Recovery, Recrystallization, and Grain-Growth

Iqra Zubair Awan

[1]*Institute Charles Gerhard, Ecole Nationale Superieure de Chimie,
34040 - Montpellier, France.*
iqrazubair@gmail.com*

Summary: This is a brief review of the important phenomena of recovery, recrystallization as well as grain-growth. The three mentioned phenomena are the mechanisms by which metals and alloys fix the structural damage introduced by the mechanical deformation and, as a consequence, in the physical and mechanical properties. These rehabilitation mechanisms are thermally activated. For this process, the materials have to be heated and any such heat-treatment is meant to reduce deformation-induced break is termed annealing. Other or different heat-treatments lead to recovery and recrystallization. It is rather strange that, though these phenomena are extremely important in metallurgical science and engineering, not so much work has been done as that in corrosion and shape memory technologies. An attempt has been made here to summarize all important aspects of these phenomena for the benefits of students of metallurgy, chemistry and solid state physics.

Introduction

The dynamic phenomenon of recovery, recrystallization and grain-growth play a very important role in Physical Metallurgy, Chemistry of the Solid State and Solid State Physics. Hence this review for the benefit of students and teachers of the above-mentioned disciplines. Humphreys and Hatherly (referred to later) have stated that: "Recrystallization is related to annealing phenomena which happen during thermomechanical processing of materials has accepted as being mutually of technological importance and scientific interest. The phenomena include most widely studied in metals and there be a vast literature spanning over about 150 years. Metallurgical research in this field is mainly determined by the necessities of industry, mainly for metal forming processes for better products."

Dierk Raabe [1] of Max-Planck Institute fuer Eisenforschung, Düsseldorf, Germany has written an excellent article in his book, Physical Metallurgy, Fifth Edition, 2291-2397 (2014) on Recovery and Recrystallization: phenomenon, Physics, Models, and Simulation. He has mentioned the earliest investigators who had worked on deformation of metals and related phenomena. The works of Kalirher [2], Sorby [3] 1886 and 1887, Stead [4] 1898, Rosenhain [5] 1914, Ewing and Rosenhain [6] 1899a and 1899b, Alterhum [7] 1922, Carpenter and Elam [8] 1920, Czochoralski [9] 1927, Burgers and Louwerse [10] 1931 and Burgers [11] 1941. Surprisingly, the excellent work of Erich Schmidt and Walter Boas [12] 1935/1950 has not been mentioned.

Terminologies

"The mechanism in which metals and alloys repair the structural damage caused by deformation, and incidentally through the resulting changes in the physical as well as mechanical properties and the fix mechanisms which are induced by thermal treatment, are term as annealing, recovery, recrystallization and grain growth. We will define these terminologies one by one. These structural changes take place in annealing after cold deformation during hot working. Since all these structural changes require diffusion, they depend on thermal activation to cause rearrangement of dislocations and grain limitations [13, 14].

Annealing

"The term annealing is a very general term used in heat treatments of metals and alloys. It includes any cycle of heating and cooling of metallic materials, irrespective of temperatures, rates and terms involved in the cycle. Depending upon the material being treated and the object of treatment, annealing operations may involve any of a very broad range of heating rates, soaking temperatures and soaking times and cooling rates. These methods can be used for stress-relieving, recrystallization and grain growth.

During annealing, the material is heated to a temperature not far below the critical seized for a adequate period of time to achieve the desired changes meant for the heat treatment, and then cooled to room temperature on a desired time. The purpose of annealing is to eliminate partially or completely the strain hardening produced by earlier mechanical forming operations, so that it can be put into service in a relatively soft, ductile condition. It must be realized that annealing involves recrystallization in which a combination of cold working and subsequent heating causes new stress-free crystals in a matrix which is itself stress free. A full anneal treatment is a process intended to reduce the metal treated to its softness possible condition. The term, "annealing" is generally used, without qualifications or further description, as full anneal. The full anneal is widely used on steels to remove the effects of cold working, eliminate residual stress and improve the machinability of medium and high carbon grades.

"Cold work is supposed to be a plastically deformation of metal by temperature that are low relative to its melting point. A rough rule-of-thumb is to assume that if esthetics deformation carried out on temperature lesser to its one-half M.P it corresponds to cold functioning Mainly the energy depleted in cold work appears in the form of heat but a fixed fraction is stored in the metals as strain energy related with a variety of lattice defects, dislocations, vacancies, formed by the deformation. The total energy stored depends on the deformation process and several other variables such as composition of the metal, the speed and temperature of deformation. The data from the work of Gordon [15], shows that the stored energy increases with increasing temperature, however on a decreasing rate, the fraction of total energy stored decreases with increasing deformation. The latter effect is shown in Fig. 1 "[15]. The quantity of stored energy can be significantly increased by increasing the magnitude of deformation, lowering the deformation temperature and through alloying an uncontaminated metal."[13]. "We know that cold working increases greatly the quantity of dislocations in a metal. A soft annealed metal be able to have dislocation densities of 10^6-10^8 per cm square, and heavily cold worked metal up to 10^{12}. These dislocations represent crystal defects with associated lattice strains; - increase the dislocation density increases the strain energy of the metal "[13].
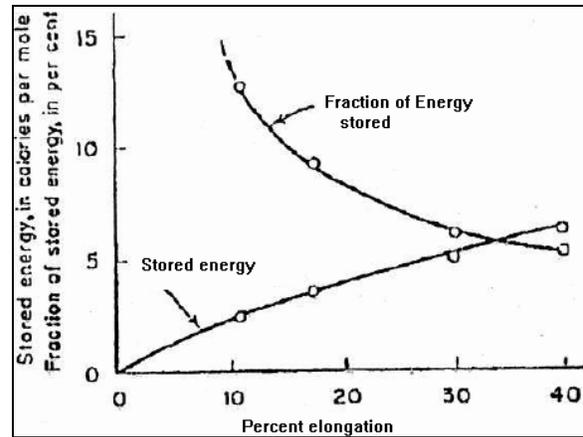

Fig. 1:   (a) Stored energy of cold work and fraction of the total work of deformation remaining as stored energy for high purity copper plotted as functions of tensile elongation (From Data of Gordan, P., Trans.AIME, 203 1043 [1955]). [14].
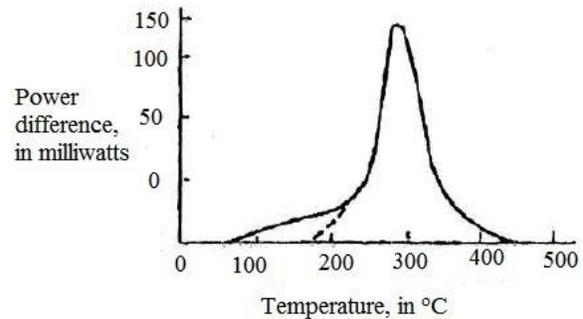

Fig. 1:   (b) Anisothermal anneal curve. Electrolytic copper. (from Clarabrough, H. M., Hargreaves, M. E., and West, G. W., *Proc*. Roy. Soc. Lomdon, 232 A, 552 [1995]). [14]
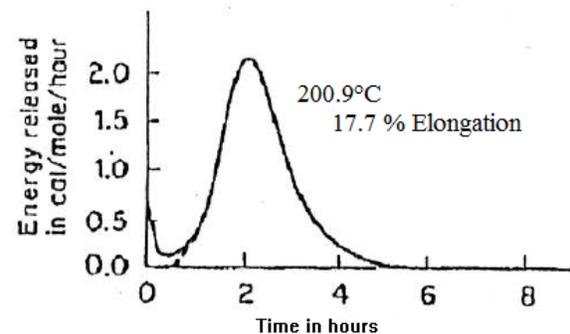

Fig. 1:   (c) Isothermal anneal curve. High purity copper. (From Gordon, P., Trans AIME , 203 1043 [1955].) [14].

"Retained energy is caused by point defect in cold work. A screw dislocation can cut another screw dislocation and generate a close-packed line of vacancies and interstitials similar to it glides with the point defect produced depending on the relative Burgers vectors of the intersecting dislocations Because strain energy related to vacancy is smaller than with related to interstitial atom the vacancies are formed in larger number than interstitial atoms [13].

Cold work metal may soften due to liberated energy of deformed metal is grater than that of annealed metal that is equal to stored strain energy Therefore distorted metal on heating speedily return to soften state [15].

The measurement of stored energy in a sample has been described nicely by Clarebrough *et.al* [16] and reproduced by Reed-Hill [14]. They have used anisothermal and isothermal annealing methods to do it. Fig. 2 and 3 show the experimental results. Both the anisothermal and isothermal curves. (Figs. 2, 3) illustrate maxima corresponding to the huge energy releases. Metallographic examination of these samples show the growth of a totally new set of strain-free crystals which grow at the rate of the novel roughly distorted crystals. This method is termed as recrystallization and is due to rearrangement of the atoms into crystals by a lower free energy [13-16].
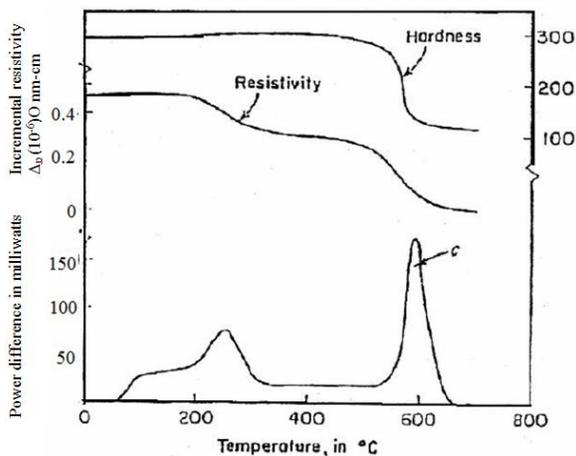

Fig. 2:   Anisothermal- anneal curve for cold worked nickel. At the top of the Fig curves are also drawn to show the effect of annealing temperature on the hardness and incremental resisitivity of the metal. (From the work of Clarebrough, H.M. Hargreaves, M. E. West, G.W., Proc. Roy. Soc., London, 232A 252 [1955]) [14].
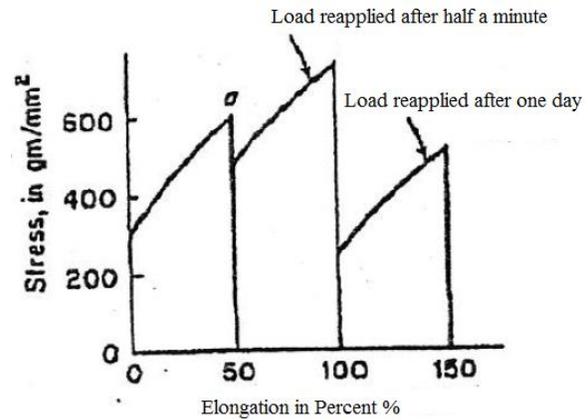

Fig. 3:   Recovery of the yields strength of a zinc single crystal at room temperature (After Schmid, E., and Boas, W., Kristallplastizitat, Julius Springer, Berlin{1935}

"According to Reed-Hill [14], the major energy release of the curves (Figs. 2 and 3), correspond to recrystallization but both the curves show that energy is released before recrystallization. The dashed lines systematically, on both curves, delineate the recrystallization portion of the energy release. The areas below each solid curve that lie to the left and over the dashed lines correspond to an energy release not related to recrystallization. In the anisothermal anneal curve, this release of strain energy starts at temperatures well below those at which crystallization starts. Similarly, in the isothermal anneal curve, it initiate at the beginning of annealing cycle also it is almost complete before recrystallization starts. The fraction of the annealing cycle that occurs before recrystallization is called recovery. Reaction that takes place in recovery stages are significantly maintin progress of recrystalization in those areas which are still not converted into crystals." [14].

"We have so far cursoraly discussed annealing, recovery, recrystallization and grain growth. But before discussing these phenomena in detail, it is important to define the third state of annealing, viz., and grain growth. When recrystalization is completed during annealing, grain growth occurs. In grain growth, certain of the recrystallized grains continue to grow up in size, however only at the expense of new crystals, which should disappear. The important aspects of these phenomena are now discussed.

We understand that "annealing, recovery, recrystallization and grain growth are microstructural

changes that happen during heating (annealing) later than cold plastic deformation or during hot working. These 3 mechanisms are sometimes referred like restoration processes as they resolve the microstructural configuration to a lower energy level" [17]. In addition to the fine works by Reed-Hill [14] and Wu Hsun /Makins/ Semiatin [17], 3 other excellent works by Mittemeijer [18] and Humptereys, F.J., M. Hatherly [19] and Raabe [1] are a must for Physical Metallurgy/Solid State Chemistry students and teachers. The following discussion is based on the works of the above mentioned authors.

Recovery

In cold work changes takes place in both physical and mechanical properties. Plastic deformation increases strength, hardness and electrical resistance but it decrease ductility." [14]. "The initial change in structure and properties that occur upon annealing a cold-worked metal is considered the beginning of recovery. As recovery progress, the following structural changes occur in sequence. [18].

1. The disappearance (annealing out) of point defects, vacancies and their clusters.
2. The eradication and reorganization of dislocations.
3. Polygonization (Sub-grain formation and Sub-grain growth).
4. The arrangement of recrystallization nuclei energetically capable of more growth." [18]

"These structural changes do not engage high angle boundary migration. Consequently, during this stage of annealing, the quality of the deformed metal essentially does not change." [18]

"In the recovery phase of annealing, the physical and mechanical properties tend to recovery their original values. That the variety of physical and mechanical properties does not recover their values at the same speed indicates complicated nature of the recovery method. Fig. 4 shows another anisothermal curve corresponding to the energy released on heating cold work polycrystalline nickel. Point C defines the region of recrystallization. Plotted on the same diagram are curves showing change in electrical resistivity and hardness as a function of temperature. It is visible that the resistivity is completely recovered before the start of recrystallization. But, the most important change is in the hardness simultaneously with recrystallization of the matrix. [18].
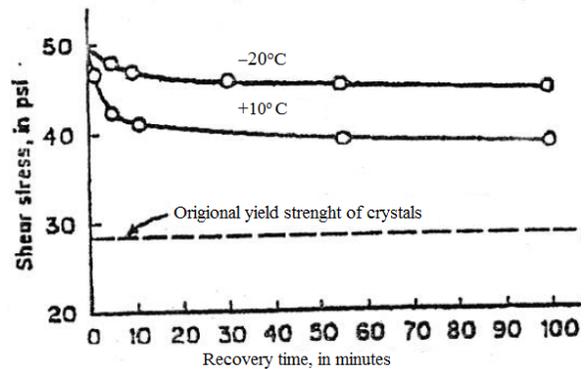

Fig. 4: Recovery of the yield strength of zinc single crystal at two different temperatures. (From the data of Drouard, R., Washburn, J., and parker, E.R., Trans. AIME, 197 1226 {1953}.) [14]

Recovery in Single crystals

"According to Reed-Hill [14], the density of the cold worked state is directly related to the complexity of the deformation that produces it. The lattice distortions are simpler in a single crystal distorted by simple glide than in a particular crystal deformed by many glide (simultaneous slip on numerous system), and lattice distortions might be still more dense in a polycrystalline metal. If a single crystal is deformed by simple glide (slip on a single plane) in a manner that does not bend the lattice, it is quite possible to completely recuperate its hardness not including recrystallization of the specimen. Actually it is impossible to recrystallize a single crystal deformed simply by easy glide, even if it is heated to temperatures as high as the melting point. Fig. 5 shows schematically a stress-strain curve used for a zinc single crystal strained in tension at room temperature when it deforms by basal (plane) slip. Basal plane is a plane perpendicular to the principal axis in monoclinic, rombohedral, tetragonal and trigonal/hexagonal crystal systems. It is parallel to the lateral or horizontal axis in crystals. A difficulty arises for crystals having hexagonal symmetry in that some crystallographic correspondent directions will not have the same position of indices. This is circumvented in utilizing a four-axis, or Miller-Bravais, coordinate system. The three a1, a2 and a3 axes are all enclosed within a single plane (called the basal plane) and are at 120° angles to one another. The z axis is perpendicular to this basal plane.
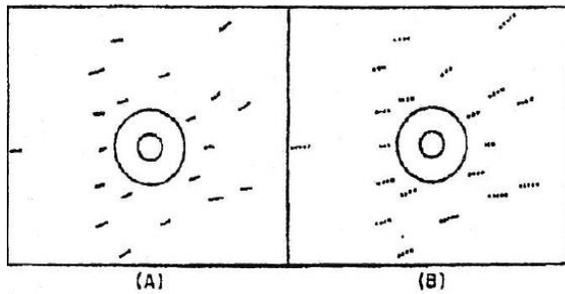
Fig. 5: Schematic Laue patterns showing how polygonization breaks up asterated X-ray reflections into a series of discrete spots. The diagram on the left corresponds to reflections from a bent single crystal, that on right corresponds to the same crystal after an anneal that has polygonized the crystal. [14].

"We have already mentioned that in cold work changes happen both in properties of metal. While plastically deformed material is studies by X-ray diffraction techniques, the X-ray reflections turn into characteristic of the cold worked state. Laue patterns of deformed single crystals show pronounced asterism corresponding to lattice curvatures. In the same way, Debye-Scherrer photographs of deformed polycrystalline metal exhibit diffraction lines that are not pointed, but broadened, in agreement with the complicated environment of the residual stresses and deformations that continue in undesirable crystalline metal following cold working. [14]. before we discuss recovery phenomenon we would again state that "plastic deformation is achieved principally by passage of dislocations through the lattice. In the early stages of deformation, the dislocations are relatively long, straight, but few. With increasing deformation more dislocations from other slip systems are generated, causing interactions among the various dislocations. These dislocations and clusters of short loops tend to tangle and to align themselves roughly to broaden boundaries. Fig. 6A and 6B, demonstrate the above-mentioned processes.

Differently annealing after plastic deformation completely restores the original state, Mittemeijer [19] has explained that only a fractional restoration of the material properties as sooner than the plastic deformation is realized. During the rearrangement/partial eradication of the dislocations in the method of recovery, the grain margins in the material do not progress; the recovery procedure occurs more or less homogenously during the matter, in flagrant contrast with recrystallizaton, characterized through the sweeping of the high-angle

grain margins throughout the deformed matrix. This procedure, thus, takes place clearly heterogeneously. It is to be understood that recovery is induced, later than the plastic deformation (by cold work), via annealing at a suitable, elevated temperature, definitely lower half of the melting temperature in Kelvin. However, if the deformation occurs at high temperatures, as in hot rolling, recovery processes already run while the material is still being deformed. One then speaks of active recovery.
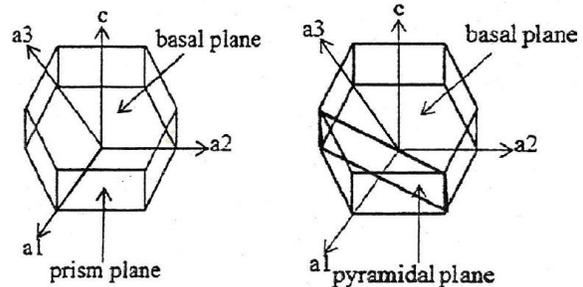

Fig. 6: Schematic representation of the hexagonal ice l_h until illustrating the a_1, a_2, a_3 and axes and the basal, prism and pyramidal planes.

Dislocation annihilation and rearrangement

Mittemeijer [19] explains that, "The heavy force for the moving of the dislocations leading to different pattern and/or to a fractional annihilation of dislocations in a reduction of the strain energy included in the strain fields of a dislocations. This decrease of the stored energy in the substance actually decreases the motivating force for the (largely) following recrystallization (recovery and recrystallization may overlie." [19]

"The total destruction of dislocations can occur through many different mechanisms. Dislocation can transfer through glide beside a single slip plane, by cross-slip and by climb. Obviously (edge) dislocations of conflicting sign on the similar slip plane can turn into annihilated via gliding to contract, Figs. 7 and 8 [18]. If these two metal dislocations of conflicting sign are of boundary type and on two different glide planes, their feasible annihilation requires a mixture of climbs with glide processes, Fig. 8. That means that dislocations annihilation occurs only at high temperatures. Two initial dislocations on two dissimilar glide planes, if screw type, are annihilated via cross-slip [19].
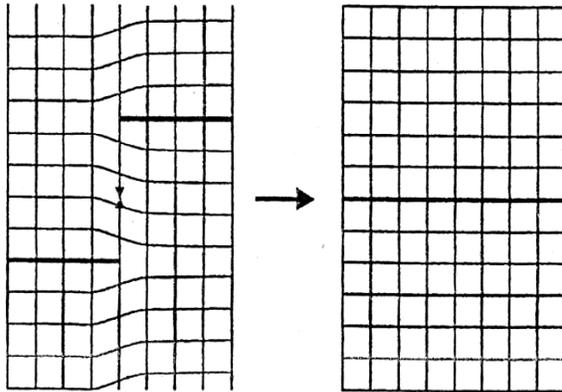
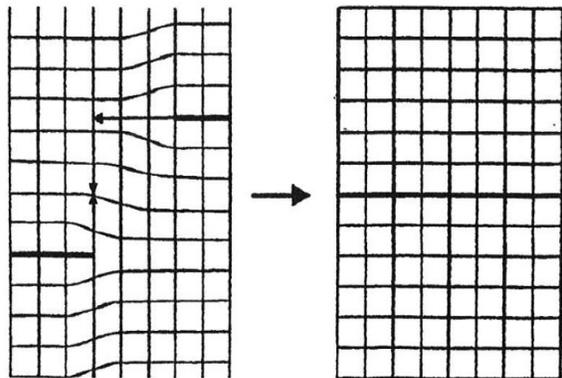Fig. 7: Annihilation of two (edge) dislocations of opposite sign by glide [18].

Fig. 8: Annihilation of two edge dislocations of opposite sign by climbs and glides [18].

"It must be appreciated that energy stored in a metal by plastic deformation makes it thermodynamically unstable. Self-diffusion provides a mechanism for the restoration of atoms in deformed region to low energy positions and to enable the metal to approach toward the truly stable condition of minimum free energy. Since diffusion involves the independent activation and migration of individual atoms, it can, and in this case does, occur without producing detectable changes in the grain size of the metal or in its visible microstructure – except, perhaps, for a progressive reduction in the intensity of the deformation bands revealed by etching. It does, however, accomplish a progressive elimination of local crystallographic distortion, and so of both macroscopic and microscopic residual stress, a reduction in hardness and strength, and slow restoration of the metal to its original properties, viz., plasticity, electrical resistivity, magnetic permeability, corrosion, resistance, etc. These changes in the structural sensitive properties of the metal, and the diffusion process by which they occur,

and jointly referred to as recovery of the cold-worked metal. It must be noted that like all diffusion process, recovery occurs at a rate that is restricted largely via the temperature of the metal and its initial state of internal energy. The temperature dependence of the recovery rate in the plastically cold-worked metal, and the same amount of recovery may occur in one minute at, say 380 °C that must require 100 years at room temperature [20].

"One of the most remarkable aspects of the recovery process is that the changes in properties it produces are not always concurrent. For example, in cold-rolled steel, the first detectable result of recovery is the relief of macroscopic and microscopic residual stresses, which is often essentially complete before these are detectable changes in its hardness, strength or plasticity properties. This makes possible the commercial heat treatment called a low-temperature stress-relieving anneal, intended to reduce the level of residual stress, improve corrosion resistance, and increase magnetic permeability etc., without appreciably reducing the strength and hardness of a cold-worked metal. So that the extent to which recovery has progressed may be accurately controlled. The light anneal is usually accomplished by prolonged heating at a relatively low temperature." [21, 22]. It is usually followed by slow cooling of the material to avoid development of a new generation of residual stresses by non-uniform thermal contraction during cooling.

Coming back to the explanation of dislocation, annihilation and rearrangement by Mittemeijer [19]. Dislocation may also glide beside a slip plane and upon colliding with a grain boundary be incorporated into the grain-boundary structure. Thus the dislocation as an isolated defect could lose its individuality by local atomic shuffles in the grain border, in association through the loss of atomic energy and in this sense annihilation of the dislocation has occurred as well. The release of strain energy can also be realized by rearrangement of the dislocations within a single grain of the matter. If the numbers of dislocations of conflicting sign are unequal, complete dislocation annihilation via any of the first two processes mentioned above is not possible. The presence of imbalanced numbers of dislocations of single slip plane: a bent grain outcome by an excess of edge dislocations of the similar type. Upon annealing, these edge dislocations can strive for measures in 'walls' and thus form low-angle tilt limits, Fig. 9. This arrangement takes place by climbs and short range glide, Fig. 10.
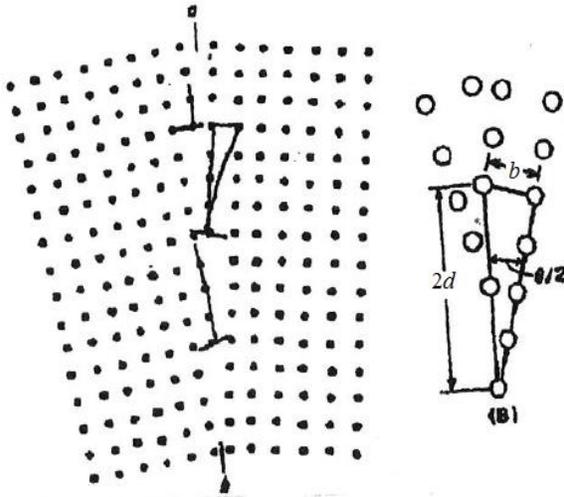
Fig. 9: (a) Dislocation model of a low angle grain boundary. (b) The geometrical relationship between ϋ, the angle of tilt, and d, the spacing between the dislocations [14].
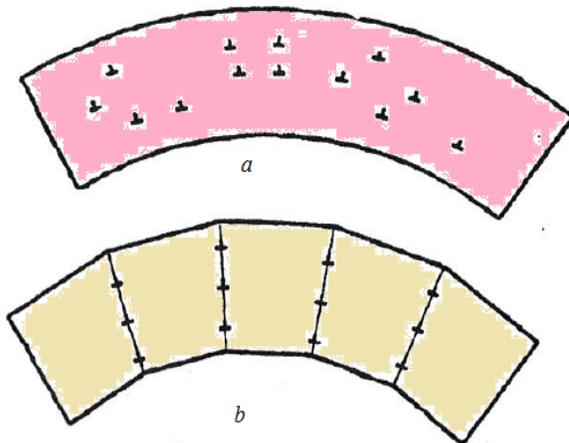

Fig. 10: (a) Bending of a single grain experiencing glide along a single slip plane: a curved grain results by an excess of edge dislocations of the same type. (b) upon annealing, these edge dislocations can strive for arrangements in walls, by climb and short-range glide and thus form low angle tilt boundaries polygonization of a bended grain by rearrangement of edge dislocations. [18].

Polygonization and subgrains

"A particularly simple form of structural recovery is observed while a crystal is bent in such a way that only one glide system operates, and is then annealed. The crystal breaks up into a fig of strain-free, small, subgrains, every preserving the local orientation of the original bent crystal, and separated by plane sub-boundaries which are normal to the glide vector of the active glide plane. This procedure or phenomenon is termed polygonization, as a smoothly curved lattice vector in the crystal turns into fraction of a polygon. For full description, see Hibbard and Dunn [23]. In polygonized metals, the sub-boundaries can be revealed by etching. They are able to be seen in the form of thick rows of pits, Fig. 11. You have to understand polygonization procedure in conditions of the dislocation distribution. When a crystal undergoes plane gliding, it is followed by possible for all dislocations, both positive and negative, to pass right through the crystal and out at the surface. In bend-gliding, a numeral of excess dislocations of one sign must stay in the crystal to contain the plastic (permanent deformation of the material without break by the temporary application of force) curvature. This strain implies the presence of a lot of extra lattice planes by the outer surface, which terminates at border dislocations inside the crystals. When the bent crystal is annealed, these dislocations reorganize themselves in walls or tilt margins normal to the Burgers vector, since in this position they largely relieve each other's elastic strain fields. Fig. 12A shows the structure of a tilt boundary in additional detail. Afterward stages of polygonization happen by progressive merging of pairs of nearby boundaries; the driving force for this comes from the progressive decrease of the boundary energy per dislocation in the boundary, as the disorientation angle Ө increases. The rate of this process, sketched in Fig. 12B, is also restricted by dislocation climb, since the dislocations in the two merged boundaries will not be uniformly spaced unless some dislocations scale". [13].

Fig. 11: Bent, annealed and etched single crystal of aluminium, showing sub-boundaries. X70 (After Cahn {1949}.) [13]
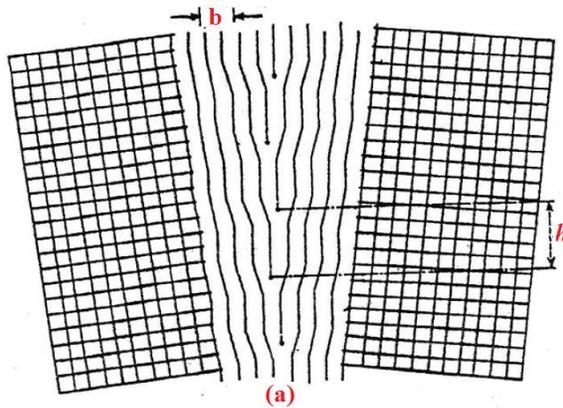
Fig. 12: (a) structure of a tilt boundary. The two subgrains are mutually titled through an angle Θ. [13]
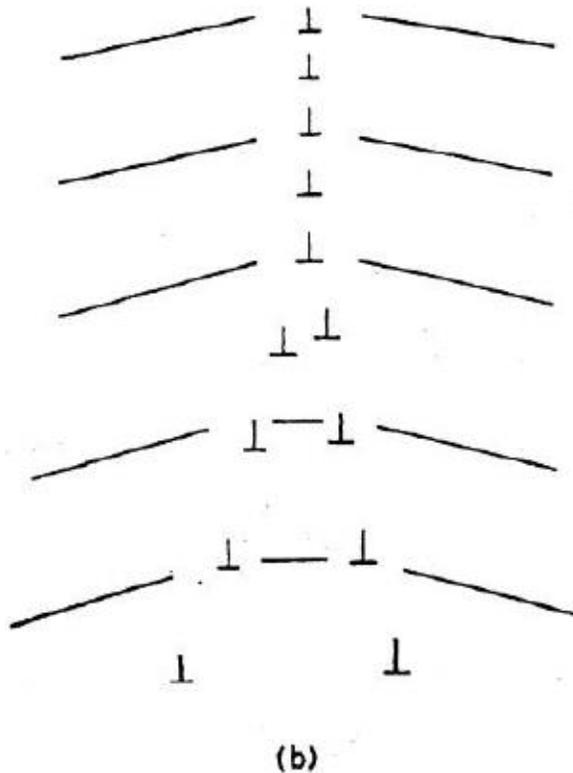

(b)

Fig. 12: (b) Tilt boundaries meeting at a "triple point". The tilt axis relating the orientations of the two subgrains is normal to the plane of the Fig.

According to Mittemeijer [19] rays that the process of polygonization is exposed in X-ray diffraction outline by the replacement of powerfully broadened reflection, observed after the plastic deformation, by a sequence of neighbouring separate spots, observed upon succeeding annealing. In simple

words, when a bent crystal is annealed, the bent crystal breaks up into a numeral of closely related small ideal crystal segments. This process is also termed polygonization [22, 23].

Dislocation movements in polygonization

According to Reed-Hill [14], "An edge dislocation is able of moving either by slip on its slip plane or by climb in a way perpendicular to its slip plane. Both are essential for polygonization, as shown in Fig. 13, where the indicated vertical progress of all dislocation represents a climb, through the horizontal progress a slip. Polygonization results decrease of strain energy. We may also state that the strain field of dislocations grouped resting on slip planes produces an effective force that makes them shift into sub-boundaries. This force exists at every temperature, except at low temperatures edge dislocation cannot climb. However, because the dislocation climbs depends on the progress of vacancies (an activated procedure through energy temperature), the speed of polygonization increases speedily by temperature. However, increasing the temperature also aids the polygonization process in another manner, for the movement of dislocations by slip also becomes easier at elevated temperatures. This fact is detectable by the fall of the critical resolved shear stress for slip with rising temperature. The elevated the temperature, the further complete is the polygonization process. When every of the dislocations include dissociated themselves from the slip planes as well as have aligned themselves in low-angle boundaries wherever two or more sub-boundaries join to form a single boundary. The angle of rotation of the sub-grains across the boundary has to grow in this process. The coalescence of sub-boundaries results from the fact that the strain energy linked with a combined boundary is below that associated by two separated boundaries. The progress of sub-boundaries that occur in coalescence is simple to understand, for the boundaries are arrays of edge dislocations which, in turn, are entirely capable of movement either through climb or glide by high annealing temperature. By time or temperature the rate of coalescence decreases so as to the polygonization process reaches a new or less stable state with widely spaced, about parallel, sub-boundaries. The process is complicated actually that slip occurs on a numeral of interconnecting slip planes and lattice curvatures are complex bend and vary with position in the crystal. The effect of such a complex deformation on the polygonization procedure is shown in Fig. 14 [21] which reveals a sub-structure resulting from deforming a only crystal in a tiny rolling mill after being annealed at 1100 °C.

The boundaries in Fig.14 are sub-boundaries and are not grain boundaries. No crystallization has taken place in this highly polygonalized single crystal. Such sub-structures are usually formed in cold worked polycrystalline metals that have been annealed at temperatures high enough to cause polygonization, but not high enough, or long enough to cause recrystallization. [14].
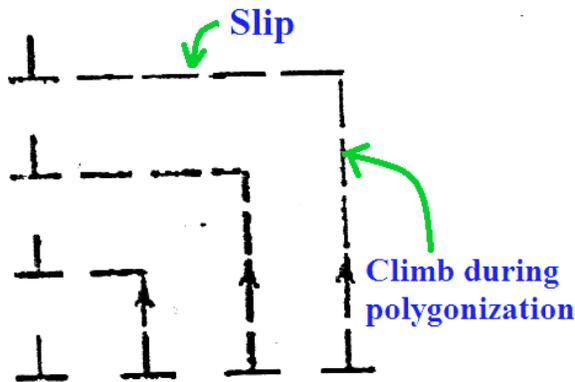

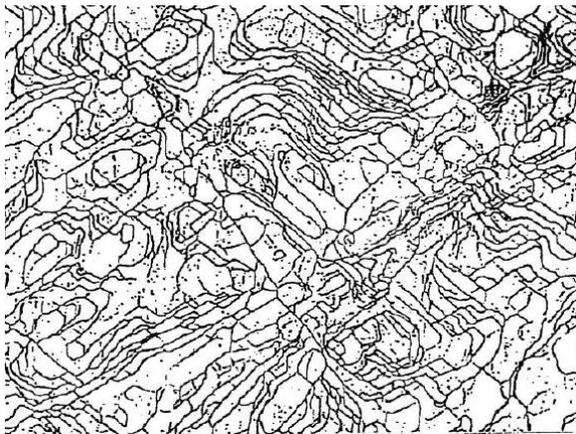Fig. 13: Both climb and slips are involved in the arrangements of edge dislocations. [14]

Fig. 14: Complex polygonized structure in a silicon-iron crystal that was deformed 8 percent by clod rolling before it was annealed 1hr at 1100°C. (Hibbard, W.R., Jr., and Dunn)

Recovery phenomenon at high and low temperatures

"Polygonization is rather a complicated process which cannot be easily expressed in terms of a simple rate equation similar to that used to describe the recovery process after easy glide. Because polygonization involves dislocation climb, relatively high temperatures are required for rapid polygonization. In distorted polycrystalline metals high-temperature recovery is measured to be basically a matter of polygonization and eradication of dislocations. [14].

"At lower temperature other processes are of greater importance. At these temperatures, recovery process is considered as mainly a matter of reducing the amount of point defects to their equilibrium rate. The most important defect is a vacancy which may contain a finite mobility even at moderately low temperatures.

Dynamic recovery

"High temperature effects the progress of dislocations into subgrains this process can actually start at high temperature so that metal undergo dynamic recovery. The tendency for dislocations to form cell/sub-structure is quite strong in pure metals and it can occur at very low temperatures". [14]. at more elevated temperatures, the cause of dynamic recovery naturally becomes stronger, as the mobility of the dislocations increases by increasing temperature. The product is that the cells/sub-grains tends to form at small strains, the cell walls become narrower/ thinner and much more sharply definite and the cell size become bigger. Dynamic recovery is, therefore, is frequently a strong factor in the deformation of metals below hot working conditions. [14]

"Dynamic recovery has a strong result on the shape of the stress-strain curve. This is because the movement of dislocations from their slip planes into walls lowers the average strain energy associated with the dislocations. The result of this is to make it less complicated to nucleate the additional dislocations that are needed to more strain the metal. Dynamic recovery thus tends to lower the valuable rate of work-hardening." [14]

"It must be appreciated that the function of dynamic recovery is not the similar in every metals. Dynamic recovery occurs most strongly in metals of high stacking-fault energies as well is not readily observed in metals of extremely low stacking-fault energy. This happens in alloys like brass. The dislocation structure resulting as of cold work in these latter resources often shows the dislocations still aligned beside their slip planes". [14]

"The correspondence between the ability of a metal to undergo dynamic recovery and the magnitude of its stacking-fault energy suggests that the primary mechanism concerned in dynamic recovery is thermally activated cross-slip. It is important to note the basic underlying difference

between dynamic recovery and static recovery, such as occurs inside annealing after cold work. In static recovery, the movement of the dislocations into the cell walls occurs as a result of the interaction stresses among the dislocations themselves. In dynamic recovery, the functional stress causing the deformation is added to the stresses acting among the dislocations. As a result dynamic recovery effects might be observed at very low temperatures, and at these temperatures the applied stresses are able to be very large." [14].

Raabe [1] and Mittemeijer [18] have also, more or less, given the same explanation and description of recovery and dynamic recovery.

Kinetics of recovery

Not much literature is available on the kinetics of recovery. Mittemeijer [19] and Raabe [1] have discussed briefly the kinetics of recovery. Hibbad and Dunn [23] have also discussed it comprehensively in a paper published in 1957. Li [24, 25] has given a comprehensive review of recovery in regard to both microstructure and properties. But Mittemeijer's explanation, though brief, is good enough to comprehend the kinetics of recovery. "According to him, the recovery procedure occurs more or less homogenously through the material, and the theory of heterogeneous transformation has no direct importance for recovery. For homogenous reaction, the possibility for the transformation to occur is the same for all locations in the virginal scheme explained by Mittemeijer [19], Humphreys and Hatherly [20] and Li [24]. Accordingly, the transformation speed decreases monotonically from t = 0 onwards. The direction for the degree of transformation has been excellently explained by Mittemeijer [19 – Pages 429-431]. According to him, the recipes intended for the determination of the effective, in general activation energy of the homogeneously happening recovery is the similar for heterogeneous transformations." According to him, recovery be capable of a complex process where a variety of sub-processes can contribute simultaneously (viz., the unraveling of the effects of nucleation growth) with impingement modes on the overall kinetics of mixed transformations. Also sub-processes may occur repeatedly, prohibiting a direct application." [19]. Balluffi *et al* [26] have thoroughly discussed the rate of recovery of a given defect which is annealing out thermally activated motion which is often expected to follow a rate equation of the form

$$-dn/dt = f(n, q1, q2 --- qn) \exp\{-E/KT\} \qquad (1)$$

where n = the total average fault concentration, t = the time, T = the temperature, K = the Boltzmann Constant, and E = the activation energy. The q's represent the properties of the solid leading to defect annihilation (e.g., the spatial distribution of sinks). The kinetics is generally measured by measuring a microscopic physical property p. If p = p(n), independent of t, T, and the q's, and if the q's are temperature independent,

$$-dp/dt = F(p, q1, q2 ----qn) \exp\{-E/KT\} \qquad (2)$$

Equation (2) can be integrated in the form

$$\int_{p_o}^{p} \frac{dp}{F(p, q1, q2 --)} = \int_{o}^{t} \exp\left\{-\frac{E}{KT}\right\} dt = \Theta \qquad (3)$$

According to eq. (3), p is then a function only of the temperature-compensated time Θ, *i.e.*,

$$p = p(\Theta) \qquad (4)$$

According to Balluffi *et al*. [26] there are at least 3 methods for determining the important quality E from these equations using isothermal or isochronal annealing data.

Method I. The times required to reach a given value of p during isothermal annealing are measured at different temperatures. Hence

$$\Theta_1 = t_1 \exp\{-E/KT_1\} = \Theta_2 = t_2 \exp\{-E/KT_2\} \qquad (5)$$

Method II. The isothermal annealing temperature is suddenly changed and the corresponding change in annealing rate is determined. In this case, the defect state is presumed to be the same at the two temperatures and accordingly,

$$\frac{\left(\frac{dp}{dt}\right)_1}{\frac{dp}{dt}_2} = \exp\left\{\frac{-E}{K}\left(\frac{1}{T1} - \frac{1}{T2}\right)\right\} \qquad (6)$$

Method III. If both the isothermal and isochronal annealing date are available for initially identical specimens, the relation

$$\ln\Delta t_1 = \text{Const} - E/KTi \qquad (7)$$

can be used, where Δt_1 is the increase of isothermal annealing time necessary to produce the same

property change increase as is observed in the i-th isothermal annealing pulse at Ti. These methods can only be used when it can be shown that equation (4) and many situations can be visualized in which this relation would not be obeyed. According to Balluffi *et al.* [26] much more information can be obtained by investigating the annealing spectrum over a range of temperature by measuring during heating at a constant rate α = T/t Under such condition, the integration of eq. (3) becomes more difficult. As an illustration of the behavior to be expected, eq. (2) may be readily integrated with it and takes the simple form.

$$-dp/dt = K_0 p^y \exp\{-E/KT\} \qquad (8)$$

where K_0 is a constant, and r is an integer.

When r = 1 (first order kinetics, the result is

$$\ln p/po \cong -K_0 (E/\alpha K)(Kt/E)^2 \exp\{-E/KT\} \qquad (9)$$

A typical annealing spectrum obtained at a constant heating rate, as in Fig. 15, from which it is visible that the annealing occurs over a restricted temperature range centered at Ta where the rate is a maximum. It is convenient to plot dp/dt Vs T, as in Fig 15, yielding an annealing peak at T = Ta. Further analysis shows that Ta should occur at about:

$$Ta \cong \frac{E}{K \ln\left(\dfrac{K_0 E}{K_0 C}\right)} \qquad (10)$$

and that the peak width at half maximum, ΔT is given by

$$\frac{\Delta T}{Ta} \cong 2.5 \frac{KTa}{E} \qquad (11)$$

It is seen that the peak is broadened and occurs at a higher temperature as α increases. Higher order kinetics can be analyzed within a similar way yielding comparable results. [26].

In such cases it is found that the peak position, Ta, depends upon the initial concentration of defects and that the peak width, ΔT, increases rather with the order. [26]
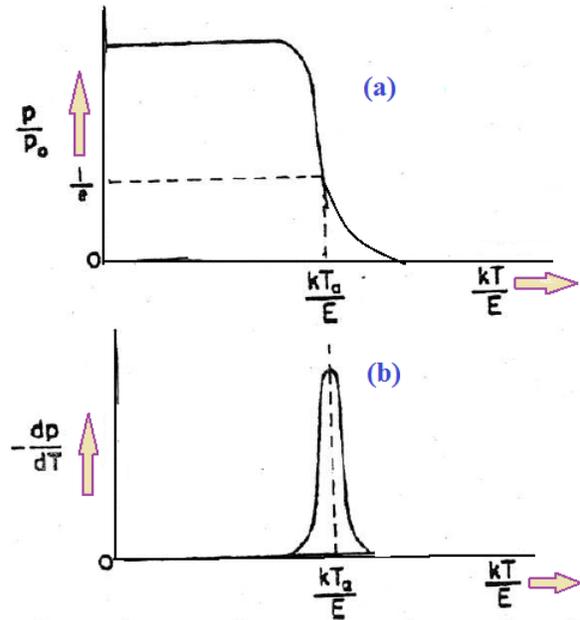

Fig. 15: (a) schematic diagram of annealing of a physical property *p* as a function of temperature during heating at a constant rate. (b) Annealing rate peak as a function of temperature during heating at a constant rate. The diagram is for first order kinetics. Here T_a is the temperature of the maximum annealing rate, and *E* is the activation energy. [26].

Influence of recovery on recrystallization

R.A. Vandermeer and Paul Gordon [27] have discussed the influence of recovery on recrystallization. According to them, the growth of a recrystallized grain into a stained matrix is forced by the difference in the stored energy of deformation that is present on the two sides of the migrating grain boundary. If the stored energy amount per unit volume of unrecrystallized material is continuously reduced during the course of recrystallization through competing recovery phenomena, after that the immediate rate of growth (boundary migration) should also decrease. That this is, in fact, the explanation for the observed decrease in growth rates with time was shown by making isothermal calorimetric measurements of the discharge of stored energy through annealing. [27]. The authors have pointed out that, from the calometric measurements it was found out that the processes of recovery and recrystallizaton overlapped in aluminium-bases alloys and that they competed by each other for the available stored energy of cold work. The retardation of recrystallization in simultaneous recovery is

probably a common phenomenon. Leslie *et al* [28] had observed that the rate of growth of recrystallized grains in iron and iron-manganese alloys decreased by time during recrystallization. They suggested that the probable reason for this decrease was a lowering of the residual strain energy per unit volume of the distorted metal because of recovery happening simultaneously with recrystallization. They presented softening curves based on hardness capacity showing that much recovery was taking place.

The control of recovery on recrystallization is, however, certainly more pronounced in some metals than in others. The manner in which the retardation of recrystallization by simultaneous recovery is influenced by impurities is not well established. This occurrence may be rationalized by the hypothesis that the addition of impurities or trace elements decreases the recrystallization rate more rapidly than the recovery rate [28].

Metarecovery and orthorecovery

Vandermeer and Gordon [27] have mentioned the work of Cherian *et al* [29] who observed 2 different recovery processes in polycrystalline, commercially pure aluminium before strained 9.2% in tension at room temperature. Once a recovery treatment at low temperatures, less than 100 ºC,and these investigators found an important lowering of the initial flow stress upon re-straining. After an additional strain of regarding 4%, the flow stress was restored to essentially that value which would have been obtained if the specimen had not been specified recovery anneal. This Cherian *et al*. [29] called "Meta recovery". At higher recovery temperatures, 150 ºC – 200 ºC, they observed not only met recovery but also an enduring decrease in the flow stress of the improved material relative to that of unrecovered material at the similar total strain for all strains. This they called "Ortho recovery". The kinetics of meta- and ortho recovery observed by Cherian *et al*. [29] were reported to be distinctly different, and the indications were that ortho recovery occurred more slowly than metarecovery. They found activitation energy for ortho recovery of 33,000col/mole.

According to Perryman [30], metallographic and X-ray observation on the underside of polycrystalline specimens of aluminium distorted at room temperature rose reliable with the idea that there were two stages of recovery. Perryman [30] observed that recovery at temperatures as low as room temperature formed much more different sub-structures than were present directly after

deformation. Perryman [30] found that recovery at high temperatures, approaching recrystallization temperatures formed structural changes much similar to those observed at the lower temperatures:

According to Vandermeer and Gordon [27], that the retarding affects of isothermal recrystallization was associated by a decrease in the linear speed of growth of the recrystallized grains as a function of annealing time.

(a) Calometric measurements of the progress of stored energy showed that recovery and recrystallization overlie to a large amount in the aluminium alloys. the motivating force for recrystallization was constantly reduced in recovery.

(b) Increasing impurity / trace element content were found to favour recovery over recrystallization.

(c) It was recommended that the two recovery stages corresponded to (a) the decrease in dislocation density inside sub-grains and the rearrangement of the dislocations located in the sub-grains boundaries into more stable configurations, and (b) gradual growth of those sub-grains that subtended small misorientation angles respecting their neighbours.

(d) That the calometric measurements on Al + 0.0068 at % Cu had revealed two different stages in the energy release in recovery annealing. That was in agreement with the earlier investigations on recovery in aluminium.

(e) Gay *et al*. [31] were led to conclude as of their X-ray microbeam studies that recovery was associated with a rearrangement of dislocations in the sub-grain boundary regions as well as with the possible progress of dislocations from the sub-grain inside to the boundaries.

Recovery of various properties

Recovery of mechanical properties

Cahn [13] has discussed in some detail the recovery of mechanical, physical, electrical, electromagnetic properties and of microstructures.

He has shown that nickel, brass, copper, every metal through low stacking-fault energy, showed no drop in hardness in stress release or recovery, and so little climb as well as rearrangement of dislocations can occur, particularly in brass and copper. As the flow stress and hardness are a purpose of the concentration and disposition of dislocations, this immutability of dislocation structure proceeding to recrystallization implies that the mechanical properties as well remain fixed. minor dislocation

movements have been postulated to explanation for significant stored energy release through recovery in some metals such as silver, but such rearrangement ought to be sufficiently slight to disappear the yield stress unchanged. In other metals, such as aluminium as well as α-iron being the most significant, dislocations can climb simply and equally these soften during a recovery anneal. Under favorable conditions, the total work-hardening may be recovered with no intervention of recrystallization. [13]. Cahn [13] has further mentioned that works of various investigators. The general rule, according to him, is: the better the deformation, the lesser is the fraction, R, of the work-hardening that be able to be recovered via a standard recovery anneal.

　　　"Crystals of hexagonal metals, Zn, Cd are exceptions, once exceptions. These can recover totally even later than huge strains by simple glide. Fig. 16 and 17 show the softening isothermals for polycrystalline iron pre-strained in tension. One can observe that the rate of softening is highest at the beginning as well as steadily diminishes. Perryman [30] has decreased the kinetics of recovery in their dependence on pre-strain, pre-straining temperature and annealing temperature. Fig 18 (a) exhibits partial recovery of polycrystalline pre-strained 30% and annealed at 225 °C [13]. And Fig 18 (b) shows aluminium pre-strained 0, 30, or 40%, in tension and recovered at room temperature etc.
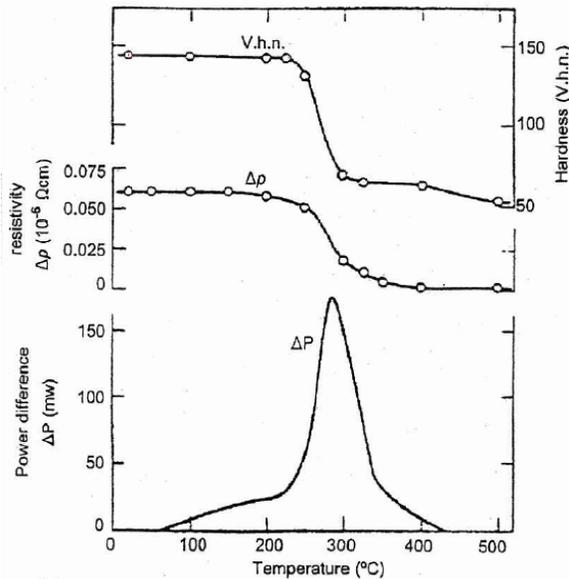

Fig. 16:　Changes in hardness, release in energy, and change in electrical resistivity upon isothermal heating of a copper rod (Clarebrought *et al.,* 1955.) [1]
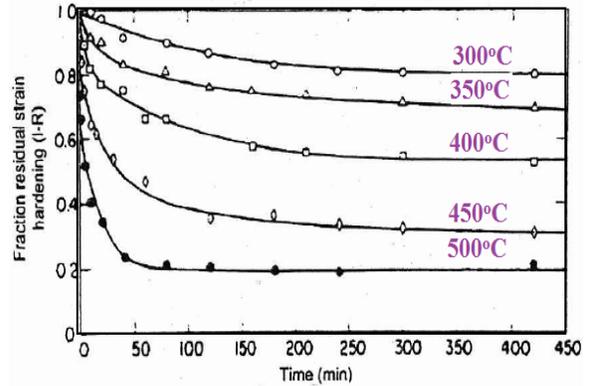

Fig. 17:　Recovery kinetics of deformed iron expressed in change in the relative flow stress (Leslie *et al.* 1963) [1].

　　　"Mechanical properties of distorted metals or alloys for instance flow stress, hardness and ductility normally recover monotonically towards the standard characteristic of the fully annealed state with the exception of brass, Cu-Al solid solutions and further similar alloys that are subject to slight anneal-hardening when annealed at temperatures excessively low to cause recrystallization, Fig. 19. A relationship of the form $\alpha = Kt^{-1/2}$ has been found between the cell diameter t and the yield stress α in aluminium, Fig 20. This relationship is similar to the Hall-Petch equation which defines the yield stress of polycrystalline steels as a function of grain diameter.
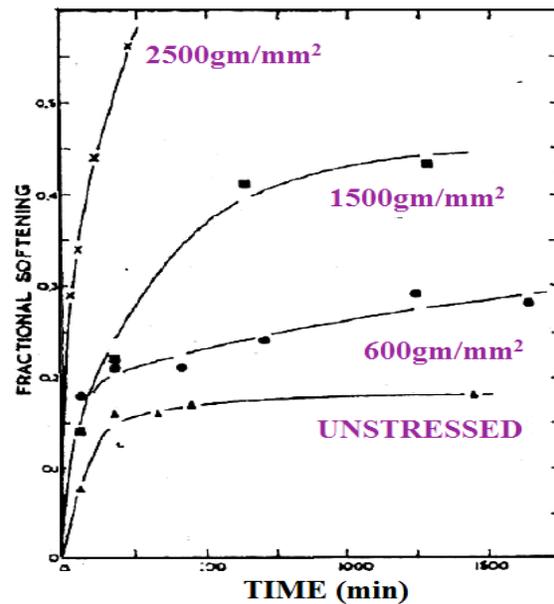

Fig. 18:　(a) Fractional recovery of polycrystalline alluminium prestrained 30% and annealed at 225°C (after Thornton and Cahn{1961}.) [13].
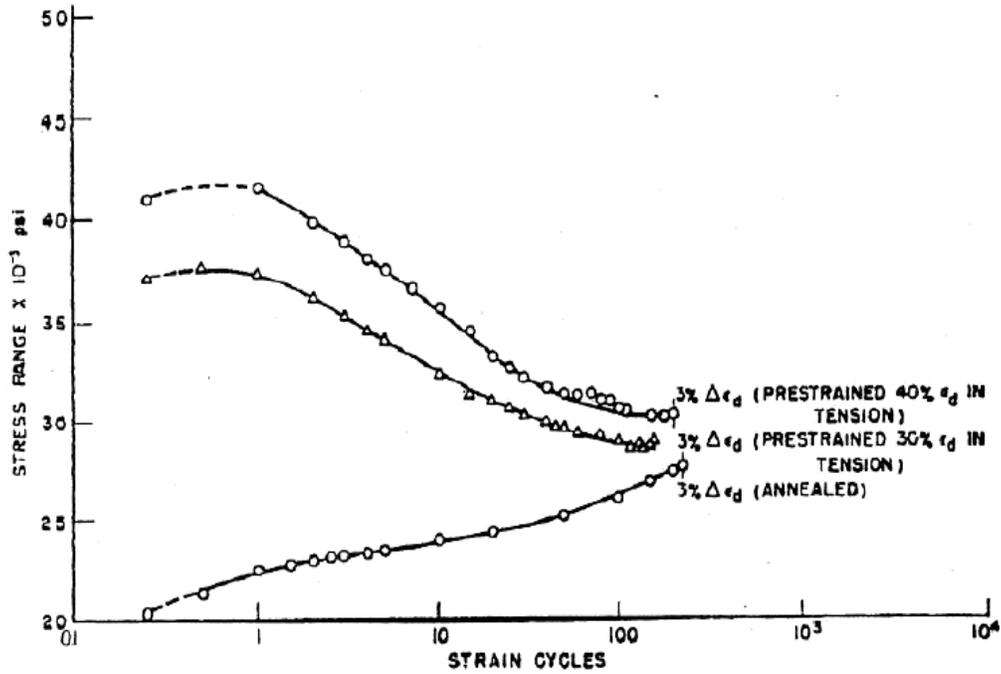
Fig. 18: (b) Aluminium prestrained 0%, 30%, or 40% in tension and recovered at room temperature under simultaneous imposed fatigue cycles over a 3% range of strain. $\Delta\varepsilon_a$. The stress range required to maintain this strange range adjusts itself automatically as the materials softens or hardens. (After coffins and Travernelli {1965}.) [13]
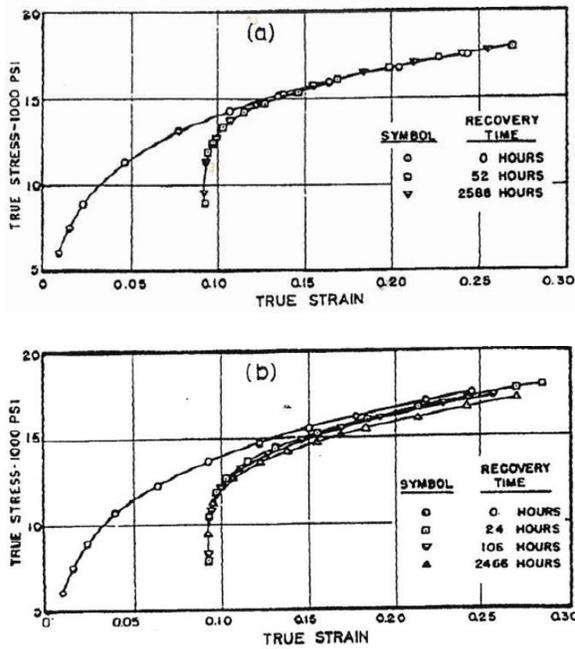

Fig. 19: Aluminium alloy deformed in tension and recovery annealed (a) at 100°C (metarecovery), and (b) at 150°C (orthorecovery). (After Cherian *et al.* {1949}). [13].
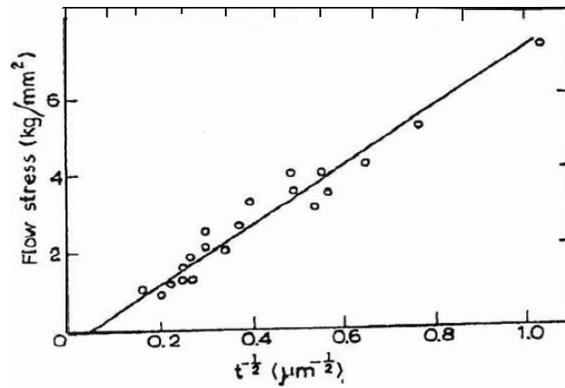
Fig. 20: Correlation between cell size and flow stress in aluminium. (After Ball {1957}.) [13].

Recovery of electrical properties

"Plastic deformation faintly increases the electrical resistivity. Lot of work has been done to the stages in which the electrical resistivity proceeds to its entirely annealed value. Fig 21 shows approximately the six stages into which the process is usually divided. For detailed discussion please consult Cahn [13], Baluffi *et al.* [26] and Clarebrough *et al.* [16].

Fig. 21: The stages in the annealing out of imperferctions produced in copper by various methods. (After Koehler *et al*. {1957}.) [13].

Electromagnetic properties

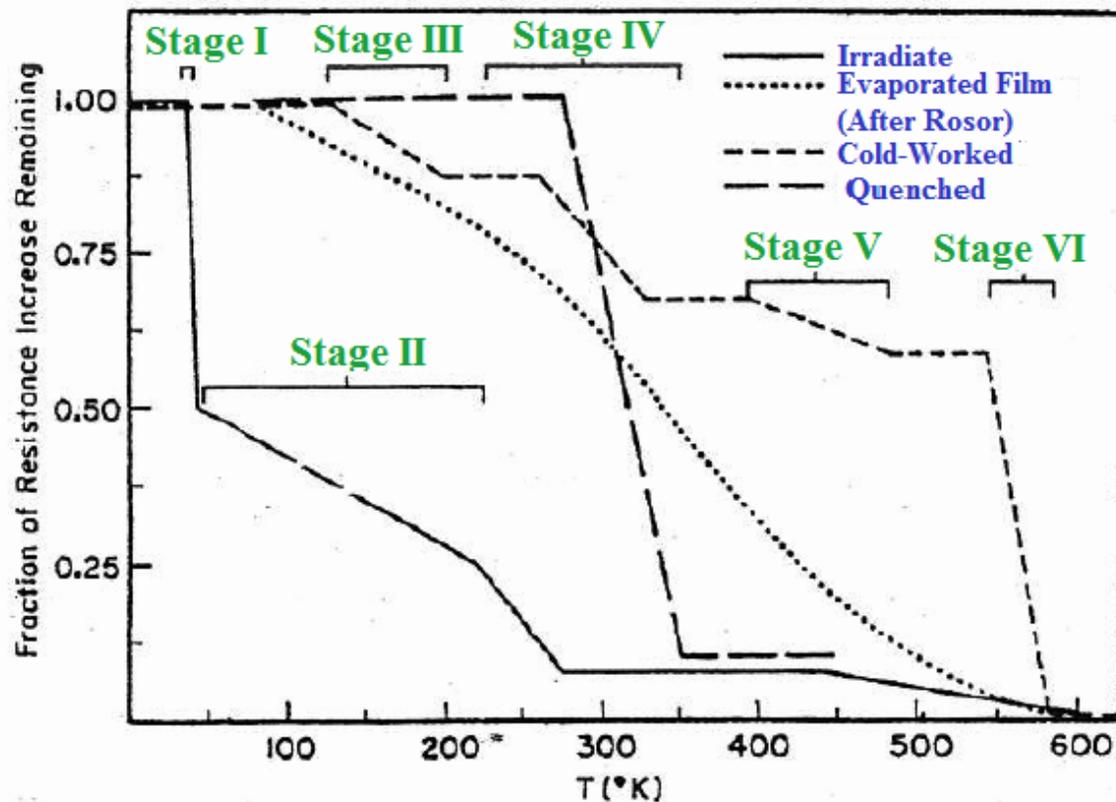According to Raabe [1], plastic deformation slightly increases resistivity, just as Cahn [13] has mentioned. For more details, please consult Gay et.al. [31], Beck and Hu [32], Haessner [33] Humphreys and Hatherly [20], Doherty et al. [34], Doherty [35].

Recovery of textures

"The mechanical properties of metals mostly depend on their microstructure as well as texture (preferred orientations) which are consequence of equally composition and thermo-mechanical treatment to which they are exposed. As recrystallizaton is characterized by the formation as well as motion of new high-angle grain boundaries, recovery conserve the existing grain structure with the original deformation texture that was formed during cold working. For detailed discussion, please consult Raabe [1]. The recovery process affects simply in-grain dislocation as well as cell substructure and it can lead to the inheritance of the new cold-working texture of the heat-treated material". [1].

Recovery of stored internal energy

"When a part of metal is plastically distorted, a definite quantity of external work has to be expended in the procedure. A small part of this work is retained as stored energy, in addition to annealing the metal is progressively released in the form of heat. Cahn [13] has discussed it in detail and has referred to other investigations. Fig. 22, form Clarebrough *et al*. [16] exhibits some interesting results. For detailed treatment see Cahn [13].

Recovery of other physical properties

"A number of other physical properties are modified / changed by plastic deformation and then recover towards their original values during heat treatment. From the kinetics and magnitude of this recovery, deductions can be drawn about the migration and interaction of defects. Cahn [13] has briefly discussed this phenomenon and has referred to a few works on this topic. For detailed information and references, please consult Cahn [13].
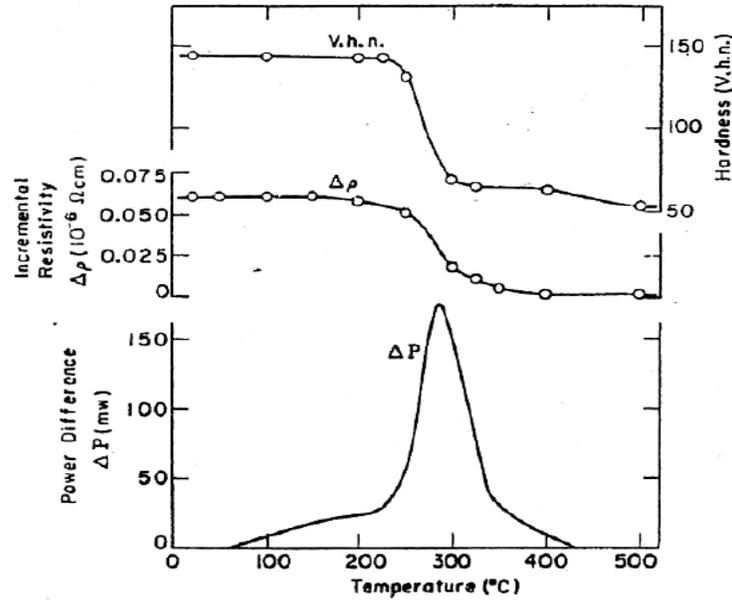
Fig. 22: Power differential, representing released energy, during the uniform heating of plastically twisted pure copper rod. Recovery of resistivity and hardness are also shown. (After Clarebrought *et al*. {1955}.) [13].
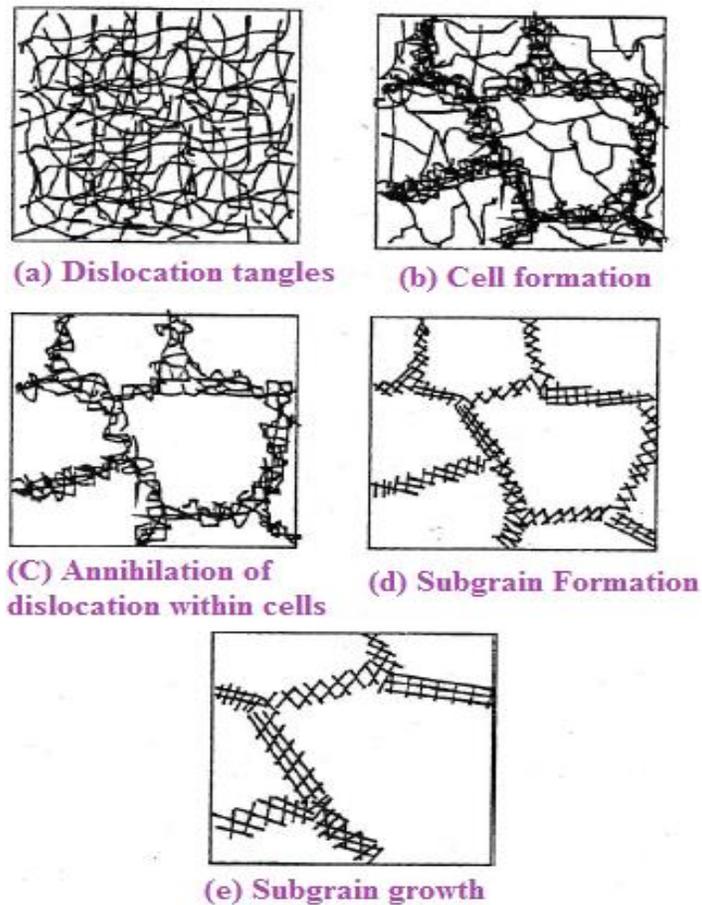

Fig. 23: Various stages in the recovery of a plastically deformed material. [19]

Humphreys and Hatherly [20] have mentioned recovery kinetics of many other complex structures.

Measurements of recovery kinetics

Humphreys and Hatherly [20] have discussed this process in detail. In simple terms, they have said that: "Recovery is a comparatively homogeneous process in conditions of both space as well as time. Once viewed on a scale which is larger than the cell or sub-grain size, most areas of a sample are changing in a comparable way. Recovery progresses regularly with time and there is no readily individual beginning or ending of the process." "Experimentally, recovery is often considered by changes within a single parameter, such as hardness, yield stress, resistivity or heat evolution. If the change on or after the annealed condition is X_R, then the kinetics of recovery are determined from experiment. It is frequently complicated to obtain a fundamental insight into recovery from such analyses, initially because the association of the parameter X_R to the microstructure is generally very complex and secondly because recovery may involve a number of concurrent or consecutive atomistic mechanisms (Fig. 23) each with its own kinetics.

Empirical kinetic relationships

Experimental results have been analyzed in conditions of numerous different empirical relations among X_R and t. The two most usually reported isothermal relationships, which we will refer to as types 1 and 2, are as follows:

Type 1 kinetics

$$\frac{dX_R}{dt} = -\frac{C_1}{t} \qquad (1)$$

which, on integration gives

$$X_R = c_2 - c_1 \ln t \qquad (2)$$

where c_1 and c_2 are constants

Apparently, this form of relatrionship cannot be valid during the early stages of recovery ($t \to 0$) when $X_R \to X_0$ or at the end of recovery ($t \to \infty$) when $X_R \to 0$.

Type 2 kinetics

$$\frac{dX_R}{dt} = -c_1 X_R^m \qquad (3)$$

Which, on integration gives?

$$X_R^{-(m-1)} - X_0^{-(m-1)} = (m-1)c_1 t \qquad (4)$$

For m > 1, and

$$\ln(X_{R)} - \ln(X_0) = c_1 t \qquad (5)$$

For m = 1

The relationship between the amount of recovery and temperature was found over a wide range of conditions to be:

$$R = c_1 \ln t - \frac{Q}{KT}$$

Recrystallization

We have just discussed recovery. "Recovery and recrystallization are two mainly different phenomena. Recovery and recrystallization were identified as important phenomena for more than 100 years and they are well-documented in the metallurgical solid state chemistry and solid state physics literature. "The process of recrystallization of plastically deformed metals and alloys is of fundamental significance in the processing of metallic alloys in support of two reasons. The first is to soften and restore the ductility of the material hardened by low temperature deformation (less than 50% of the absolute melting temperature). The second is to organize the grain structure of the last product [33]. This subject has been nicely reviewed by Humphreys and Hatherly [20] and Haessner [33]. Doherty *et al.* [34] have mentioned that within all structural transformations, there are two alternate types of transformation as initially recognized by Gibbs. In the first of these, Gibbs I, typically called nucleation and growth – the conversion is wide in the magnitude of the structural change but is, initially, spatially restricted with a sharp boundary between the old and the new structures. The second type of transformation, Gibbs II, frequently described as continuous or homogeneous, the transformation is initially small in the magnitude of the structural change, but it occurs throughout the parent structure. One annealing plastically deformed materials, both dislocation recovery, that takes place before as well as during recrystallization and too normal grain growth are Gibbs II transformation which occur uniformly during the sample while recrystallization also abnormal grain growth are Gibbs I transformations." [35].

It must be noted that all investigations, reports start with the basic facts that: "When a cold-

worked metal is held at a high enough temperature for a long enough time, and without interference by other physical changes in it, the process of recovery becomes at last complete. But, usually, however, a temperature high enough to accomplish it within a useful short period of time is also high enough so that a second and quite different process called recrystallization occurs, to eliminate distortional energy from the material before it can be dissipated by the slower process of recovery. Recrystallization is the appearance, within a mechanically-worked metal/alloy of a new generation of crystals, identified with the old crystals in composition and crystal structure, but differing from them is being entirely free of stored energy, residual stress and strain-hardening produced by earlier plastic deformation. Usually, the new crystals also differ from those they replace in size, shape and orientation. Like recovery, recrystallization is a result of the spontaneous effort of a distorted crystal structure to reduce its free energy by dissipating as heat the distortional energy stored in it during previous deformation. Unlike recovery, recrystallization involves the two important and well-known processes of nucleation of new crystals within the solid metal, and growth of these new crystals to an observable size. [22]. Smith [36], Hu and Makin [18], Doherty *et al*. [34], Reed-Hill [14], Mittemeijer [19], Cahn [13], Raabe [1], Lee and Han [37], Humphreys and Hatherly [20] etc. etc. have all used more or less similar explanations of the phenomenon of recrystallization, *i.e*. when a metal is plastically defrost at low temperature relative to its melting point, it is supposed to be cold worked. The majority of energy spent in cold working turns into heat but a definite fraction is stored in the material as strain energy, residual stress, stored energy etc. That cold working (or even hot working under certain conditions) increases greatly the number of dislocations. Plastic deformation also creates point defects, vacancies and interstitials which retain energy. And that the higher temperature, the shorter the time needed for recovery and recrystallization. And that the motivating force for recrystallization comes from the stored energy of cold work. In those cases in which polygonization is fundamentally complete earlier than the start of recrystallization, the stored energy can be assumed to be limited to the dislocations in polygon walls. The removal of the sub-boundaries is an essential part of the recrystallization process." [14]. and that the recrystallization process occurs after heavy plastic deformation and heating at elevated temperature. Recrystallization removed the dislocations and dislocation-free grains are formed within the deformed or recovered structure and release of stored energy." [20]. Both Cahn [13] and Raabe [1] have

discussed and differentiated between primary recrystallization and secondary recrystallization. Whereas both Cahn [13] and Raabe [1] have discussed recrystallization and factors that retard or increase recrystallization, extensively, Mittemeijer [19] has paid more attention on the phenomenon of grain-nucleation and grain-growth.

We will briefly discuss the (1) Primary Recrystallization and related processes, and (2) Secondary Recrystallization. We will discuss the two kinds of recrystallization one by one and their related processes.

Primary Recrystallization

As mentioned earlier, Cahn [13] and Raabe [1] have discussed Primary Recrystallization phenomenon in some details. Cahn [13] has quoted six laws, due to Burke and Turnbull [38] which play important roles in primary recrystallization:

(a) A minimum deformation is needed to begin recrystallization.
(b) The lesser the deformation, the higher is the temperature required to initiate recrystallization.
(c) Increasing the annealing time decreases the temperature necessary for recrystallization.
(d) The final grain mass depends upon the degree of deformation and to a smaller degree upon the annealing temperature, in general being smaller the greater the degree of deformation and the lower the annealing temperature.
(e) The larger the original grain size, the larger the amount of deformation necessary to give comparable recrystallization temperature and time.
(f) The quantity of deformation required to give equivalent – deformational hardening increases with rising temperature of working as well as, by implication, for a given degree of deformation a higher working temperature entails a coarser recrystallized grain size and a higher recrystallization temperature.
(g) And to these six laws, Cahn [13] added the 7th law.
(h) Fresh grains do not grow up into deformed grains equal or a little different orientation.
(i) An eight law, not strictly concerned with primary recrystallization:
(j) Constant heating after primary recrystallization is complete, causes the grain size to increase.

All these laws and theories have been adequately discussed by Cahn [13] on page 1149. According to him, the difference is grain size of a

recrystallized metal as a purpose of pre-strain as well as temperature is apt to be rather complex. This information is sometimes gathered, represented within a single view diagram, a recrystallization diagram. Fig. 24 shows such a diagram for electrolytically refine iron published in 1941 by Burgers [11]. Later more complex and refined diagrams were published by the Russian scientists. "A recrystallization diagram is at best an approximate: for example, where new grains nucleate at the boundaries of the distorted grains [27] and [39], the final grain size must obviously depend on the grain size before plastic deformation.
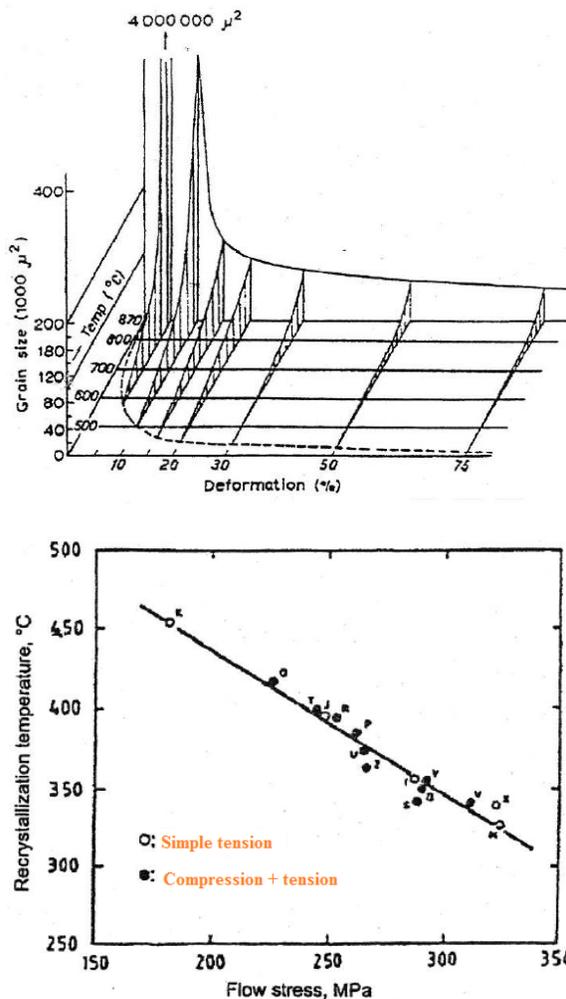

Fig. 24: Recrystallization diagram of electrically refined iron (Burgers, 1941) [13].

"It must also be remembered that resources with low stacking-fault energy can store a high fraction of the internal elastic strains due to the limited cross-slip as well as climb capabilities of the dislocations during recovery (and recrystallization)

and metals that have a higher stacking-fault energy, e.g. Al, Fe, can considerably decrease the deformation energy hence as well the remaining motivating force for primary recrystallization through the preceding recovery period." [1]

"According to Raabe [1], main static recrystallization proceeds by the formation as well as motion of new high-angle grain boundaries. In recrystallization, no new deformation is imposed. The procedure follows Johnson-Mehl-Avrami-Kolgomorov (JMAK) sigmoidal kinetics and typically leads to a refinement the microstructure (KJMA). Grain structures resulting as of primary static recrystallization characteristically consist of equiaxed crystals. The motivating force is provided in the stored deformation energy, *i.e.* primarily by the long-range elastic stresses associated with the dislocation and subgrain structure that was formed during plastic deformation. The motivating force can hence be approximated as being proportional to the stored dislocation density, the shear modulus, the magnitude of the Burgers Vector. The mechanical properties (hardness, yield strength) decay at first slowly during the incipient recovery (incubation or nucleation stage) and then very rapidly, *i.e.* sigmoidally, when the newly produced grains remove up the deformation microstructures. Final impingement of the rising crystals leads to ending of transformation. The ending of primary recrystallization is accompanied by a competitive grain growth" [1].

Kinetics of Primary Recrystallization

Cahn [13] has extensively discussed the kinetics of primary recrystallization covering some kinetic experiments, the overlapping of recovery and recrystallization, the retarding surface of recovery on retardation, annealing textures, nucleation of primary grains (we will discuss grain-nucleation and grain-growth later on), cell growth, enlargement of primary grains with the role of impurities/ trace elements on grain refinement/ growth (see Khan 39), special orientations, vacancies in grain boundaries, primary recrystallization of two-phase alloys etc etc. Mittemeijer [18] has discussed the kinetics of recrystallization as a whole. "According to him the majority of the kinetic analyses performed of recrystallization adopt an approach similar to that for heterogeneous phase transformation kinetics. "Nucleation", growth and impingement are illustrious as three commonly overlapping mechanisms. This framework can lead to the classical Johnson-Mehl-Avrami equation (also discussed above by Raabe), describing the degree of transformation (fraction

recrystallized) as a task of time at constant temperature. To emphasize the limited validity of the standard JMA equation, the basic assumptions made in its derivation are, isothermal conversion, either unpolluted site saturation at t = 0 or pure continuous nucleation, high driving force in order that Arrhenius-type temperature reliance for the nucleation and growth are certain, and randomly dispersed nuclei which produce isotropically". Mittemeijer [19] has pointed out the limitations of JMA analysis and advised not to blindly apply it.

Secondary Recrystallization

This topic has been discussed both by Raabe [1] and Cahn [13]. First let us see what Raabe [1] has written about it.

"He says that secondary recrystallization refer to a particular grain-growth phenomenon where a very small quantity of grains grow to a dimension that exceeds the standard grain size by means of one order of magnitude or else more in terms of the grain diameter. However, secondary recrystallization is a somewhat confusing term. When shed into a more appropriate expression, it is occasionally also referred to as discontinuous grain coarsening or else discontinuous grain growth. It does not explain the sweeping of the deformed microstructure such as encountered in primary static recrystallization but as an alternative it refers to the extensive growth of a few large grains in an otherwise recrystallized grain structure. It has, thus, only a phonological resemblance to primary static recrystallization become some text describes the process in terms of a nucleation stage where a number of the grains grow first widely in the incipient stage of secondary recrystallization (pseudo nucleation) and a subsequent "growth" stage where these enormous crystals sweep the other regularly sized crystals. This latter growth stage does not need a definite additional driving force but the simple fact that some grains are much better than others as topologically sufficient that the local curve provides wide further growth of these candidates. In physics terms, secondary recrystallization is a grain growth method where a small quantity of grains grows widely and can assume a size that is more than 100 times larger than that of the average size of the grains that surround them. The reason for this behavior can be restricted back-driving forces, inhomogeneous microstructures, and in homogeneities in the grain-boundary properties in terms of energy and mobility. A typical example is the discontinuous grain growth of enormous Goss-oriented grains in Fe-3wt% Si soft magnetic steels". [1].

According to Cahn [13] "When the primary recrystallization is complete, the grain structure is not so far stable. The main motivating force, linked with the retained energy of deformation, is spent, but the material still contains grain boundaries which have fixed interfacial energy. This situation is at best metastable, and ideal thermodynamic stability is only attained when the sample has been converted into a monocrystal.

"When the annealing of an firstly distorted sample is continued long beyond the stage when primary recrystallization is complete, the even tone of grain growth may be broken up by the unexpected extremely rapid growth of some grains only, to dimensions which may be of the arrange of centimeters, while the rest of the grain stay small and are finally all swallowed by the huge grains. This is termed as secondary recrystallization. Cahn [13] has explained 7 general characteristics, the main being that: (1) the secondary structure, when complete, occasionally has a prominent texture. Such a texture constantly differs from the previous primary texture, and (2) A well-defined minimum temperature must be exceeded for secondary recrystallization. The largest grains are normally produced just above this temperature; at higher annealing temperatures smaller secondary grains product, and (3) The motivating force of secondary recrystallization, and the large grains are well launched, normally arises from grain boundary energy (just as does normal grain growth); under special circumstances, the surface energy of the grains can contribute. Fig. 25 exemplifies several of these regularities. assume a primary structure contain a grain of approximately twice the average diameter and that the interfacial energy of the boundaries separating this grain from its smaller neighbors (γs) is the similar as the energy of the boundaries among these neighbors themselves (γp), then the sides of the large grain have to be convex here as shown in Fig 26, Cahn [40] if the triple grain junctions are to be in equilibrium. The configuration, which is clearly recognizable in Fig 27, Cahn [40] is obviously instable, since the curved boundaries will become straight and in the process will distress the equilibrium at the triple points, which therefore migrate outwards. The large the grain growth, the more sharply will its bounding faces turn into curved and the process is thus self-sustaining. At a convinced size a grain has reached "breakaway point", and, after that grows rapidly, getting the stage of the grain at the top of Fig 27. There is no mystery, then, about the continuation of secondary growth once it has happening; only the initial stages are hard to understand. Cahn [13] has further discussed qualitative and quantitative analyses of the growth of grains and has quoted the work of May and Turnbull [41].
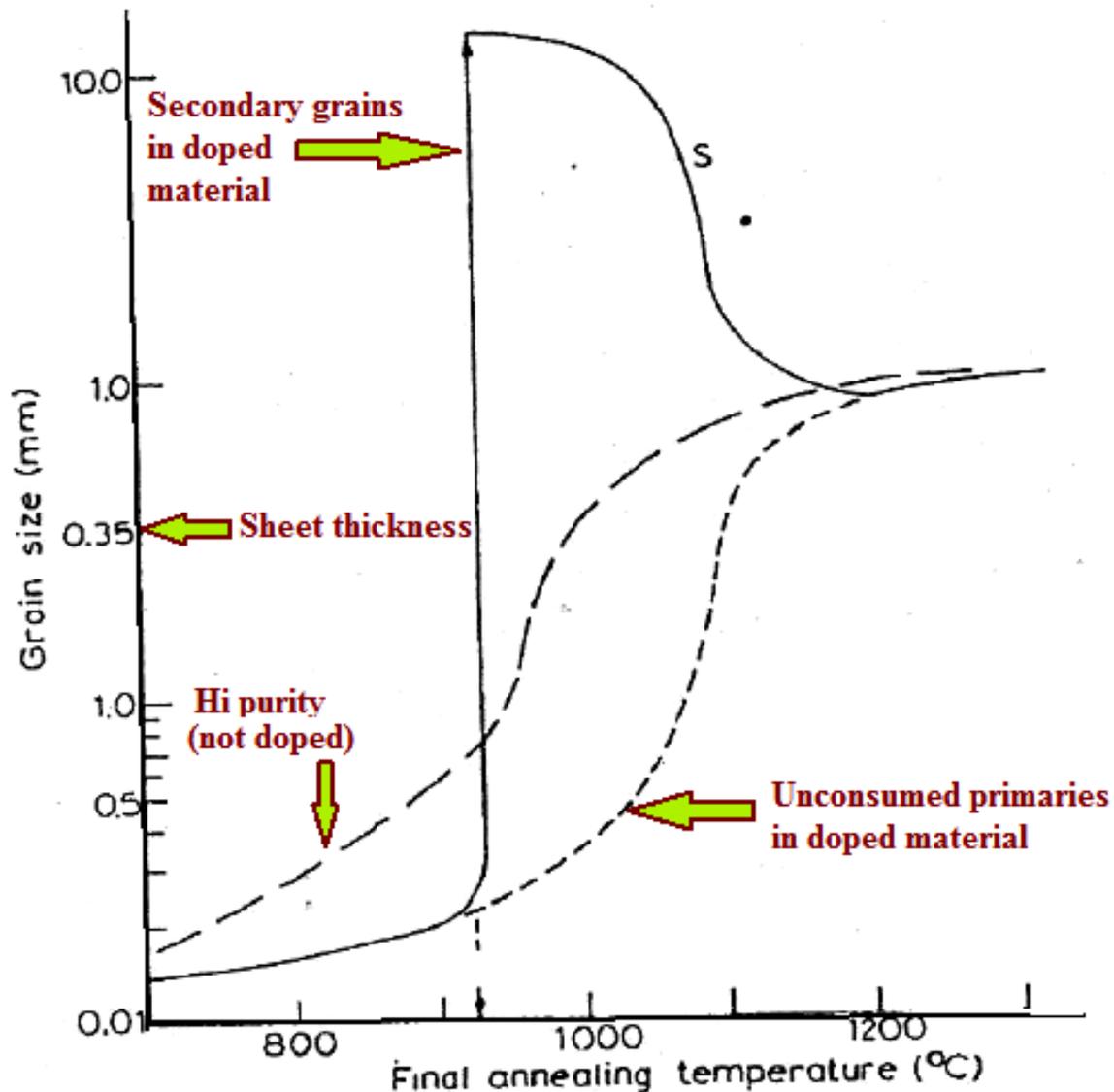
Fig. 25: Grain size (long scale) as a function of temperature for pure and MnS-doped 3% silicon-iron, cold-rolled at 50% to 0.5mm thickness, annealed 1 hour. The drop in curve S is due to an increase in the number of secondary grain at higher temperatures, until above 1100°C all grains grow equally to a size limited by the sheet thickness. (After May and Turnbull {1958}.) [13].

Cahn [13] has mentioned 3 processes that might permit a primary grain to grow the breakaway (see Cahn, p.1189). On the basis of these three processes, Cahn concluded that it was clear that the nature of the disperse phase is vital in calculating secondary crystallization; if there is a extremely sharp primary texture with a little rogue grains, a appropriately dispersed second phase is very important for secondary recrystallization. If the dispersion drag is insufficient, a large amount normal grain growth occurs; if it is too huge, the secondary grains cannot grow at all. Thus in Fig. 26 the primary grain size of the pure alloy increases more quickly than that of the alloy (Fe + 3% Si) doped with Mgs dispersion; the latter under goes secondary recrystallization, the former does not." [13]. Calvet and Renon [42]. Cahn [13] has also discussed in point surface-controlled secondary recrystallization and has quoted the works of Detert [43] and Walter and Dunn [44] and may other investigators. It is beyond the scope of this evaluation to go into details about it.
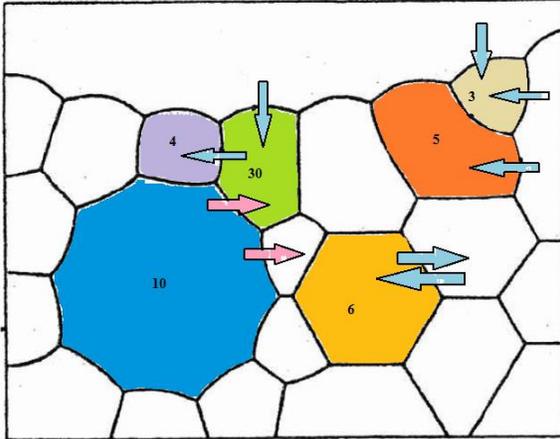
Fig. 26: Schematic diagram showing effect of the number of sides of grain on the curvature of grain boundaries. Six sided grains are in two dimensional equilibrium; smaller grain disappear, larger grains grow. (After Coble and Burke {1963}.) [13]
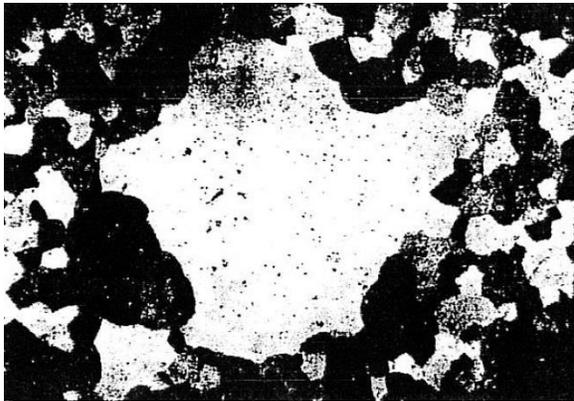

Fig. 27: Incipient recrystallization in zinc sheet. X 100 [13]

"Cahn [13] has further discussed the relationship between secondary recrystallization and sintering and has pointed out the relationship between pores and particles, the disappearance of pores due to diffusion of vacancies to the nearby grain boundary [44]. If, however, secondary recrystallization takes place, after that all the pores in a secondary grain have been crossed only once by an affecting boundary, and moreover one that is moving rapidly. For effective sintering, then, secondary recrystallization has to be prevented".

Kinetics of Secondary Recrystallization

We have already discussed primary and secondary crystallizations and also the kinetics of

primary recrystallization. We have earlier explained [13] that when the annealing of an initially deformed sample is long beyond the range when primary recrystallization is complete, the even tone of grain growth may be interrupted by the unexpected very quick growth of some grains only to abnormal growth. These large grains eventually swallow the small grains. This is known as secondary recrystallization. The drawing force of secondary recrystallization normally arises from grain boundary energy below special circumstances; the surface energy of the grains can contribute [13]. For secondary recrystallization to occur, it is necessary that normal growth is initiated or hindered. The kinetics of the secondary recrystallization usually follows a proportional growth with the time, Fig. 28. It is in accordance with the Avranic equation [46, 47]

$$R = G (t - t_0)$$

where t_0 is the nucleation time, G is the growth rate dR/dt. Initially, the nuclei of the defect-free crystals form, called the nucleation period, followed by their growth at a constant rate until all the smaller deformed crystals have been consumed.
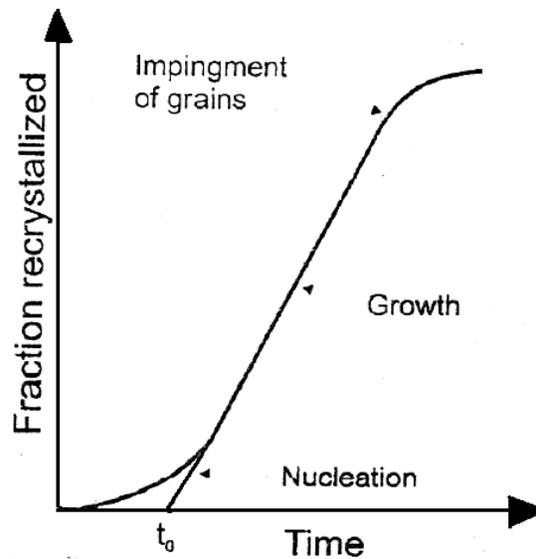

Fig. 28: Variation of recrystallized volume fraction with time [17].

It is usually accepted that any helpful model have to not only account for the initial state of the material but also the continually altering relationship among the growing grains, the deformed matrix and any second phase or other microstructural factors. The situation is further complicated in dynamic

systems where deformation and recrystallization happen simultaneously. Accordingly, it has generally proven not possible to produce a precise predictive model for industrial processes without resorting to wide experiential testing. But in it are many difficulties as it may require industrial equipment that has not yet actually been manufactured. It is to be appreciated that most of the investigations have concluded that the kinetics of primary recrystallization and secondary recrystallization are to some extent similar. "The smaller the grain size of primary recrystallized specimen just before the onset of secondary recrystallization, the longer for the evolution of secondary recrystallization. The time for the evolution of secondary recrystallization gets good agreement with the calculated time on the assumption that the coarse grain, which could decrease the driving force for secondary recrystallized grains to evolve, in the primary recrystallized specimen could control the kinetics of secondary recrystallization. The kinetics of secondary recrystallization is considered to be controlled mainly by the grain size distribution of the primary recrystallized specimen. [46]. Fig. 29 [18, 28] illustrates the kinetics of secondary recrystallization for cube texture formation in Fe + 3% Si alloy during isothermal annealing at 1050 °C. The characteristics of this curve for secondary recrystallization are quite similar to those for primary recrystallization.
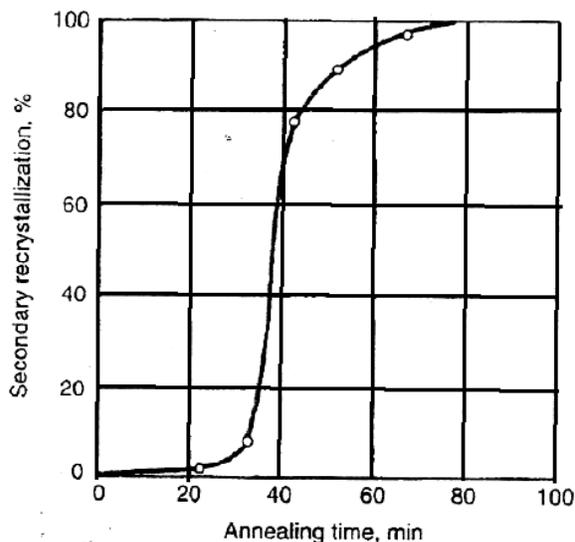

Fig. 29: Kinetics of secondary recrystallization for cube texture formation in Fe-3Si during isothermal annealing at 1050°C (1920°F). The characteristics of this curve for secondary recrystallization are quite similar to those for primary recrystallization. [17].

Annealing and Recrystallization Textures

"When a price of metal is deformed by several directional process, for instance wire drawing or rolling, the basic grains obtain a preferred orientation or texture; the grains estimated, usually with a good deal of scatter, to an ideal orientation. In an intense case the whole sample might turn into a pseudo-single crystal. Occasionally several ideal orientations co-exist, in order that the total scatter is greater. The actual orientation distribution of grains is termed a deformation texture. When such a material is recrystallized it again acquired a texture, which may be identical to the preceding deformation texture but further frequently is quite different – this is an annealing texture. X-ray diffraction data obtained as of the sample is used to plot pole figs to illustrate the annealing/ recrystallization texture." [13]. for detailed description phase see Cahn [13], and Raabe [1].

Deformation Textures

"According to Humphreys and Hatherly [20], a lot of mechanical and physical properties of crystals are anisotropic, and consequently the properties of a polycrystalline combined will depend on whether the individual grains or subgrains, which include the sample, are erratically oriented or be likely to have some in crystallographic orientations of the crystallites inside a polycrystalline aggregate is known as the texture of the material. The orientation that takes place during deformation is not random. They are a consequence of the actuality that deformation occurs on the majority favorably oriented slip or twinning systems and it follows that the distorted metal acquires a preferred orientation or texture. If the metal is then recrystallized, nucleation occurs preferentially in regions of exacting orientation. The capability of the nucleus to grow may as well be inferred by the orientations of next regions in the microstructure. Together these features, nucleation and growth, make sure that a texture also develops in the recrystallization material. Such textures are called recrystallization textures to differentiate them from the interrelated but rather different deformation textures from which they expand". Reference [20] contains much information on illustration of textures, deformation textures in FCC metals, pure metal textures, alloy textures, deformation textures in BCC metals, and deformation textures in HCP metals rolling textures, and factors which influence texture development".

Formation of Annealing Twins

According to Cahn [13], a normally observed characteristic of recrystallized structures in certain fcc metals (Cu, Pb alloys; austenitic steels etc) in the existence of copious parallel-sided annealing double lamellae, as shown sketched in Fig. 30. These lamellae are always boundary by {111} planes or coherent boundaries --- instantaneously subsequent to primary recrystallization there are a small number of lamellae per grain, but their quantity increases through the progress of grain growth. (For explanation see ref. 13). Fig. 31 [19] shows the original annealing twins in annealed 70:30 brass. Some coherent (CT) and incoherent (IT) twin boundaries are noticeable. Humphreys and Hatherly [19] have also explained the mechanism of twin formation, viz. (1) by growth faulting (2) by boundary dissociation (3) during recovery (4) during recrystallization and (5) during grain growth.
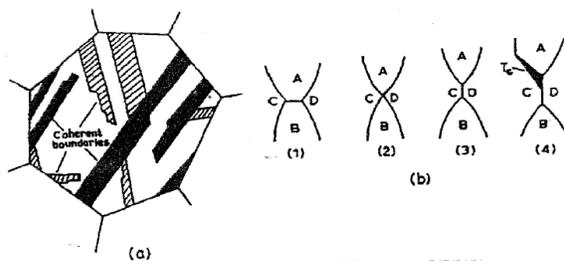

Fig. 30: (a) Annealing twin lamellae in a face- centered cubic metal grain. (b) Stages in the development of at twin during grain growth. [13]

Driving Forces of Recrystallization and Grain Growth Phenomena

Raabe [1] says that from a thermodynamic point of view, the majority of recrystallization and grain-growth phenomena can be properly characterized as non-equilibrium transformation. In each type of process, a motivating force acts on a grain-boundary section. The free-enthalpy alteration of the system is then linked with the discharge of stored energy per volume that former swept by the morning grain boundary.

"In standard, two types of motivating forces can be distinguished. This first one is a stored volume energy that acts uniformly in every portion of the affected grain that is being swept by a moving grain boundary. During such cases, the stored energy can frequently be simplified and written in scalar form. A most important (though simplified) example is the distinction in the stored dislocation density across a moving border. It must be emphasized though that for some experiments, *i.e.* in the case of differences in elastic or magnetic energy, the tensorial character of these mechanisms must be in use into consideration. The second class of motivating forces is of a configurational, *i.e.* a topological character. This means that the driving force and the discharge in the stored system energy depend on the accurate local arrangement of the defects that are removed or else rearranged through a moving grain boundary. A characteristic example is constant grain growth, *i.e.* competitive grain coarsening, where the motivating force depends on the local curvature and energy of the grain boundary but not on the size of the entire grain. Similarly, in continuous grain growth, the local grain boundary portion that moves in the direction of its center of curvature in order to decrease its total length does not "know" the size of the grain that it encompasses. This means that in this case only the local capillary driving force matters. It is conversely not a constant force that acts uniformly in each volume portion of the similar crystal but it differs all over in the crystal and polycrystal depending on the local grain boundary configuration in terms of the curvature and grain-boundary intersection lines and points." For detailed quantitative analysis, please consult Raabe [1].

Sub-grain Rotation during Recrystallization

James C. M. Li [51] has mentioned the chance of sub-grain rotation during recrystallization. He has analyzed thermodynamically and kinetically the option of the rotation of a sub-grain with respect to its neighbours as a usual process in recrystallization. He found that it was dynamically possible if the of rotation favored the removal of low-angle boundaries more than that of high-angle boundaries, the removal of twist and asymmetric boundaries above the tilt and symmetric boundaries, and the elimination of the large-area boundaries over that of the small-area boundaries. As the rotation direction had two degrees of freedom, there existed a rotation among the two degrees of freedom such that directions whose two degrees of freedom satisfied such a rotation provided no driving force for rotation. Further directions would supply free energy to rotate within once sense or else opposite. Kinetically one of the next two processes inside one boundary might be rate controlling: the cooperative movement of the dislocations in the boundary, as well as the cooperative diffusion of vacancies in the lattice. The rotation of a sub-grain favored the removal of one of the boundaries which contributed the major fraction of motivating force as well the largest portion of resistance. This caused the sub-grain to combine with the new sub-grain separated by that boundary. The probable time required for individual coalescence compared favorably with the experimental rate of sub-grain growth in aluminium. Figs 32-34 illustrate this phenomenon. Many investigators have confirmed this observation.
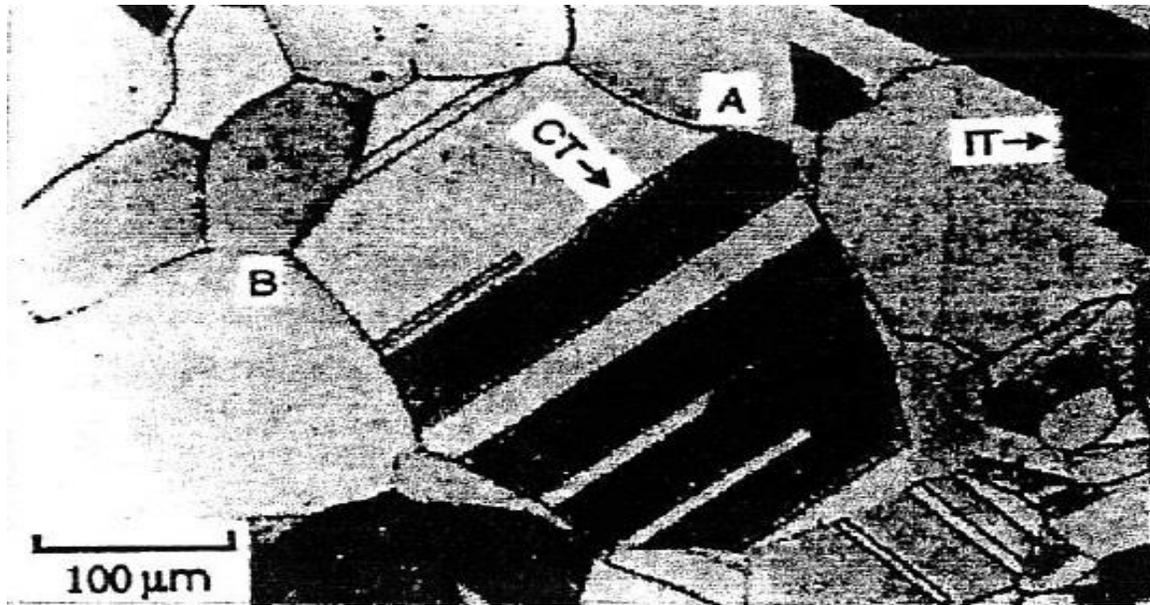
Fig. 31: Annealing twins in annealed 70:30 brass. Some coherent (CT) and in coherent (IT) twin boundaries are marked. [19]

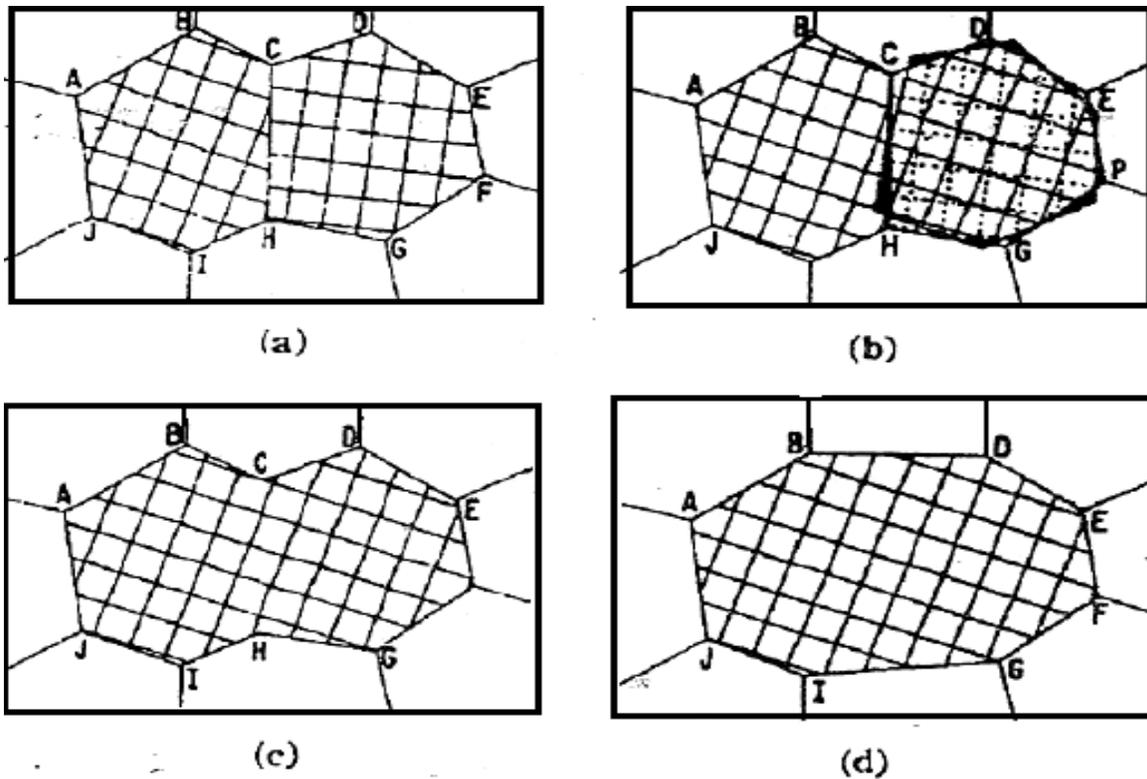Fig. 32: Coalescence of two subgrains by rotation of one of them.

a.   The original subgrain structure before coalescence.
b.   One subgrain is under rotation.
c.   The subgrain structure just after coalescence.
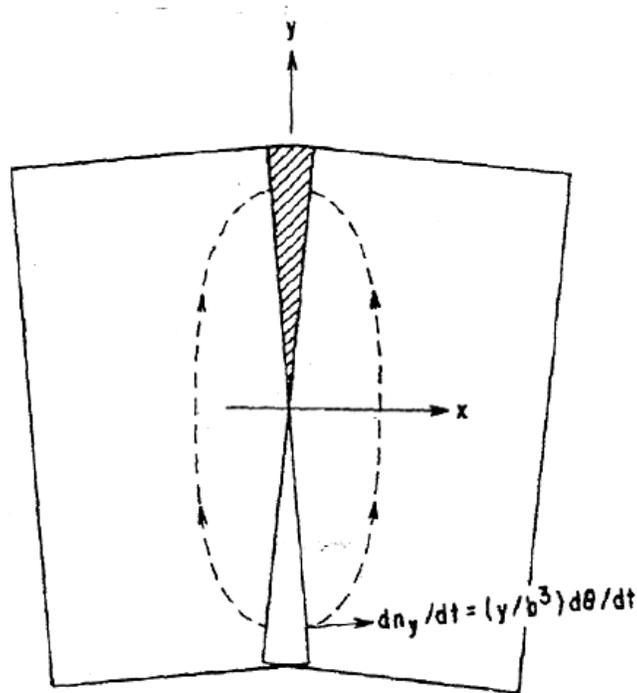d.   The final subgrain structure after some subboundary migration. [49]

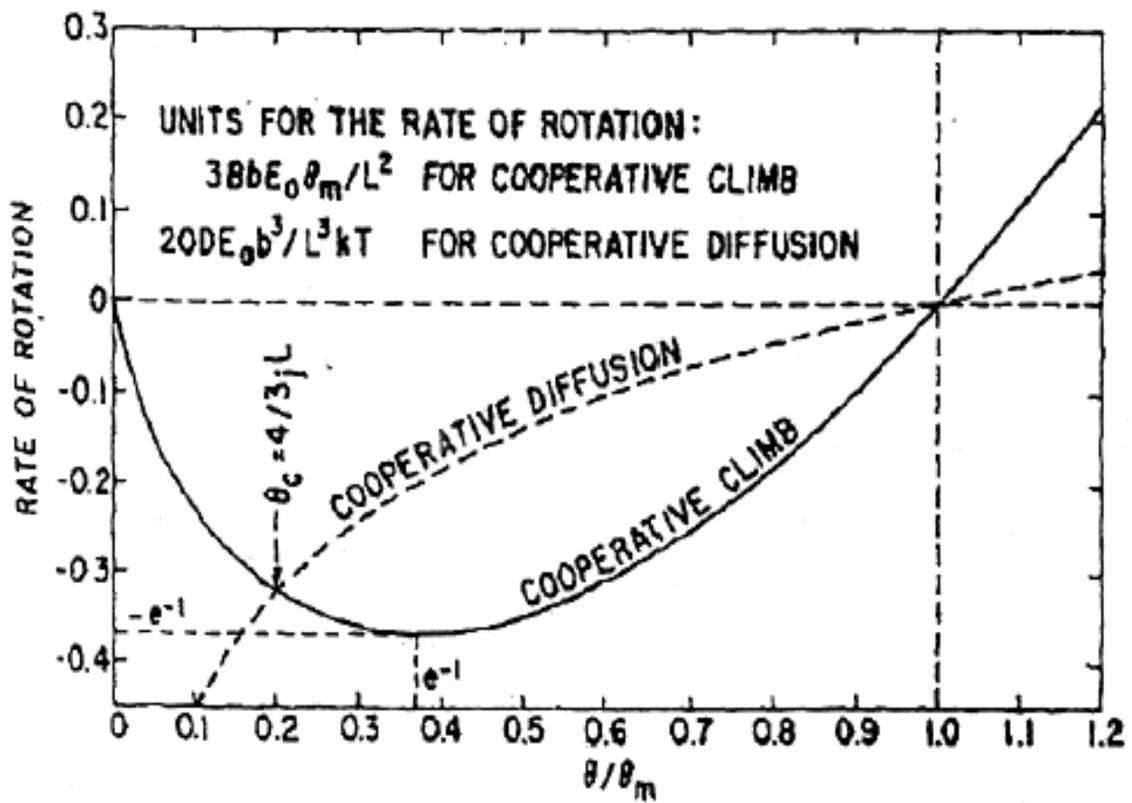Fig. 33: Cooperative diffusion of vacancies in sub-grain rotation. [49]

Fig. 34: The rate of rotation of the angle tilt boundary. [49].

Dynamic Recrystallization

"The term dynamic recrystallization refers to all recrystallization phenomena that happen through the plastic deformation. See for references [1]. "Dynamic recrystallization is often experimental during hot deformation of prepared alloy. It appears to be quite comparable to the process which occurs in metals and is to be predictable in materials where recovery is slow and there is maintenance of huge amounts of stored energy. For grain boundary passage and dynamic recrystallization evidence, please refer to [19]. The physical mechanisms responsible for dynamic are comparable in a lot of aspects to those controlling static (primary) recrystallization. Single crystal studies have exposed that twining plays a significant role in dynamic recrystallization, yet in polycrystalline materials twining can play an important role in the propagation of dynamic recrystallization, mainly in low stacking fault energy alloys. Doherty *et al.* [34] have discussed texture development during dynamic recrystallization, post-dynamic recrystallization, effect of solute elements on precipitation and on dynamic recrystallization, geometric dynamic recrystallization. Dynamic recrystallization is thermally activated as well as its progress obeys the Avrami equation.

An extensive, detailed review article on dynamic recrystallization titled, "A Review on Dynamic Phenomena in Metallic Materials by Huang and Loge" [50] was accepted for publication in Materials Design on 3 September, 2016 (ref. PH – S0264 – 1275 (16) 31175-3, DOI:10.1016/j.madjes, 2016.09.012, Ref JMADE 2262 as a PDF file). Its introduction and summary are reproduced here for students, researchers, teachers for further consultation.

Introduction

Mainly the metallic parts during their processing cycle, subjected to hot deformation, through which dynamic recrystallization (DRX) frequently takes place. The final microstructure and mechanical properties of the alloys are mostly determined by the recrystallization and linked annealing phenomena, and the research on recrystallization can date back to 150 year ago [1]. The rapid progress of the DRX theory from 1960s was summarized by McQueen in 2004 [18]. Several important factors can have a major effect on DRX; these contain the stacking fault energy (SFE), the thermo-mechanical processing (TMP) conditions, the initial grain size, chemistry and microchemistry of

the material in terms of solute level and second phase particles etc., which is also the reason why a vast quantity of related works can be found in literature.

During hot deformation, discontinuous dynamic recrystallization (DDRX) is often observed for low SFE materials, where nucleation of new strain-fee grains occurs and these grains grow at the rate of regions occupied of dislocations. Cell or subgrain structures by low angle grain boundaries (LAGBs) are formed through deformation for materials with high SFE due to the proficient dynamic recovery, they gradually progress into high angle grain boundaries (HAGBs) at larger deformations, a process which is known as continuous dynamic recrystallization (CDRX) [1]. Besides DDRX and CDRX, another comparatively new theory of geometric dynamic recrystallization (GDRX) was also experimental on deforming aluminium to big strains at elevated temperatures. In this case, the deformed grains turn into extended with local serrations but stay distinguishable through deformation to large strains unless their thickness is below 1-2 subgrain size, at which time the developed serrations become pinched off and equiaxed grains with HAGBs are formed. Substantial grain refinement is thus obtained during the grain elongation as well as thinning. Numerical models are developed for these three types of DRX, most of them are focused on DDRX [3], with sparse models on CDRX [4] and mostly unexploited GDRX models [5].

It should be prominent that there is no strict separating line among these three types of behaviors. For example, CDRX was observed through rolling of fine-grained 304, austenitic stainless steel which is of low SFE [6], DDRX was reported for high SFE high purity Al [7], CDRX and DDRX can even co-exist during hot working of Mg-3Al-1Zn [8] or duplex stainless steel [9]. When designing new alloys, the adding together of alloying elements to the base material may adjust it's SFE [10] and thus changes the recrystallization mode. Even designed for the same material, altering the TMP conditions [11] or initial grain size [12] can also direct to the transition from DDRX to CDRX. Meanwhile, CDRX and GDRX can as well operate along with, e.g., in Zircaloy-4 [14]. Due to the increasing requirements on the formability of the metallic parts, products which were before formed at room temperature, where DRX is in general irrelevant, are currently often processed at warm or else hot temperatures [14, 15]. It is not simple to identify which type of the three DRX processes is in use at certain TMP conditions since they share some similarities and can

take place concurrently and/or transitionally. There is really a hot debate between researchers within the recrystallization field on whether CDRX or GDRX should be responsible for the grain modification of aluminium [16], the center issue is whether the HAGBs experiential after big deformation are altered from LAGBs, microshear/ deformation bands, or original HAGBs.

The recrystallization phenomenon in common was reviewed in 1997 by the top experts in this area [17], since then the EBSD technique, which can offer invaluable information on the development of the crystallographic orientations and facilities the understanding of different DRX processes, has been widely spread. CDRX and DDRX were recently reviewed in an outstanding and extensive review paper [18], however, GDRX was not covered and the related DRX numerical models were only in brief described. The evaluation of the three types of DRX processes was only infrequently mentioned in a few articles [19-21]. This short review paper covers seminal basic works as well as very recent contributions to all the three DRX processes. It differs from the above mentioned review papers since it updates key aspects on DRX from affecting factors, characterization methods to mechanisms and mathematical models.

The main objective of this paper is to offer a short review of the different types of DRX observed through *hot deformation* for different types of metallic alloys, *i.e*., DDRX, CDRX and GDRX. The review offered is proposed to equip the beginners in metallurgy with a brief insight into the DRX phenomenon. For more details on this topic, the interested readers are referred to the classic textbook on recrystallization [1] and the two outstanding but longer review papers [17, 18]. In Section 2, the terminologies used in this field are initially summarized, together with the key factors influencing the DRX processes, as well as the corresponding characterization methods. The transitions among the various types of DRX processes are only briefly discussed due to deficiency of literature data. From Section 3 to 5, more details on the three types of DRX are given, including their mechanisms and related numerical models. Finally, in Section 6, further studies inside the DRX field are recommended.

Dynamic recrystallization

Terminology

In recrystallization processes there are different phenomenological categories, a lot of them are interrelated and the borderlines among them are often nuclear. It is favorable to recall all these processes, even though a number of them will not be covered in this evaluation work. In addition, there are also some terminologies in recrystallization field which are worth mentioning before going into the details of DRX.

Defects like dislocations with interfaces increase in deformation which makes the material thermodynamically unstable. When deforming metal at eminent temperatures, thermally activated processes ten to remove these defects in order to decrease the free energy of the system. The microstructure and also the properties can be moderately restored to their original values before deformation by recovery during annihilation as well as rearrangement of dislocations. Recovery usually brings relatively homogeneous microstructural changes and it generally does not involve the migration of HAGBs among the deformed grains [17]. Analogous recovery processes may take place in annealing or else throughout deformation, which are known as static recovery (SRV) and dynamic recovery (DRV), respectively.

The "development of a new grain structure in a distorted material by the formation and migration of HAGBs determined by the stored energy" introduced by plastic deformation is termed as recrystallization [17]. Recrystallization may take place heterogeneously with apparent nucleation and growth stages, and in this case it is described as a discontinuous process. Differently, it can as well take place homogeneously such that the microstructures change progressively with no clear nucleation and growth stage, exhibiting a continuous character. Static recrystallization (SRX) refers to the recrystallization process through annealing while that occurred in deformation at elevated temperatures is called dynamic recrystallization (DRX). During the initial stages of annealing, fine dislocation-free crystallites are produced by SRV, these nuclei grow and use the strain hardened matrix, determined by the stored energy associated with the dislocations and/or sub-boundaries. This process is the most studied and widely used recrystallization process, which can be classified as discontinuous static recrystallization (DSRX). SRX can also take place consistently without apparent nucleation and growth stage, a good illustration is when Al alloys with particle-stabilized subgrain structure are annealed at elevated temperatures. Fine particles precipitate out along grain/subgrain boundaries, gradual subgrain growth

takes place during subsequent annealing due to the coarsening of dispersoids (reduced pinning force) and LAGBs misorientations increase progressively until they are transformed into HAGBs. In this way, a new microstructure develops homogeneously throughout the distorted matrix, and is thus labeled as continuous static recrystallization (CSRX) or extended recovery.

Back to DRX, the definitions of these three types of DRX processes is given in the introduction section, *i.e.*, discontinuous dynamic recrystallization (DDRX), continuous dynamic recrystallization (CDRX) and geometric dynamic recrystallization (GDRX). It will not be reiterated here, but more information will be given for each DRX process in later sections. It is important that, if the straining is stopped after the critical strain for DDRX although the annealing temperature does not drop adequately fast, the recrystallization nuclei formed in the material will produce with no incubation time into the matrix with higher stored energy. This phenomenon is known as metadynamic recrystallization (MDRX) or post-dynamic recrystallization (PDRX) [18], which will not be further discussed in this review due to the space limitation.

It has been realized for many years that the nuclei of recrystallization are not produced by random atomic fluctuations as for the case of phase transformations, but small volumes which previously exist in the distorted microstructure [1]. For these small volumes, frequently subgrains, to effectively become growing new grains, it is necessary to have a high angle misorientation, as well as an energy advantage. The classical theory of strain induced grain boundary motion (SIBM) projected by Beck and Sperry [13], which involves the bulging of part of pre-existing HAGB that is associated with a single subgrain possessing great size to provide the energy benefit, more confirmed this. For SRX, this energy benefit can be built from side to side SRV by subgrain growth [13] or subgrain coalescence. The time preferred to form the large subgrains such that the motivating force appropriate to the stored energy is enough to overcome the boundary curvature, according to the classic work of Bailey and Hirsch [13], is termed as incubation time to indicate the initiation of recrystallization *during SRX*, This definition is, on the other hand, no longer convenient for DRX, thus the critical strain (ε_{cr}) or critical dislocation density (ρ_{cr}) is often used to define the onset of recrystallization. After defining the key terminology within the recrystallization domain, the factors influencing DRX are now summarized in the following section.

Summary

The three DRX processes, *i.e.*, DDRX, CDRX and GDRX, were taking place in dissimilar TMP conditions *at high temperatures* for a variety of metallic materials have been reviewed. The terminologies used in DRX field were summarized, collectively with the key factors influencing the DRX processes, in addition to the investigational techniques to characterize them. An importance was particular on the mechanisms and the presented numerical models.

DDRX, moreover known as conventional DRX, is evidently the most widely studied DRX process. DDRX operates in low to medium SFE materials at high temperatures (T > 0.5 Tm). At this stage, generally DDRX initiates by strain of grain boundaries along with a necklace structure composed of fine recrystallized grain are produced. DDRX might also take place at or near further heterogeneous sites such as shear bands, kink bands; however these sites are more commonly produced at low deformation temperature or else high strain rates. It is obvious that the new grains might be formed by twinning at high strain levels. In general, the stable state of DDRX in conditions of flow stress, recrystallized grain size as a purpose of deformation temperature as well as deformation strain rate is well established. The DDRX at changeable TMP conditions, still, needs more investigation. Because it is apparent from experiments that different DRX mechanisms can perform together or in series with changing TMP. It becomes further complex if precipitation, which affects nucleation of recrystallization as well as grain boundary migration, takes place at the same time as with deformation. There exist many of numerical models of DDRX; the majority of them work well in some preferred situations. On the other hand, none of these models is deemed as a general model that takes all key ingredients of DDRX into account and is capable to model DDRX in changeable conditions for particle-containing materials.

In terms of CDRX, it was establish that this DRX process takes place in all metals as well as alloys irrespective of their SFE when the deformation temperature is moderately low (T < 0.5 Tm), in severe plastic deformation. At high temperatures, however, CDRX is most often observed in high SFE materials. The mechanism of CDRX is not also understood as for DDRX, the main matter is how the LAGBs increase their misorientation and convert into HAGBs. At low deformation temperatures, it appears that LAGBs can increase their misorientation consistently by the misorientation saturates at relatively low values, and it is the formation of microshear bands or kink bands that ultimately leads to the formation of HAGBs. At high deformation temperatures, microshear bands or kink bands become less significant, the transformation of

LAGBs to HAGBs is done moreover by the equal increase of misorientation or progressive lattice rotation by grain boundaries. Experimentally, the arrangement of HAGBs also depends on the grain orientation, and there exist a number of stable orientations, in which the increase in misorientation is not enough to form new HAGBs even at large deformations. A whole systematic study on CDRX by varying original grain sizes, grain orientations, chemical compositions, deformation temperatures and strain rates etc., providing quantitative data on flow stress, crystal size and misorientation evolutions, is still missing. CDRX are further complex situations where precipitation or fine second-phase particles are involved, or in varying TMP conditions, also needs further development. In terms of numerical modeling, only a sparse amount of CDRX models are found in the literature. This is partly due to be short of an apparent image on the CDRX process mechanisms.

As compared to the two DRX processes mentioned above, GDRX appears rather simple. The geometric shape alter due to deformation was studies under different deformation modes such as hot rolling, hot torsion and plane strain compression. It is concluded that GDRX takes place once the thickness of the HAGBs reaches 1-2 subgrain size distance. This frequently means, even it is not clearly stated, that GDRX occurs concurrently for each grain. Since the rate of GDRX is infinitely dependent on the thickness of HAGBs, the grain morphology is of greater significance as compared to DDRX and CDRX. This includes the local variations such as serrations, and the global shape, *i.e.*, spherical, cubic or any new shapes. HAGB migration suitable to dislocation density variation or grain boundary energy, also the introduction of new HAGBs, needs to be considered to clarify the discrepancies experimentally observed in conditions of the evolution of the thickness of HAGBs. Correspondingly to the further two DRX processes,

precipitation or fine second-phase particles influence the GDRX process through retarding grain boundary migration. Dissimilar models are developed, furthermore at this stage, physically based models combining the general geometric shape modify with the local grain morphology (serrations) and HAGB migration, as well as the formation of new HAGBs are still missing. Since most of the above mentioned phenomena are already modeled, it seems realistic to combine them into a single model.

In summary, it can be said that although some of the aspects related to DRX still need further investigation, the existing knowledge obtained from both experiments and simulations can already provide valuable help on controlling the microstructure evolution during thermomechanical processing. Fig. 35-37 illustrates the various modes of dynamic recrystallization. All the referenes mentioned in 5, 5.1, 5.2, 5.2.1 and 5.3 can be found in the original review article by Huang and Logé [50].
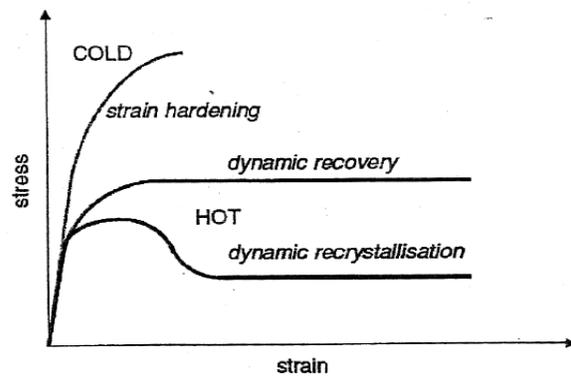
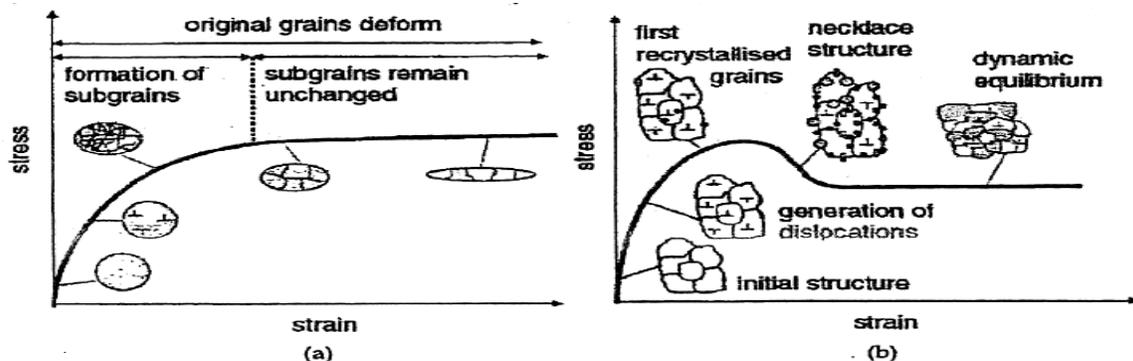Fig. 35: Typical flow curves during cold and hot deformation [57].

Fig. 36: Evaluation of the microstructure during (a) Hot deformation of a material showing recovery. (b) Continuous dynamic recrystallization (CDRX).
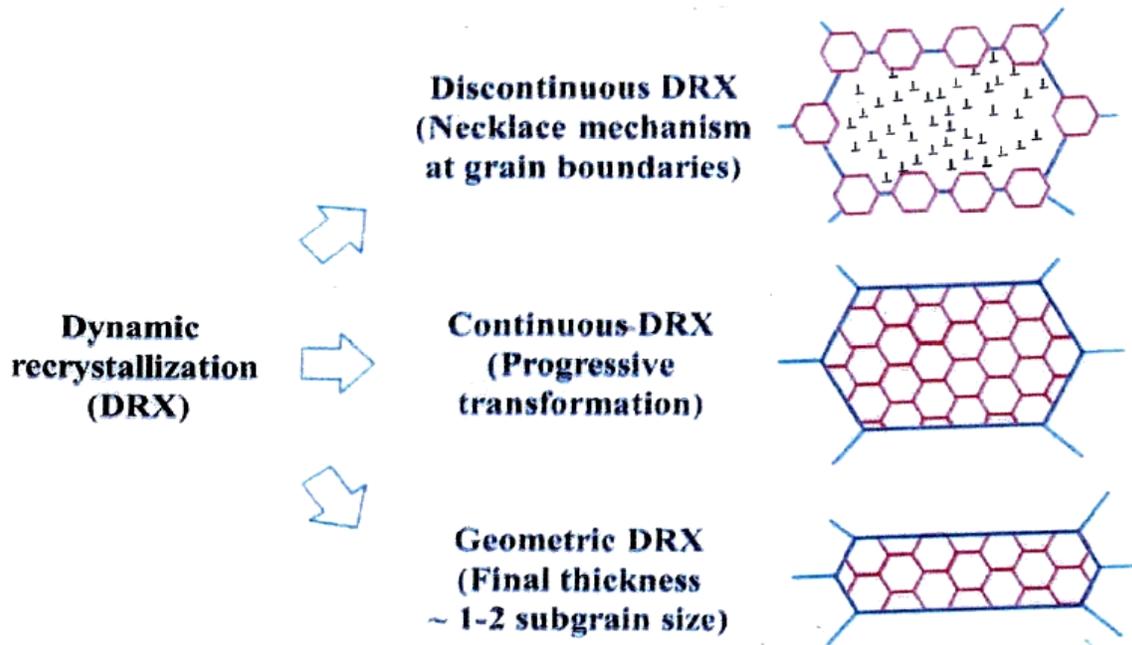
Fig. 37: Graphical abstract. [48].

Grain Growth

Structure of Metals

Very often we presume that the reader knows everything we are talking about. It is not so simple. Sometimes the old books contain excellent explanations/definitions of certain simple terms. Before discussing grain growth, we are taking the liberty of defining some grains-related terms.

"Metals are crystalline, *i.e*. they consist of small crystals – They are thus polycrystalline. The crystals in these materials are normally referred to as its grains. Since the crystals/grains are small, optimal microscoping, using up to 1000 magnifications, is used to study the structural features associated with the grains. This is known as sub-structure. Furthermore, there is the basic structure inside the grain themselves, *i.e*. the atomic arrangements inside the crystals known as the crystal structure". [14].

We will not be discussing the theory of metals and related topics such as dislocations and ship phenomena, but later grain boundaries and all related phenomena.

Nucleation of sub-grains/cells

We know that the grains in metals and alloys play very important roles in determining their various properties. As we have mentioned earlier, all metals/alloys are made up of crystals/grains. Their shapes and sizes depend on the composition, on the method of preparation (e.g. casting, rolling, swaging, temperature, cooling rate etc.) of the materials. The important factor is diffusion which depends on the processing temperature of the material. The significant direct result of a temperature rise in a solid material is that it increases the diffusion rate of self-diffusion and for the diffusion of the atoms (alloying or impurities) that may be present. Basically, diffusion is simply the spontaneous change in location of individual atoms within an existing material. Any process or reaction that involves spontaneous rearrangement or redistribution of individual atoms is done solely by diffusion. The role of diffusion is important in almost all metallurgical processes such as grain growth, recovery, recrystallization etc. And any action that influences diffusion rates directly affects all such processes and hence is of vital interest to the material scientists, solid state chemist and solid state physicists. The probable mechanism of diffusion also depends on vacant lattice points in the material. It means that diffusion is accomplished by successive movements of individual atoms, and the movement of any given atom in a solid necessary involves certain energy of activation. And whenever activation energy is involved, the Maxwell-Boltzmann probability formula applies. (For detailed description please refer to ref.21). It must be understood that physical and chemical changes accomplished by diffusion can be tremendously accelerated by relatively small increases in temperature. For any activitation energy per mole is required to initiate diffusion, is based

upon the average behavior of a large number of atoms. Individual atoms within a group/cluster of atoms vary widely in energy at any given moment, so that some of the atoms present will require for their activation a relatively large increment of energy, others a small one, and a few none at all. Any factor that increases the initial energy of a specific atom correspondingly diminishes the additional energy required to activate it for diffusion. Therefore, the potential energy available from a plastically deformed area will be activated, increasing the diffusion rate locally. It is thus important to remember that diffusion is much more rapid in a deformed high energy region of a grain boundary than through the less important crystals which meet to form the grin boundary and diffusion rates are significantly higher for a deformed metal than for the same undeformed metal free of the internal energy stored by deformation. In 2.3 and 4.1.1 we have briefly discussed the formation of sub-grains/cells and polygonization structures. Doherty *et al* [34] have discussed in some detail the nucleation of sub-grains and grain growth during recrystallization.

"According to them, through deformation, energy is stored in the material in the form of dislocations. This energy is free in recovery, recrystallization and grain growth when the material is annealed at an adequately high temperature. Recrystallization occurs by strain-induced boundary migration of only a few grains. These a small number

of grains grow become very large at the expense of the small grains in the matrix.

The growth of newly formed strain-free grains at the expense of the polygonized matrix is obtained by the relocation of high-angle boundaries. The motivating force for recrystallization is the remaining strain energy in the matrix following recovery. This strain energy exists as dislocations mainly in the sub-grain boundaries. Thus, the various factors that control the mobility of the high-angle boundary or the driving force for its migration will influence the kinetics of recrystallization and consequently the grain structure. In connection with the driving force for recrystallization, a fine sub-grained matrix has high strain-energy content than does a coarse sub-grained matrix. Accordingly, recrystallization (also grain size) occurs faster in a fine-sub-grained matrix than in a coarse-sub-grained matrix. During recrystallization, continued recovery may occur in the matrix by sub-grain growth, resulting in a reduction of the driving energy for recrystallization and therefore a decrease in the recrystallization rate (and retardation of grain growth). From driving energy considerations, it is understandable that the tendency for recrystallization (and grain growth) is stronger in heavily deformed than in moderately or lightly deformed materials. For a given deformation, the finer the original grain size, the stronger the tendency for recrystallization. Fig. 38 [17] shows effects in low carbon steels.
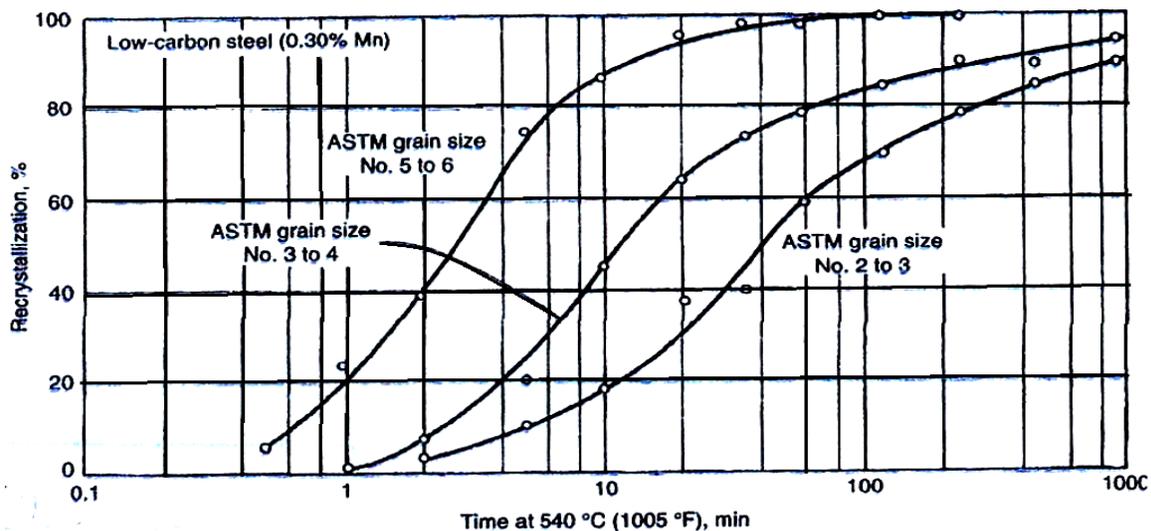

Fig. 38: Effect of penultimate grain size on the recrystallization kinetics of a low- carbon steel, cold rolled 60% and annealed at 540°C (1005°F). Note the incubation time is shortened as the penultimate grain size before cold rolling is decreased. [17]

"Sometimes a section of the initial boundary of grain results grain-boundary nucleation from the

region of a high dislocation content and is observed in large-grained materials deformed at low and

medium strains. This bulging mechanism of nucleation for recrystallization is a consequence of the strain-induced boundary migration. As shown in Fig. 39 [17], the nucleus may be produced by subgrain growth to the exact of the original grain boundary in grain boundary migration to the right as well as subgrain growth to the left forming a fresh high angle boundary, and by grain boundary migration to the right and subgrain growth to the left but without forming a new high angle boundary.
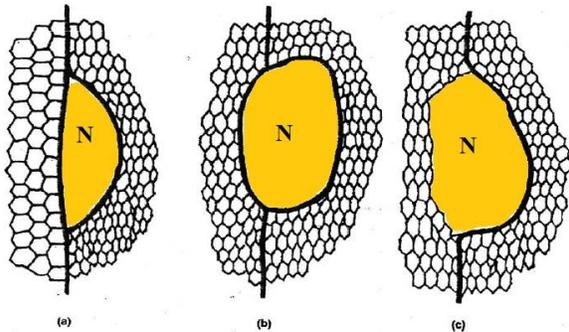

Fig. 39: Schematic showing three types of grain-boundary nucleation and the growth of the nucleus (N) at the expense of the polygonized subgrains. [17]

Grain Growth

According to Cahn [13], "when primary recrystallization is complete, the material have grain boundaries having fixed interfacial energies, irrespective motivating force of retained energy is used up and the grain structure is not so far stable. Thus situation is at best metastable, with ideal thermodynamic stability is simply attained as the sample has been converted into a monocrystal. The situation is closely similar to foam of soap bubbles (Fig. 26) which progressively coarsen along with become a single bubble (as discussed earlier in 4.2). Fig 26 served to clarify the unsteadiness of grain structures on geometrical foundation. Initially assume that in two dimensions, the grains were collection of perfectly regular hexagon. The sides of the grains would next be flat and all triple points would be in equilibrium, as grain boundaries (in two dimensions, strictly grain edges) then all meet at 120°. Since all boundaries are assumed to be high-angle boundaries, as a result of equal energy, the triangle of forces at the triple point is stable. On the other hand, if a "rogue grain" has simply, say 5 sides, the standard internal angle will exceed 120° unless the sides are curved convex outwards – then the sides should be unstable, and in search of shorten themselves during straightening, they will concern the triple point

equilibria at the apex, therefore the grain in progressively consumed. The smaller the quantity of sides, the sharper is the curvature of the sides for a certain grain size, and more rapidly is the procedure of absorption. Once a rogue grain has vanished, the new neighbors find themselves out of equilibrium, and the process continues. Grains with over 6 sides grow, those with smaller amount vanish. In a soap-bubble array, it has been experimentally that the bubble diameter D follows the time law:

$$\frac{dD}{dr} = \frac{1}{D}$$

The rate of boundary migration is inversely proportion to the mean radius of curvature and that this term is proportional to D". For detailed discussion, please consult Cahn [13].

In short, the phenomenon of grain growth is rather complicated and a lot of work has been done over many decades. There are many text books, conferences proceedings, research articles and review articles. They have been mentioned with references.

As stated earlier, the procedure of recrystallization of plastically distorted metals and alloys play very significant roles in the making, shaping and usage of materials. Furthermore, during deformation, energy is stored in the material in the form of dislocations, stacking-faults, vacancies etc.

The main point to remember in the process of grain growth is self-diffusion. Whenever a grain boundary exists in a metal, the high energy offers a sufficient thermodynamic reason for grain growth to occur – to eliminate the boundary and so to reduce free energy in the metal. The tendency toward grain growth exists in all polycrystalline metals at any (reasonable high) temperature. The rate at which it actually occurs varies from unobservable show to tremendously rapid according to environment and internal structure.

"In the fifties, four variables were considered of interest in their influence upon the rate of grain growth [21].

(a) *Difference in original grain size*:

Difference size between adjacent grains provides the diffusion-rate differential for active grain growth: the greater the size difference, the higher is the rate differential, and rapid the growth of the larger grain.

(b) *Elevated temperatures*:

A substantial increase in temperature increases diffusion rates accelerates grain transfer between adjacent grains, so increases the rate of growth of the larger grain.

(c) *Initial grain size*:

"If there is a similar distribution of grain sizes, the rate of grain growth at any given temperature is higher if average grain size is fine than if it is coarse."

(d) *Extent of cold work*:

"Energy stored in a metal is effective in increasing the rates of diffusion and grain growth and is somewhat selective in this process. During cold working, the smaller crystals within the aggregate are formed to strain harden more rapidly than the larger ones, and so to acquire a disproportional great share of the total energy stored in the metal. The resulting disproportional-energy differential reinforces surface-energy differences in promoting growth of the larger grains at the rate of smaller ones.

So far so good for the earlier knowledge: There is a lot of literature available on the grain growth, both normal and abnormal. But the process starts first with cell formation. We explain it again briefly. According to Cahn [13], the division of dislocations, vacancies in a plastically deformed metal is not identical. It was established that grains in different deformed polycrystalline metals consist of sub grains or cells, the interiors of which contain a moderately low dislocation concentration which the small-angle cell boundaries (cell walls) sharpen after then the cells gradually grow larger as their interiors happen to more drained of dislocations. After recovery the cells are of same dimension in all direction. Among all metals dislocation with more than a few different burgers vectors are offered to form cell wall and sharpness varies. There is a direct selection among the stacking fault energy and therefore the degree of dissociation of dislocations with their capability to climb, and the sharpness of cells walls. The cell formation is a rather compound form of polygonization and the cell boundaries are comparable to sub-boundaries formed by polygonization. Fig 40 shows a family of etched in an annealed aluminium sample. Cell structures are formed mainly in effect through creep and fatigue. In these cases the functional stress equally produces the deformation and helps the dislocations to climb into the cell walls. Cell walls are extreme mobile than

average large-angle grain boundaries. It is probable that cell walls in metals move simply in the influence of internal stress in all grain and that small-angle tilt boundaries progress readily in the control of a suitable sheer stress, creating thus a small plastic strain. The recovery of yield stress/hardness of a metal through progressive annealing behind plastic deformation is about certainly due to cell structure/size.

Fig. 40: Subgrains in annealed aluminium sheet. X100. (After LaCombe and Beaujard {1947}.) [13]

A general association has been established between yield strength and cell size:

$\sigma = \sigma_o + kd^{-1/2}$, where σ = yield stress

d = sub-grain or cell size, σ_o = friction stress when dislocations glide on the slip plane and
k = stress concentration factor.

Such a relationship was found by Khan [52-56] between yield stress and martensite plate size (sub-structure) in Cu-Al-Zn martensites and the growth of plates had followed the time law. These papers were described by Prof. Walter S. Owen of MIT and Prof. R. W. Cahn of Sussex/Cambridge as excellent and most indigenous works. Some other aspects pointed out by Cahn [13] of relationship between yield stress and cell/subgrain structure are:

(a) Metals of low stacking fault energy, wherein dislocations cannot readily climb, do not form pointed cell structures; also no recovery of yield stress is observed.

(b) A stress applied in recovery anneal accelerates recovery of mechanical properties, also it accelerates the increase in growth of the cell structure.

(c) The mean sub-grain size and misorientation inside a variety of grains in rolled iron depends on the orientation of individual grains relative to the rolling plane as well as direction, etc.

Role of grain boundary migration in grain growth

In brief, we know that grain boundaries has significant role in grain growth properties of engineering materials also this process has an important function in recovery, recrystallization and grain growth. Grain boundaries have been classified as low angle grain boundaries, high angle grain boundaries etc. Humphreys and Hatherly [20], Mittemeijer [19], Cahn [13], Raabe [1], Doherty *et al*. [34, 35], Burke and Turn-bull [37], Aust and Rutter [55], Lücke and Stüwe [56], Hu Hsun [17] etc etc have thoroughly discussed this process and can be consulted for detailed information. This process can be summarized in these words: "Recovery may be defined as the low angle boundary migration and during the nucleation of recrystallization, and Recrystallization may be defined as the motion of large angle boundaries. Humphreys and Hatherly [19] have extensively discussed the various theories of the mobility of low angle as well as large angle margins, and all the various factors that effect both these migrations. The various factors can have both

negative and positive effects on grain growth (or retardation) and subsequently on the mechanical and physical properties of engineering materials.

Humphreys and Hatherly [20] have also discussed (Chapter 6) the sub-grain coarsening, the motivating force for sub-grain growth, experimental measurements of sub-grain coarsening, sub-grain growth in boundary migration, sub-grain growth through rotation and coalescence, the cause of particles on the rate of sub-grain growth and the particle limited and particle stimulated sub-grain size and growth and the abnormal grain growth. It is this last phenomenon that we briefly discuss.

Abnormal grain growth

"The process in which microstructure of grains after recrystallization becomes unstable and formation of some grain might extremely small Fig 41 (a+b) and even greater is called abnormal grain growth". It is a significant way for producing large-grained materials for processing of Fe-Si alloys used for electrical applications. The motivating force for abnormal grain growth is generally the reduction in grain boundary energy". For further information please see [19]. Fig 42 (a) shows abnormal grain growth in Al - 1% Mg + 1% Mn alloy annealed at 600ºC [19] and a polycrystalline Zr Specimen showing individual grains with pronounced grain boundaries, Fig. 42(b).
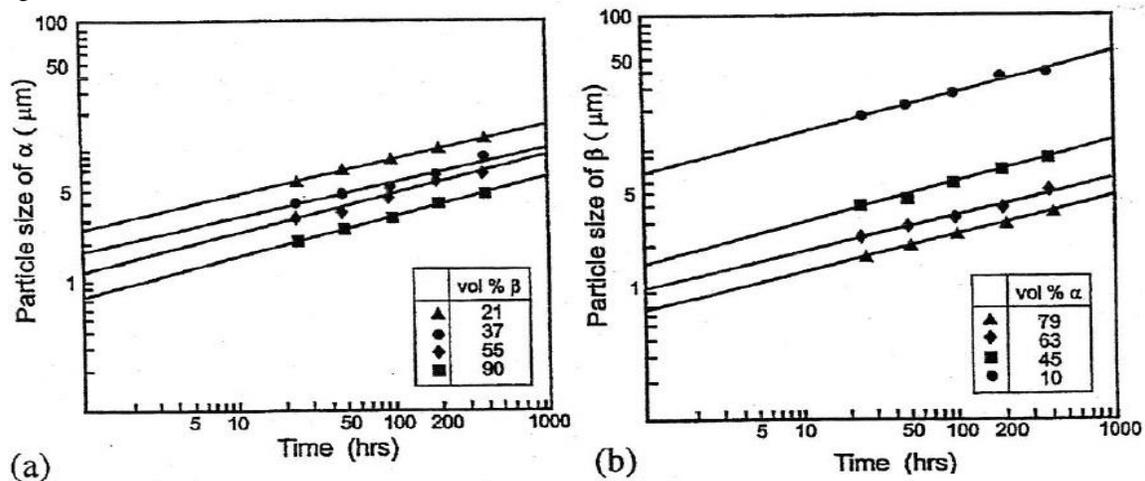

Fig. 41: (A) Growth kinetics of the phases in a series of Ti-Mn alloys containing different volume fractions at 973K, (a) α phase, (b) β phase. (After Grewel and Ankem) {1989}.) [19].
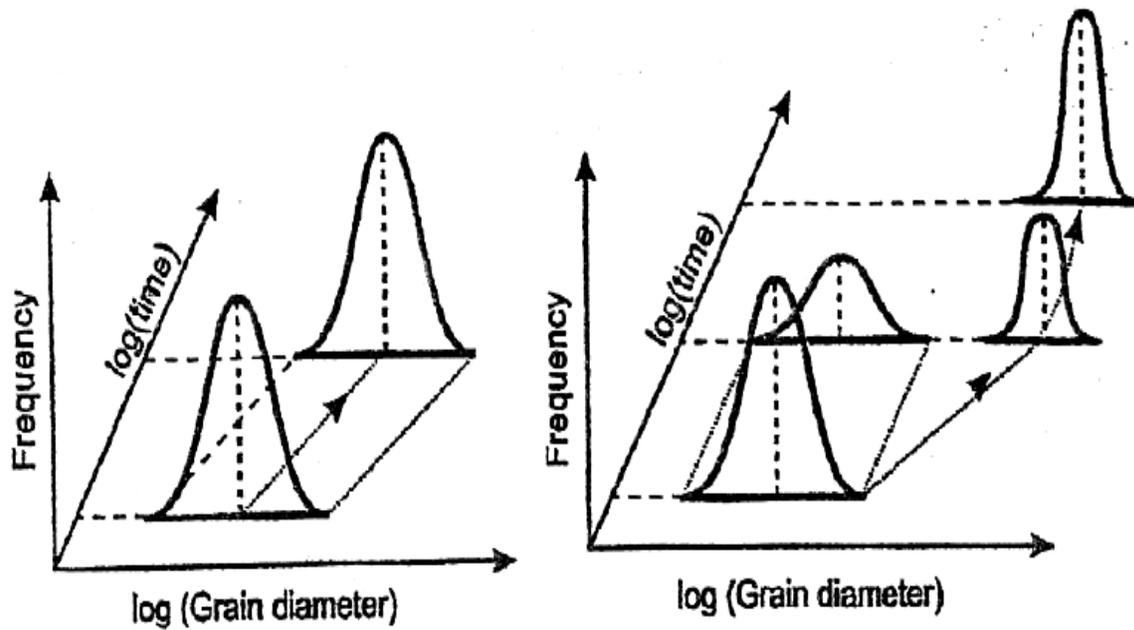
Fig. 41: (b) schematic representation of the change in grain size distribution during (a) normal size growth. (b) Abnormal grain growth. (After Detert {1978}.) [19].

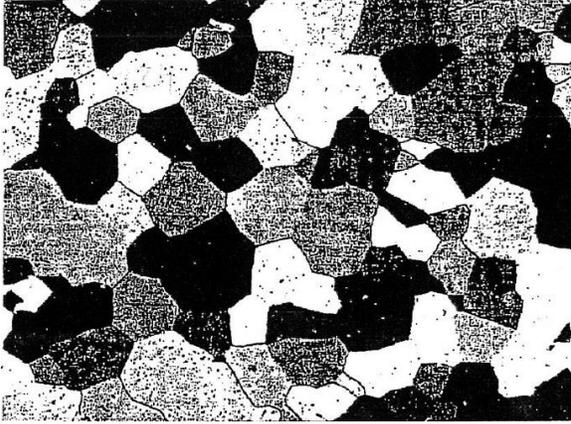Fig. 42: (A) Abnormal grain growth in Al-1% Mg-1%Mn annealed at 600°C. [19].

Fig. 42: (b) Polycrystalline zirconium specimen photographed with polarized light. In this photograph, individual crystals can be distinguished by a difference in shading, as well as by the thin lines representing grain boundaries. X350 (Photomicrograph by E. R.. Buchanan).

Computer Modeling and Simulation

Humphreys and Hatherly [20] have also given details of many simulation methods, viz., Monte Carlo simulations, Cellular automatic, Molecular dynamics, Vertex simulations, Computer Avrami models, Neural net work modeling etc etc.

Measurement of Recrystallization

Humphreys and Hatherly [20] have mentioned the following techniques and associated parameters:

Techniques for measuring recrystallization

(a) Optical microscopy
(b) Transmission electron microscopy
(c) Scanning electron microscopy
(d) Electron backscatter diffraction
(e) X-ray diffraction
(f) Ultrasonics
(g) Property measurements

Driving pressure for recrystallization

(a) Calorimetry
(b) X-ray diffraction
(c) Electron microscopy and diffraction

Fraction recrystallized

(a) Microscopically methods
(b) EBSD methods

Nucleation and growth rates

(a) Nucleation of recrystallization
(b) .2   Growth rates

Grain and subgrain size

(a) .EBSD measurements
(b) .Calculation of size
(c) .Precision of measurement

Grain boundary character distribution

(a) Disorientation angle
(b) Disorientation axis
(c) Boundary plane

Grain boundary properties

(a) Boundary energy
(b) Boundary mobility

Parameters of two-phase alloys

(a) Particle size
(b) Volume fractions
(c) Interparticle spacing
(d) Particle distributions

Summary

Scientists and engineers are always fascinated by subject of recovery, annealing, recrystalization and grain growth due to their industrial applications. The treasure is subjected after 150 years in this field. This has been achieved by the efforts of hundreds of scientists and engineers and costs worth hundreds of millions of dollars (if not billions). But being always curious and unsatisfied, the scientists and engineers are still work in this field. A lot of engineering and physical properties of materials depends on the processing of cold-worked metallic materials and their heat treatments. We would like to quote Cahn [13] for the brief definition of these processes.

"*Recovery* usually implies those changes which do not occupy the sweeping of the distorted material by migrating high-angle grain margins. Thus, in this stage, a deformed crystal (or a polycrystalline aggregate) retains its basic identity, even though the density and also division of defects inside it change.

Recrystallization is the term applied to those stages during which the crystal orientation is changed, possibly more than one time, by the way of high-angle grain boundaries through the material.

Primary Recrystallization is the process in which novel grains are nucleated then grow at the rate of the deformed material until this is entire consumed. At this stage the migrating boundaries will have impinged on each other. Primary recrystallization is then complete and the metal has a minimum recrystallized grain-size which is dependent on such factors like its composition, the degree of preceding deformation and the annealing temperature.

Grain Growth involves the more migration of grain boundaries, by a concentrated rate, through the recrystallized material; thus producing a structure which consists of a few of grains, the average size of which is bigger than that at the end of primary recrystallization.

Normal Grain Growth is the other common form of grain growth in which rate of boundary migration is about identical all over the material, with the consequence that, at any moment, the grains are approximately similar in size.

Abnormal Grain Growth is the situation in which grains becomes extremely large at the rate of residue due to migration during grain growth is stopped.It is occasionally called "secondary recrystallization" or "coarsening".

Due to indication of return of various properties to values which are characteristics of undeformed state irrespective of mechanism the term "recovery" is frequently useful in a special, and more literal sense. The reference state is at all times the completely recrystallized (so-called "fully annealed") condition in which none of the deformed structure is retained." [13]. Figs 43 to 46 illustrate these phenomena graphically.

The works of Cahn, Beck, Mittemeijer, Raabe, Doherty and Hatherly and their colleagues etc and a number of invaluable books, Conferences and Workshop Proceedings till 2017 have all contributed enormously to the development of this treasure of knowledge.
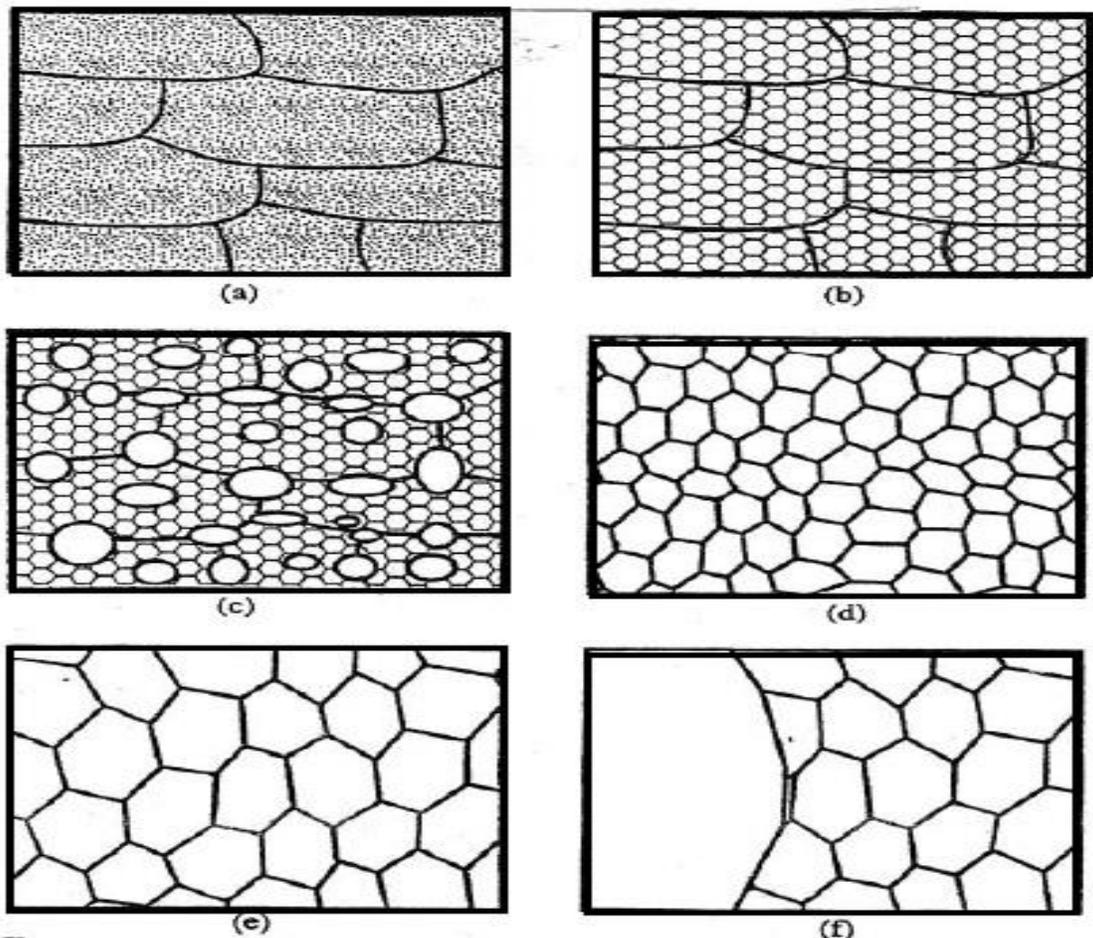

Fig. 43: Schematic diagram of the main annealing processes: (a) deformed state, (b) Recovered (c) Partially recrystallized, (d) Fully recrystallized, (e) Grain growth and (f) abnormal grain growth. [19].
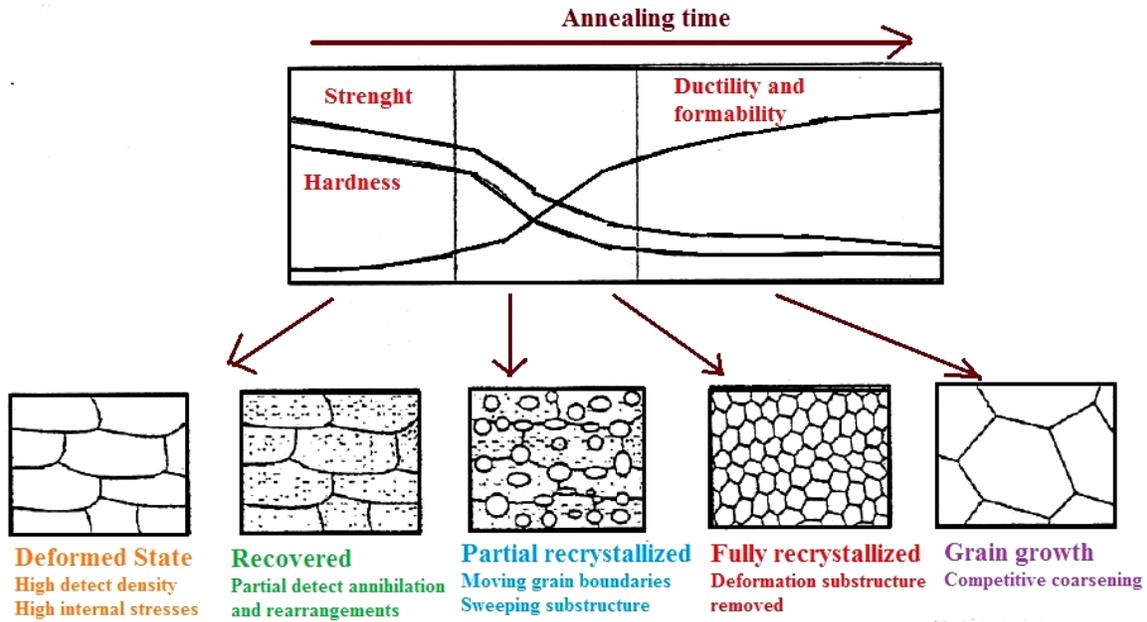
Fig. 44: Schematic diagram illustrating the most generic steps during recovery, static primary recrystallization and grain growth.
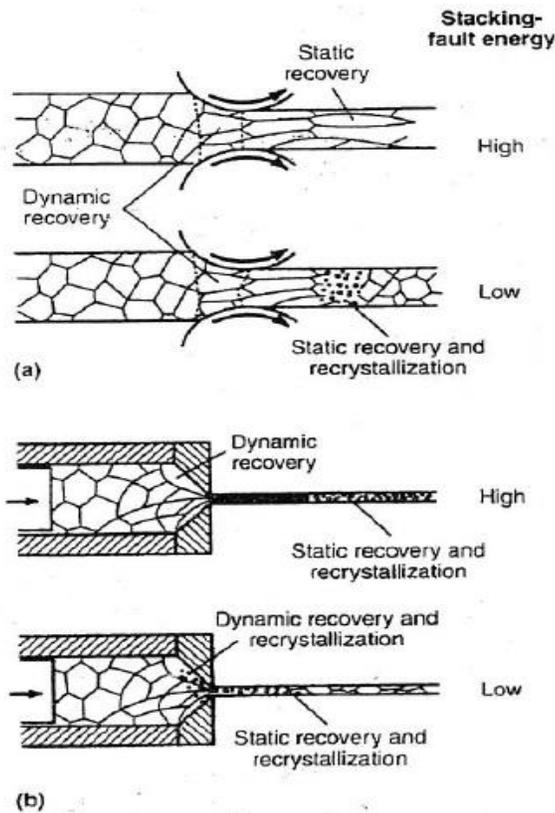

Fig. 45: Hot working effects on microstructure. (a) Rolling with a thickness strain of 50%. (b) Extrusion with a strain of 99%. [17].
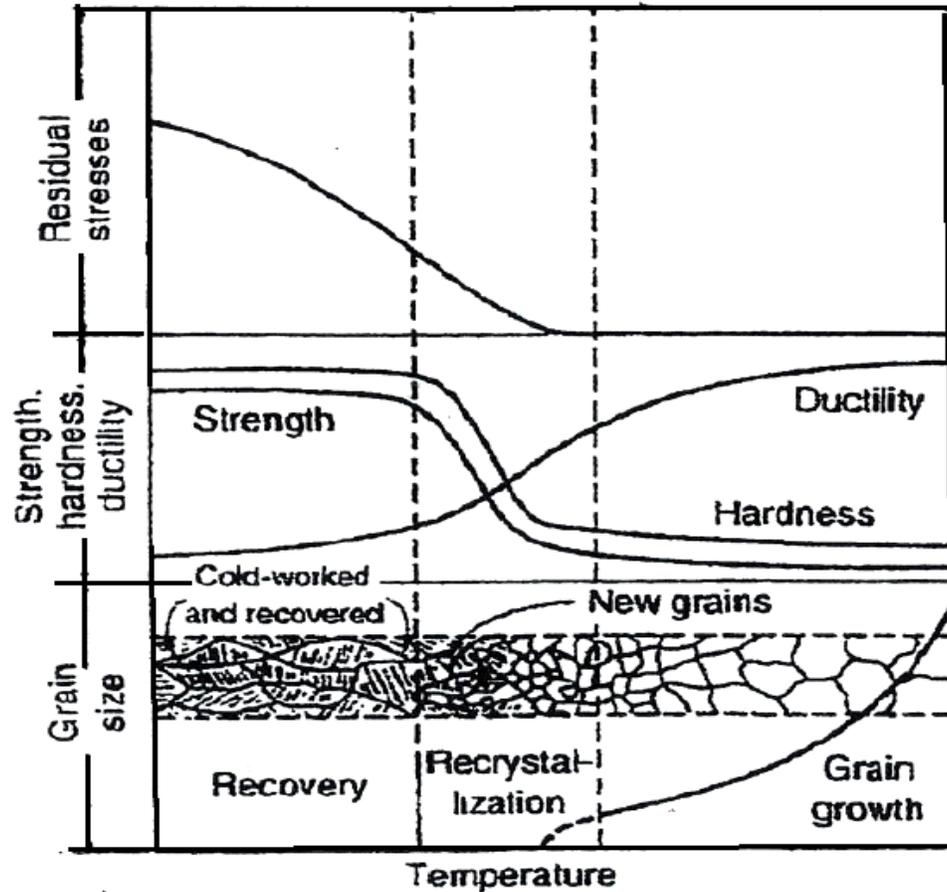
Fig. 46: Effect of annealing temperature on recovery, recrystallization and grain growth of a clod- worked structure. Grains in the recovery region are still strained while recrystallized grain (shown as unlined areas are strain-free). [17].

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