Foil. Marker = 0.1 μ

FIGURE 6. Dark Field of Ion Bombarded Nickel Showing Black-White Strain Contrast. Foil Normal $\sim$ [001]. Marker = 0.1μ

FIGURE 7. High Magnification of Nickel Ion Bombarded at 370°C . Circles Indicate Possible Stacking Fault Tetrahedra. Marker = 0.05μ

FIGURE 8. High Magnification of Nickel, Ion Bombarded at 370°C Illustrating the Prismatic Dislocation Loops. Foil Normal $\sim[001]$. Marker = 0.05μ

FIGURE 9. Voids and Dislocation in Nickel, Neutron Irradiated at 260°C to a Fluence of $2 \times 10^{20}\text{n/cm}^2$ ($E > 0.1\text{ meV}$). Marker = 0.1μ

FIGURE 10. Clusters and Dislocation Loops in Nickel Neutron Irradiated at 50°C to a Fluence of 3×10^{19} n/cm² ($E > 0.1$ meV). Marker = 0.1 μ

FIGURE 11. Nickel, Ion Bombarded at 370°C Showing Voids. Micrograph Represents Region Approximately 1.2 μ from Bombarded Surface. Marker = 0.1 μ

FIGURE 12. Line of Constant Interaction Energy Between Two Dislocation Loops Having the Loop Orientation as Burgers' Vector as Shown

FIGURE 13. Ni-270, Ion Bombarded at 370°C to 10 dpa Illustrating the Defect Lattice. Marker = 0.1 μ

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