1986

# A Study of Dislocations in Crystals using the Technique of Chemical Etching

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## Recommended Citation

Dennis, Ramona R., "A Study of Dislocations in Crystals using the Technique of Chemical Etching" (1986). *Honors Theses*. 113.
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Senior Independent Study
The Honor's Program
Spring, 1986

# A Study of Dislocations in Crystals using the Technique of Chemical Etching

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# THEORY OF CRYSTALLINE STRUCTURES

The beauty of many of the crystalline solids has attracted man's attention from the earliest times. Eventually the science of crystallography developed from studies of their shapes. The word crystal stems from the Greek word "krystallos" meaning "clear ice."[1] The word was first applied to describe the clear quartz crystals found in the Swiss Alps. The crystals were thought to be formed from water under conditions of extreme cold.

The earliest recorded crystallographic observation was in 1597 by Andreas Libarius. He noticed that crystals of different substances often have characteristic shapes. He suggested that the salts in mineral water could be identified by examining the shapes of the crystals deposited after the water had evaporated. Almost a century later the work of two Danish men, Niclaus Steno and Erasmus Bartholinus, led to the realization that while the faces of different crystals may vary in size and shape, the angle between corresponding pairs of faces is the same. Steno postulated that a crystal must grow from a nucleus by deposition of material in layers, so that a flat face remains parallel to itself as it grows outward.

Toward the end of the Eighteenth century the French crystallographer J.B.L. de Romé de I'isle discovered that all crystals of any one substance exhibit the same interfacial angle regardless of the facial sizes. He was somewhat successful in attempting to derive these angles from a few geometrical "primitive forms."[2] Abbé René Jüst Hauy, a colleague of I'isle, emphasized the elementary shapes into which some crystals, such as calcite, could be cleaved. He showed that the many different shapes exhibited by crystals of the same substance could all be constructed by regular repetition of a small fundamental unit of a characteristic size and shape.

The Nineteenth century saw a steady development of crystallography in the light of Haüy's observations. The study of the external shapes of crystals continued to advance with the development of detailed relationships between the observable geometry of faces and the sizes and shapes of the internal units. Speculations about the constitution of units continued during this period until they were rationalized by the foundations of modern atomic theory at the beginning of this century. Crystalline solids were then recognized as regular, repeating arrangements of the constituent atoms to fill the whole volume of a crystal. Subsequent developments in the present century have made it possible to determine the size and shape of these units and the relative dispositions of the various atoms within them.

An ordered arrangement of the constituent atoms into a repetitive three-dimensional pattern is the prime characteristic of the ideal crystalline state. With this conception, advancing experimental methods freed the study of crystalline materials from the restrictions imposed by the necessity of a regular external shape. Thus, solids which have never formed natural external plane faces are just as representative of the crystalline state of matter. Soon it became clear that the strict atomic regularity of the perfect crystalline structure is really only one extreme limit of the structural arrangements that are to be found in solids. Solid matter can also merge by degrees into an amorphous state in which the whole volume is occupied by a random jumble of the constituent atoms with no repetitive regularity.

The orderly arrangement of crystals is a consequence of the attractive binding forces which form permanent attachments between the constituent atoms. The development of perfect regularity corresponds to a minimization of the energy of the solid at the change of state in which it is formed. Wave mechanical treatments of the interactions of atoms in close proximity can predict quantitatively the minimum energy configurations for some systems of simple atoms. The calculations become too difficult to be handled quantitatively in more complicated systems, but they can be extended in a roughly qualitative manner.

The symmetry in the arrangements of the atoms is largely a consequence of the type of bond they form. Thus, the general classifications of bonds and how they determine the local arrangements of atoms will be discussed. The bonding mechanisms responsible for the formation of crystals are ionic, covalent, metallic, hydrogen, and van der Waals.

The easiest type of bond to understand is the ionic bond. It results from an electrostatic attraction between a positively charged ion (cation) and a negatively charged ion (anion). An ionic bond is favored when an atom with low ionization energy and an atom with high electron affinity combine. Sodium Chloride is an excellent example of an ionic compound. Little energy is required to remove an electron from sodium to form a cation and the ion has achieved a relatively stable electron configuration. Significant energy is released when an electron is added to chlorine to form an anion, and it has also achieved a relatively stable noble gas electron configuration. Considerable energy is also released when the two atoms are brought together and their opposite charges attract each other. Both of the atoms have also achieved a stable noble gas configuration through the transfer of an electron from sodium to chlorine.

A covalent bond is formed by the attraction between two atoms as a result of sharing one or more outer shell electrons by the two atoms. Attractive forces arise because each electron is attracted by both positively charged nuclei. Repulsive forces arise because the two negatively charged electrons repel each other. There are a total of four attractive forces and only two repulsive forces, and the attractive forces predominate. This type of bonding generally occurs between the smaller atoms that need one or two electrons to fill their electron shells to a noble gas electron configuration.

Metallic bonding occurs because metals have relatively low ionization energies and all of the atoms are alike. Since these atoms have a relatively loose hold on their electrons, a metal lattice forms. This consists of a regular array of positive ions immersed in a cloud of highly mobile outer shell electrons. Metallic bonding results from attraction between the positive ions and the cloud of negative electrons.

A hydrogen bond occurs when very strong dipole-dipole attractions occur between hydrogen atoms bonded to small, strongly electronegative elements and nonbonding electron pairs on other such electronegative elements. Dipole-dipole attractions occur when a partially charged atom is attracted to another atom of opposite charge.

An accurate account of the nature of van der Waals forces (or London forces) requires the use of quantum mechanics. However, for the purpose of this paper a simpler explanantion will suffice. The average distribution of charge in a nonpolar molecule over a period of time is uniform. However, because electrons move, the electrons and thus the charge may mot be uniformly distributed at any given instant. Electrons may be slightly accumulated on one side of the molecule. As a consequence, a small temporary dipole will occur. This temporary dipole in one molecule can induce opposite dipoles in surrounding molecules. It does this because the negative (or positive) charge in a portion of one molecule will distort the electron cloud of an adjacent portion of another molecule causing an opposite charge to develop there. These temporary dipoles change constantly, but the net result of their existence is to produce weak attractive forces between nonpolar molecules, and thus make possible the existence of their liquid and solid states.

The smallest subunit of a crystal lattice that can be used to generate the entire lattice is called a unit cell. All crystal lattices can be generated by the fourteen types of unit cells shown on the following page.[3] Only the metal, Polonium, exhibits the simple cubic lattice structure.

The crystalline structure of ionic solids often depends upon the difference in size of the cations and anions. The different packing arrangements for NaCl and CsCl are a direct consequence of the fact that cesium ions are larger than sodium

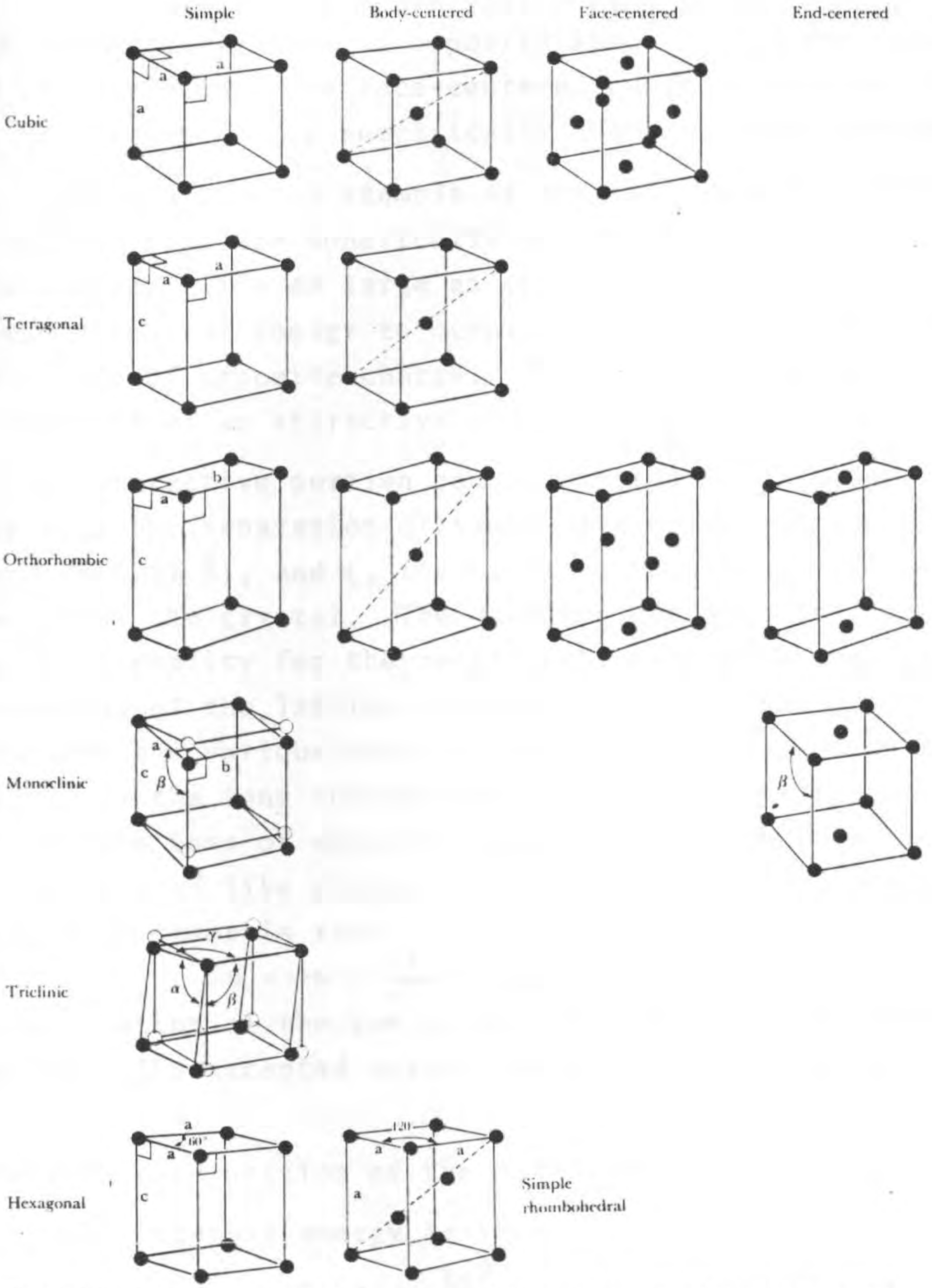


**Figure 13-28** The 14 unit cells that generate all the possible three-dimensional crystal lattices. The terms cubic, tetragonal, orthorhombic, and so on are described more fully in Table 13-3. We shall deal with only the cubic lattices in this chapter.

ions. Thus, NaCl is a face-centered cubic which occurs when an ion of either kind has six nearest neighbors of the other kind. CsCl is a body centered cubic which occurs when each ion has eight nearest neighbors of opposite kind. Since the crystals used in this study were face-centered cubic structures, they will be discussed more specifically than any other structure.

NaCl is a typical example of the face-centered cubic. The $Na^+$ and $Cl^-$ ions are spherically symmetric, with the $Cl^-$ ion approximately twice as large as the $Na^+$ ion. This causes the minimum potential energy to occur when each ion is surrounded by six ions of opposite charge. The potential energy of each ion consists of an attractive and a repulsive portion.

The attractive portion can be written $E_p = -\alpha \frac{ke^2}{r}$ where r is the separation distance between neighboring ions (which is 2.81 Å), and $\alpha$, the Madelung constant, depends on the geometry of the crystal. The Madelung constant is a constant of proportionality for the relationship stated above. It is independent of the lattice dimensions, but it has different values for the various types of ionic structure. Each ion is affected by the ions surrounding it so that $\alpha$ is the summation of 6 for the ions of opposite charge at a distance of r plus 12 for the ions of like charge at a distance of $2\sqrt{r}$, etc. The Madelung constant is thus

$$\alpha = +6 - \frac{12}{\sqrt{2}} + \frac{8}{\sqrt{3}} - \ldots$$

The calculation of the sum is carried out until it converges to a number. The accepted result for a face-centered cube is $\alpha = 1.7476$.

The repulsive portion of the potential energy is $E_p = + \frac{A}{r^n}$. The total potential energy is then

$$E_p = -\alpha \frac{ke^2}{r} + \frac{A}{r^n}$$

A can be found by considering the equilibrium separation $r = r_0$ such that the force $F = - dE_p/dr$ is zero. By differentiating and setting $dE_p/dr = 0$, then $A = \alpha ke^2 r_0^{\,n-1}/n$. Thus, the total potential energy of an atom in a crystal is

$$E_p = -\alpha \frac{ke^2}{r_o}\left[\frac{r_o}{r} - 1/n\left(\frac{r_o}{r}\right)^n\right]$$

At $r = r_o$

$$E_p = -\alpha \frac{ke^2}{r_o}\left(1 - 1/n\right)$$

n can be approximated from the dissociation energy of the crystal (the energy needed to break the crystal up into atoms). The dissociation energy of NaCl is 184 kcal/mole, so n is approximately 9.[4]

Thus far, the discussion of crystals has been concerned only with perfect crystalline structures in which the atoms reside at specific sites in the structure. To be realistic, we must recognize that the locations of these lattice sites represent only mean positions for the atoms. At any finite temperature an atom has thermal motion whose amplitude and direction depend on its electronic configuration and the restoring forces determined by its binding into the general structural arrangement.

Real crystals cannot have such perfection of atomic arrangement, and it has become increasingly clear that for some properties structural imperfections can be just as important as the overall atomic pattern. Departures from this regular pattern are known as defects and the general types will be discussed.

The study of defects begins most conveniently with the classification of the different types of defects. When a crystal is imperfect, the individual deviations from perfection are not spread uniformly throughout the crystal, but are concentrated in well-defined regions. The distorted material in these regions is surrounded by material that is distorted only slightly or not at all. Three modes of general classification are point, line, and plane defects. These classifications are based on the shape of the distorted region. A point defect is when the distortion is mostly confined to a small more or less spherical volume. A line defect is when the distortion occurs along some straight or curved line in the crystal. A plane defect is when the distortion is confined to an approximately thin layerlike region.

The simplest example of a point defect is a vacancy, or Schottky defect, a lattice site that should have an atom but does not. This defect may occur by the removal of an atom of an otherwise perfect crystal or it may occur during growth by a failure of the growing lattice to incorporate an atom at one of its normal sites. This extra atom is called an interstitial. Another type of point defect is a substitional defect. This is an atom of an abnormal type at a normal lattice site.

A small displacement of an atom from a normal lattice site to a nearby normally unoccupied site will simultaneously introduce both a vacancy and an interstitial. This is a Frenkel defect. This introduces an extra strain at both locations which increases the potential energy of the crystal.

A realization that real crystals must contain line imperfections grew from explanations advanced for the mechanism of slip, one of the important processes in the plastic deformation of crystals. It has long been known that when a crystal is permantly deformed by an applied stress, plastic flow takes place by the movement of lamellae of the crystal relative to each other. This movement occurs on particular parallel planes spaced at relatively narrow intervals by the translocation of whole blocks of structure between active planes. Early studies of crystalline deformation showed that displacements always take place on crystallographic planes (known as slip or glide planes) along a lattice direction within the plane.[5] The particular slip system that is operative is related to the crystal, its structure, and the method of deformation.

At first sight, the atomic mechanism of a process in which blocks of structure glide over one another on well-defined slip planes presents no real problems. It can be likened to a stack of playing cards being subjected to a force from an arbitrary side. However, this simple picture encounters difficulties when any attempt to calculate the stresses necessary for the onset of the slip are made. If perfect blocks of structure are required to move bodily over one another, the initiation of movement calculated requires the stresses to be greater than those actually measured. Actual measurements in the laboratory showed these theoretical calculations to be $10^3$ to $10^4$ times greater than the stresses actually required.[6]

An acceptable slip mechanism had to account for these stresses as well as retaining the features of simple block movement that account for the observed geometry. It was recognized that the forces could be reduced if the atomic forces acting across the slip plane could be overcome one at a time rather than simultaneously over the whole area. Eventually a mechanism was postulated in which slip is caused by the movement through the structure of line defects, known as dislocations.

At first, it might be thought that a simple example of a line defect would be a series of regularly spaced point defects such as a line of vacancies. However, this does not occur. Such a line would be broken up by thermal motion into a disconnected set of individual point defects. Dislocations always end on the surface of a crystal. No localized internal rearrangement of the atoms can ever cause the defect line to terminate or be separated into disconnected sections having ends inside the material.

There are two basic types of line defects, edge dislocations and screw dislocations. An edge dislocation corresponds to an extra plane of atoms introduced in one part of the crystal. Initially, consider a lattice composed of a regular array of cubic blocks with adjacent faces cemented together as in figure 1.[7] If the lattice is deformed by placing a cut along

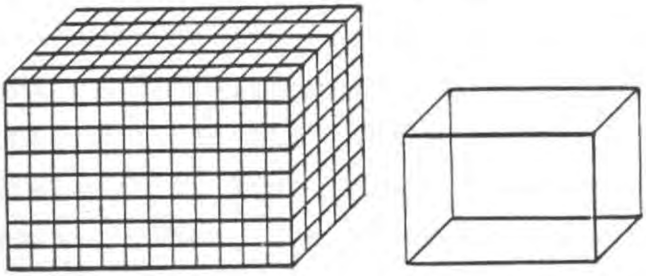


Figure 1

plane P as in figure 2[7], then inserting into this cut an extra plane of blocks. This plane of blocks is slid wedgewise part way into the crystal, but not all the way, resulting in the

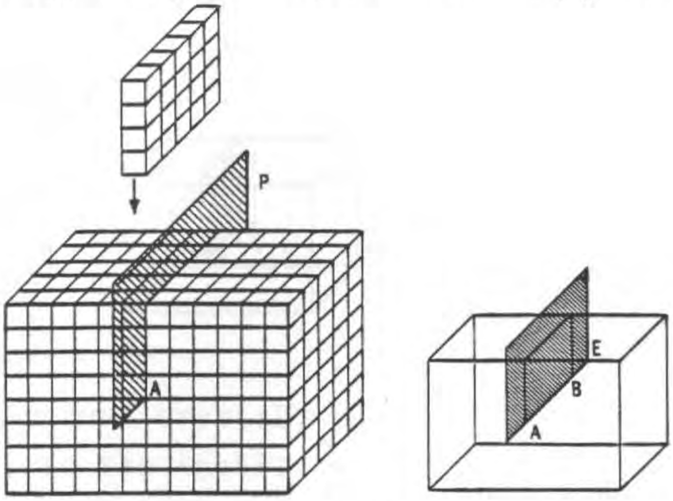


Figure 2

configuration demonstrated in figure 3.[7] The lower edge of this plane is the line AB running through the crystal. Notice that the line enters the front face of the crystal and exits the back face. The extra plane can make a clean undistorted fit everywhere within the lattice except along the bottom edge which is line AB. Thus, the distorted region is small in cross section and infinitely long in comparison to its thickness.

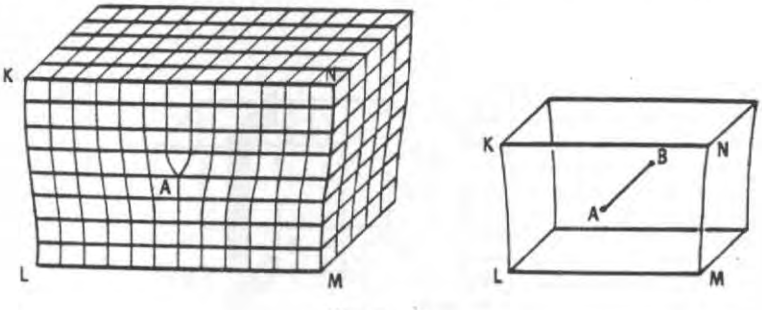


Figure 3

Another example of a line defect is shown in figures 4 and 5[7] to demonstrate that the dislocation cannot end within the crystal. A dislocation line travels from A to C and it would seem that it ends at C. Although AC does end at C, the dislocation line does not because the back edge of the plane CD forms a vertical dislocation line terminating at D on the upper surface of the crystal. Thus, the dislocation enters at A, bends at C, and exits at D.

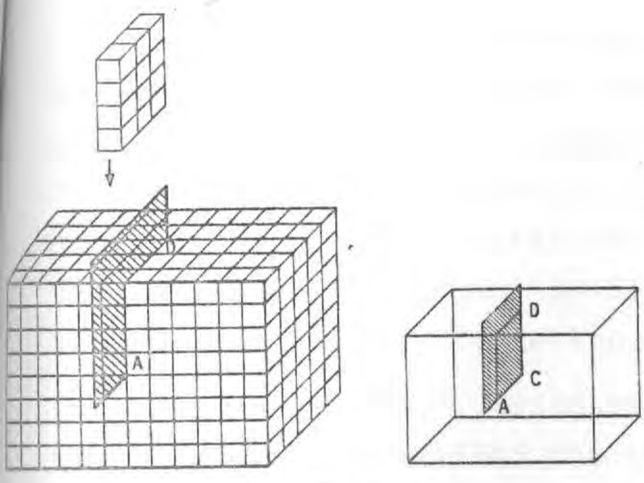


Figure 4

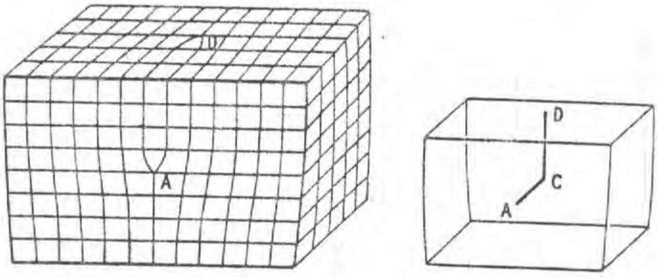


Figure 5

In an actual crystal, this edge of blocks A to C to D corresponds to the inability of the atoms to complete all of the bonds for their lattice positions. This results in a row of "dangling bonds"[7] along the dislocations. These unused bonds cause the crystal to slip along these lines.

Screw dislocations are somewhat harder to visualize and illustrate. Again consider a lattice of uniform cubes, cemented face to face. A cut is made part of the way through the crystal along the front face extending to line RS as demonstrated in figure 6.[7] Rather than inserting an extra plane of blocks, the crystal is sheared along the cut by pushing downward on top of the crystal on one side and upward on the bottom of the other side of the crystal as indicated. If the two sides are slid by one block, the configuration in figure 7[7] is acheived. Once again the blocks line up except for line RS.

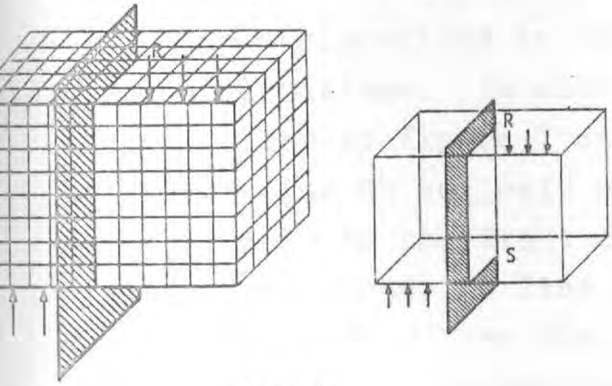


Figure 6

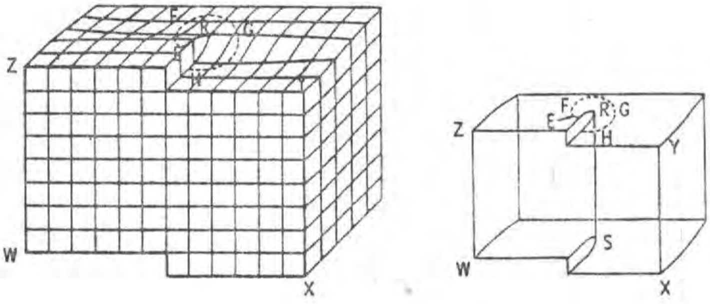


Figure 7

The name "screw dislocation" results from the fact that if we were to trace a circular path EFGH around the dislocation, remaining always on the upper surface of the crystal, we would find that we had returned to the block just under our starting point. The entire crystal which was originally a set of layers one on top of the other has been converted to one layer of blocks in a helical formation.

Another method of illustrating screw dislocations is to consider a cylinder like the one shown in figure 8.[8] If the cylinder is cut along a plane ABCD and displaced as shown, then a screw dislocation results. The atoms thus have a helical arrangement around the axis of the cylinder, which is the dislocation line.

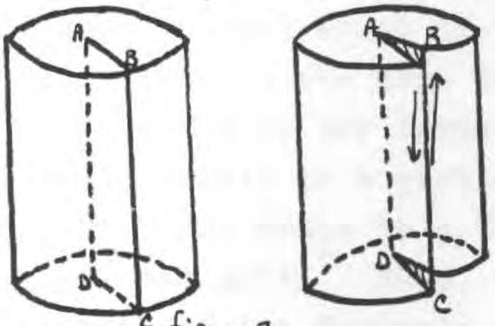


figure 8

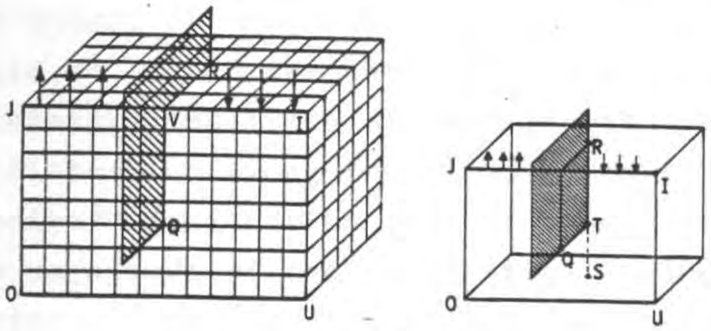


Figure 9

Many dislocations in crystals have features of both edge and screw dislocations. An example would be to consider figure 9.[7] It is similar to figure 7 except the cut does not extend the full length of line RS but only as far as T, then it runs horizontally along line TQ to the front of the crystal. As before, a screw dislocation extends to line RT as illustrated in figure 10.[7] A close inspection of the diagram shows that the equivalent of an extra layer of blocks has been inserted horizontally into the crystal to the right of QT. Thus, the configuration around Q is identical to that of A in figure 3 with the exception that the extra layer of blocks has been inserted from the right instead of the top. The complete dislocation RTQ enters the crystal as a screw dislocation at R, proceeds as a screw along RT, turns the corner at T, and emerges at Q as an edge dislocation. Thus, the dislocation has not terminated inside the crystal, although it did change its type and direction.

A useful concept in describing the way in which dislocations interact and combine is the Burger's vector for the dislocation. This vector, like all vectors, has direction and magnitude. It is the direction in which the shear displacement occurs.

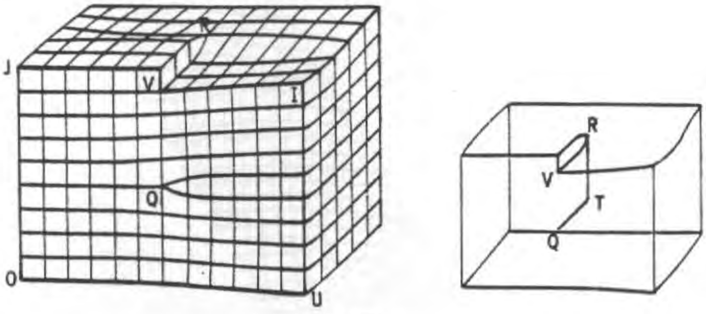


Figure 10

Considering figure 10[7], the amount of shear along the cut plane is just equal to the amount of shear required to displace the plane by the edge length of one of the unit cubes. Obviously, this could be any larger intergral multiple of this length. That would result in a greater distortion along the dislocation line RTQ. This would be a "stronger" dislocation than if the shear is only one unit.[7] Thus, the magnitude of the dislocation is the length of its Burger's vector. Note that since the same shear has produced both the screw dislocation RT and the edge dislocation TQ, the Burger's vector is the same for both. Hewever, observe that RT is parallel to the Burger's vector, while TQ is perpendicular to it. In general, a dislocation parallel to its Burger's vector is a screw, one perpendicular is an edge, and one inclined at some arbitrary angle is a combination of edge and screw dislocations.

A superficial look will be considered for the third class of defects. The plane defects of interest are of two types, grain boundaries and twin planes. The latter is a special case of the former. A grain boundary is the interface between two contiguous crystals which are not of exactly the same orientation as shown in figure 11.[7] At such boundaries, the crystal lattice layers come together at an angle. As a result, the material around the boundary is strained or deformed. Inspection shows that this boundary is equivalent to an array of edge dislocations.

In general, any grain boundary can be represented by such an array of dislocations lying in the boundary surface. The spacing and character (edge, screw, or mixed) of these dislocations depend on the amount and direction by which the two crystals differ in orientation.

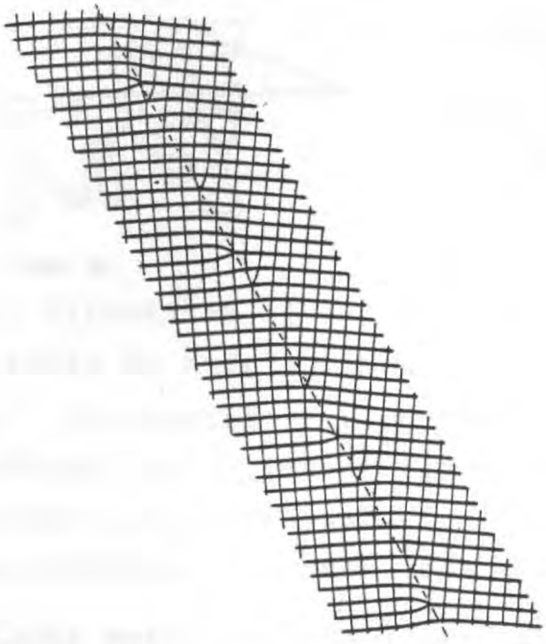


Figure 11

Grain boundaries are usually a consequence of a growth process in which crystallization begins independently at many points. The orientation at each of these independent crystals or grains will then be more or less random. When they finally grow together a random grain boundary results. There are certain special grain boundaries at which the atomic fit is generally very good. These boundaries have an atomic arrangement that is regular over the entire boundary area, and are almost always plane, in contrast to random grain boundaries which may be either plane or curved. Because of this regularity, these are usually called coherent boundaries.

The best example of a coherent boundary is the twin plane. A twin plane can sometimes be described as a grain boundary such that the atomic arrangement within each grain is a mirror image of that in the other as illustrated in figure 12.[7] Twins of this type are relatively rare. More frequently the arrangements are related as follows. A crystal can be cut into along the twin plane, one half is turned 180° relative to the other around an axis perpendicular to the plane, and the two halves are reunited. This is illustrated in figure 13.[7]

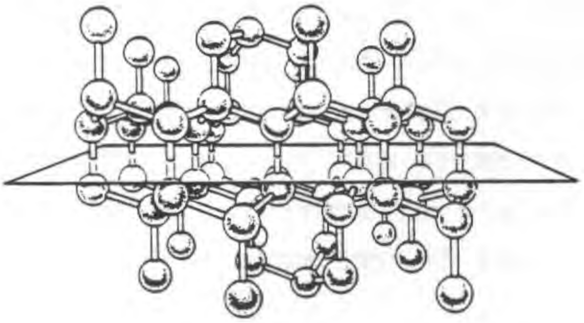


Figure 12

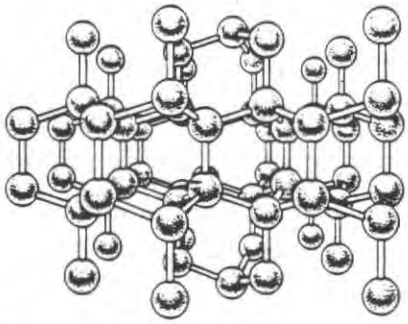


Figure 13 Twinned diamond structure.

So far the discussion of defects has been purely theoretical made plausible by models. How can these defects be proven experimentally? The methods for conclusively demonstrating the presence of defects are different for each class of defects. They are generally straightforward for plane defects, less direct for line defects, and even less direct for point defects.

Three methods most frequently used are crystal cleavage, x-ray reflection, and chemical etching. Many crystals cleave easily along the principle planes in the crystal. If cleavage shows one set of natural faces in one section, while another set of similar faces are rotated 180° from the first about some axis, it is good evidence that there is a twin plane perpendicular to the observed axis.

X-ray diffraction techniques are a powerful tool in the sutdy of the structure of crystals. When a collimated beam of x-rays is allowed to pass through a crystal and fall on a photographic plate. A regular array of spots, called a Laue pattern, are observed on the plate in addition to the central beam. These correspond to the planes within the crystal so any irregularities in the crystalline structure would be evident in the Laue pattern.

When a crystal is attacked by a suitable solvent the initial dissolution often takes place in a manner which is visibly related to the underlying structure. The solvent will react more quickly at points of imperfection on the crystal surface. After contact with the solvent for a period of time the crystal faces will show a number of pits or cavities where solution has been most pronounced. The pits are usually bounded by sloping planes along which the solvent has acted most rapidly, and solution cavities of this kind which have definite shapes are partly dependent on factors such as the nature of the solvent and its concentration, but the symmetry of their shape and of their attitude on different crystal faces may be considered an indication of the symmetry of the underlying structure.

# EXPERIMENTAL PROCEDURES AND ANALYSIS

The crystals used in this study were all face-centered cubic structures. Magnesium oxide, Potassium chloride, and Lithium flouride crystals were used although the Lithium flouride was used only once.

Magnesium oxide was worked with first. An etching and polishing procedure was located from literature. The crystals were polished in boiling orthophosphoric acid for three minutes, rinsed in room temperature acid, methanol, ether, and distilled water. Etching was performed with a solution of five parts saturated $NH_3Cl$, one part concentrated $H_2SO_4$, and one part distilled water. The crystals were etched at room temperature for 15 minutes then rinsed in methanol and finally ether.[9]

This procedure and some variations of it were attempted. The polishing procedure was too vigorous and had to be revised. This meant that some of the cleanliness of the crystal surface was sacrificed in order to etch. However, etching these crystals revealed nothing. It was then suggested that a different type of crystal be used.

Since KCl and LiF crystals of high purity were available, etching procedures for each of them were located from the literature. The KCl procedure was actually for NaCl, but because of their similarities the procedure was thought likely to work for KCl. Two LiF procedures were available. However, the chemicals used in the recommended procedure were not available. Thus, it became a choice between using a procedure for LiF which the literature itself referred to as indistinct or using the NaCl procedure for KCl.

The procedure for NaCl from the literature was as follows. The best etch pits are obtained when the solution consists of 4g of ferric chloride per liter of glacial acetic acid. The crystals should be etched for approximately 30 seconds with moderate agitation. After etching, the crystal should be rinsed in acetone and dried in a jet of warm air.[10]

A number of variations were attempted using this procedure. The crystals were etched for varying time periods, form 30 seconds to 10 minutes. They were also subjected to different types of stresses. A #8 shot was dropped on one from different heights. This was done in an attempt to produce rosettes on the crystal surface. A rosette is composed of dislocation lines radiating from a central point (the impact point). Other crystals were tapped with a #4 finishing nail and a needle in other attempts to produce dislocations. One crystal exhibited pits that remotely resembled indistinct etch pits.

An examination of the literature suggested that the ferric ion concentration was the deciding factor in the production of etch pits.[11] Therefore, the ferric ion concentration was varied around the original solution. This produced no effect.

At this time, an attempt was made to etch a LiF crystal. The less preferred procedure used a dilute solution of $FeCl_3$ ($1.5 \times 10^{-4}M$).[11] The crystal was etched for 2 minutes in varying ferric ion concentrations with no result.

A further search of literature revealed an ethcing procedure of adding saturated $FeCl_3 \cdot 6H_2O$ to glacial acetic acid by volume 25 to 1.[12] Based on this, a solution of 4g of $FeCl_3$ per 100 ml of acetic acid was prepared because it was obvious that the previous procedure was much too weak. This solution was too strong, so the solution was diluted to half the previous concentration of $FeCl_3$ with acetic acid. This produced pits which once again resembled etch pits but weren!t quite distinct.

It was decided that the Magnesium Oxide crystals should be attempted again because the procedure for them was specifically for MgO. A variation in the time of etching would be the deciding factor. Seven specimen were selected and polished for five minutes in room temperature orthophosphoric acid, rinsed in methanol, ether, and distilled water. The etching solution consisted of five parts saturated $NH_3Cl$, one part concentrated $H_2SO_4$, and one part distilled water. The crystals were etched for times varying between 10 and 40 minutes. One crystal was removed every five minutes then rinsed in methanol and then ether.

The crystal which had been etched for 20 minutes had very distinct pyramidal etch pits. Reproduction of these pits was necessary. The remaining six crystals were etched for times varying between 17 and 22 minutes with one crystal being removed every minute. Once again the only crystal that etched was the 20 minute crystal.

These results raised a number of questions. Why did only the 20 minute crystal etch? Why did the etching procedure not work as it did for the experimental work for the article from which the procedure came? What is the difference between NaCl and KCl such that the etching procedure for NaCl did not work for KCl?

It is apparent that slight differences in the etching technique cause one crystal to etch while another does not. These differences could be due to the purity of the crystal or the solutions, temperature, volume of the solution used for etching, etc. The preparation of the solution itself probably causes a great difference. Any and all of these factors would affect the time for etching to take place. Thus, it should be expected that the etchs differ from one another.

An explanation for the difference in potassium chloride and sodium chloride is most probably due to the differences in the atoms themselves. It has already been noted in this paper that the difference in the sizes of Na and Cs causes one to form a face-centered cubic structure while the other forms a body-centered cubic lattice. Sodium chloride and potassium chloride are both face-centered cubic structures, sodium and potassium are both members of the alkali metals group, and the two are frequently used interchangeably in every day chemistry.

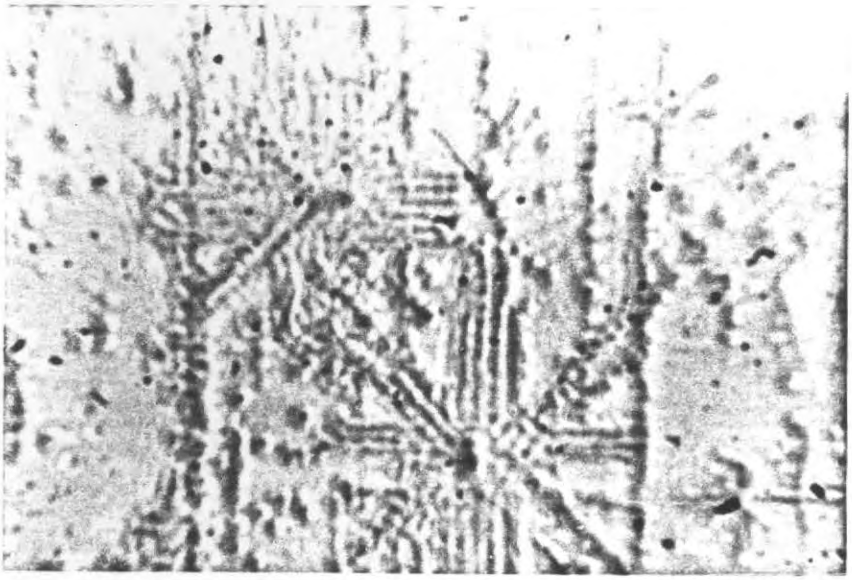
It may be that only more variation in the time and concentration of the reaction are necessary for the potassium chloride crystal to etch. It may also be that the closeness in size of potassium and chlorine causes a steric hinderence to the overall reaction. $FeCl_3$ was used for LiF and NaCl crystals. When examining the relative sizes of these atoms, it is apparent that they have nearly the same ratio of sizes. That is, the ratio of lithium to flourine and sodium to chlorine is very nearly the same.

Some photographs of the defects found on the crystal are shown. Photograph #1 is of two rosettes on the surface of the crystal. Rosettes are an indication of the stress imparted to the crystal by the impact of some object on the crystal. Photograph #2 is a group of defects along a straight line. Often these lines of defects will be at right angles to one another following the slip or glide planes within the crystal. Frequently, this type of defect occurs when the crystal is subjected to a compressing force.

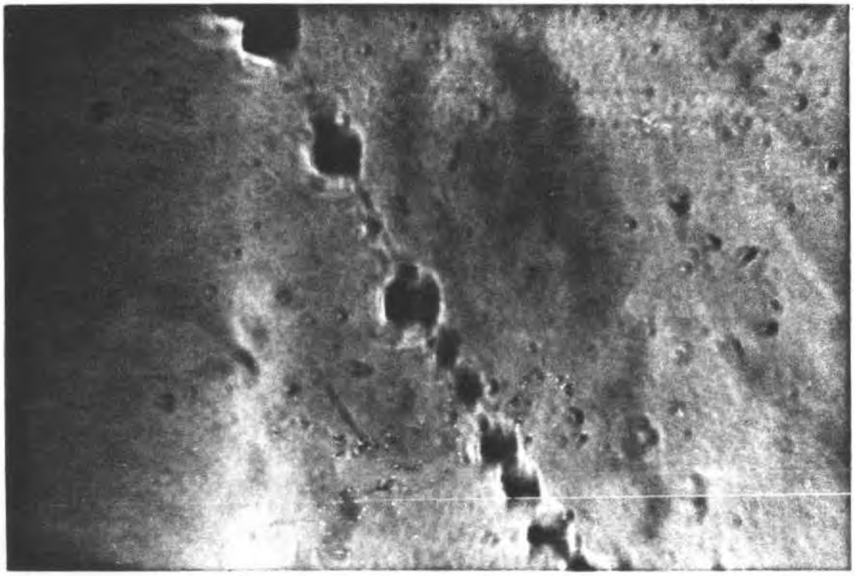
The purpose of this study was to investigate the nature of crystal defects. It centered around one particular laboratory technique- chemical etching. Etching reveals a vast amount of information about a crystal. The shape of the etch pits is representative of the structure of the crystal. The location and orientation of the pits on the surface of the crystal can reveal the type of stress it has been subjected to as well as its composition.

In summation, only the MgO crystals exhibited etch pits. These crystals were previously stressed at Wake Forest University. Most of the experimental work was used in attempting to repeat the procedure from literature. It is usually difficult to repeat experimental procedures because critical factors may be omitted.

The results of the successful etch were quite fortuitous. The crystal contained many examples of the different stress patterns that etching reveals. Included among them are line defects, rosettes and slip or glide planes.



Photograph #1



Photograph #2

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