

STRUCTURAL METALS—UNDERSTANDING THROUGH MATERIALS SCIENCE

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The micro and macro levels of materials behavior are being increasingly tied together by understanding growing out of materials science. In this discussion the elasticity and strengthening mechanisms of structural metals are shown to be areas where strong bridges between micro and macro level behavior can be made. The structural steels, including the newer high strength steels, are discussed in terms of this materials science understanding. Possibilities for the control of the elastic moduli of structural metals are explored.

INTRODUCTION

In recent years the materials science approach has led to a more basic understanding of the principles underlying the behavior of materials of interest to civil engineers. With this information supplementing the more traditional materials testing approach, the micro and macro levels of materials behavior have been related to various degrees in all structural materials. The understanding of the relation between mechanical properties and the detailed structure and composition of materials has led to innovations in design and the development of improved and new materials. Control of structure and composition at the micro level can result in the improvement of properties at the macro level.

Civil Engineering materials can be grouped into three classes, based on similarities at the micro level—ceramics, organics and metals. Ceramic materials, which have complex atomic level structures and are often amorphous (non-crystalline), include the glasses, ceramics, and Portland cement concrete. The organic materials, such as wood, plastics, rubbers and asphalts, are characterized by long chain molecules at the micro level. Metals, such as steels and aluminums, are characterized by the regular crystalline arrangement of their atoms, which are held together by a particular kind of bonding.

The following discussion will be limited to structural metals, using them as examples of how properties at the macro level can be related to micro level composition and structure.

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ELASTICITY

The elasticity of metals is an example of a relatively easy bridge between the macro and micro scales. Consider the case of two atoms, attracted to each other by electron sharing but repelled from each

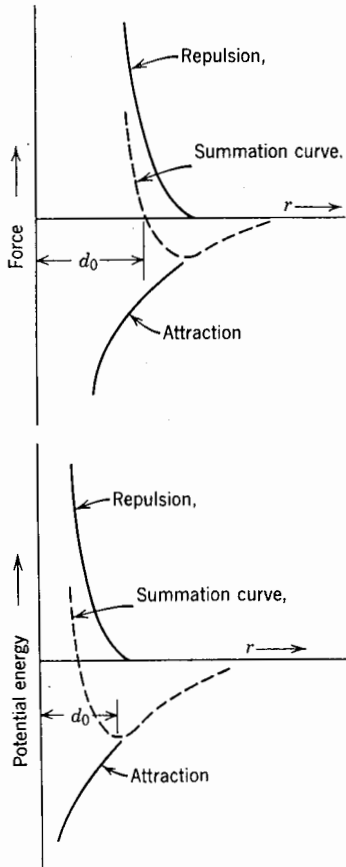


FIG. 1.—QUALITATIVE CURVES SHOWING (A) FORCE AND (B) POTENTIAL ENERGY AS FUNCTIONS OF INTERATOMIC SPACING, r

other as they approach closely enough to begin to overlap in space. Fig. 1a shows schematically how the attractive and repulsive forces combine to produce a stable situation where the atoms have zero force between them at an equilibrium interatomic spacing. The equilibrium spacing, d_0 , is shown in Fig. 1b to correspond with the minimum of

potential energy. since the equilibrium position is one of minimum energy, this spacing between the atoms will be stable. If a disturbing force were to cause a small displacement of the atoms from the equilibrium spacing, they would return to the stable energy minimum upon removal of the force.

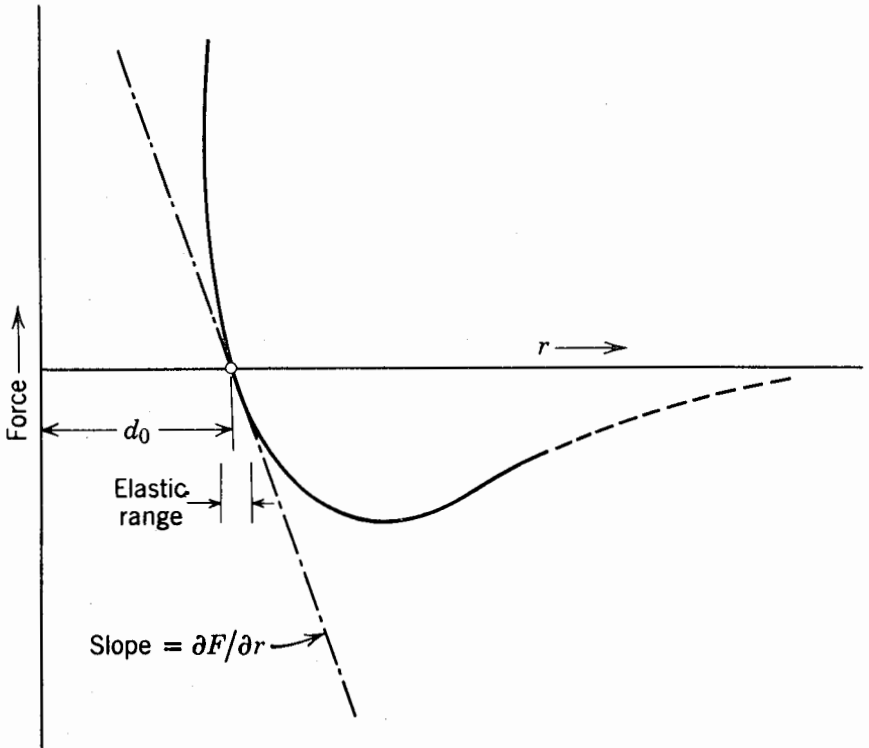


FIG. 2.—SUMMATION CURVE FOR FORCE VS. INTERATOMIC SPACING IS NEARLY LINEAR FOR SMALL DISPLACEMENTS FROM THE EQUILIBRIUM SPACING, GIVING RISE TO MACRO LEVEL LINEAR ELASTICITY

The same kinds of curves, drawn in Fig. 1 for two atoms, apply to large aggregations of atoms. Thus, in crystalline groupings of atoms the individual atoms are regularly spaced at an equilibrium distance from each other. The equilibrium interatomic spacing in this more general case is determined by the minimum of potential energy of the entire system consisting of many atoms.

In many materials of engineering importance the force-interatomic

spacing curve is nearly linear in the vicinity of the equilibrium spacing. This direct proportionality of force and deformation for small displacements gives rise to the modulus of elasticity on the macro scale. Fig. 2 shows how the force summation curve can be approximated by a

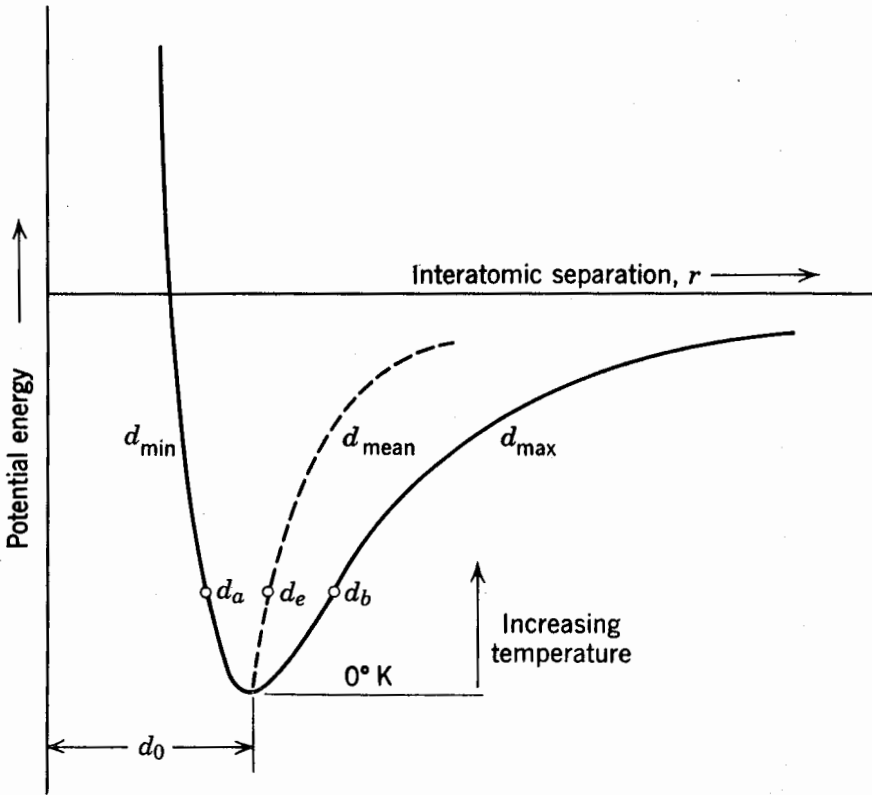


FIG. 3.—CHANGE IN INTERATOMIC SPACING WITH INCREASING TEMPERATURE; ASYMMETRY OF POTENTIAL ENERGY CURVE GIVES RISE TO MACRO LEVEL THERMAL EXPANSION (AFTER WULFF)

straight line at the atomic level. The integration of this force-displacement relationship over the large number of atoms present in a crystal-line matrix results in the elastic modulus observed in conventional materials tests.

Having related atomic level potential energy relationships to the mechanical response of materials, it is easy to extend these concepts to

the thermal response. At temperatures above 0° K, thermal energy causes atoms to vibrate about their equilibrium positions. As shown in Fig. 3, the atoms oscillate between the minimum and maximum interatomic spacings possible at any given energy level. The mean position represents the average interatomic spacing at a given temperature. Because the potential energy curve is asymmetrical, the mean position increases as the temperature increases. This increase in average interatomic spacing at the atomic level corresponds with the thermal expansion observed at the macro level. If the energy is increased by thermal

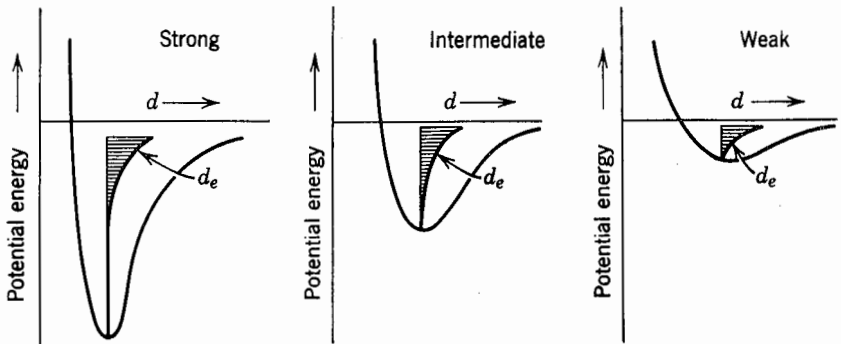


FIG. 4.—RELATION OF ELASTIC MODULUS, COEFFICIENT OF THERMAL EXPANSION AND MELTING POINTS TO EACH OTHER AS A FUNCTION OF ATOMIC BONDING STRENGTH (AFTER WULFF)

addition until the depth of the potential energy well is reduced to zero, the material melts.

The elastic modulus, coefficient of thermal expansion and melting point can thus be seen to be related to each other, since each is based upon potential energy-interatomic spacing relationships at the micro level. Modulus of elasticity is inversely proportional to the radius of curvature at the bottom of the potential energy well. The melting point is directly proportional to the depth of the well. The coefficient of thermal expansion is proportional to the degree of asymmetry of the energy curve. The way in which these three quantities vary with the relative strength of atomic bonding is shown in Fig. 4. A strongly bonded material has high modulus of elasticity and high melting point, with a low coefficient of thermal expansion. A weakly bonded material has relatively low values of modulus and melting point, with a high thermal expansion coefficient.

STRENGTH AND FRACTURE

The strength of materials, observed on the macro scale, can also be related to micro level structure and composition. Early calculations for the strength of a crystalline material considered the relative shear of two planes of atoms, as shown in Fig. 5. Taking atomic bonding

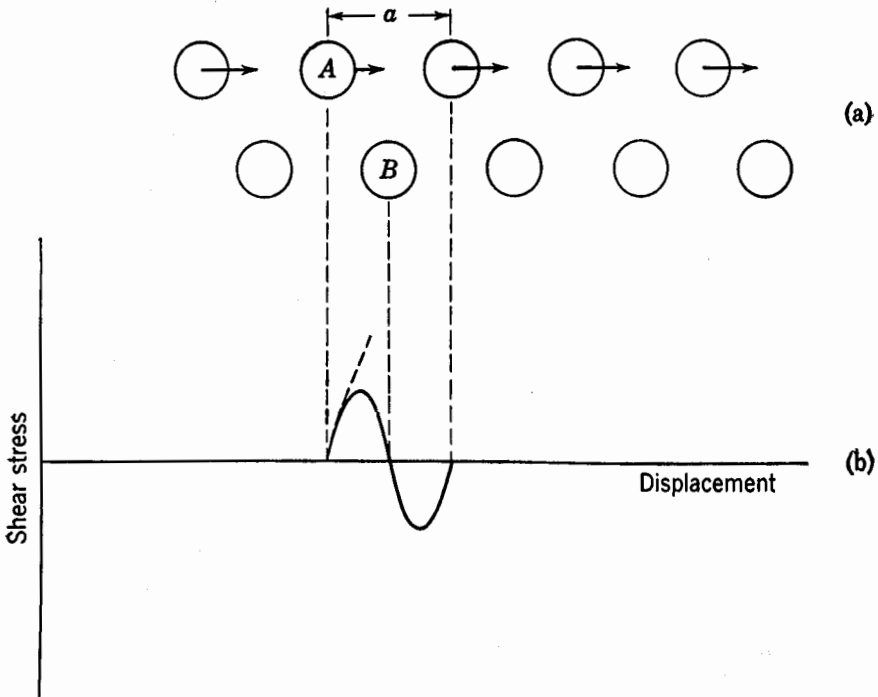


FIG. 5.—RELATIVE SHEAR BETWEEN PLANES OF ATOMS: (A) SCHEMATIC REPRESENTATION OF TWO PLANES OF ATOMS, (B) SHEAR STRESS AS A FUNCTION OF DISPLACEMENT OF THE ATOMS FROM THEIR EQUILIBRIUM POSITIONS

forces into consideration, theoretical strength was calculated as that stress required to permanently displace a whole plane of atoms from their equilibrium positions. Using this block-slip model, the elastic limit, B , is found to be approximately one-sixth the shear modulus, G , or $G/B \approx 6$. This theoretical value can be compared with actual observed G/B ratios in Fig. 6. An obvious discrepancy exists, with real materials having much lower strengths than those predicted by the block-slip atomic bonding model.

The dislocation model of slip in a crystal, introduced in the 1930's, resolves the discrepancy between calculated and observed strengths. In this model slip occurs in a stepwise manner caused by the presence of atomic scale imperfections, called dislocations, in the crystalline matrix. An edge dislocation can be visualized as an extra half-plane of atoms in a crystal. Dislocations are grown into crystals as accidents of growth, and multiply as the crystalline material is stressed. As shown schematically in Fig. 7, the dislocation moves under an applied stress by successively breaking and making new bonds with the extra half plane

	Shear Modulus G (dynes/cm ²)	Elastic Limit B (dynes/cm ²)	G/B
Sn, single crystal	1.9×10^{11}	1.3×10^7	15,000
Ag, single crystal	2.8×10^{11}	6×10^6	45,000
Al, single crystal	2.5×10^{11}	4×10^6	60,000
Al, pure, polycrystal	2.5×10^{11}	2.6×10^8	900
Al, commercial drawn	$\sim 2.5 \times 10^{11}$	9.9×10^8	250
Duralumin	$\sim 2.5 \times 10^{11}$	3.6×10^9	70
Fe, soft, polycrystal	7.7×10^{11}	1.5×10^9	500
Heat-treated carbon steel	$\sim 8 \times 10^{11}$	6.5×10^9	120
Nickel-chrome steel	$\sim 8 \times 10^{11}$	1.2×10^{10}	65

FIG. 6.—COMPARISON OF SHEAR MODULUS AND ELASTIC LIMIT (AFTER MOTT)

in a stepwise manner. An analogy can be drawn between this process and the moving of a floor rug: instead of pulling on one edge of a rug to move it a distance horizontally, one can effect a similar translation by making a hump in the edge of the rug opposite the desired direction of movement then running the hump across the rug. The stepwise movement takes less force, but finally achieves the same result. This is true in the crystal also, as show in Fig. 7b. The permanent deformation resulting from the passage of a dislocation through the crystal is similar to that which would result from the block slip of two portions of the crystal, but the required force for the former mechanism is considerably less.

The detailed strain field around an edge dislocation is shown schematically in Fig. 8. A region of tension exists on one side of the dislocation, with a region of compression on the other side where the extra atomic half-plane distorts the lattice. Foreign atoms of sizes different from the host matrix atoms tend to migrate to dislocation sites, replacing or fitting between the host atoms to partially relieve these strain

fields. This results in a decrease in strain energy, and a dislocation with such an atmosphere of foreign atoms tends to be locked against movement. The stress required to move a locked dislocation would be higher than that required to move a dislocation in a pure matrix, since the energy to move the dislocation out of its lower energy

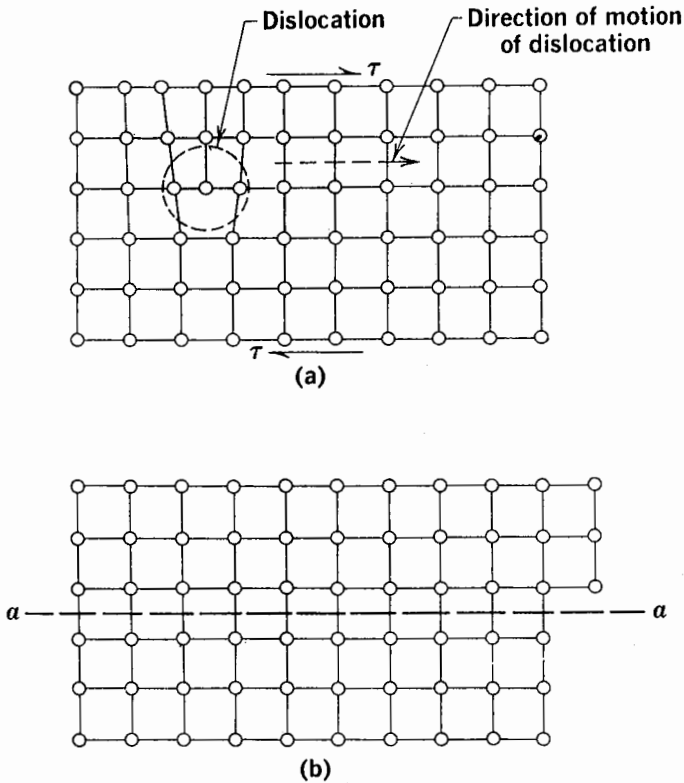


FIG. 7.—MOVEMENT OF AN EDGE DISLOCATION THROUGH A CRYSTAL, CAUSING SLIP

atmosphere and back into the pure matrix would have to be supplied. Once the dislocation has been broken from its atmosphere it can be kept in motion through the matrix by a somewhat lower applied stress. The upper and lower yield stress phenomenon in carbon steels is a macro level reflection of the dislocation unlocking phenomenon.

Once moving in a small single crystal, a dislocation travels to the edge of the crystal and pops out as plastic deformation. This is illus-

trated in Fig. 9 for the movement of three types of dislocations. A pure screw dislocation in a shear-type imperfection often found in crystals. A general dislocation consists of components of the pure edge dislocation described above and the pure screw dislocation. As shown in Fig. 9, the movement of any of these types of dislocations results in permanent deformation at the edge of the crystal.

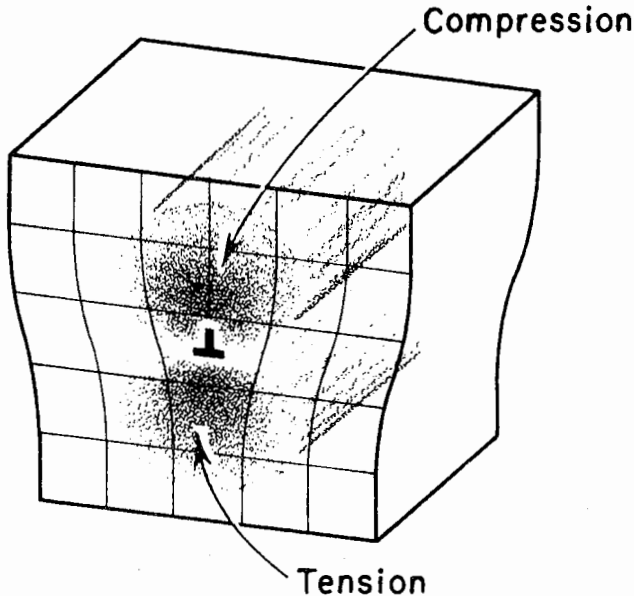


FIG. 8.—COMPRESSIVE AND TENSILE STRAIN FIELDS AROUND AN EDGE DISLOCATION

In polycrystalline metals, however, dislocations are not usually able to get a crystal surface and instead tend to pile up grain boundaries within the polycrystal. Fig. 10 shows how edge dislocations, generated by a multiplication source, pile up at an obstacle such as a grain boundary. Gross yielding of the material occurs when dislocations pop through the grain boundary, affecting adjacent grains.

Dislocation interactions can also inhibit the movement of dislocations through the matrix. The intersection of two dislocations can result in a jog in the dislocation line, as shown in Fig. 11. In certain situations such a jog anchors the dislocation, inhibiting further motion. The interaction of a dislocation with second-phase particles lying in

the slip plane can also inhibit movement of the dislocation under a given stress. Fig. 12 illustrates how a dislocation line must be forced past such second-phase particles, requiring a higher stress than that required for motion through a pure matrix. These dislocation interaction mechanisms, fully operative in the plastic range of loading where large scale movements of dislocations occur, lead to work hardening of

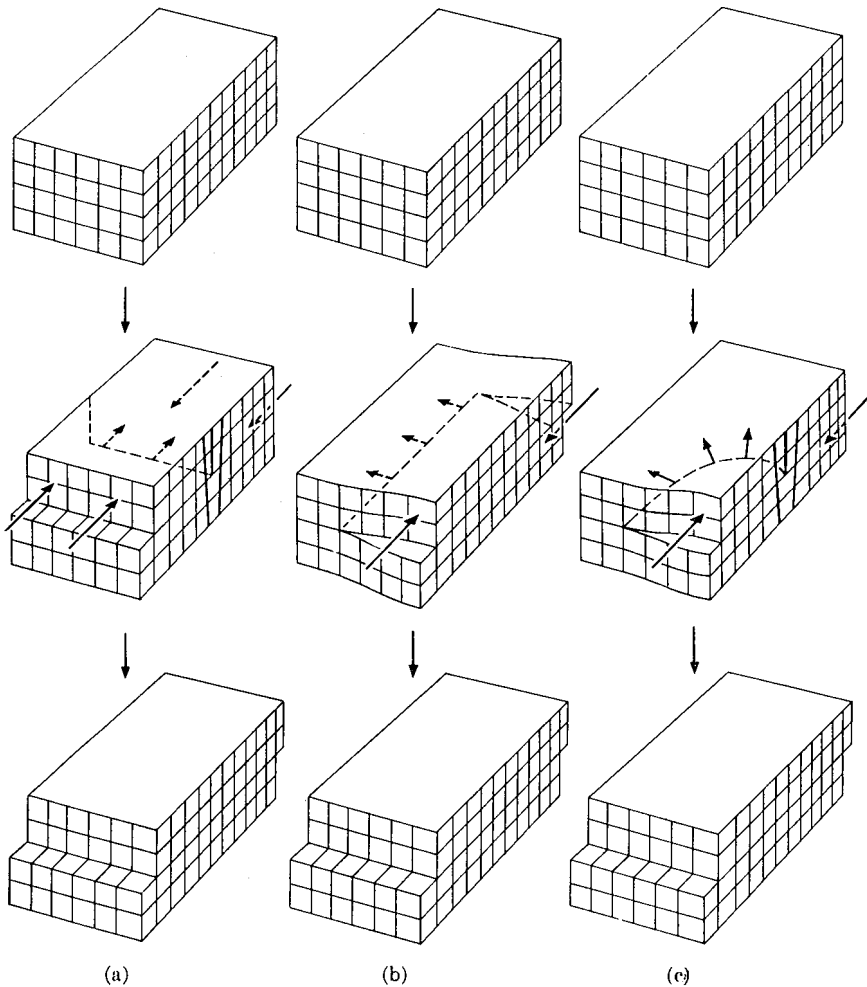


FIG. 9.—PLASTIC DEFORMATION CAUSED BY THE MOVEMENT OF (A) PURE EDGE, (B) PURE SCREW, AND (C) MIXED DISLOCATIONS THROUGH A CRYSTAL

the material. Thus in general, an increased stress is required to cause continued deformation in the plastic range.

Dislocation interactions can also lead to crack formation, thus contributing to the fracture of materials. Fig. 13 shows how dislocations can combine to open a crack at the intersection of slip bands. The extra half-planes of these edge dislocations lead to a large tensile buildup which results in localized crack formation.

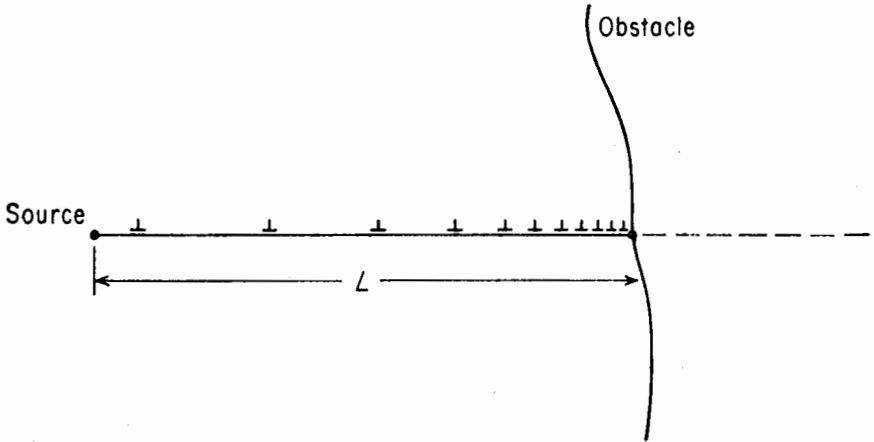


FIG. 10.—PILE-UP OF DISLOCATIONS AT AN OBSTACLE

DISLOCATION OBSERVATIONS

Although dislocations are an atomic scale phenomenon, evidence of their existence can be observed using microscopy techniques. Surface traces of dislocations can be optically observed after being marked by etch pitting, and bulk material observations of dislocations can be made using thin foil electron transmission microscopy.

Dislocation lines cannot end inside a crystal, so they either form closed loops or run out at a surface of the crystal. As an edge dislocation line cuts a surface, it is still surrounded by the localized tensile and compressive stresses which are caused by the presence of the extra half-plane of atoms. Fig. 8 is a schematic view of the atomic arrangement in a surface layer of atoms as an edge dislocation cuts through it. There is a finite strain energy stored in the portion of the crystal immediately surrounding the dislocation, due to the atomic misfit. When the entire surface is attacked by a strong etching solution, preferential

etching occurs at the sites of dislocation due to the higher energy states of the atoms there. An example of such etch pits, these in lithium fluoride, is shown in Fig. 14a. The arrangements of etch pits in a line through the center of this micrograph is a typical configuration called

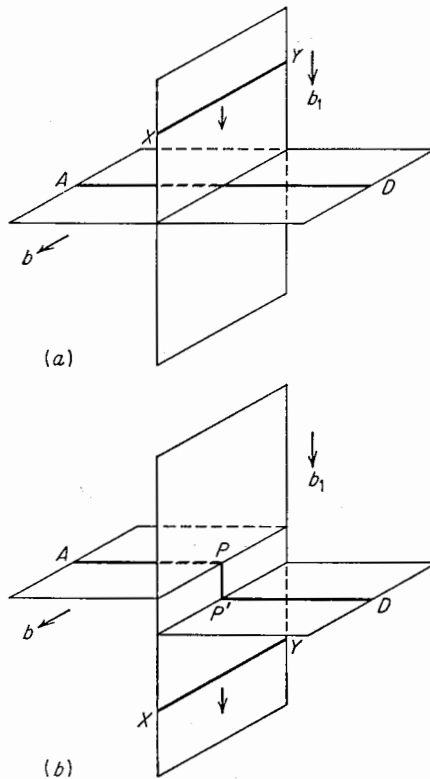


FIG. 11.—INTERSECTION OF TWO DISLOCATIONS, RESULTING IN THE FORMATION OF A DISLOCATION JOG (AFTER READ)

a low-angle grain boundary. The corresponding schematic representation of a low-angle grain boundary in Fig. 14b shows how a small change in crystal orientation occurs across such an arrangement of edge dislocations. The usual grain boundary observed at low magnification after metallographic etching is a large-angle orientation change which can be pictured as a complex arrangement of many more dislocations.

In order to observe dislocation arrangements in the bulk of the material, a thin section specimen is carefully cut from it. The section is cut in such a way that little mechanical damage is incurred by the specimen, then is thinned to 10^{-5} to 10^{-6} ch thickness by chemical

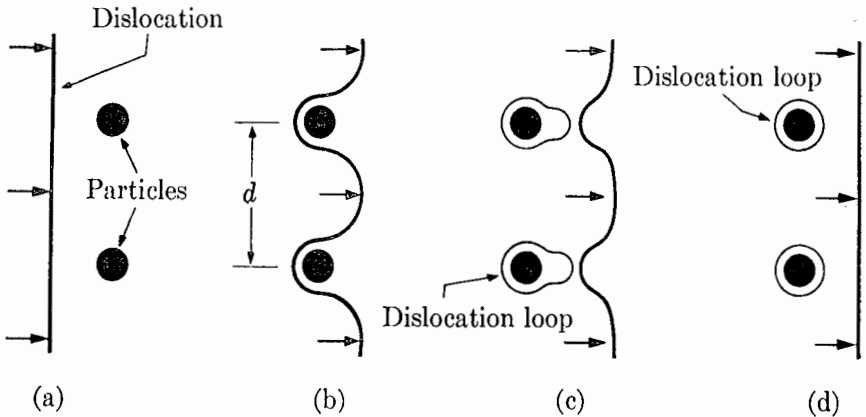


FIG. 12.—INTERACTION OF A DISLOCATION WITH SECOND-PHASE PARTICLES IN ITS SLIP PLANE

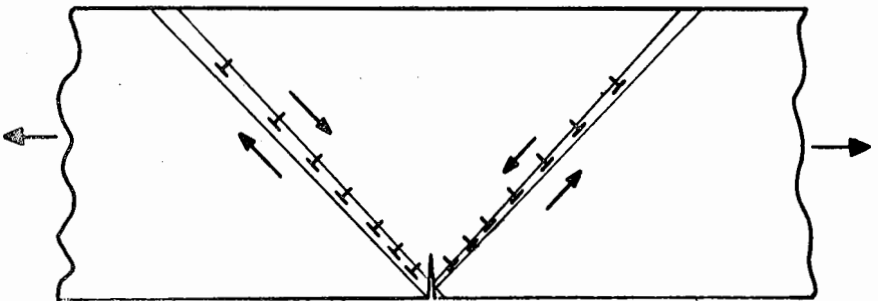


FIG. 13.—FORMATION OF A CRACK DUE TO THE COMBINATION OF EDGE DISLOCATION SLIP BANDS

polishing. In this form electrons accelerated through a high potential can be transmitted through the thin foil in a conventional electron microscope. Electrons hitting the areas of crystalline imperfection surrounding dislocations are diffracted out of the electron beam, so dislocations show up as dark lines on the fluorescent screen struck by the

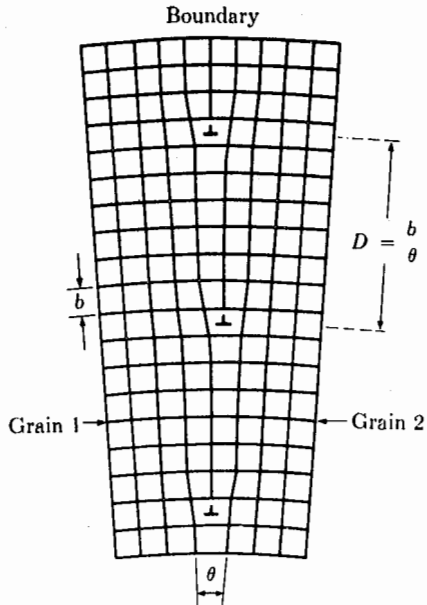
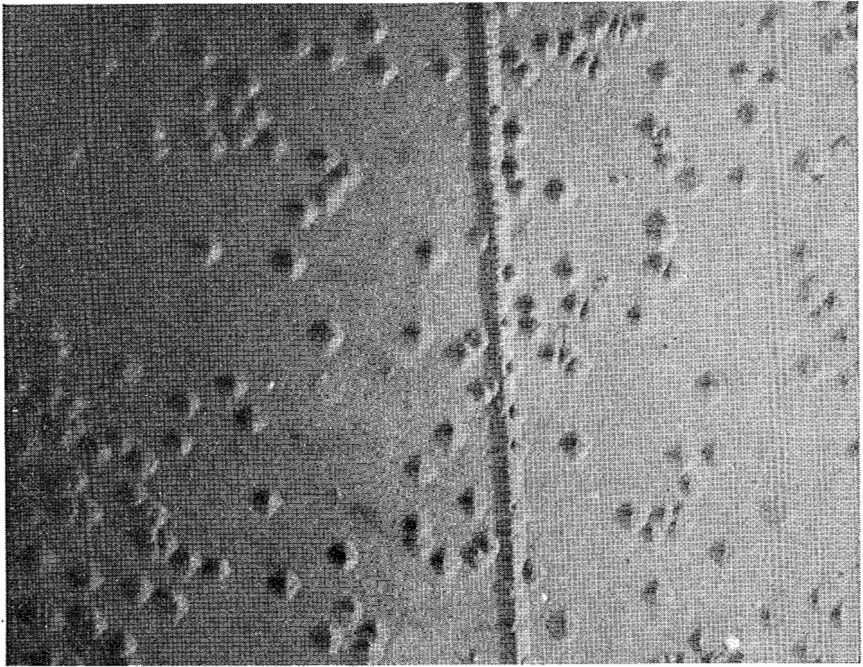


FIG. 14.—(A) ETCHED SURFACE OF LITHIUM FLUORIDE SHOWING ETCH PIT PATTERN CORRESPONDING TO LOW ANGLE GRAIN BOUNDARY. (B) SCHEMATIC REPRESENTATION OF LOW ANGLE GRAIN BOUNDARY OF EDGE DISLOCATIONS

undisturbed transmitted electrons. Fig. 15 shows dislocation lines in a thin foil of aluminum which has been prepared from a cyclicly loaded specimen.

The observation of dislocations by such techniques as etch pitting and thin foil transmission electron microscopy allows checks to be made on atomic scale models employed to explain the mechanical behavior of crystalline materials.

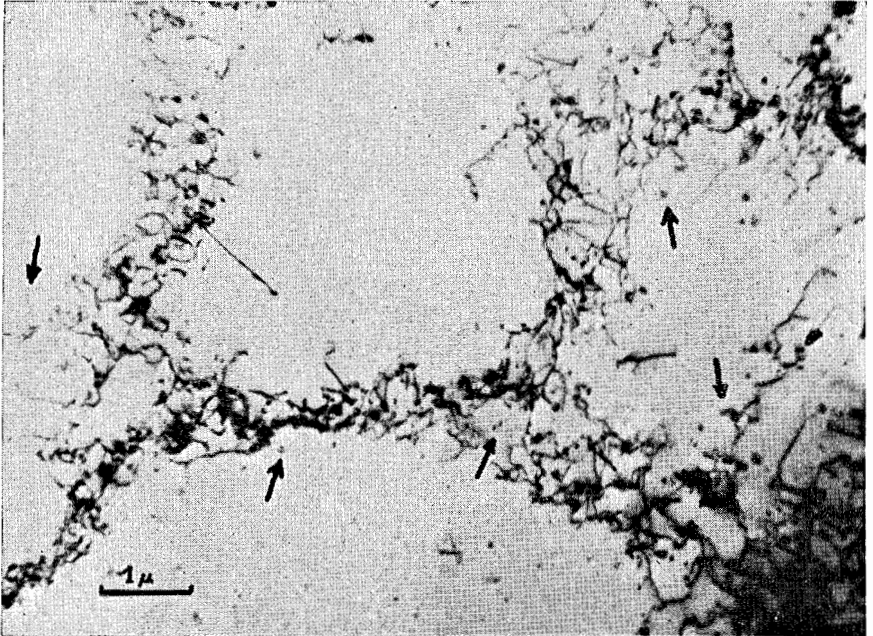


FIG. 15.—DISLOCATION STRUCTURE IN THE BULK OF AN ALUMINUM SPECIMEN, REVEALED BY THIN FILM ELECTRON MICROSCOPY TECHNIQUES. (FROM R. N. WILSON AND P. J. E. FORSYTH, ROYAL AIRCRAFT ESTABLISHMENT TECHNICAL NOTE MET, 311, JUNE, 1959)

STRESS-STRAIN CURVE

With the above introduction to dislocation theory, it is now possible to relate the micro and macro levels through the several stages of a full-range stress-strain diagram which represents the behavior of a metal subjected to a one-time loading to failure. In the stress-strain curve shown in Fig. 16, typical for a mild steel in tension, the micro-macro tie can be made at four stages: the elastic region, the yield point, the work hardening range, and fracture.

The macro level in the elastic portion of the curve is characterized by a linear relation between stress and strain and by fully recoverable deformation. This elastic behavior is based upon small changes in inter-atomic spacing with force at the micro level, as shown in Fig. 2. As long as the atoms return to their original equilibrium positions when

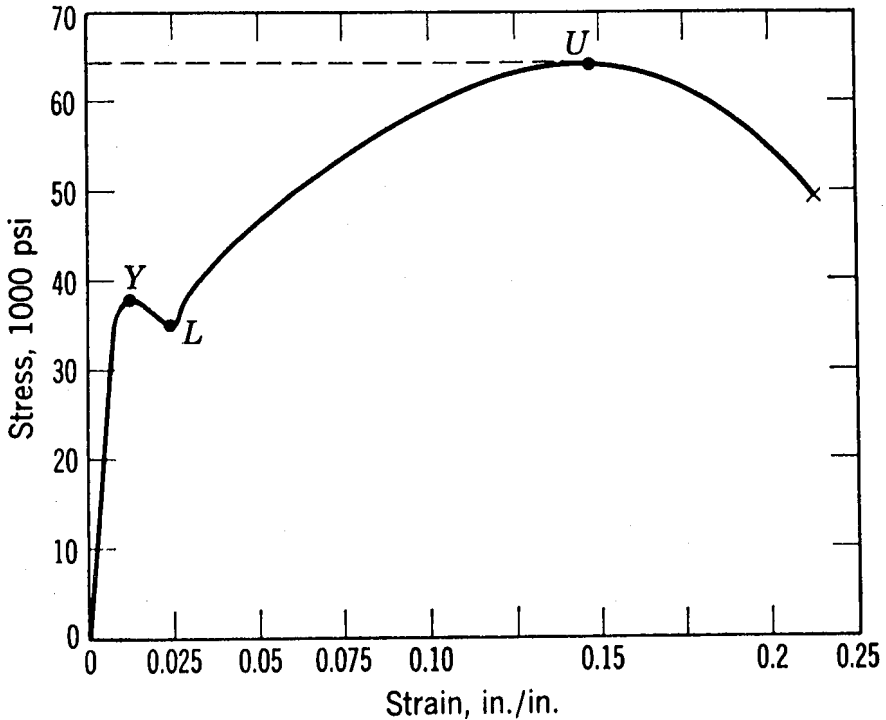


FIG. 16.—FULL RANGE STRESS-STRAIN DIAGRAM FOR A MILD STEEL
LOADED TO FAILURE IN TENSION

the applied stress is removed, reversible elasticity is observed at the macro level.

When the stress is increased beyond the elastic limit of the material, plastic flow or yield occurs. At the macro yield point, dislocations at the micro level have begun to move across slip planes and pop through the grain boundaries. The summation of such atomic level rearrangements results in the plastic strain associated with post-yield behavior. In the particular case of low carbon steel, the yield phenom-

enon consists of an upper and a lower yield point. These correspond with the initial breaking away of dislocations from their surrounding locking atmospheres of solute atoms and the subsequent easier continued movement of the dislocations across the slip planes.

As the dislocations continue to be moved after the yield point, dislocation multiplication and intersections begin to make it more difficult for continued movement. This increased stress required for continued plastic straining appears as work hardening at the macro level. The work hardening phenomenon continues right up to fracture, and results in increased hardness and strength of the material. On a true stress-strain curve the stress would continue to rise until fracture occurred, but on the engineering curve shown in Fig. 16 nominal stress decreases as the material decreases in area due to localized necking.

Microcracks formed at the head of dislocation pileups and at other stress concentrators begin to grow at high stresses. Such cracks eventually join together and grow to a critical size for rapid propagation to macro level fracture.

STRENGTHENING MECHANISMS

Several strengthening mechanisms are employed to raise the strength levels of structural metals. These mechanisms—solid solution alloying, precipitation hardening, grain size control, and work hardening—can each be understood in terms of dislocation mechanics.

Solid solution alloying begins with the random distribution of relatively small numbers of alloying element atoms throughout the metal matrix. Since it represents a lower strain energy state than does the random solution, these solute atoms regroup to form atmospheres around the dislocations present. This formation of an atmosphere is shown schematically in Fig. 17 where the two large solute atoms shown are only representative of the large number that may constitute the atmosphere. Once anchored by an atmosphere of solute atoms a dislocation requires a higher stress to move it, so the presence of the alloying element atoms has the effect of strengthening the host metal matrix.

If the alloying element atoms are formed into small second phase particles throughout the metal matrix instead of being in solution, a different kind of strengthening occurs. Second phase particles can be dispersed through the host matrix by a process known as precipitation hardening. The metallic alloy is first raised to a temperature high

enough for the alloying element to be distributed randomly, then the material is quenched to a low temperature to lock in this random arrangement. The metal is then raised to a moderate temperature just high enough to give diffusional mobility to the alloying atoms. These

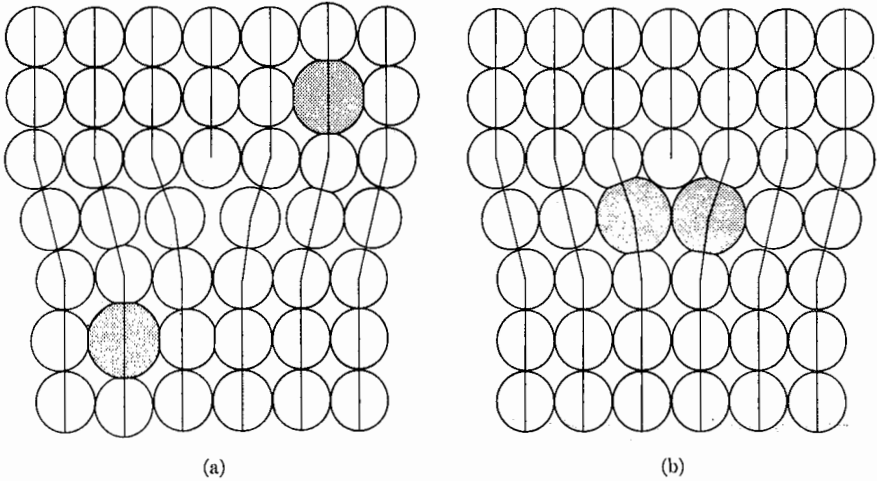


FIG. 17.—FORMATION OF A LOCKING ATMOSPHERE OF SOLUTE ATOMS AROUND AN EDGE DISLOCATION

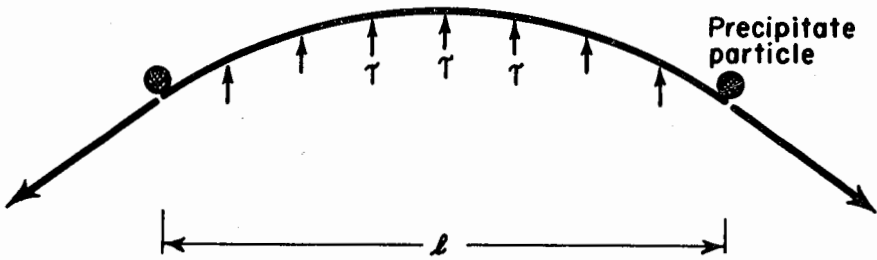


FIG. 18.—BLOCKING OF DISLOCATION MOVEMENT BY CLOSELY SPACED PRECIPITATE PARTICLES

atoms, since they are less soluble in the host matrix at relatively low temperatures, precipitate out as a second phase. If the heat treatment is properly programmed, many small precipitates are formed at close spacings. As shown in Fig. 18, such precipitate particles serve to restrain the movement of dislocations which are being driven by an applied stress. The stress required to force a dislocation past precipitate particles

at a spacing l is proportional to $1/l$, so the more finely spaced the precipitate particles the more the metal is strengthened. It is important to note that a precipitation hardened alloy subjected to a high temperature, such as occurs locally in welding, can loose its strength as small

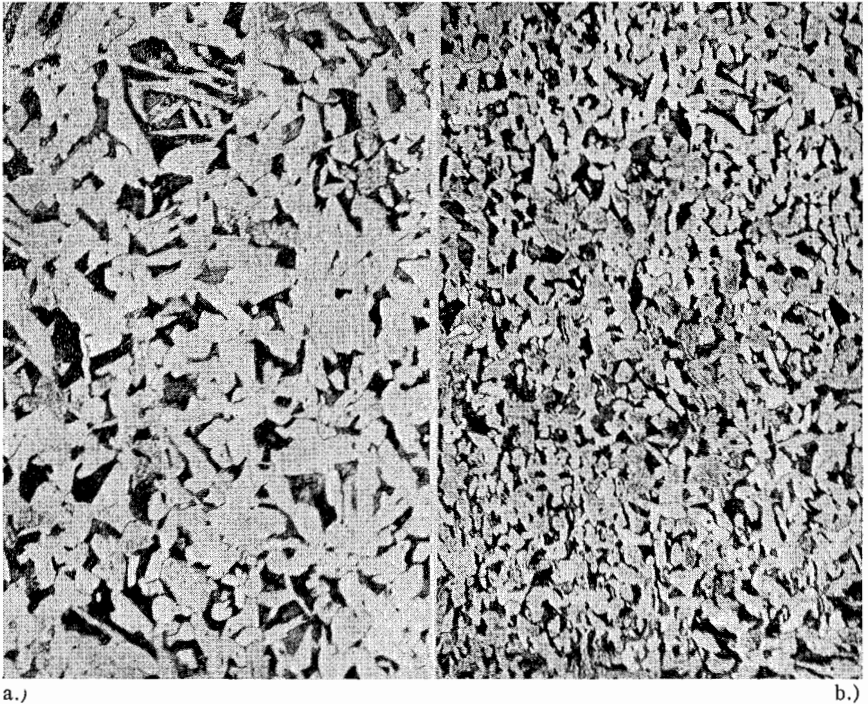


FIG. 19.—MICROSTRUCTURE OF STRUCTURAL STEELS AT 100 X: (A) ASTM-A36 STEEL,
(B) A SIMILAR STEEL WITH SMALL AVERAGE GRAIN SIZE

precipitate particles merge into fewer large particles at greater spacings.

Grain boundaries, the discontinuities between crystalline regions of different orientations in polycrystalline metals, are also effective in inhibiting dislocation movements. Fig. 19a shows a typical micrograph of an A36 steel, etched and shown at 100X magnification to reveal the grain structure. Polycrystalline metals with a small average grain size exhibit higher strength than those with larger grain size, with strength being proportional to $d^{-1/2}$, where "d" is the average grain diameter.

Thus, the steel shown in Fig. 19b with a smaller average grain size than that in Fig. 19a, would show higher strength than the virgin A36 steel.

If a metal is strained well past the yield point and into the plastic range, dislocation interactions result in work hardening of the material. If it were unloaded from a point on the work hardening portion of the stress-strain curve, the metal would show a yield strength approximately equal to the stress from which it had been unloaded, if it were then loaded a second time. Thus, work hardening can be employed to strengthen metals, using dislocation interactions to inhibit plastic deformation due to dislocation movement.

These several strengthening mechanisms have in common the inhibiting of dislocation movements under an applied stress. Each of these mechanisms, and often combinations of them, are used to strengthen various commercially available structural metals.

HIGH STRENGTH STEELS

The strength levels of structural steels have been increasingly raised in the past couple of decades as the strengthening mechanisms described above have become more understood and more used. The stress-strain curves sketched in Fig. 20 show the wide range of structural steels commercially available today.

The lowest curve shown in Fig. 20 represents ASTM A36, the carbon steel which is used in most standard structural applications today. In this steel carbon is the main alloying element, and its presence and subsequent dislocation locking develop the observed strength levels. No special heat treatment is applied, but microstructure and grain size are roughly controlled in the hot rolling process where the material is formed into structural shapes.

The second curve represents the high strength, low alloy steels. These somewhat higher strength steels, with yield strengths in the 50,000 psi range, were added to the AISC building code in 1961 to give an option to the standard carbon steels. In the past few years these steels have been heavily used in buildings and other constructed facilities. The high strength, low alloy steels derive their strength levels from the addition of other alloying elements to the usual carbon. These alloying elements produce a solid solution strengthening of the iron matrix. The third curve in Fig. 20 represents a similar class of steels, the columbium-vanadium bearing steels. These steels employ small percentages of columbium and/or vanadium as alloying elements.

As in the case of the structural carbon steels, neither the high strength, low alloy steels nor the columbium-vanadium bearing steels are given a special heat treatment in their production.

In order to raise strength levels higher, structural steel producers

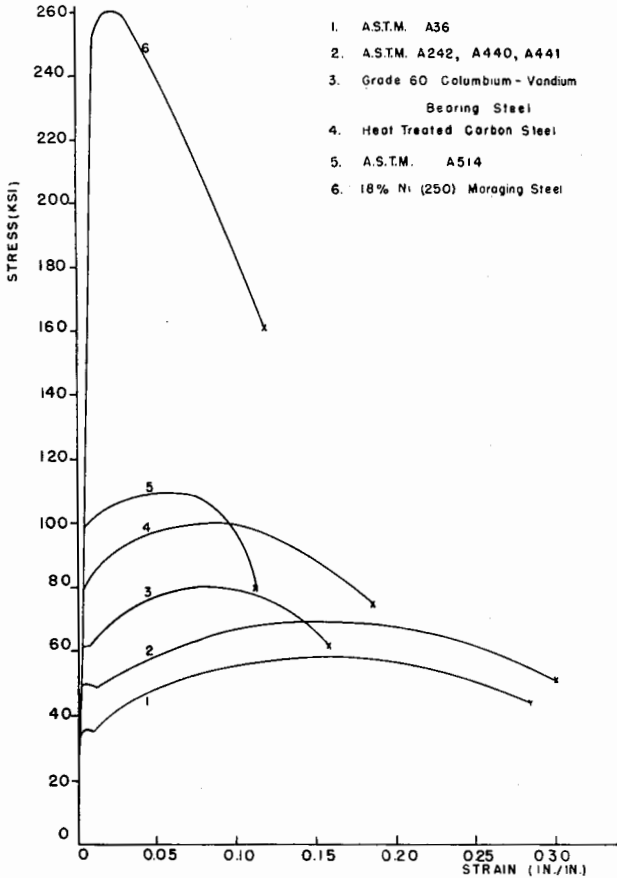


FIG. 20.—TYPICAL STRESS-STRAIN CURVES FOR SEVERAL COMMERCIALY AVAILABLE STRUCTURAL STEELS

have devised heat treatments which will produce precipitation hardening of the basic iron matrix. The fourth curve in Fig. 20 represents a heat-treated carbon steel, produced by *quenching* the steel from a high temperature then *tempering* it at a moderate raised temperature to allow precipitate particles to form. The fifth curve shows a heat-treated

constructional alloy steel, produced by quenching and tempering a steel to which both carbon and other alloying elements have been added. The heat-treated constructional alloy steels, available in the 100,000 psi strength range since the early 1950's, have recently been designated ASTM A514. Both the heat-treated carbon steels and

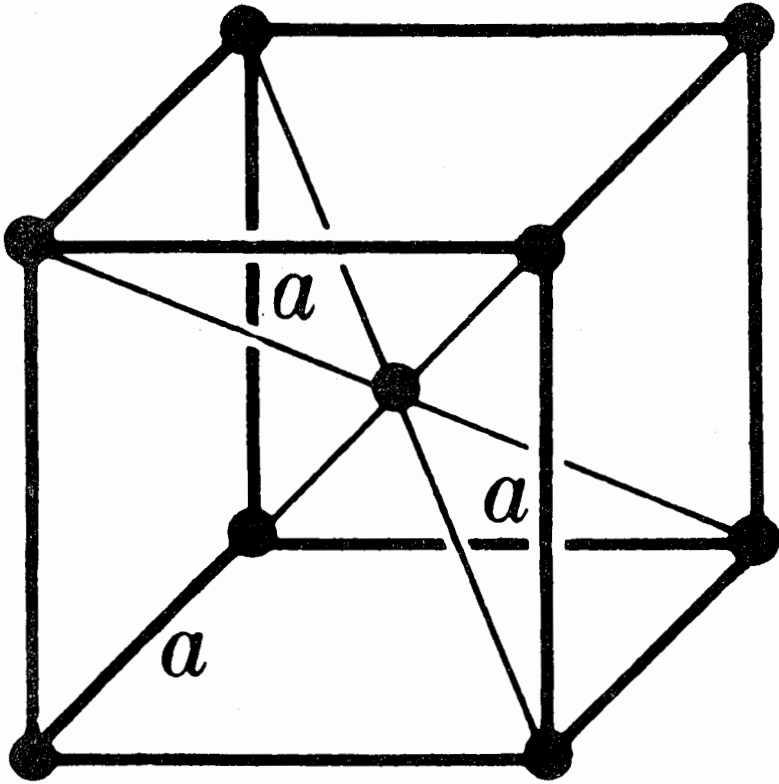


FIG. 21.—BODY-CENTERED CUBIC ARRANGEMENT OF ATOMS IN A CRYSTAL

heat-treated constructional alloy steels rely on carbon-based precipitation for their precipitation hardening strengthening.

The highest stress-strain curve shown in Fig. 20 represents the maraging steels. These steels are unique among structural steels in that they do not employ carbon as an alloying element. Instead of carbon, relatively large percentages of nickel are employed to produce strength levels in the 200,000 to 300,000 psi range. Though a special

quenching and tempering heat treatment, the maraging steels are precipitation hardened by a finely dispersed nickel-based precipitate. The close spacing of these precipitate particles results in very effective blocking of dislocations, thus allowing the development of high strength levels.

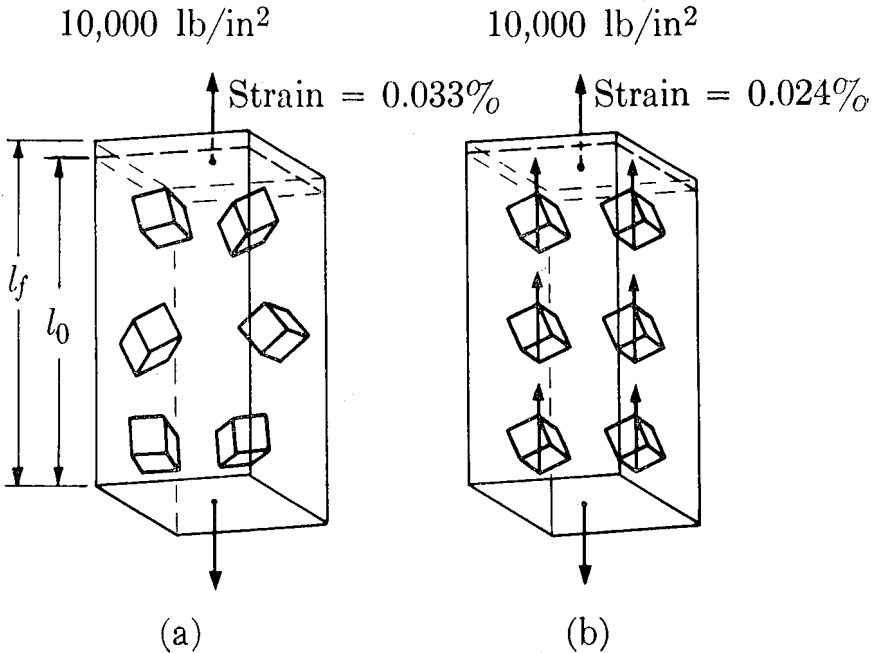


FIG. 22.—EFFECT OF PREFERRED ORIENTATION IN A POLYCRYSTALLINE IRON BAR ON ITS ELASTIC BEHAVIOR: (A) RANDOM ORIENTATION, (B) PREFERRED ORIENTATION WITH BCC CUBE DIAGONALS ALONG THE AXIS OF THE BAR (AFTER GUY)

STIFFNESS CONTROL

Although the strength levels of the steels shown in Fig. 20 range over almost an order of magnitude, the modulus of elasticity of each of the steels shown is essentially the same. This fact often creates a problem in the use of higher strength steels in structural applications in that deflection and stability considerations become limiting factors. The usefulness of the higher strength steels would be enhanced if their moduli of elasticity could be increased as strength levels are improved.

The elastic modulus is determined by the amount of strain ob-

served at various imposed stresses. This macro level strain arises from elastic stretching of atomic bonds and from small recoverable movements of dislocations at the micro level. Stiffness can be enhanced by

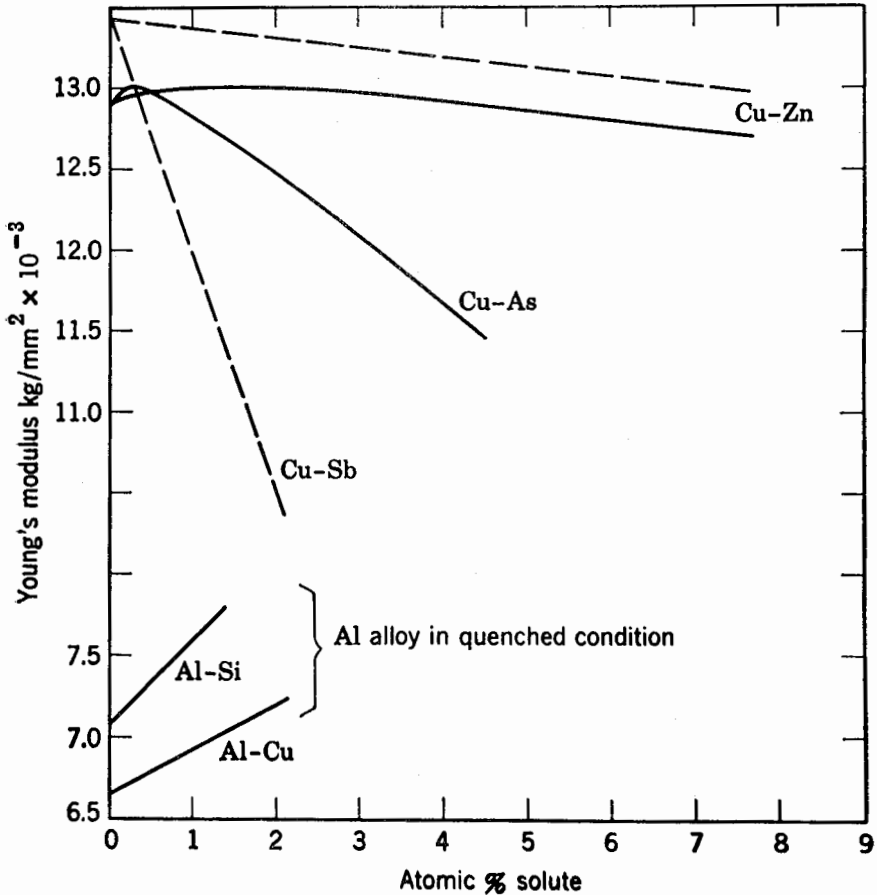


FIG. 23.—EFFECT OF SOLUTE ATOMS ON ELASTIC MODULUS (AFTER McLEAN)

improving the effectiveness of the atomic bonds and by more effectively blocking even the small movements of dislocations.

Atoms in a metal have a characteristic crystalline arrangement. Iron atoms in steel at normal temperatures have the body-centered-cubic arrangement shown in Fig. 21. Aggregates of atoms in this form

show different amounts of deformation under equal forces applied in various directions. For example, if a crystal of the bcc arrangements shown in Fig. 21 were loaded alternatively with the same stress on the cube faces and across the diagonal of the cube, different amounts of strain would result. Bonding across the cube diagonal would be seen to be more effective than that across cube edges, so strain would be

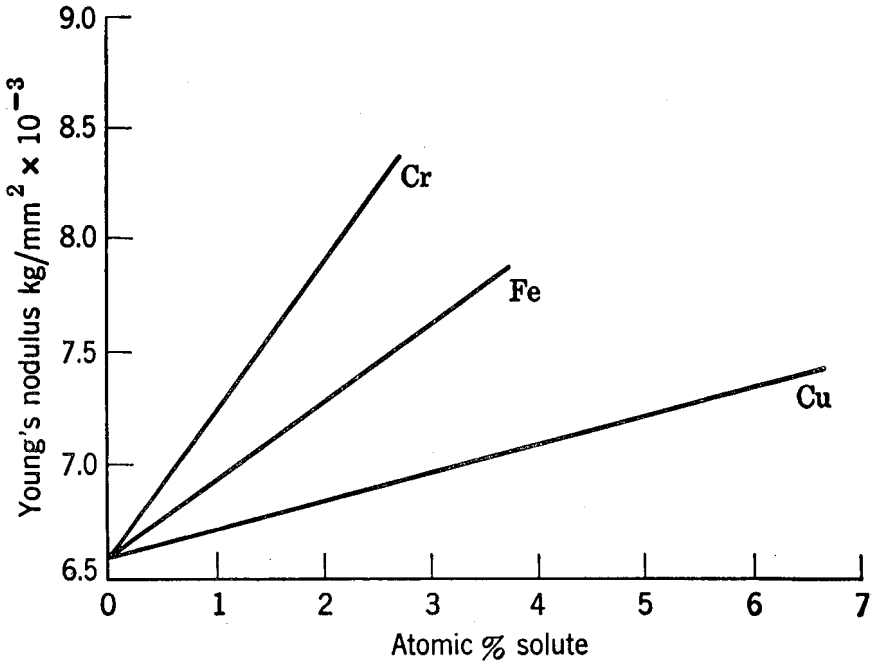


FIG. 24.—EFFECT OF SECOND PHASE PARTICLES ON THE ELASTIC MODULUS OF ALUMINUM (AFTER DUDZINSKI)

lower across the diagonal. If a large single crystal of iron were oriented with the bcc diagonal in the direction of loading, a minimum value of elastic modulus would be observed. In a polycrystalline metal it is often possible to produce preferred orientation of the crystalline structure by rolling operations. The effect of preferred orientation in an iron bar is shown schematically in Fig. 22. Here the elastic behavior of a polycrystalline alloy is enhanced if the normal random orientation of the crystal axes is changed to a preferred orientation along the bcc cube diagonal.

The elastic modulus can also be affected by the addition of solute atoms to the basic metallic matrix. Fig. 23 shows how the moduli of some aluminum and copper systems change as a function of the addition of certain solute atoms which form a solid solution. The effect of the addition of solute atoms which form hard second phase particles, due to a heat treatment processing, is shown in Fig. 24.

Evidently the modulus of elasticity of certain metals can be changed by atomic level manipulations. To date the ability to change modulus in desired directions has not been perfected for commercial quality metals such as the high strength steels. If an answer to this stiffness problem is found, it will be based largely on the increased understanding of the macro-micro relationships made possible by an ever developing materials science.

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