Letters to the Editor

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On Spin Dependent Nuclear "Radii"

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IN applying the Breit-Wigner scattering formula to low energy neutron scattering ($J=I\pm\frac{1}{2}$) the nuclear radius, R, enters as a potential scattering amplitude. At thermal energies the total scattering cross section, σ_s, is given by:

$$\sigma_s/4\pi=g_+|R+a_+|^2+g_-|R+a_-|^2$$

where the + and − refer to states having total angular momenta J_+ and J_-, respectively. The a_+ and a_- are sums of all resonance amplitudes of states having J_+ and J_-, respectively, and the g_+ and g_- are statistical weight factors corresponding to the J.

It is possible that the apparent size of the nucleus (i.e., R, the range of the infinitely repulsive potential) may depend upon the incoming neutron spin orientation. This would yield an R_+ and an R_- and require σ_s to be given by

$$\sigma_s/4\pi=g_+|R_++a_+|^2+g_-|R_-+a_-|^2.$$

Similarly, the incoherent (spin effect) thermal scattering cross section, σ_{inc}, would be given by

$$\sigma_{\text{inc}}/4\pi=g_+g_-|R_++a_+-R_--a_-|^2.$$

For isotopes in which there are no significant resonance contributions to the thermal scattering cross section, this reduces to

$$\sigma_{\text{inc}}/4\pi=g_+g_-(R_+-R_-)^2.$$

It is an observed fact that materials having no significant low energy neutron resonance scattering also have no incoherent scattering.[1] From this it immediately follows that the *off-resonance* radii, R_+ and R_-, are equal. However, such may not always be the case for materials having low energy neutron resonances. In the cases of vanadium[2] and manganese[3] it is possible to fit observed cross-section data using only a single nuclear radius. In the cases of sodium[4] and cobalt[5] a single radius fit does not seem possible. This implies that the radius appearing in the resonance formula may not, in general, be given by the $A^{\frac{1}{3}}$ rule or even be given uniquely; rather, it should be treated as a double parameter.

[1] Compilation by E. O. Wollan, unpublished. (See odd Z or odd N—e.g. Be, F, Al, Nb, Bi.)
[2] M. Hamermesh and C. O. Muehlhause, Phys. Rev. **78**, 175 (1950).
[3] Harris, Hibdon, and Muehlhause (to be published).
[4] Hibdon, Muehlhause, Selove, and Woolf, Phys. Rev. **77**, 730 (1950).
[5] Harris, Muehlhause, and Thomas, Phys. Rev. **79**, 11 (1950).

The Generation of Vacancies by Dislocations

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SEVERAL processes have been observed which require the generation of large numbers of vacancies within crystals: (1) The darkening of the alkali halides by x-rays and cathode rays.[1] Apparently somewhat over 10^{18} and probably as many as 10^{19} F-centers per cc can be generated in this way; (2) Gyulai and Stepanow have observed[2] an increase in ionic conductivity during plastic flow in NaCl that implies the generation of the order of 10^{18} vacancies per cc; (3) The Kirkendall effect[3] (relative motion of markers during diffusion experiments by a process presumably related to plastic flow) in metals implies the generation of at least 10^{20} and probably as high as 10^{22} vacancies per cc within a small volume. The Kirkendall effect has not yet been observed in single crystals, so the generation of vacancies may occur at grain boundaries; however, it seems reasonable at this stage, to assume that it is a relatively universal phenomena and will be found in diffusion within well-made single crystals. Both process (1) and (2), particularly the latter, seem to require the generation of far more vacancies than can be explained by assuming the only source is the group of vacancies frozen-in on cooling from the melting point,[4] unless the dislocations formed by these can act as catalysts for very extensive generation.

Apparently it is necessary to determine a mechanism by which a very large number of vacancies can be generated within a small volume of a crystal, for example within a volume of the order of mosaic dimensions (10^{-13} cm^3). A clue concerning such a mechanism is given by the following consideration. Suppose the crystal contains a small "patch" or disk of interstitial atoms which are arranged as part of an extra interstitial plane, perhaps five or so atoms in linear dimensions, which is in registry with its neighbors. This disk constitutes a small general dislocation ring having a finite projection on a plane normal to its Burgers vector. It can grow by taking atoms from the neighboring completed planes, between which it is sandwiched, thereby forming vacancies. Given one such disk more or less, on each plane, in the volume under consideration, an almost unlimited number of vacancies could be generated, assuming the temperature is adequate to furnish the activation energy which should be nearly the same as that required to form a vacancy near the surface. Isolated disks of this type cannot be expected to occur very generally in a crystal in which vacancies are the predominant lattice defect. Even if a relatively small number of Frenkel defects were produced at elevated temperatures, the vacancies which are formed in greater numbers would annihilate the interstitial atoms during cooling. On the other hand, a dislocation which runs through a slip plane passing across the volume apparently can provide the equivalent of a large number of disks of interstitial atoms. This dislocation may, for example, be one of the type which promote crystal growth through Frank's mechanism,[5] in which its Burgers or screw character is of importance. In general a dislocation of this type will meander through the slip plane, changing its character from Burgers to Taylor-Orowan (edge) type and back again along its path. Each segment having Taylor-Orowan type may act as a source of vacancies. The generation of vacancies will cause the dislocation line to move out of the slip plane by growth of the "extra" plane. The line will loop around the edge of this plane. In other words, each kink in the dislocation line at which it possesses a strong component of Taylor-Orowan Character has the properties of a disk of interstitial atoms. Should a very large number of vacancies be required, as during a diffusion process in which the Kirkendall effect is prominent, there seems to be no reason why a large number of kinks of this type should not occur. For example, growth of the extra planes at the kinks initially present will have the effect of producing motion of the portions of the dislocation in the slip plane in a direction parallel to the extra planes (normal to the Burgers vector). Additional kinking can occur at points where the dislocation becomes stuck. Such sticking may occur at F-centers or V-centers, when these are being generated, or at regions of inhomogeneity during the diffusion process.

It also seems possible to employ the endless screw character of a Burgers dislocation to generate an endless spiral of interstitial planes from a single small disk of interstitial atoms which form a part of an interstitial plane that is in registry with its neighbors. The first time such a disk has grown around the dislocation line, the two ends which meet in a plane passing through the axis of the screw are separated by a plane of the screw, equal in thickness to the Burgers vector of the screw. There will be a tendency for the two ends to glide toward one another, causing, in effect, a split of the intervening plane, each edge of the extra plane joining one-half of the split. This process can occur even if the Burgers vectors associated with the extra plane and with the screw dislocation are different. However, if the extra plane involves only a few atoms, an activation energy will be required to produce the split in the intervening plane. Should this not be supplied, the two ends may continue growth past the meeting point without joining the regular lattice, in which case a three-dimensional spiral extra plane can form. This spiral resembles somewhat the spiral surfaces employed by Frank[5] in his treatment of crystal growth, although the dislocation considered here is profoundly different in the sense that the layers of the spiral are separated by the planes of the original screw during the early state of formation, when the radius of a given arm of the spiral is of the order of a few atomic dimensions. Once the spiral has grown sufficiently in the direction normal to the axis of the initial Burgers dislocation, the planes will assume the normal characteristics of dislocations and can move in the direction of the Burgers vector associated with them in such a way as to establish a more nearly equilibrium configuration. The edges of the spiral extra plane, which constitutes the dislocation line, may be expected to spiral on the surface of a double "cone" formed by abutting the bases of two cones whose apexes point in opposite directions and whose common axis lies along the axis of the original Burgers dislocation. We shall call these dislocations *spiral prismatic dislocations* since they possess a strong component of what the writer[6] has termed prismatic character. The formation of a dislocation of this type, and its growth by the borrowing of atoms from neighboring planes, evidently can yield an almost unlimited supply of vacancies. It should be added that the initial interstitial disk from which the spiral dislocation can be generated may be replaced by an edge dislocation which lies in a plane cutting the initial Burgers dislocation obliquely and which grows in such a manner as to intersect the Burgers dislocation by addition of atoms to the extra plane associated with the edge dislocation.

[1] F. Seitz, Rev. Mod. Phys. **18**, 384 (1946); Estermann, Leivo, and Stern, Phys. Rev. **75**, 627 (1949); Casler, Pringsheim, and Yuster, J. Chem. Phys. **18**, 887 (1950); H. U. Harten, Zeits. f. Physik **126**, 619 (1949).
[2] Z. Gyulai and D. Hartly, Zeits. f. Physik **51**, 378 (1928); Z. Gyulai, Zeits. f. Physik **78**, 630 (1932); A. W. Stepanow, Zeits. f. Physik **81**, 560 (1933); also a coordinating paper by the writer to appear in the Physical Review.
[3] A. D. Smigelskas and E. O. Kirkendall, Trans. A. I. M. E. **171**, 130 (1947); L. S. Darken, Trans. A. I. M. E. **175**, 184 (1948); J. Bardeen, Phys. Rev. **76**, 1403 (1949); F. Seitz, Acta Crys. **3**, 335 (1950). The experimental work has been extended in a very important way by da Silva and Mehl (as yet unpublished work) which the writer has had the privilege of reading.
[4] F. Seitz, Phys. Rev. **79**, 890 (1950).
[5] F. C. Frank, discussion of the Faraday Society, No. 5 (1949), p. 48.
[6] F. Seitz, Phys. Rev. **79**, 723 (1950).

Spiral Prismatic Dislocations and the Origin of Slip Bands

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SPIRAL prismatic dislocations, closely related to those described in the previous letter, apparently can be formed by condensation of vacancies. For example, the vacancies present in metals and salts at the melting point may condense[1] along dislocations having a strong component of Burgers character and generate prismatic dislocations of this type. It is perhaps easiest to visualize the process of formation by assuming that the vacancies first condense along a short length of the dislocation, perhaps because of a component of edge character which is mixed with the Burgers character, and then condense in the direction of planes normal to the axis of the dislocation, dissolving away neighboring planes of the spiralling lattice. It should be possible in a metal, such as copper, to generate 10^{12} precipitates of this form per unit volume, each of which possesses of the order of 10^7 vacancies. In the energetically most stable form of a precipitate-dislocation of this type the vacancies would be contained in one plane corresponding to one revolution or less of the spiral. However, the spiral may have a number of revolutions if the vacancies condense during a period of cooling which endures for a sufficiently short time that the equilibrium arrangement is not achieved. Let us assume that the dislocation spirals n times on each half of the double "cone" on which it is distributed and that the radius of the base of the cone is r. We shall assume for simplicity that the thickness of each plane of the lattice is one atomic distance a and that the height of one of the cones is $x=na$. The number of vacancies associated with the cone is then approximately

$$N_v=2\pi(r/a)^2(x/a)/3. \tag{1}$$

n is related to the spacing d between two successive revolutions of the spiral in the direction normal to the axis of the spiral by the equation

$$d/a=r/n=(r/a)/(x/a). \tag{2}$$

Substituting in (1), we obtain

$$r/a=(3N_vd/2\pi a)^{\frac{1}{3}}. \tag{3}$$

If x is small compared with r, corresponding to the most stable limit of the configuration, the energy of the dislocation[2] is, in order of magnitude,

$$\mathrm{E}=\pi r^2E_0\theta f \tag{4}$$

in which $E_0=Ga/4\pi(1-\sigma)$, where G is the shear modulus and σ is Poisson's ratio, θ is $2x/r$ for the limit of interest to us in which x is small compared with r and f is a slowly varying function of x/r, which we shall assume is of the order of unity. The energy per vacancy in the spiral dislocation is

$$\epsilon=(3/4\pi)[G/n_0(1-\sigma)](a/r)f \tag{5}$$

which, with the use of (3), may be expressed in the form

$$\epsilon=(3/4\pi)[G/n_0(1-\sigma)](2\pi a/3N_vd)^{\frac{1}{3}}f. \tag{6}$$

In these equations, n_0 is the number of atoms per unit volume, so that $G/n_0(1-\sigma)$ is an energy of the order of 10 ev for a typical material. If N_v is 10^7 and r takes its maximum value of 1.7×10^3a, corresponding to a single revolution of the spiral, (5) is of the order of 5×10^{-3} ev or about 100 cal./mole of precipitated vacancies. On the other hand, if d/a is 100, this is multiplied by about 1.8. Hence there is no radical difference in the energy per vacancy in a spiral having one revolution and one having eight revolutions. In other words, spiral prismatic dislocations, formed by precipitation of vacancies, which contain of the order of 10^7 vacancies have substantial probability of occurring in a form in which the spacing between neighboring arms in the direction normal to the axis of the cone is of the order of 200A.

Spiral dislocations of this type have the following two important characteristic properties:

(1) They may contribute to the mosaic structure of the crystal. The distortion introduced into the lattice by such a spiral is the same as if the volume of the double cone on which the spiral winds were removed from the lattice, the material were drawn together and the surface rejoined. This process would cause the portions of the crystal on either side of the cone to be rotated through an angle θ, relative to one another. θ is of the order of 10^{-2} radians if d is $100a$. Hence the contribution to the mosaic structure is by no means negligible. It may be noted that each spiral of this type has many of the characteristics required of a mosaic boundary. For example, it should be relatively stable when the crystal is annealed since the energy of instability is relatively small.

Presumably neighboring spirals will interact with one another as a result of the elastic deformation each spiral induces in the medium. The writer visualizes the spirals as forming a primitive type of lattice structure in which the distance between spirals is comparable to the linear dimension r of a given spiral.

(2) The spirals, when used in connection with Frank and Read's "static" mechanism[3] for the multiplication of dislocations,