Annealing Textures

Three stages:

1. Recovery
2. Recrystallisation – most significant texture changes
3. Grain Growth
Cold worked 85%

Cold worked 85% + stress relieved at 300°C for 1 hr

Cold worked 85% + stress relieved at 316°C for 1 hr
**Phenomenology of recovery and recrystallisation**

(a) deformed state

(b) recovered state (subgrains)

(c) partially recrystallised state nucleation + growth

(d) recrystallised state

(e) continuous grain growth

(f) discontinuous grain growth or, abnormal grain growth or, exaggerated grain growth

Secondary recrystallisation
Recrystallisation and texture

Metallographical investigation:
- recrystallised volume fraction $X(t)$
- recrystallised grain size $D_{RX}$
- recrystallised grain shape (aspect ratio)
- nucleation density, nucleation sizes

however, no crystallographic information

recrystallisation (nucleation and growth):
formation and motion of high-angle grain boundaries
lead to characteristic changes in crystal orientation, or texture

the recrystallised grains inherit the orientations of their subgrains / nuclei

hence, texture analysis gives valuable information on nucleation and subsequent growth of recrystallisation
Recrystallization of two grains deformed with the identical strain path.
Formation of recrystallised nuclei

- In recrystallisation, an entirely new set of grains is formed.
- New crystals appear at points of high-lattice strain energy such as slip-line intersections, deformation twin intersections, and in areas close to grain boundaries.

Nucleation occurs at points of strong lattice curvature.

Conditions for the formation of a nucleus

- A region of a crystal can become a nucleus and grow only if its size exceeds some minimum value (~ a diameter greater than about 15 nm) for it to be able to expand.
- It becomes surrounded, at least in part, by the equivalent of a high-angle grain boundary.

- Required because the mobility of an arbitrary low-angle grain boundary is normally very low.
Mechanism for the formation of recrystallised nuclei in polycrystals

- If a difference in dislocation density exists across a grain boundary
  - A portion of the more perfect grain might migrate into the less perfect grain under the driving force associated with the strain energy difference across the boundary.

- Forward movement of the grain boundary to form a bulge.

- Boundary movement should effectively sweep up the dislocations in its path,

  Creation of a small, relatively strain-free volume of crystal.

- If this bulge exceeds the critical nucleus size, both primary conditions for the formation of a nucleus would be satisfied.

- Bailey and Hirsch
**Other Mechanisms**

As a result of polygonisation, it might be possible to produce a subgrain capable of growing out into the surrounding polygonised matrix.

- Subgrain coalescence, or the combination of subgrains to form a strainfree region large enough in size to grow.
- Elimination of a subgrain boundary must result in a relative rotation of the two subgrains that are combined.

- Cahn and Beck

![Diagram](image)

- It can be assumed that both climb and slip are involved in the process. Since climb involves vacancy motion, moderately high temperatures are required.

- An extension of this process will eventually result in a small grain surrounded by a high-angle boundary of large enough size to grow progressively into the polygonised matrix that still surrounds it.
Two subgrains with different orientations about to coalesce. To remove the sub-boundary there must be a relative rotation of the two subgrains. For simplicity, assume that this rotation takes place in only one subgrain.

Surface energy effects should now straighten out the cusped sections $BCD$ and $GHI$ at the top and bottom of the combined subgrain.

This rotation shows the two subgrains united.

**Subgrain coalescence**

- an effective movement of dislocations out of the original boundary $CH$ (separating the two subgrains) and into the remaining surface of the combined subgrains. This tends to make this surface one of high energy.
Nucleation of recrystallised grains on shear band
Texture evolution during recrystallisation

Relation between crystal orientation in deformed and recrystallised metals

- A deformation texture may change into an identical recrystallisation texture
- A deformation texture may transform into a different recrystallisation texture (the most common case)
- A deformation texture may transform into a random recrystallisation texture

Nature and degree of perfection of recrystallisation texture depends on:

- the type of plastic working
- the degree, temperature and time of last deformation stage
- the degree of intermediate deformation
- the temperature of intermediate annealing
- the content in alloying elements and impurities
Recrystallisation texture: General observation

- When annealing is done below a definite temperature limit, recrystallisation texture is qualitatively identical with the deformation texture with relative variation in volume fraction of different components.

- It is understood that all recrystallisation nuclei are already present in a deformed matrix in the form of subgrains.

- Deformation texture heavily deformed cold-rolled copper contains the main orientation \( \{112\} \langle 111 \rangle \), plus traces of \( \{100\} \langle 100 \rangle \) orientation, which may be regarded as nuclei of cube texture of recrystallised metal.

- Explains why the degree of deformation, temperature and other conditions of plastic deformation can affect the recrystallisation texture.

- A more perfect deformation texture produces a clearer recrystallisation texture.
The recrystallisation texture is often crystallographically related to the initial deformation texture

- Lattice of recrystallised grains of f.c.c. metals is tilted through ~ 30-40° around the common <111> axis relative to the lattice of deformed grains.
- Texture of primary recrystallisation in b.c.c. metals may be assumed to be obtained by tilting the lattice through an angle of roughly 25-30° on the <110> axis.
- In recrystallised h.c.p. metals, the lattice is tilted through 30° around the <0001> direction, which is common for both deformed and recrystallised grains.

Recrystallisation texture depend on a great number of factors - much more numerous than the factors affecting deformation textures.

Formation of recrystallisation textures depends on

- the initial structure of deformed metal, and
- the ratio of migration rates of various high-angle boundaries, which in turn may be affected by many factors.
The Theories of Recrystallisation Textures

- Two theories
  - theory of oriented nucleation
  - theory of oriented growth
Theory of oriented nucleation

Nucleation:

 Formation of essentially strain free, low dislocation density regions that was capable of rapid growth in to the adjacent deformed material.

 Nucleus should have within 1 to 2°, the local orientation that existed in the deformed volume.

 If the nucleus was to grow there must be across part of its boundary of the necessary large misorientations in the deformed material such as grain boundaries, deformation bands, shear bands etc.

 In low SFE materials, the formation of annealing twins would drastically alter the orientation of the growing nucleus.

 From texture point of view, the significant event is the development of a nucleus with the orientation of a grain forming part of the final texture.

 From microstructure point of view, it is the development of small dislocation reduced volume in the deformed structure.
Nuclei of the required orientation are more profuse than those generated by a purely random nucleation process and it was assumed that as these grow, texture evolves.

The hypothesis of oriented nucleation proceeds from the assumption that some or other orientation is being determined directly at the stage of formation of recrystallisation nuclei.

Sub-grains of a definite orientation grow larger and more perfect and thus serve as recrystallisation nuclei.

The hypothesis of oriented nucleation is helpful in cases where a deformation texture changes to a qualitatively identical recrystallisation texture.
Theory of oriented growth

Recrystallisation texture is determined by the original orientation of the nuclei having the highest rate of growth in the deformed matrix.

In heavily deformed cold-rolled copper, such nuclei are subgrains with the \{100\}[100] orientation. Though these nuclei are very less than the nuclei with the principal \{112\} [111] orientation, they can grow much more rapidly under definite conditions and thus determine the formation of a cubic texture in recrystallised metal, which replaces all other orientations.

Cold rolled copper

annealed at 600 °C

<1 sec

\{112\}[111] and \{110\}[112] - typical of rolling texture,

+ \{100\}[100] and \{110\}[111].

1 sec to 10 minutes

\{100\}[100] – strengthens
\{112\}[111], \{110\}[112] and \{110\}[111] - weakens

As primary recrystallisation and grain growth proceed in copper, grains of a definite orientation grow preferably gradually replacing all other orientations.
The hypothesis of oriented grain growth agrees well with the fact that the lattice of recrystallised grains is rotated relative to that of deformed grains on a common axis through an angle within an interval of 25° - 40°.

High-angle boundaries of the greatest mobility are found in this very interval of misorientation angles of crystals on both sides of a boundary.
Cold-drawn copper wire - double axial $(111> (100)$ texture,
Upon primary recrystallisation - its texture will be identical,

- Number of grains having $(100)$ orientation diminish with increasing temperature of annealing.

\[
\{112\}[111] \rightarrow \{100\}[100] \\
\{110\}[112] \rightarrow \{113\}[112]
\]

**Effect of alloying and impurities**

- Small additions and impurities in the solid solution may have an extremely strong effect on the nature and perfection of recrystallisation texture.

- They may
  - preclude the formation of some or other recrystallisation texture or
  - improve the perfection of a texture.

- Their effect on recrystallisation texture depends on
  - the degree of deformation,
  - annealing temperature,
  - on the concentration of an impurity,
  - and other engineering factors.

- Total effect cannot be described by simple relationships.
Role of second phase particles on recrystallisation

• Many industrial alloys, particularly those of iron and Al, contain large second-phase particles and thus PSN is a likely recrystallization mechanism in many engineering alloys.

• Unlike most other recrystallization mechanisms the nucleation sites are well defined regions of the microstructure which can be altered by alloying or processing. Thus the number of potential nuclei and hence the recrystallized grain size may be controlled.

• The orientations of the recrystallization nuclei produced by PSN will in general be different from those produced by other recrystallization mechanisms. We thus have the potential to exercise some control over the recrystallization texture.

• The interaction of dislocations and particles is temperature dependent, PSN will only occur if the prior deformation is carried out below a critical temperature or above a critical strain rate.
The most important effects which second-phase particles have on recrystallization are:

- The stored energy and hence the driving pressure for recrystallization may be increased.

- Large particles may act as nucleation sites for recrystallization.

- Particles, particularly if closely spaced, may exert a significant pinning effect on both low and high angle grain boundaries.

- The size, distribution and volume fraction of second-phase particles in an alloy are determined by alloy composition and processing, and by varying these, the metallurgist is able to control the microstructures and textures produced during recrystallization.
• Annealing behaviour of a particle containing alloy is dependent on the volume fraction, size, shape and spacing of the particles.

• The effect of the particles on the overall dislocation density. This provides the driving pressure for recrystallization.

• The effect of the particles on the inhomogeneity of deformation in the matrix. This may affect the availability and viability of the sites for recrystallization.

• The nature of the deformation structure in the vicinity of the particles. This determines whether particle stimulated nucleation of recrystallization (PSN) will be possible.
Fig. 9.1. The room temperature hardness of extruded and annealed dispersion-strengthened copper alloys, (after Preston and Grant 1961).
Fig. 9.2. The effect of particle size on recrystallization in Al–Si alloys reduced 50% by cold rolling and annealed at 300°C. (a) The time for 50% recrystallization, (b) The grain size after recrystallization, (data from Humphreys 1977).
Fig. 9.3. The effect of interparticle spacing on the recrystallization of Al–Cu single crystals reduced 60% by rolling and annealed at 305°C. (a) The time for 50% recrystallization, (b) The grain size after recrystallization, (data from Doherty and Martin 1964).
Recrystallization of Al retarded by small particles
PARTICLE STIMULATED NUCLEATION OF RECRYSTALLIZATION

• Recrystallization originates at a pre-existing subgrain within the deformation zone, but not necessarily at the particle surface.
• Nucleation occurs by rapid sub-boundary migration.
• The grain may stop growing when the deformation zone is consumed.
Fig. 9.14. 111 pole figures of (a) the deformation, and (b) the recrystallization texture of a 65% deformed single crystal of Al containing 4.6 μm Si particles. The deformed and recrystallized orientations are related by a rotation of ~40° about an axis close to the marked <112> direction, (Humphreys 1977).
• Although the orientation spread of the nuclei is greater than that of the deformed matrix, the PSN nuclei are not randomly oriented.

• Although the nuclei are not randomly oriented with respect to the deformed material, the spread of orientations around what may already be a weak texture may, when averaged over the whole polycrystalline specimen, result in an almost random distribution of nuclei orientations.
Secondary recrystallisation

- In grain growth - all grains grow larger more or less uniformly and the metal may be characterized by an average grain size index.
  - *continuous grain growth*

- Certain conditions of annealing may produce a structure composed of
  - a multitude of relatively small grains of ~ the same size and
  - an appreciably smaller number of very large grains attaining a few centimeters in diameter
  - *discontinuous grain growth*

- Most grains grow in size very slowly or virtually do not grow at all,
- some individual grains grow to a large size by consuming their smaller neighbours.

**secondary recrystallisation**

Its kinetics is very similar to that of primary recrystallisation.
Main factors that contribute to the “Abnormal grain growth” are:
Second phase particles, texture and surface effects

Fig. 11.19. Abnormal grain growth in Al–1%Mg–1%Mn annealed at 600°C.
Secondary Recrystallisation and Texture

- Grains grown through secondary recrystallisation often have a preferable crystal orientation
  - their texture is always unlike that on primary recrystallisation or grain growth (being more sharp, i.e. more perfect).

- If a cube texture has been formed in copper or aluminium upon primary recrystallization, then the texture of secondary recrystallisation will form in a manner that:

  The lattice of recrystallised grains will be rotated relative to the lattice of grains of stabilized matrix through an angle of 38° on the common [111] axis.

  - KronbergWilson relation

- The nature and perfection of secondary recrystallisation textures depend on the conditions of plastic working, intermediate and final annealing stages, including annealing atmosphere, the thickness of annealed sheets, and other factors.
Transformation texture
Table 1  Orientation relationships between $\gamma$ and $\alpha$

<table>
<thead>
<tr>
<th>Orientation relationship</th>
<th>Lattice correspondence relation</th>
<th>Number of alternatives</th>
<th>Total number of variants</th>
</tr>
</thead>
<tbody>
<tr>
<td>Bain (B)</td>
<td>{001}<em>\gamma \parallel {001}</em>\alpha, \langle110&gt;<em>\gamma \parallel \langle110&gt;</em>\alpha</td>
<td>3</td>
<td>3</td>
</tr>
<tr>
<td>Kurdjumov–Sachs (K–S)</td>
<td>{111}<em>\gamma \parallel {011}</em>\alpha, \langle011&gt;<em>\gamma \parallel \langle111&gt;</em>\alpha (twin related variants)</td>
<td>4</td>
<td>24</td>
</tr>
<tr>
<td>Nishiyama–Wassermann (N–W)</td>
<td>{111}<em>\gamma \parallel {011}</em>\alpha, \langle112&gt;<em>\gamma \parallel \langle011&gt;</em>\alpha</td>
<td>3</td>
<td>12</td>
</tr>
</tbody>
</table>
Table 2  The 24 variants of Kurdjumov–Sachs orientation relationship; symbols \( \gamma \) and \( \alpha \) refer to fcc and bcc crystals, respectively.

<table>
<thead>
<tr>
<th>Symbol</th>
<th>Variant</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>1. ( (111)<em>\gamma \parallel (110)</em>\alpha ) [( 011 ] ( \gamma \parallel [{111 } ] )</td>
</tr>
<tr>
<td></td>
<td>3. ( (110)<em>\gamma \parallel (111)</em>\alpha ) [( 011 ] ( \gamma \parallel [{111 } ] )</td>
</tr>
<tr>
<td></td>
<td>5. [( 011 ] ( \gamma \parallel [{111 } ] )</td>
</tr>
<tr>
<td></td>
<td>1. ( (\bar{1}11)<em>\gamma \parallel (110)</em>\alpha ) ( [{110 } ] ) ( \parallel [{111 } ] )</td>
</tr>
<tr>
<td></td>
<td>3. ( (\bar{1}01)<em>\gamma \parallel (111)</em>\alpha ) [( 011 ] ( \gamma \parallel [{111 } ] )</td>
</tr>
<tr>
<td></td>
<td>5. ( (011)<em>\gamma \parallel (111)</em>\alpha ) [( 011 ] ( \gamma \parallel [{111 } ] )</td>
</tr>
<tr>
<td></td>
<td>1. ( (\bar{1}11)<em>\gamma \parallel (110)</em>\alpha ) ( [{110 } ] ) ( \parallel [{111 } ] )</td>
</tr>
</tbody>
</table>

2 Stereographic plot showing positions of \( \{002\} \) \( \alpha \) poles for each of the 24 K–S variants in \( (111) \), single crystal matrix after Ref. 13. See Table 2 for codes of symbols and numbers.
# BCC to Hexagonal transition

The 12 Orientation Variants According to the Burgers' Relationship During $\beta - \alpha$ Transformation

<table>
<thead>
<tr>
<th>No. of variants</th>
<th>Correspondence planes $\beta$ phase $\alpha$ phase</th>
<th>Correspondence directions $\beta$ phase $\alpha$ phase</th>
</tr>
</thead>
<tbody>
<tr>
<td>V1</td>
<td>$(011) \parallel (0001)$</td>
<td>$[\overline{1}11] \parallel [2\overline{1}0]$</td>
</tr>
<tr>
<td>V2</td>
<td>$(011) \parallel (0001)$</td>
<td>$[1\overline{1}1] \parallel [2\overline{1}0]$</td>
</tr>
<tr>
<td>V3</td>
<td>$(011) \parallel (0001)$</td>
<td>$[1\overline{1}1] \parallel [2\overline{1}0]$</td>
</tr>
<tr>
<td>V4</td>
<td>$(011) \parallel (0001)$</td>
<td>$[1\overline{1}1] \parallel [2\overline{1}0]$</td>
</tr>
<tr>
<td>V5</td>
<td>$(011) \parallel (0001)$</td>
<td>$[1\overline{1}1] \parallel [2\overline{1}0]$</td>
</tr>
<tr>
<td>V6</td>
<td>$(011) \parallel (0001)$</td>
<td>$[1\overline{1}1] \parallel [2\overline{1}0]$</td>
</tr>
<tr>
<td>V7</td>
<td>$(011) \parallel (0001)$</td>
<td>$[1\overline{1}1] \parallel [2\overline{1}0]$</td>
</tr>
<tr>
<td>V8</td>
<td>$(011) \parallel (0001)$</td>
<td>$[1\overline{1}1] \parallel [2\overline{1}0]$</td>
</tr>
<tr>
<td>V9</td>
<td>$(011) \parallel (0001)$</td>
<td>$[1\overline{1}1] \parallel [2\overline{1}0]$</td>
</tr>
<tr>
<td>V10</td>
<td>$(011) \parallel (0001)$</td>
<td>$[1\overline{1}1] \parallel [2\overline{1}0]$</td>
</tr>
<tr>
<td>V11</td>
<td>$(011) \parallel (0001)$</td>
<td>$[1\overline{1}1] \parallel [2\overline{1}0]$</td>
</tr>
<tr>
<td>V12</td>
<td>$(011) \parallel (0001)$</td>
<td>$[1\overline{1}1] \parallel [2\overline{1}0]$</td>
</tr>
</tbody>
</table>
BCC to Hexagonal transition

The schematic diagram of the variant selection from (001)[100] of β phase through orientation variant V11.