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| journal or publication title | Science reports of the Research Institutes, Tohoku University. Ser. A, Physics, chemistry and metallurgy |
| volume | 6 |
| page range | 125-129 |
| year | 1954 |
| URL | http://hdl.handle.net/10097/26621 |

# Geometry of the Formation of the Dislocation Network by the Condensation of Vacancies*

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(Received January 8, 1954)

**Synopsis**

When ring dislocations of the Seitz type are formed on the (110) atomic planes in the face centred cubic lattice solidified from the melt, the hexagonal and three-dimensional network can be formed by the growth and reactions of the ring dislocations.

The dislocation network of the hexagonal form was first predicted by Mott(1), first observed by Hedges and Mitchell(2) and studied by some investigators(3)(4). The formation of the network, however, is the problem not yet developed. This paper is intended to step towards the solution of this problem, mainly discussing the geometry of the formation of the network by the condensation of vacancies.

Concerning this problem, there are two cases which seem to be worth consideration, namely the case in which the crystal is annealed after being subjected to plastic deformation and the case in which the crystal is solidified from the melt. In the former case, the starting state is such that, there are a number of excess dislocations of the operated slip systems, namely the piled-up dislocations, and the deformed and destroyed original dislocation network. There are also numerous excess vacancies and interstitials which were produced during the plastic deformation. When the crystal is annealed, the dislocations in it will move or climb and react by the aid of vacancies until a regular dislocation network is reformed. The main processes must be the annihilation and shortening of the dislocations so as to decrease the density of dislocations from about $10^{10-12}$ lines/cm$^2$ to about $10^8$ lines/cm$^2$, but the annihilation or shortening process itself can not produce the new dislocation network geometrically. It may be worth noticing that the nodes of the original dislocation network will not suffer such severe changes by the plastic deformation as the segmental dislocation lines of the network, so they may play an important role in the formation of the new network together with the nodes newly born during cold-work (or during annealing provided that recrystallization does not take place). In this paper, however, the former case will not be refered to. An idea on the case of

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* The 744th report of the Research Institute for Iron, Steel and Other Metals.

(1) N. F. Mott, Proc. Phys. Soc., B **64** (1951), 729.

(2) J. M. Hedges and J. W. Mitchell, Phil. Mag., **44** (1953), 223.

(3) T. Suzuki and H. Suzuki, to be published.

(4) N. Thompson, Phil. Mag., **44** (1953), 481.

recrystallization was put forward recently by T. Suzuki and H. Suzuki(3). So, we shall confine the discussion to a mechanism of formation during the solidification of the crystal.

Teghtsoonian and Chalmers(5) proposed the mechanism of the formation of regular arrays of dislocation lines making subboundaries, but it may be inadequate to regard this as the origin of the hexagonal dislocation network. Another mechanism of formation of the dislocation lines was proposed by Seitz(6), namely the formation of dislocation rings by the aggregation of vacancies. Ring dislocations can be produced by the aggregation of excess vacancies into sheets on some simple atomic planes, forming disc-shaped holes in these planes whose sides then collapse. Their Burgers vectors are prependicular to these planes. The precise mechanism of formation of the ring dislocation is uncertain at present. All that can be said is that the energy of a spherical void of aggregated vacancies is proportional to $n^{\frac{2}{3}}$, where $n$ is the number of vacancies, and the energy of the ring dislocation is proportional to $n^{\frac{1}{2}}$ approximately, as shown in Fig. 1. $n_0$, the number of vacancies corresponding to the point of intersection of two curves is roughly estimable in usual metals, but little knowledge will be obtained from such an estimation*. When $n$ exceeds $n_0$, the ring dislocation will become more stable than the spherical void.

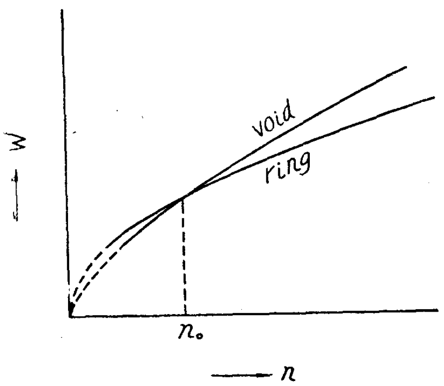

Fig. 1. The energies of the spherical void and the ring dislocation formed by $n$ vacancies.

As far as the face-centred cubic lattice is concerned, it is enough to consider five types of dislocation rings as illustrated in Fig. 2. The ring of type A is formed by two-layer condensation of vacancies on the (100) plane. Type B is by two-layer condensation on the (110) plane. Type C is by three-layer condensation on the (111) plane, but the Burgers vector of this ring is very large. So the ring of type D is rather likely to be formed. The ring of this type is made

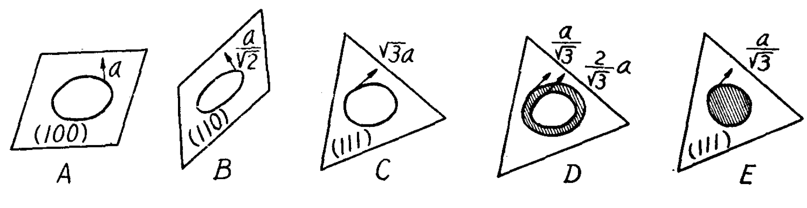

Fig. 2. Five types of ring dislocations in the f.c.c. lattice.

(5) E. Teghtsoonian and B. Chalmers, Canad. J. Phys., **29** (1951), 370.
(6) F. Seitz, Phys. Rev., **79** (1950), 723, 890.

* If we put $const{\cdot}Gb^2l$ as the energy of the ring dislocation and $4\pi r^2 S$ as that of the spherical void, where $G$ is the modulus of rigidity, $b$ the Burgers vector, $l$ the length of the ring, $r$ the radius of the void, and $S$ the surface energy of the void, equating both energies, i.e., $const{\cdot}Ga^3 n_0^{\frac{1}{2}} = const{\cdot}Sa^2 n_0^{\frac{2}{3}}$, we have $n_0^{\frac{1}{6}} = const{\cdot}Ga/S$, where $a$ is the lattice constant. As $S$ is comparable with $Ga$, it follows $n_0 = (const)^6$, which means that the estimation of $n_0$ largely depends on the methods of approximation and the experimental values and is hardly determinable.

by three-layer condensation of vacancies at the inner part and one-layer condensation at the outer part, that is, the part of hexagonal layer (hatched area). Type E is formed by one-layer condensation on the (111) plane, and has the hexagonal layer surrounded by the ring. The rings of the type D and E are the sessile dislocations in their true sense[7]. The Burgers vector of each ring is shown in Fig. 2.

Among these five types, the ring of the lowest energy will be produced. Approximately, the energies of the rings may be compared by calculating the line energies and the energies of hexagonal layers of rings made from the same number of vacancies. Then, the energy of the dislocation ring of type A is given by

$$W_A = \frac{G\sqrt{n}\,a^3}{4\sqrt{\pi}\,(1-\nu)} \cdot \log \frac{r_\infty}{r_0} \tag{1}$$

where $G$ is the rigidity modulus, $n$ the number of vacancies, $\nu$ the Poisson's ratio and $a$ the lattice constant. And the energies of the others are represented briefly as follows.

$$W_B = 0.59\, W_A \tag{2}$$

$$W_C = 2.3\, W_A \tag{3}$$

$$W_D = 1.3\, W_A + H \cdot s \tag{4}$$

$$W_E = 0.44\, W_A + H \cdot S \tag{5}$$

where $H$ is the energy of the hexagonal layer per unit area, $s$ and $S$ are its areas. Type E looks the most favourable, but it may not be so, because the ring dislocation of this type always produces and leaves the hexagonal layer in the crystal after it grows and settles down into the stable configuration, and the energy of hexagonal layer increases proportionally to $n$. So the ring of type B may be the most favourable to be formed.

After the ring dislocation is formed, the excess vacancies will condense on the dislocation and make it climb. Climbing means the growth of the ring in the present case. During the climb motion, the ring dislocation can meet another ring which is in another similar atomic plane, and may be able to react with it to form a new dislocation line on the line of union. The meeting is not always favourable for the reaction of two rings. Fig. 3 represents the various kinds of

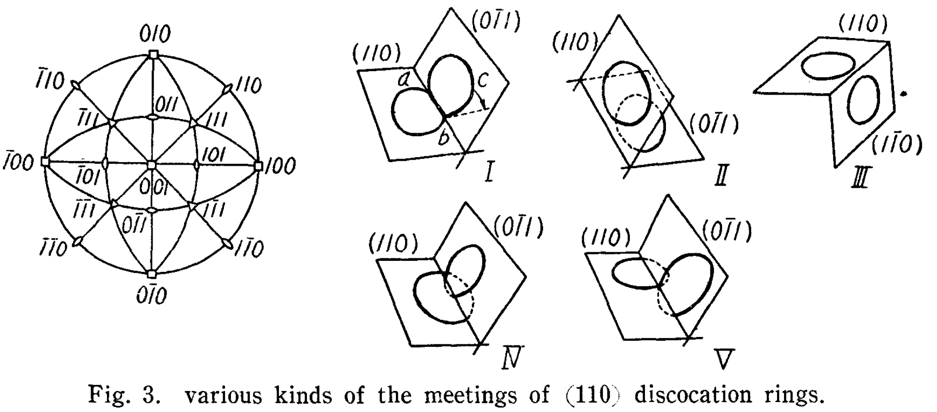


Fig. 3. various kinds of the meetings of (110) discocation rings.

(7) F.C. Frank, Phil. Mag., **42** (1951), 809.

meetings of (110) dislocation rings (type B). In the case of 3-I, the reaction can be performed to make the part *ab* illustrated in the figure, but in the cases of 3-II and 3-III, the rings do not atract each other and the combination does not take place. These unfavourable cases will be discussed later together with the cases of 3-IV and 3-V in which the rings contact and react with each other at only one point.

The meetings and reactions of the rings are similar to the Cottrell-Lomer reaction(8) in regard to their Burgers vectors but different from it regarding the planes on which they exist. For instance, in the case of Fig. 3-I, the reaction produces a dislocation with a $<101>$ vector and $<\bar{1}11>$ line direction, which then continues the climb motion partially in the (101) plane until it lies into the $(\bar{1}\bar{1}1)$ or $(\bar{1}11)$ plane which contains the $<101>$ vector. Thus, the resultant dislocation is known as the dislocation of normal type in the face-centred cubic lattice. So, this composite dislocation must dissociate into two partial dislocations of the Heidenreich-Shockley type(9). As a result of the dissociation the velocity of condensation of vacancies on it will decrease very largely. The part of the ring which did not react, as the part *c* in Fig. 3-I, can also change into the normal dislocation by partially climbing and lying into another (111) plane which contains the Burgers vector of the ring, as shown by the dotted line. For instance, the (110) ring can change itself into the dislocation with the $<110>$ vector and the $(1\bar{1}1)$ or $(\bar{1}11)$ plane partially. Similarly the $(0\bar{1}1)$ ring may become the dislocation with the $<0\bar{1}1>$ vector and the (111) or $(\bar{1}11)$ plane. The ring which does not react thoroughly with others will become a prismatic dislocation as pointed by Seitz(6).

Thus, we can obtain the dislocation network by connecting such reactions and climbings as can be seen in Fig. 4. The resultant network is composed of normal dislocations and has the hexagonal but three dimensional structure. By annealing for a long time, every small portion of the network will settle down into the regular and fairly stable form such as studied by T. Suzuki and H. Suzuki(3) and Thompson(4). For the formation of such a regular network, however, it seems necessary to make the dislocation nodes climb and the segmental dislocations climb or, maybe, slip for fairly long distance. Otherwise, the irregular portion of the network or of the dislocation itself will remain in the crystal.

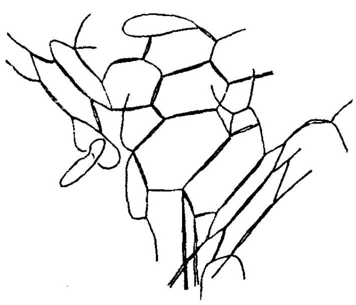

Fig. 4. The hexagonal and three dimensional network in the middle of formation.

In the cases of 3-II and 3-III, the reaction does not take place, but two rings will pass through each other in the manner as illustrated in Fig. 5, leaving two

(8) W. M. Lomer, Phil. Mag., **42** (1951), 1327.
(9) R. D. Heidenreich and W. Shockley, Report of a Conference on the strength of solids, University of Bristol, England, Physical Society, London, 1948.

unstable small rings interlinked. In the cases of 3-IV and 3-V, the rings will contact and react along very short portions and will probably make an irregular network, as seen in Fig. 6-I and 6-II, if the tendency of the unreacted portions to settle down into the regular dislocation lines is strong, or they will pass through each other in a similar manner to the cases of 3-II and 3-III, as in Fig. 6-III, leaving a small intricate ring which will subsequently vanish, if the growth of the rings is rapid enough to suppress the above tendency.

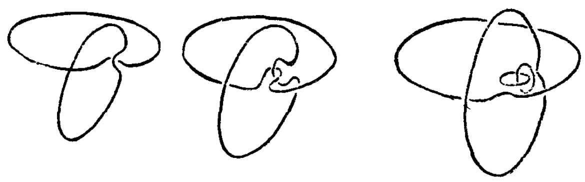

Fig. 5. The ring dislocations passing through each other after an unfavourable meeting.

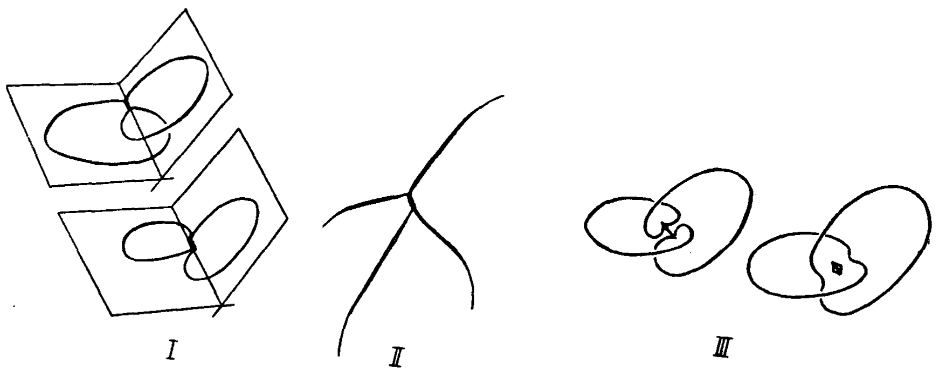


Fig. 6. The rings in the point-like contact with each other and their resultant configurations after growth.

From the dislocation rings of type A a dislocation network with hexagonal structure can be constructed in an interesting but complicated way. However, it may be meaningless. From dislocation rings of type C, D and E, no regular dislocation network can be constructed.

I am grateful to Drs. T. Suzuki and H. Suzuki and Prof. F. C. Frank for advice and many helpful discussions. I also thank to Prof. M. Yamamoto for his encouragement.