RADIATION DAMAGE IN REACTORS*

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ABSTRACT

Fuel, cladding, and structural materials, moderators, coolants, pressure vessels, and other components of nuclear reactors are bombarded by fast and thermal neutrons (with fluxes up to and above $10^{15} \text{ cm}^{-2} \text{ sec}^{-1}$) at elevated temperatures. Fuel elements are also exposed to fission fragments; all components are exposed to high fluxes of gamma rays. As a result the physical properties of materials are modified, usually in undesirable ways. Brittleness, altered creep, and swelling which leads to complex deformations and induced stress are common problems. A broad understanding of these effects in terms of displaced atoms, point defects, dislocations, disordered regions, voids, and transmutation-induced impurities has been developed. Much empirical information has become available, and means for lessening certain effects have been found. Knowledge of the details of atomic mechanisms, particularly the quantitative aspects, in the broad range of materials of interest is still inadequate, and the body of empirical data, particularly at the higher levels of exposure, needs to be extended. Bombardment of thin specimens with charged particles from accelerators has been a useful tool for simulating high neutron exposure, but the degree to which the simulation is adequate needs further investigation.

INTRODUCTION

Among the various materials problems associated with nuclear reactors, there is one that continues to be of special interest to physicists, namely the influence on solids of nuclear radiation. Beginning with the deleterious effects of fast neutrons on graphite moderator materials, which were first encountered in the early 40's, such problems have been under continuous study for about thirty years. So much has been learned in this time that all problems may appear to be solved, but this is far from true. Indeed, as will be detailed below, major developments continue to occur.¹ In this paper

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†Much of the discussion that follows applies also to materials problems of controlled thermonuclear devices, where damage to walls and structural components by energetic ions and fast neutrons is expected to present serious problems. See Refs. 14 and 15.

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a short survey of effects of current interest will be given. No attempt will be made to treat the subject exhaustively.

The principal cause of radiation effects in reactors is the flux of neutrons throughout the interior of the reactor, and the fission fragments which irradiate fuel and blanket materials. Other charged particle components of the radiation exist but are of relatively minor significance. Gamma rays, which are intense in the core, deposit heat and, in the case of insulating materials, may also produce a part of the radiation damage. The neutrons originate in the fission process where they are born with a spectrum of energies ranging from zero to somewhat above 10 MeV, with a mean energy of about 2 MeV and most probable energy of about 1 MeV. The actual spectrum at any point in the reactor departs from this, generally toward lower energies, depending on the amount of moderation and absorption that has occurred. The greatest intensity of flux, as encountered in the core of a power producing reactor, may be somewhere between 10^{14} and 10^{16} neutrons/cm² sec, with a mean energy of the order of 1 MeV. At the pressure vessel in a power reactor the flux of fast neutrons may be in the range 10^9 to 10^{11} neutrons/cm² sec.

The interactions of neutrons with matter fall into two categories, nuclear scatterings, which produce energetic knock-on atoms (typically with energies of tens to hundreds of keV), and nuclear reactions which produce transmutations. The transmutation-produced impurity atom of perhaps greatest technological significance is helium, produced by (n, α) reactions. Numbers vary enormously with conditions, but typically 1 to 100 atomic parts of helium per million will be produced in stainless steel components in the core of a reactor in a few years of operation. Reactions of neutrons with Ni, Fe, Cr, and such impurities as nitrogen and boron all play a role in helium production. Hydrogen is also produced in substantial quantities by (n,p) reactions, but because of its higher diffusivity it escapes more rapidly and therefore usually is of less significance.

The other products of neutron interaction, namely energetic knock-on atoms, are the starting point for most of the damage to the crystalline lattice. Such a knock-on initiates a collision cascade in which, typically, hundreds of atoms in a region some hundreds of Angstroms in dimensions participate. The initial product of such a cascade is a large sprinkling of atoms that have been displaced from their lattice sites to interstitial positions, and an equal number of vacancies scattered over the region of the cascade. Some of these point defects will find themselves in small clusters, such as di-vacancies, tri-vacancies, di-interstitials, etc., which have distinctive behavior. Immediately afterward, thermally induced rearrangement, annihilation, and migration sets in, so that the

final state depends upon temperature, impurity content, and other variables. Fig. 1 illustrates a typical cascade, as found by Beeler¹⁶ from a somewhat simplified computerized model.

Fig. 1. Typical collision cascade in α -iron as calculated from a binary collision model by Beeler. Three-dimensional trajectories of atoms are shown projected onto a (001) plane. The short heavy line with the kink in the middle is the primary knock-on that initiated the cascade (arbitrarily assumed to start with 5 keV). Trajectories of secondary knock-ons are shown as heavy dotted lines emanating from the primary track. Thereafter alternately dashed and solid lines represent the trajectories of successively higher order knock-on atoms.

Other computerized calculations^{17,18} have given realistic information on cascades in simple crystals at low or moderate energies. All of the studies together allow one to understand reasonably well the initial state of damage following a primary knock-on event in a wide range of conditions.^{19,45} The interactions between defects and their thermally-activated rearrangements are less well agreed upon.

When the defects produced by the primary knock-on are numerous and close together, which is particularly true for energetic events in substances of high atomic weight, collective descriptions of the damage process, such as the so-called displacement spike models, may

have limited usefulness. Such models remain highly qualitative and rather controversial.

Fission fragments, being intermediate mass ions of very high energy, produce a high density of displaced atoms along thin short paths (1 to 10 μm in length). They are responsible for intense radiation effects in fuel and blanket materials. Limitations of space prevent discussion of these rather special problems here. The interested reader is referred to the standard texts.²⁻⁷

MECHANICAL EFFECTS

From early days in the study of radiation effects, it has been known that heavy exposure of metals to fast neutrons produces hardening and embrittlement.³ Typical stress-strain curves before and after irradiation are shown in Fig. 2. Irradiation raises the

Fig. 2. Stress-strain curves typical of many structural metals before and after irradiation.

yield stress and ultimate stress and reduces the ductility. Generally such effects set in at lower fluences in simple annealed metals than in metals strengthened by cold working or precipitation. In ferritic steels the ductile-to-brittle transition temperature is usually raised by irradiation, and the ductility above the transition temperature is lowered. In a general way it is clear that the defects produced by irradiation will tend to increase the yield stress by locking dislocations, an effect in pure metals similar in many respects to that of alloying. The effects in the work-hardening range and beyond are more complex, and while many models have been advanced to explain observations, these

are qualitative and rather speculative. A mass of empirical data on structural alloys has been accumulated, but is not well coordinated by quantitative theoretical understanding. Generally radiation embrittlement imposes the most serious limitations on two structural components of reactors: fuel cladding, which may be caused to leak or rupture, and pressure vessels for water reactors, where the useful lifetime of the vessel may be reduced.

Recently a valuable development has occurred. Certain residual elements, copper and phosphorus, have been found to cause radiation brittleness in pressure vessel steels. It has been found that elimination of these elements virtually eliminates radiation embrittlement at temperature and fluences of interest in pressurized water reactors and boiling water reactors.^{20,23} Fig. 3 shows brittleness before and after irradiation in three samples of steel which differ in their phosphorus and copper content.

Fig. 3. Comparison of radiation embrittlement (as measured by Charpy-V notch 5.2 kgm/cm² transition temperature) of steels with varying amounts of residual elements (see text).

The curves show Charpy-V notch energy as a function of test temperature. The upper set of curves is for a commercial sample of A302B steel plate. The unirradiated specimen shows the typical transition from brittle behavior below about -50°C to ductile behavior above about +50°C. After irradiation to 3.1×10^{19} fast neutrons/cm² at 288°C, the curve is shifted to the right by ~92°C, and the ductility at higher temperature is distinctly

impaired. The middle curves show the same tests on a special heat of similar steel with a lower content of residual elements. The increase of transition temperature with irradiation is now only 75°C, and the high-temperature ductility is almost unaffected. The bottom set of curves show results on a special batch of steel with very low residuals. The irradiated and unirradiated specimens are now essentially indistinguishable. The identification of phosphorus and copper as the elements responsible has been completed in further extensive experiments with this and several other steels (summarized in ref. 23). The reasons for this effect have not been determined, although one suggestion is that copper interacts with vacancies to form defect complexes which are more numerous and more stable than in the pure steel, and that this retards the dynamic recovery process during irradiation.

Irradiation affects high temperature creep and creep rupture properties of steels, usually increasing the creep rate (although it also can decrease it) and reducing the creep rupture strength. Irradiation provides several mechanisms by which creep can be altered, but the total picture is complex. For discussions of these effects the reader is referred to Bush,⁸ Reuther and Zwilsky,²³ Gilbert and Harding,²⁴ and King, et al.²⁵

VOIDS AND SWELLING

Of the point defects introduced into metals by irradiation, some (the interstitials) are mobile at quite low temperatures. Single vacancies become mobile around 100°C to 200°C, and small clusters of vacancies have widely varying degrees of mobility.²⁶ While some clusters are known to be stable to higher temperatures, and voids have been seen in irradiations at lower temperature, there were grounds for believing that irradiations of structural alloys at high temperatures (e.g. 500°C) would produce little damage because of the simultaneous annealing that would occur.* Thus it was a surprise when Cawthorne and Fulton in 1966 reported pronounced swelling, with the formation of microscopic voids, in stainless steel irradiated to high fluences at elevated temperature. Since this observation, which was published²⁸ in 1967, an immense amount of work in many countries has been done on all aspects of the

*Barnes and Mazey²⁷ showed in the early 1960's that bombardment of copper with α -particles from a cyclotron at certain temperatures could produce voids stabilized by gaseous helium, known as bubbles. They developed a theory for nucleation, growth, coalescence, and migration of helium bubbles in copper which later proved to be of service in understanding high-temperature helium embrittlement in structural alloys and fission-product swelling in fuels.

phenomenon of swelling and void formation.^{11, 12, 29} For the reactor designer, particularly in the case of fast reactors with their very close mechanical tolerances, this swelling phenomenon is a serious problem which requires costly modifications of design and limitations on performance. Huebotter and Bump²⁰ estimated that, to accommodate only 5% swelling in breeder reactors over the period 1970 to 2020 had a 1970 present worth of from \$860,000,000 to \$5,600,000,000 relative to the case of 15% swelling. Thus the impetus for seeking low-swelling alloys is great.

It is now known that most metals heavily irradiated at temperatures in the range approximately $0.3 T_m$ to $0.6 T_m$, where T_m is the absolute melting temperature, show the swelling phenomenon. Irradiations above or below this temperature range do not produce it. The swelling sets in after a threshold fluence, which depends on material and conditions, and increases thereafter, through the range 10^{20} to 10^{23} fast neutrons/cm². In early stages the volume is often observed to increase more rapidly than linearly with fluence above threshold, and at very high doses (experimentally achievable with ion bombardment rather than neutrons) a tendency to saturate is seen. Volume increases may ultimately reach 10 to 20%, or more. As an example, Fig. 4 shows the density change, $-\Delta\rho/\rho$, as a function of fluence in

Fig. 4. Density change, $-\Delta\rho/\rho$, as a function of fluence in aluminum and several of its alloys irradiated at about 50°C (after Jostsons et al.³¹).

aluminum and several of its alloys, irradiated at temperatures above 50°C.

Specimens that have swollen in this manner contain a large number of microscopic voids, clearly visible in the electron microscope. Fig. 5 is a rather typical electron micrograph of aluminum (1100 grade) irradiated to a fluence of 2.8×10^{22} neutrons/cm² ($E > 0.1$ MeV) at 50°C. The voids in this picture range from 100 Å to 1000 Å in diameter. Note that they tend to have cubic shapes and are distributed apparently at random.

Fig. 5. Electron micrograph of rather heavily irradiated aluminum showing voids within the grains and silicon precipitation at the grain boundary (after Jostsons et al.³²).

In the earliest stages of void production, the voids are very small. With further irradiation the number of voids per unit volume increases, and their mean size increases. It is typical for the voids to have geometrical shapes determined by low index (i.e. low energy) plans of the metal. At high temperatures of irradiation, for a given fluence the voids tend to be larger but less numerous. Thus, Fig. 6 shows the minimum void size observed in 304 stainless steel vs. irradiation temperature; Fig. 7 shows the void density for the same material and set of irradiation temperatures. It is observed that voids are absent or less numerous near grain boundaries, and that copious tangles of dislocations grow in at the same time as the voids.

Fig. 6. Size of smallest voids observed in irradiated 304 stainless steel vs. temperature of irradiation (after Appleby et al.⁴²).

Fig. 7. Density of voids in irradiated 304 stainless steel vs. temperature of irradiation (after Appleby et al.⁴²).

As a result of the many investigations of void growth and swelling in the last seven years, the qualitative nature of the phenomenon and its scope are well known. The mechanism responsible is widely believed to be the following: Vacancies and interstitials are produced in equal numbers in the radiation damage cascades, and at the temperatures in question migrate rapidly to sinks. Although this process is approximately symmetrical between vacancies and interstitials, it is not exactly so, and the imbalance is crucial. Interstitials are attracted more strongly to dislocations, where they are absorbed, than are vacancies (by reason of the larger strain field around the interstitial). Moreover, even in regions far from dislocations interstitials migrate more rapidly than vacancies.* Thus an excess of vacancies is built up after the irradiation starts, and eventually reaches a sufficiently high supersaturation to cause nucleation and growth of vacancy clusters, i.e., voids. The nucleation process itself remains controversial, and indeed several different nucleation mechanisms may operate. Homogeneous nucleation may occur, if the supersaturation is sufficiently high; heterogeneous nucleation probably is more common, and this may be brought about by existing impurities or precipitates, by small heavily damaged regions (displacement spikes, etc.) introduced by the radiation, or by residual gases in the metal and gases, notably helium, produced by the radiation. That displacement spikes and transmutation gases are not essential is proved by the production of voids in very pure metals by electron irradiation, which is incapable of producing either spikes or transmutations. On the other hand, injection of helium along with or prior to irradiations has been shown to lower the threshold exposure at which voids are formed in some metals. This shows that gases can promote void formation. A second major issue, not yet well resolved, is the question of why the small clusters of vacancies do not collapse into dislocation loops. The presence of enough helium (or other gas) in the

*If the mean residence time of a defect, before annihilation or removal is τ , and the rate of production of this defect, per unit volume, is $\dot{n}$, the mean concentration of the defect under steady state condition is $\dot{n}\tau$. The faster migration of interstitials, in addition to their attraction to dislocations, causes τ for interstitials to be less than τ for vacancies and thus lowers the relative concentration of the former.

nucleated cluster would prevent collapse.* In some cases there appears to be a geometric metastability of the small vacancy cluster which is sufficient. Growth of the void nucleus clearly takes place by the continued inflow of vacancies produced in the course of the irradiation. Also in the growth stage both an excess of vacancies over the equilibrium number and an excess over the number of interstitials continue to be necessary.

Several quantitative models of void production have been developed and are reasonably successful in describing the simpler observations, although a rather large number of parameters is required that are not well determined by independent means.

The observation, at first puzzling, that void swelling occurs only in a certain range of temperatures ($\sim 0.3 T_m$ to $0.6 T_m$) can now be understood: At too low a temperature vacancies are not mobile and thus cannot aggregate. At too high a temperature the rapid motion of vacancies to existing sinks prevents the buildup of the necessary supersaturation (moreover, the equilibrium concentration, which must be exceeded, increases rapidly with temperature). The higher the flux of radiation the higher the range of temperatures in which supersaturation can be achieved. In agreement with this, experiment shows that the upper limiting temperature for swelling is flux dependent. In ion bombardments, where the damage rate is much higher than in reactors, swelling occurs in a temperature range with a substantially higher maximum.

The swelling phenomenon, since it occurs at temperatures that are common in modern power reactors, and in materials such as stainless steels which are employed for numerous structures in and near the core of the reactor, presents problems to the reactor designer.³⁰ In the liquid metal fast breeder reactor, fuel is contained in long cylindrical stainless steel tubes, typically 1/4 in. in diameter and about 9 ft long. These fuel "pins" are formed into bundles about 5 in. in diameter, and each bundle is contained in a larger tube of stainless steel. The bundle is inserted into the reactor and removed as a unit, and a typical large reactor will

*That an appreciable amount of helium is present in voids after some irradiations has been proved in ingenious experiments of Katyal, Keesom, and Cost.³³ Aluminum was irradiated with cyclotron alpha particles to produce concentrations up to 300 ppm of helium. After annealing at temperatures above 600°C, which produced voids visible in the electron microscope, the specific heat of the aluminum sample was measured as a function of temperature from 2°K upward. A pronounced peak in the specific heat centered on 4.2°K was observed. This is clearly caused by the liquid-vapor transition of the helium in the bubbles.

contain about 500 bundles. Liquid sodium, which is the heat transfer medium, flows upward through each bundle. The bundles are supported and positioned by a pair of grids (known as grid plates) lying in horizontal planes, one above the other. Without belaboring the details, it is clear that operation of the reactor and fuel handling both require that rather close mechanical tolerances in this assembly be maintained. The fuel elements must retain their integrity for the life of a core, perhaps a year, while the grid plates and other permanent structures must retain their integrity for the life of the reactor. Since the neutron flux is not uniform everywhere, one side of a typical fuel assembly will be subjected to more fluence and hence more swelling than the other. This produces bowing of the long assembly which imposes undesirable limits on the lifetime of the core. Unequal flux on the two sides of each grid plate tends to cause buckling of the grid plate. For reasons such as these it is urgent to develop alloys with minimal swelling.

From the understanding of the mechanism sketched above, some means of accomplishing this suggest themselves: Additional sinks for vacancies, or increased mobility of vacancies, or reduction of the preferential trapping of interstitials at dislocations will reduce the excess concentration of vacancies; traps for interstitials can hold up the migrating interstitial and simultaneously constitute a trap for vacancies; perhaps some conditions can be found which hinder the nucleation of voids or even destabilize the growing voids; and finally, particular alloys may be found for which the swelling rate at desired operating conditions is lower because of altered values of all the relevant parameters.

Adda³⁴ has reported that in nickel and copper the addition of elements that reduce the stacking-fault energy minimizes swelling. One explanation is that this reduces the strain field around dissociated dislocations, reducing their preference for interstitials. In the case of addition of Al to Cu, one case in point, the mobility of vacancies is also increased, which would cause the maximum swelling to occur at a lower temperature.

Certain minor changes in the composition of stainless steels have produced substantially improved resistance to swelling.

Bramman, et al.³⁵ suggest that very stable, finely dispersed coherent precipitates, such as TiC, NbC, and γ' phase, can produce traps which promote recombination of Frenkel pairs thus reducing the concentrations of vacancies. They also show that cold working in such steels is beneficial, maximum swelling rates in 20% cold worked material being some two to three times lower than those in the same material after solution treatment. Straalsund³⁶ et al. confirm that austenitic stainless steel subjected to cold work

shows negligible void formation after irradiation to 3×10^{22} neutrons/cm² in the temperature range 370°C to 850°C. Bloom³⁷ has noted lower swelling in 10% cold-worked 304 stainless steel after similar irradiations. There is a discrepancy, however, between the microstructures reported by these various investigators, Bramman reporting that the swelling reduction from cold work is due to a reduction in size, rather than number of voids, while Straalsund and Bloom attribute the effect to a reduction by factors of 20 to 50 in the number of voids with little change in their size. The reduction in number of voids could be most readily explained on the assumption that the greater number of traps in the cold worked material reduces the concentration of vacancies attained, which inhibits the nucleation process.

Laidler³⁸ has investigated a niobium-stabilized austenitic stainless steel, AISI 348, in which the niobium carbide can be caused to precipitate in association with stacking faults to provide a high concentration of sinks for point defects (as well as improved tensile and creep properties at elevated temperature). It is hoped that in this material void formation, and swelling, can be strongly suppressed.

Farrell et al.³⁹ investigated void formation in aluminum and some aluminum alloys. They found that 6061 and 6063 aluminum alloys are much more resistant to void formation than commercial grade 99.0% purity aluminum (1100 aluminum). The former are precipitation-hardened alloys containing up to 1.2% magnesium and 0.8% silicon. They contain very finely divided coherent Mg₂Si precipitates interlaced with dislocations.

An alternative explanation for the effect of cold work is advanced by Buswell et al.⁴⁰ who argue that at high enough dislocation density, dislocations are inhibited from climbing by mutual interference which causes them to cease to perform as sinks for interstitials. They also point out that coherent precipitates will similarly inhibit dislocations from functioning as sinks for interstitials, but that incoherent precipitates provide plentiful sinks at their interfaces with the lattice. Thus coherent precipitates discourage void growth, but overaging to the stage where the precipitates become incoherent reverses the inhibition.

A curious phenomenon sometimes observed when voids are produced by radiation is the formation of a superlattice of the voids. Wiffen⁴¹ reports such effects in molybdenum irradiated at 585°C

and niobium irradiated at 790°C , in both cases to a fluence of 2.5×10^{22} neutrons/cm² (>0.1 MeV). Two of Wiffen's electron micrographs are shown in Fig. 8.

Fig. 8. Examples of superlattices of voids in (a) molybdenum irradiated at 585°C and (b) niobium irradiated at 790°C . The fluence for both samples was 2.5×10^{22} neutrons/cm², $E > 0.1$ MeV (after Wiffen⁴¹).

Although the superlattice is not perfect, its regularity is quite remarkable, and there is a clear alignment with the underlying crystal lattice. Other observations of void superlattices have been reported by Kulcinski et al.⁴³ and Evans et al.⁴⁴ The latter group bombarded molybdenum foils with 2 MeV N^+ ions at 870°C to the rather high dose whereby an average of 100 displacements were experienced by each atom. Fig. 9 shows two electron micrographs at different orientations of the resulting arrays.

Fig. 9. Superlattice of voids in molybdenum after heavy bombardment with N^+ ions at 870°C. Left photo shows projection of void array in $[010]$ direction, right shows projection in $[111]$ direction. The array is b.c.c. (after Evans et al.⁴⁴).

Considerations of elastic interactions between voids in an anisotropic medium show that such arrays minimize the interaction energy. Predictions of the superlattice constant can be made from such considerations and are in semiquantitative agreement with the observations. The process by which such a lattice is formed is not entirely clear, however, and offers an interesting field for further work that perhaps would illuminate all of the pro-

cesses by which voids are formed and grow. The metals in which such superlattices have been observed are substances of high elastic anisotropy, and for the most part body-centered cubic superlattices are found in body-centered cubic alloys and face-centered cubic superlattices are found in face-centered cubic alloys.

USE OF ION BOMBARDMENT

It has already been remarked that bombardments of thin specimens with energetic ions can produce effects similar to neutron bombardment. The ion bombardment technique, in addition to its intrinsic interest, is of importance as a means of simulating heavy exposure to fast neutrons in much less time and at lower cost, with the further advantages that the irradiated specimens are not highly radioactive, as is the case after prolonged reactor irradiations, and the ambient conditions (temperature, etc.) are more easily controlled during irradiation. At least three disadvantages to this simulation should also be recognized: the spectrum of knock-on atoms is never precisely the same as with neutrons; transmutation effects are absent; and, because of the short ranges of heavy ions of appropriate energies, the specimens must be very thin, which means that disturbance from proximate surfaces must be guarded against.

As an illustration, 5 MeV nickel ions have a range of slightly

less than 2 μm in metallic nickel, and 1.3 MeV protons have a range of only 10 μm in the same substance. MeV neutrons, on the other hand, penetrate a mean distance of tens of centimeters before suffering a collision. For comparing exposures it is conventional to calculate the mean number of times during a bombardment that each atom in the specimen is displaced from its lattice site to an interstitial position (from which it usually finds its way back into a lattice site by thermally activated processes before being displaced again). Kulcinsky et al.⁴³ have given a useful chart of this index of relative effectiveness of different types of radiation applied to nickel (Fig. 10).

Fig. 10. Maximum damage rates available by various energetic particle bombardments of nickel, measured as displacements per atom per second (after Kulcinski et al.⁴³).

The rates of production of displacements per atom per second for various kinds of radiation are shown. For the charged particle radiations the calculations are for presently attainable fluxes (in some cases limited by allowable heating of the sample). This figure shows that damage with ions can be accumulated at rates 2 to 5 orders of magnitude higher than with neutrons.

A further elaboration of the ion bombardment technique is to bombard, simultaneously or consecutively, with impurity ions and with ions of the same chemical species as the sample, thus simulating the principal effects of transmutation. This method has been employed rather frequently to introduce helium in studies of void formation.

From one point of view the ultimate in simulation would be to bombard a very thin specimen with ions of the same species provided with the same spectrum of energies as the primary knock-ons that would be produced by neutrons (allowing, of course, for the spectrum of energies of the neutrons being simulated). For elements as heavy as iron and fission neutrons, however, this presents grave difficulties, for the required ions are of such low energy that their penetration is in the sub-micron range. Thus 1 MeV neutrons on iron produce primary knock-ons in iron with a mean energy of about 30 keV. A second approach is to use protons of about the same energy as the neutrons. The energy transfer cross section for MeV protons on many nuclei is not enormously different from that of neutrons of the same energy, although it is not identical. However, unless the specimen is much thinner than the range of the proton (the order of 10 μ m), degradation of energy of the proton in the specimen will occur, producing a gradient of damage and possibly retention of hydrogen. Still another approach is to bombard with ions of higher energy, use the observations to determine essential parameters in a model of the damage process, and then use the model to predict corresponding damage by neutron irradiation. Obviously none of these approaches is ideal, but the latter may eventually prove to be the most powerful. Clearly a large field remains open for further investigation.

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REFERENCES

1. General discussions of radiation damage will be found in references 2-7. Extensive treatments of materials problems of reactors are given in references 8-13.
2. F. Seitz and J. S. Koehler, "Displacement of Atoms During Irradiation," Solid State Physics (Academic Press, Inc., N.Y., 1956) Vol. 2, p. 305.
3. G. J. Dienes and G. H. Vineyard, Radiation Effects in Solids (Interscience Pub., New York, 1957).
4. D. S. Billington and J. H. Crawford, Jr., Radiation Damage in Solids (Princeton University Press, Princeton, N.J., 1961).
5. J. J. Harwood, H. H. Hausner, J. G. Morse, and W. G. Rauch, editors, The Effects of Radiation on Materials (Rheinhold Pub. Corp., New York, 1958).
6. G. Leibfried, Bestrahlungseffekte in Festkörpern; eine Einführung in die Theorie (B. G. Teubner, Stuttgart, 1965).
7. M. W. Thompson, Defects and Radiation Damage in Metals (Cambridge U. Press, London, 1969).
8. S. H. Bush, Irradiation Effects in Cladding and Structural Materials (Rowman and Littlefield, New York, 1965).
9. "Irradiation Effects in Structural Alloys for Thermal and Fast Reactors," ASTM Symposium San Francisco, 1968 (Special Technical Publication 457, ASTM, Philadelphia, 1969).
10. Symposium on Materials Performance in Operating Nuclear Systems, M. S. Wechsler and W. H. Smith, editors, Ames, Iowa, Conference 1973 (USAEC, TN 37830).
11. "Voids Formed by Irradiation of Reactor Materials," S. F. Pugh, M. H. Loretto, and D. I. R. Norris, editors, Proceedings of the British Nuclear Engineering Society, Harwell, England, March, 1971.
12. "Radiation-Induced Voids in Metals," J. W. Corbett and L. C. Ianniello, editors, Proceedings of an International Conference, Albany, N.Y., June, 1971 (USAEC CONF-710601).
13. International Meeting of Fast Reactor Fuel and Fuel Elements, M. Dalli-Donne, K. Kummerer, and K. Schroeter, editors (Gesellschaft f. Kernforschung, Karlsruhe, West Germany, 1970).

14. Critical Questions in Fundamental Radiation Effects Research, L. C. Ianniello, editor (USAEC Report Wash 1240-73).
15. G. L. Kulcinski, "Materials Problems for Fusion Reactors," Materials Engineering Congress, Cleveland, 1972 (to be published).
16. J. R. Beeler, Jr., Phys. Rev. 150, 470 (1966).
17. J. B. Gibson, A. N. Goland, M. Milgram, and G. H. Vineyard, Phys. Rev. 120, 1229 (1960).
18. G. H. Vineyard, "Computer Experiments with Lattice Models," Interatomic Potentials and Simulation of Lattice Defects, P. C. Gehlen, J. R. Beeler, and R. I. Jaffe, editors (Plenum Pub. Corp., New York, 1972), pp. 3-25.
19. M. T. Robinson and I. M. Torrens, "Computer Simulation of Atomic Displacement Cascades in Solids in the Binary Collision Approximation" (to be published).
20. U. Potapous and J. R. Hawthorne, Nuclear Applications 6, 27 (1969).
21. L. E. Steele, Atomic Energy Review 3, 50 (1969).
22. J. R. Hawthorne, E. Fortner, and S. P. Grant, Welding Journal Research Supplement 35, 553 (1970).
23. T. C. Reuther and K. M. Zwilsky, "The Effect of Neutron Irradiation Toughness and Ductility of Steels," Proceedings of Kyoto International Conf., Oct. 25-26, 1971.
24. E. R. Gilbert and N. E. Harding, ref. 9, pp. 17-37.
25. R. T. King, K. Farrell, and A. E. Richt, ref. 10, pp. 133-152.
26. A. C. Damask and G. J. Dienes, Point Defects in Metals (Gordon and Breach, New York, 1963).
27. R. S. Barnes and D. J. Mazey, Proc. Roy. Soc. A275, 47 (1963).
28. C. Cawthorne and E. J. Fulton, Nature 216, 515 (1967).
29. A. L. Bement, Jr., "Void Formation in Irradiated Austenitic Stainless Steels," Advances in Nuclear Science and Technology, E. J. Henley and J. Lewins, editors (Academic Press, New York) Vol. 7, pp. 1-120.

30. P. R. Huebotter and T. R. Bump, ref. 12, pp. 84-124.
31. A. Jostsons, E. L. Long, Jr., J. E. Furguson, and K. Farrell (to be published).
32. A. Jostsons, E. L. Long, Jr., J. O. Stiegler, K. Farrell, and D. N. Braski, ref. 12, pp. 363-384.
33. O. P. Katyal, P. H. Keesom, and J. R. Cost, ref. 12, pp. 248-254.
34. Y. Adda, ref. 12, pp. 31-83.
35. J. I. Bramman, K. Q. Bagley, C. Cawthorne, and J. E. Fulton, ref. 12, pp. 125-141.
36. J. L. Straalsund, H. R. Brager, and J. J. Holmes, ref. 12, pp. 142-155.
37. E. E. Bloom, ref. 12, pp. 1-30.
38. J. J. Laidler, ref. 12, pp. 174-186.
39. K. Farrell, J. T. Houston, A. Wolfenden, R. T. King, and A. Jostsons, ref. 12, pp. 376-385.
40. J. T. Buswell, S. B. Fisher, J. E. Harbottle, D. I. R. Norris, and K. R. Williams, ref. 12, pp. 533-549.
41. F. W. Wiffen, ref. 12, pp. 386-396.
42. W. K. Appleby, D. W. Sandusky, and U. E. Wolf, ref. 12, pp. 156-173.
43. G. L. Kulcinski, J. L. Brimhall, and H. E. Kissinger, ref. 12, pp. 449-478.
44. J. H. Evans, R. Bullough, and A. M. Stoneham, ref. 12, pp. 522-532.
45. D. G. Doran, Trans. Am. Nucl. Soc. 17, 207 (1973).