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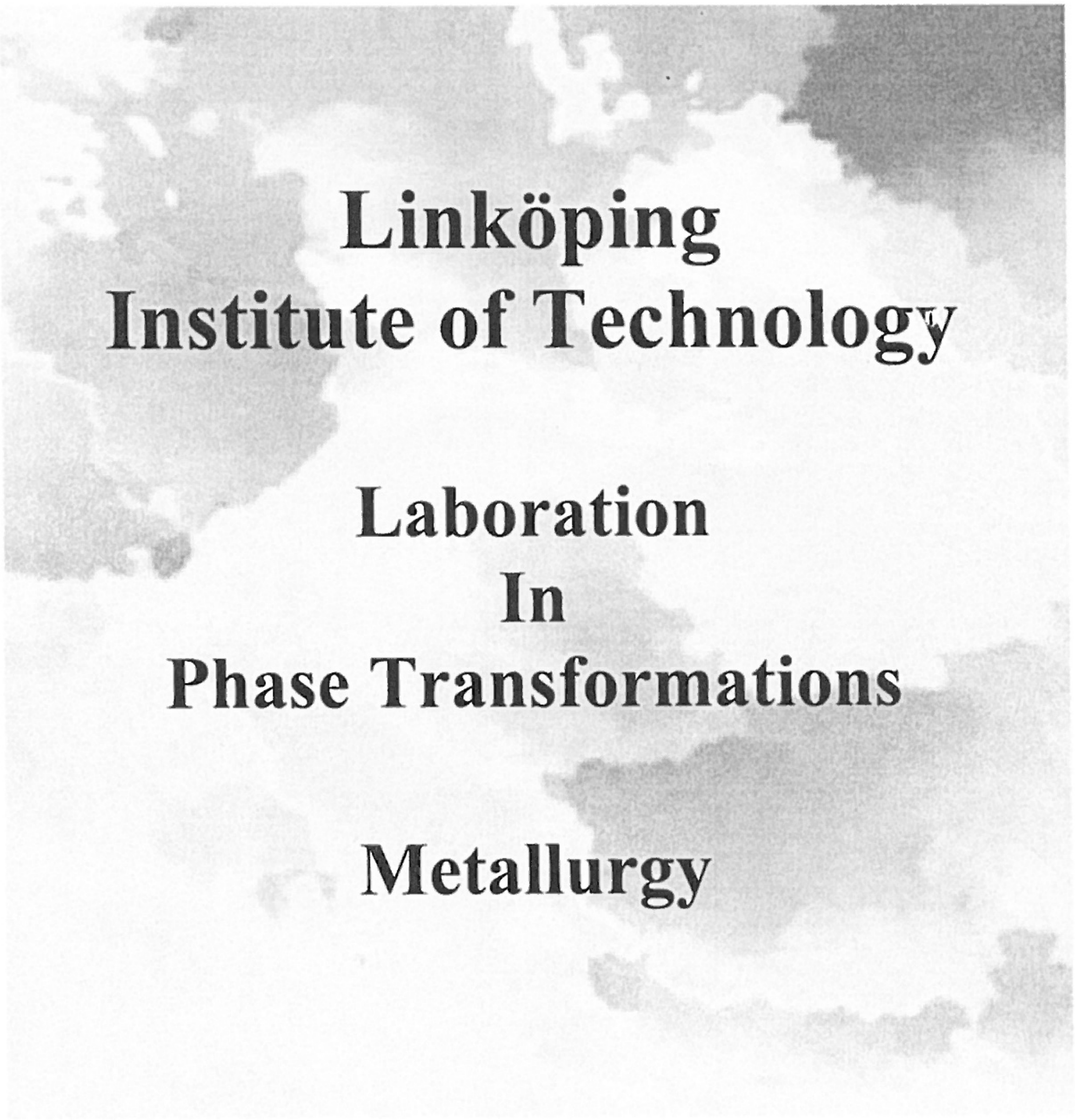
Laboration 80
Phase Transformations



Linköping Institute of Technology

Laboration In Phase Transformations

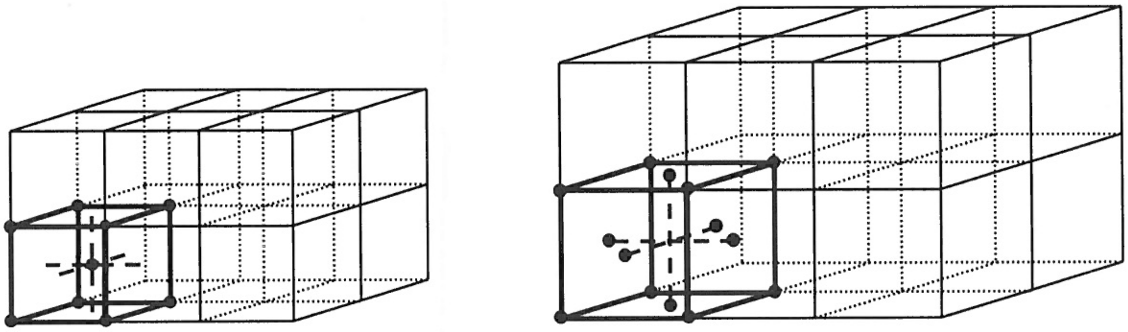
Metallurgy

Metallography
Laboration in Phase transformation

1 Basics

1.1 Single-Component Phase Diagram

A metal can be in different states or phases at different temperatures: in gaseous state, molten or in solid form. Many metals in the solid state exhibit different modifications, which exist in different temperature ranges. Every modification is characterized by a special atomic arrangement in the crystal lattice (see Fig. 1). The occurrence of one and the same element in various modifications is the so-called polymorphy (any material having multiple crystal structures) or allotropy (pure elemental solids e.g. C as graphite, diamond, fullerenes etc.).



α-iron crystal structure, body-centered cubic (bcc)

γ-iron crystal structure, face centered cubic (fcc)

***Figure 1**: Atomic arrangement for different modifications of iron.*

The transition from one modification to another (or generally one state to another) is called **transformation**. The temperature the transformation takes place at is the so-called transformation temperature.

The transformation for a metal plotted on a temperature scale results in a diagram, which for every temperature shows the state of the metal. The stable phase can be deducted from this state diagram, or phase diagram. Figure 2 shows such a diagram for one metal, in this case iron (Fe). It is therefore called a single-component phase diagram. Please note that all gaseous, liquid and solid modifications are called phases. Such diagrams are not restricted to one metal.

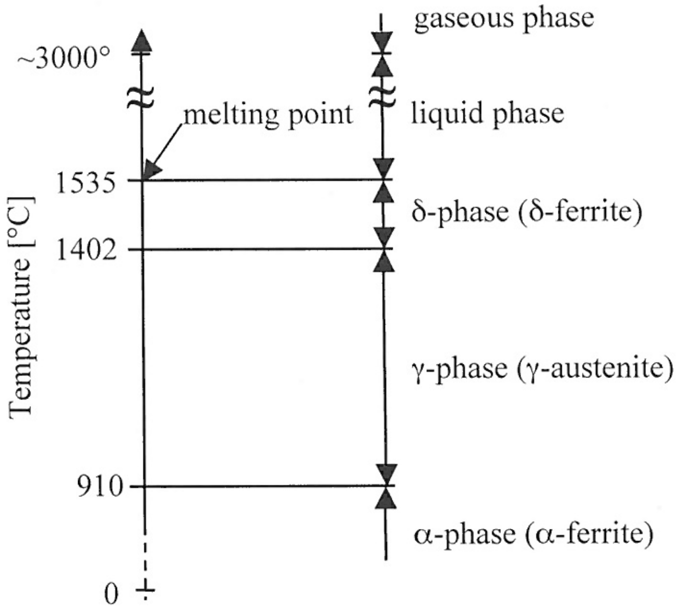


***Figure 2**: Single-component phase diagram for iron Fe*

1.2 Binary Phase Diagram

If a metal is alloyed with another element, its transformation temperature usually changes. A diagram, which shows the alloy's state for every composition is a so-called **binary** phase diagram for an alloy with **two** elements (ternary for three elements and polynary for many elements in one alloy).

1.2.1 Binary system with both components completely solvable within each other in the liquid as well as solid state

Fig. 3 shows a binary phase diagram in which both components are completely solvable within each other in the liquid as well as solid state. Components are denoted with A and B. In the left part of Fig. 3 cooling or recalescence curves are displayed for four different compositions; pure element A, 30% of element B dissolved in A, 70% of B in A and pure element of B.

A phase diagram can be constructed using different methods. One way is with the help of cooling curves from a series of different compositions for one alloy.

Let's consider the cooling curves in Fig. 3. Cooling curves in general show the temperature evaluation over time during which an alloy with one specific composition is cooled down. The cooling curve for pure element A contains one recalescence point at temperature T_1, where the temperature remains constant due to the enthalpy of fusion released with the transition from the liquid to the solid state (the melting point of A).

The cooling curve for the alloy containing 30% of B proceeds without interruption to temperature T_2. At this point the rate of the change of temperature decreases (the curve becomes less steep). This means that the cooling proceeds more slowly between T_2 and T_4. During this interval the alloy solidifies. As a consequence of the continuous and not abrupt solidification process a small but continuous amount of enthalpy of fusion is released opposing the loss of temperature during the cooling and thus decelerates the solidification.

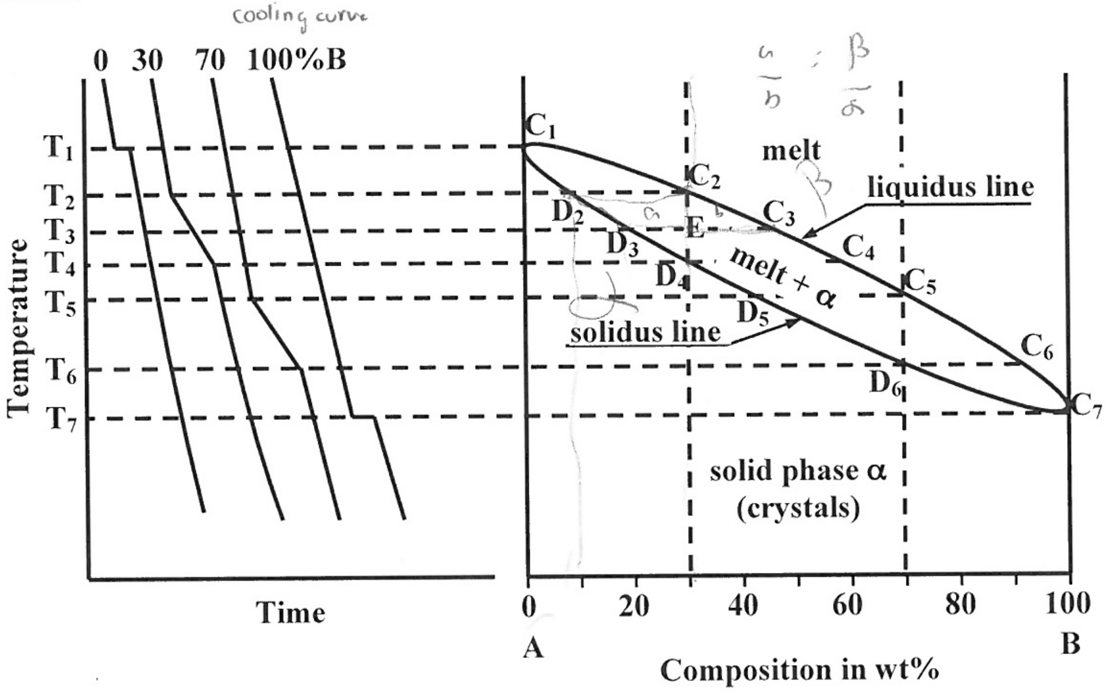


Figure 3: Phase diagram for the materials system AB with complete solubility in the liquid and solid state

An alloy usually does not solidify at a fixed temperature, as the latter example showed, but within a solidification interval. This is a consequence of the changing composition of different phases with varied temperature.

The cooling curve for the alloy with 70% of element B looks similar to the one for 30% of B. However, the solidification interval for the former alloy lies at lower temperatures, T_5-T_6. For the pure element B, the cooling curve shows similar to pure A only one recalescence point at temperature T_7.

In the phase diagram (see right part of Fig. 3) the temperatures for the start and finish of the solidification process are drawn for each composition. All points in the phase diagram denoting the start of the solidification are connected resulting in the so-called liquidus line. The same is done for the finishing points. Above the liquidus line all AB alloys are found in the molten state. Below the solidus line all alloys of AB are in the solid state, as a so-called α-phase in the diagram.

The alloy existing between the two lines consists of both, liquid phase and solid α-phase. Region between the liquidus and solidus line is thus a two-phase region, whereas the regions above and below are single-phase regions. Solidus and liquidus lines are thus borders between single and two-phase regions.

If an alloy with 30% B is allowed to cool down slowly from a high temperature, the following transformation behaviour can be found. The crystallization process starts at temperature T_2, where the first crystal grains are formed. These crystals do not consist of the same composition as the melt but contain much lower B, namely the composition marked by D_2. This is due to the fact that the solid phase has the same temperature as the melt. But at that temperature, in this case T_2, the solid phase can only contain a maximum amount of B given by the intersection of the temperature line T_2 with the solidus line, and that is D_2. This horizontal line, which intersects the liquidus and solidus line at a given temperature, e.g. D_2-C_2 or D_3-C_3, is called a konod. With decreasing temperature the precipitated phase becomes more and more enriched with element B. That in turn means that the residual liquid phase contains also more B. This may seem as a contradiction but it is not because with the beginning solidification process the phase precipitated was very rich in element A. As the temperature continuous to fall, the composition of the melt and the solid phase changes and is given by the liquidus (along C_2-C_3-C_4) and solidus line (along D_2-D_3-D_4), respectively. If the cooling process is slow enough, the composition of the precipitated crystals evens out through element diffusion.

The solidification process is completed when the solid phase reaches the same composition as the initial liquid phase, point D_4. Just before the solidification is concluded the remaining melt (very small amount) exhibits a composition marked by C_4.

The melting of this alloy proceeds the opposite way. At temperature T_4 small amounts of liquid phase appear with the composition denoted by C_4. With continuous heating the amount of liquid phase increases and the composition follows the line C_4-C_3-C_2. The composition of the solid phase changes accordingly along D_4-D_3-D_2. At temperature T_2 the melting is complete.

The amount of solid phase compared to melt at a given temperature can be calculated with the lever rule using the point of intersection between the line denoting the initial composition and the konod at the given temperature, e.g. point E at T_3. The amount of liquid phase, for instance at T3 at a composition of 30% B, is given by the length of the line D_3-E compared to the length of E-C_3.

1.2.2 Binary system with complete solubility in the molten state but restricted solubility in the solid state

The character of the phase diagram changes, if the solubility is restricted up to the solidus and liquidus line.

<u>Eutectic phase diagram</u>
Fig. 4 shows a binary eutectic phase diagram with the components A and B including both phases α and β in the solid state. In the left of Fig. 4 cooling curves are shown for the pure element A as well as different alloys.

Alloys with a B-content of up to 30%, point F in the diagram, or with B-content over 80%, point H, have similar cooling curve for their solidification as in the system of unrestricted solubility in the solid phase. After complete solidification, these alloys consist of a single phase, namely α or β, respectively. Alloys with a content of 20-30% B as well as 80-85% B reach a two-phase region α + β with continuous cooling, i.e. precipitation takes place of β and α, respectively.

Alloys with the composition between F and E (30-60% B) cool down to the temperature T_5 as described above. The composition of the melt follows the liquidus line C_4-E and composition of the solid phase the solidus line for the α phase D_4-F. Again, the solid phase is given enough time to even-out the composition through diffusion.

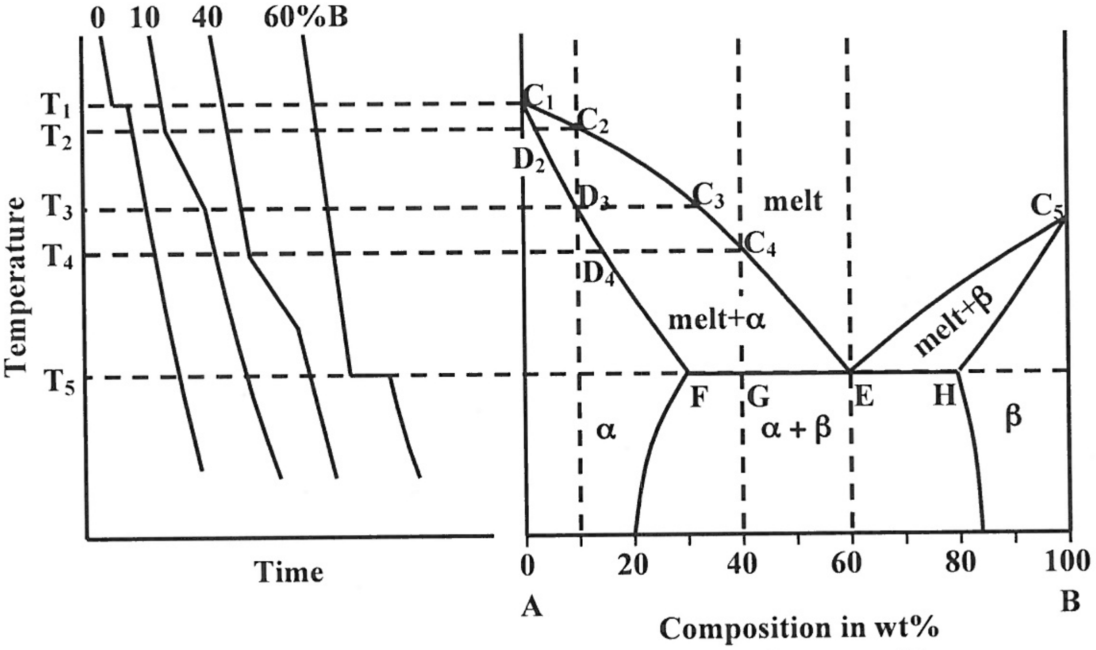


***Figure 4**: Eutectic phase diagram for materials system AB*

At the point, where the liquidus line C_1-C_2-C_3-C_4-E for the α phase intersects with the liquidus line C_5-E for the β phase, the melt is saturated with both components A and B at the same time. Consequently, the respective liquid phase precipitates α and β phase simultaneously. During this process, the temperature remains constant until the melt has completely transformed.

Point E is called **eutectic point,** because two solid phases, in this case α and β, crystallize concurrently from a "eutectic" melt and produce a close mixture with each other, a so-called **eutectic mixture**. This eutectic reaction, which takes place, can be written as follows:

melt → α + β

After solidification, an alloy of e.g. 40% B consists of α and β phase. The α phase exists as partly pure large crystals precipitated before the eutectic transition and partly as the close mixture with the β phase in the eutectic.

A melt with the composition between E and H (60 – 80% B) solidifies in the same way as the alloy with a composition between F and E, but with the difference that now β precipitates instead of α.

A melt with the eutectic composition has no solidification interval. It just exhibits a melting point and solidification point, analogue to the melt of a pure metal. The corresponding reaction from the solid state is called eutectoid reaction, which means that the transformation takes place concurrently from the solid state into two other solid phases.

1.2.3 Example

Fig. 5 and 6 show phase diagrams for the Cu-Sn system and Cu-Zn, respectively.
A magnified region of the Fe-C phase diagram can be seen in Fig. 7.

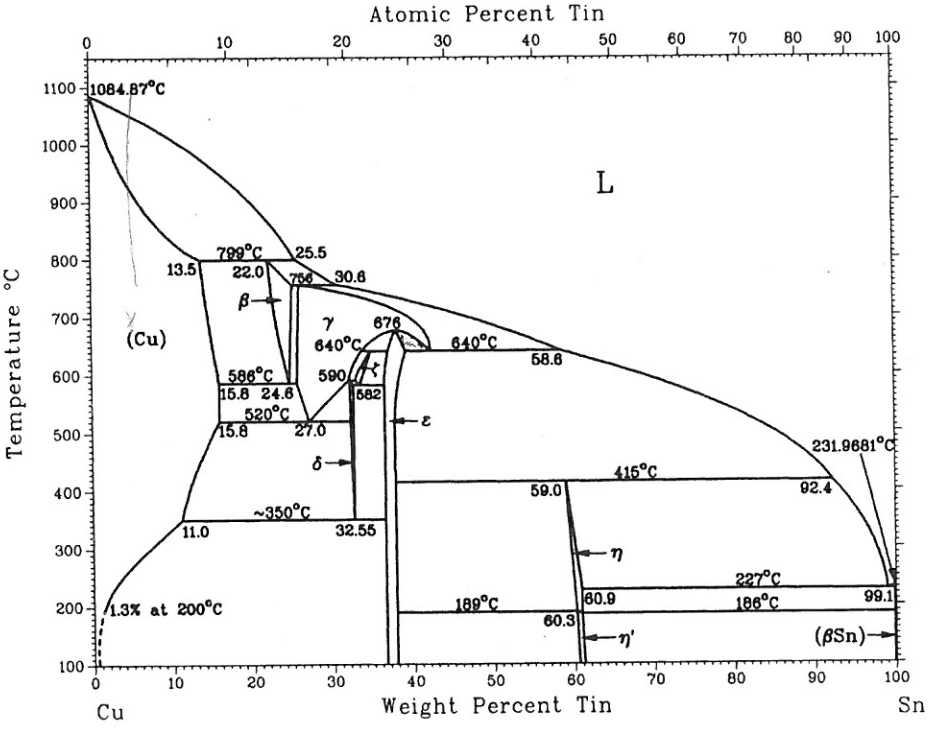


Figure 5: Phase diagram of Cu-Sn

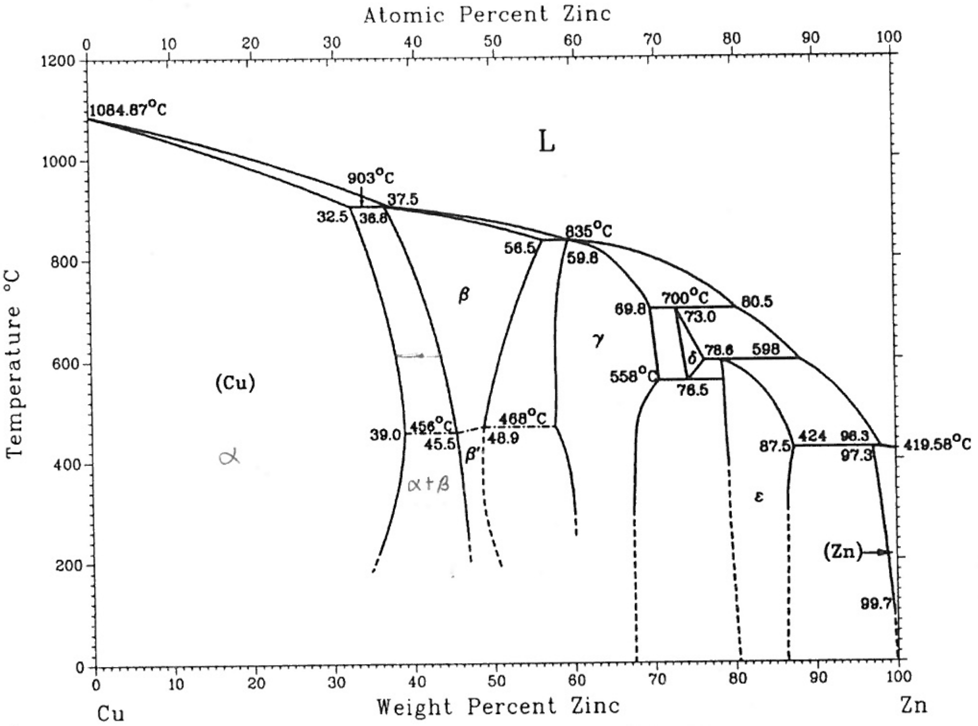
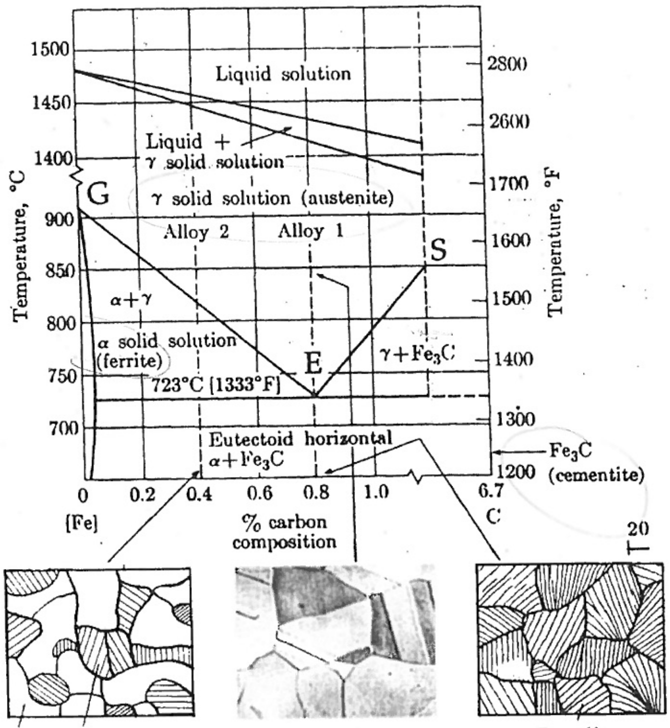


Figure 6: Phase diagram of Cu-Zn



Ferrite + Pearlite Austenite Pearlite

Figure 7: Section of the phase diagram of Fe-C

1.3 Brass – Transformation with solid solution

Cu-Zn ทองเหลือง

If Cu and Zn are alloyed to brass, several new structures in different compositional regions can occur (see Fig. 6). Some of these structures exist also in compositions, which contain small amounts of other elements e.g. Pb, Sn or Al. By varying the composition many properties can be changed and tuned, for instance colour, strength, hardness, ductility, resistance against corrosion or a combination of them.

One of the most useful (commercially) forms of brass contains about one Zn against one Cu atom. Let us consider brass, where Cu is alloyed with approximately 40% Zn (see Fig. 6). Above ~780°C the alloy consists of only one phase, the β phase, which exhibits a body-centred-cubic crystal structure. In the temperature region between ~460 and ~780°C the face-centred-cubic α phase is additionally formed.

Solid solutions and intermediate phases are temperature dependent (see Fig. 6). Consequently, transformations such as precipitations can take place. Precipitation means that during cooling down within a certain temperature interval a second phase can be formed within the crystals of the earlier formed phase. This kind of transformation often gives rise to a **Widmanstätten structure** (see Fig. 8), which is characterized by thin needles or lamellae of the second phase precipitated at or grown into the basis of the crystallographic plane system. Often, the Widmanstätten structure starts to precipitate at the grain boundaries and forms a network of rosettes or small dendritic-like grains along certain crystallographic orientations determined by the original (parent) phase.

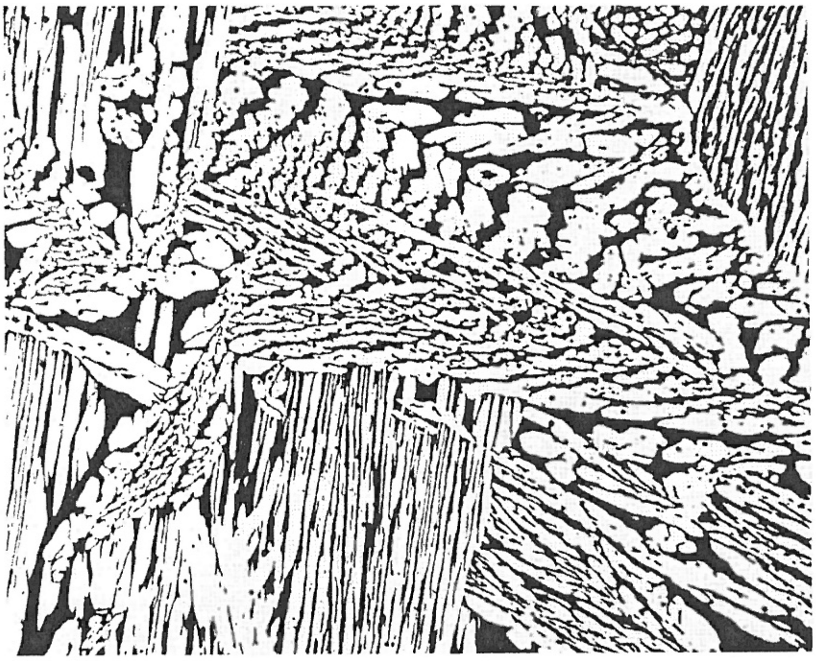


***Figure 8**: Microscopic image of a typical well-developed Widmanstätten-structure after casting. α (light) and β (dark) brass. 200x magnification.*

1.4 Pearlite – Phases and structural constituents

Consider steel with a composition of 0.8% C and the rest is Fe. Its state with different temperatures can be seen in the phase diagram in Fig. 7 along a vertical line through point E. Above 723°C this alloy consists of only one phase, **austenite**, which is a solid solution of C in Fe. Below 723°C a two phase region can be seen, ferrite and cementite. Upon cooling through point E, the austenite decomposes into the ferrite + cementite, the so-called **pearlite**.

The decomposition starts with the nucleation of a crystal of cementite. With its growth the C from the neighbouring material will be bound to the cementite resulting in a C-poor surrounding, the ferrite (see Fig. 9).

The cementite grows as thin plates. Around these the steel is deprived of its C i.e. ferrite is left. New lamellae of cementites always nucleate at a certain distance away from the preceding one. This is determined by the region of sufficient amount of C, so that pearlite (cementite + ferrite) exhibits a characteristic appearance of alternating lamellae of cementite and ferrite, whereas the thickness of the lamellae of cementite and ferrite, respectively, is constant (see Fig. 9). Please note that pearlite is not a single phase but a mixture of two phases (ferrite + cementite) with a regular arrangement and characteristic appearance. For such a mixture the collective name **structural constituent** can be applied.

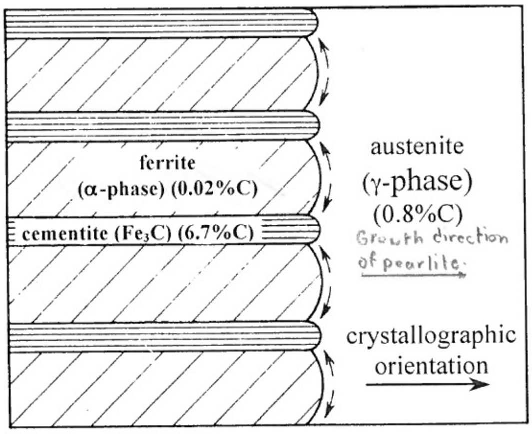


Figure 9: *Schematics of the development of cementite. First, nucleation of cementite crystal. Then, lamellae of ferrites nucleate where the C is deprived, and finally, the structural constituent pearlite has developed.*
The short arrows point in the direction of C diffusion at the solidification front.

A structural constituent is identified by a characteristic appearance. The concept comprises both, pure phases and mixtures of phases in a characteristic arrangement. A phase on the other hand, features a **uniform composition or structure with consistent properties**.

A steel at low temperature containing 0.8% of C consists of two phases (ferrite + cementite), but has only one **structural constituent** (pearlite). At higher temperatures, above

723°C, only one phase can be found, namely austenite, which at the same time is the only **structural constituent**.

1.5 Hypo- and hypereutectoidic steel

Point E is called **eutectoidic point**. Steel with the corresponding amount of C of 0.8% is a eutectoidic steel. With lower C content the steel becomes hypoeutectoidic (undereutectoidic) and hypereutectoidic (overeutectoidic) with a C content larger than 0.8%.

In the pictures in the lower part of Fig. 7 the microstructure at room temperature is shown of hypoeutectoidic (left), eutectoidic (middle) and hypereutectoidic (right) steel. With 0% C content only ferritic grains can be seen. Increased C content results in the nucleation and growth of pearlite grains. Reaching the eutectoidic composition (0.8% C), the sample consists completely of pearlite. Cementite starts to precipitate at the pearlite grain boundaries with C content above 0.8%. In the region above the line G-E-S almost the steel consists entirely of austenite.

One **grain** is a region with one crystallographic orientation of the lattice. Different grains (regions with different crystallographic orientations) are separated by grain boundaries (see Fig. 10). During etching, which is part of the sample preparation procedure (see section 3), the grain boundaries emerge due to the different etching rates with differently oriented material.

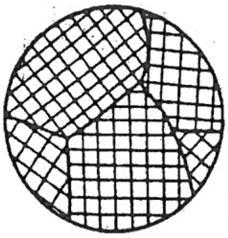


***Figure 10**: Grains with grain boundaries.*

2 Heat treatment of steel

2.1 Introduction

Due to heat treatment, the structure of steel can be drastically changed in order to bestow the steel new and favourable properties, e.g. high or low hardness, ductility, machinability, etc.

The most important heat treatment operations are:

Annealing
Hardening
Tempering

Annealing is often performed in order to soften the material and increase the ductility or to enhance the machinability especially with respect to the chipping performance during cutting. It can also be applied as part of the hardening process to transform the material into a different structure, which is necessary to harden the steel. Usually, annealing is finished by cooling down the material slowly.

Hardening often requires a heating of the material to a high temperature as the first step frequently accompanied by a structural transformation. The heating is followed by a rapid cooling (often quenching in water or oil) of the material to prevent the back transformation into the initial state structure or only allow a small amount of the material to transform. With large cooling rate, usually the material becomes very hard but also brittle.

A very brittle material can become more ductile during **tempering**. During this process, the material is moderately heated, which can allow diffusion and transformation to a certain extend. The brittleness decreases mostly at the expense of the hardness.

2.2 Martensite

To obtain martensite the steel has to be hardened. The material is heated up to a temperature, where the material (pearlite 0.8% C; or pearlite + cementite C > 0.8%; or pearlite + ferrite C < 0.8%) transforms totally or partially (for hypereutectoidic steel) into austenite. The austenite is able to dissolve large amounts of C in its fcc structure. With rapid cooling or quenching the C is forced to stay in the structure, due to the limited diffusion rate and the austenitic fcc structure transforms into a so-called body centred tetragonal crystal, where the C is "trapped" at energetically unfavourable places. That causes a high degree of distortion and large internal stresses resulting in a very hard but brittle material. The needle-shaped microstructure of the martensite can be seen in Fig. 11. Tempering to temperatures below 723°C (see Fig. 7) or slower cooling rates allow for some C diffusion and thus little transformation into cementite, which lowers the hardness and yields in a more ductile material.

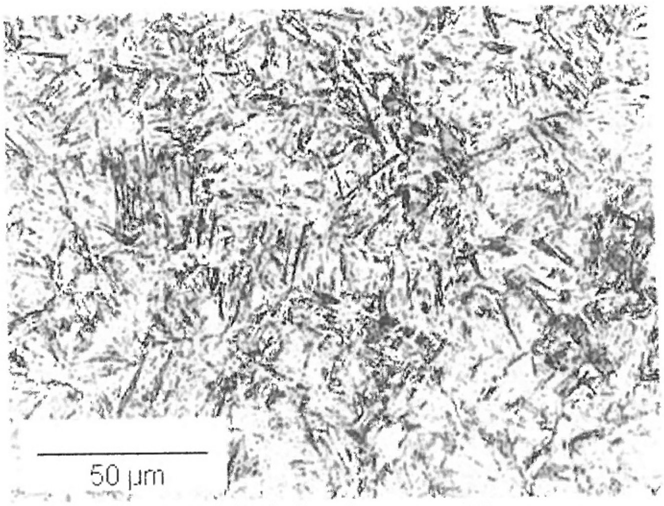


Figure 11*: Well-defined martensitic structure.*

3 Dislocations

No crystal is perfect. Every material contains defects, which can be point, line, or volume defects. These imperfections are responsible for local disturbances in the regular arrangement of the atoms (lattice). Defects can highly influence the properties of a crystalline material. The primary crystalline defects are **dislocations**, which are heavily studied nowadays. The distribution, arrangement and density of these line defects are subject of research to determine their characteristics and the influence they have on the materials properties.
Gliding is another type of plastic deformation, which can occur during deformation of a crystal. This can be observed when a part of a crystal slips or glides over the next crystallographic plane (so-called **slip or glide plane**) along a certain crystallographic orientation (**glide or slip direction**). A slip plane in combination with a slip direction in the same plane makes for a so-called **slip system**.

Some of the alkali halides, such as LiF, may be prepared in states of high purity and structural perfection. Suitable specimens may be obtained by cleavage rather than cutting, and the nature of the material lends itself to the ready detection of the presence of dislocation. Such halides are therefore very suitable for experimental studies on the properties of dislocations.

LiF forms an ionic crystal with the NaCl structure. LiF of high purity may readily be produced in single crystal form and accurately parallel faces may be prepared by cleavage on the {100} planes. The process of cleavage can cause so-called cleavage steps, which run continuously across the cleaved sample parallel to the cleaving direction. The latter can be seen in Fig. 12.

Two particular experimental techniques can be used to examine the deformation process in this crystal under the microscope. These are:

i) the **etch pit technique**, where dislocation sites are revealed on the crystals surface
ii) the optical technique of stress-birefringence for revealing the distribution of strain in the crystal.

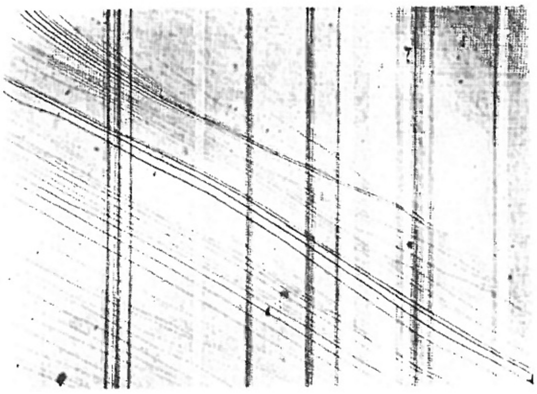


***Figure 12**: Cleavage steps (upper left corner to lower right corner) and slip steps (vertical) in a crystal*

The first technique depends on the energy of a crystals surface being higher at the point of emergence of a dislocation, because of the strain energy associated with the dislocation. It is thus possible to reveal the site of the dislocation by immersing the crystal in a solution in which the crystal is slightly soluble. With suitable solutions the region of the crystal very close to the dislocation is dissolved more rapidly than the remaining, more regular, surface. This leads to the production of dislocation etch pits on the surface; the size and shape of the pits depend upon the solution used, the time of etching, the type of crystal structure and the crystal plane which is etched.

The stress-birefringence technique relies on the fact that some transparent crystalline substances become doubly refracting when elastic strains are introduced. Under suitable illumination with polarized light, these strains are rendered visible as contrast effects. In this way, the slipped regions of a crystal, which are in a strained condition due to the elastic distortions associated with the dislocations, may be detected.

An etched surface can be seen in Fig. 13, where lines of closely spaced pits run in random directions, which link together to form a network. These lines delineate sub-grain boundaries. The growth process is not quite perfect and the material comprises a number of slightly misoriented regions or sub grains. The misorientation between sub grains (<<1°) is accommodated by a close array of dislocations which are revealed as a row of pits. The larger the misorientation between the sub grains the greater is the density of pits in the boundary. Other randomly distributed pits occur within the sub grains at single grown-in dislocations.
The second feature of the etched specimen is the presence of a general background attack (see Fig. 13), caused by the higher solution rate of small impurity precipitates.

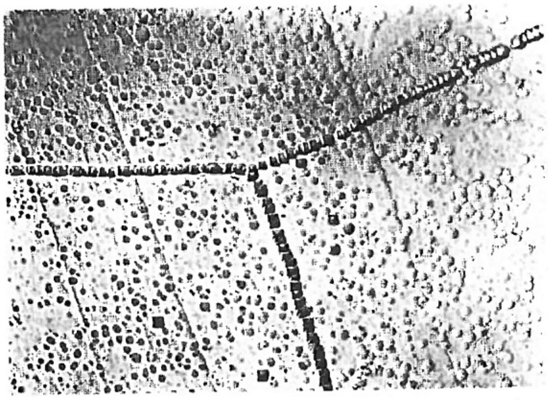


***Figure 13**: Dislocation etch pits along sub-grain boundaries, together with other small pits at precipitates. 500x.*

4 Sample preparation

For microscopic investigations the so-called inner surface has to be excavated. This can be done through either cutting or depth grinding. Otherwise, there is a risk that the visible structure is not representative for the entire material but only the prepared surface region. Also, grinding has to be done gently to prevent excessive heat in the material to prevent any structural change.

Subsequent to grinding, the surface is polished and etched. A non-etched but polished surface does not show much contrast and structure. However, pores and or inclusions of, e.g. slag, are best viewed without etching process.

During etching, various structural constituents are affected differently. The etchant nitric acid (HNO_3 in alcohol) leaves, for instance cementite, unaffected but acts on ferrite.

A pearlitic sample observed through the microscope shows a bright and dark pattern (see Fig. 14). With sufficiently high magnification both cementite and ferrite appear bright, but the boundaries between them can be seen as dark lines. The lamellae of the cementite are very thin, i.e. with lower magnification they just seem to be dark lines.

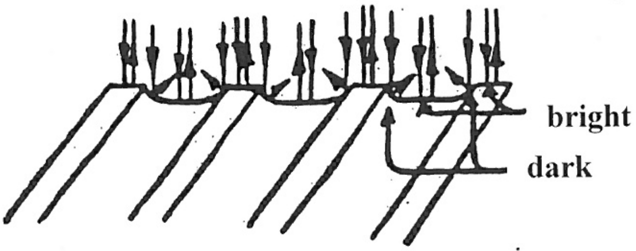


***Figure 14**: Reflection in the microscope of a pearlitic sample.*

Etching evokes grain structures even in single-phased samples or regions. Depending on the etchant, one can divide the etch process into **grain boundary etching** and **etching on the face of the grain**.

The latter etching process depends on the fact that the etchant affects metal surfaces with different lattice arrangements to a different degree. Surfaces with the highest atom density are etched slowest. The result is that certain etchants etch down the grain surface until as much surface parts resistant against etching as possible are facing the liquid (etchant). This can be schematically seen in Fig. 15.

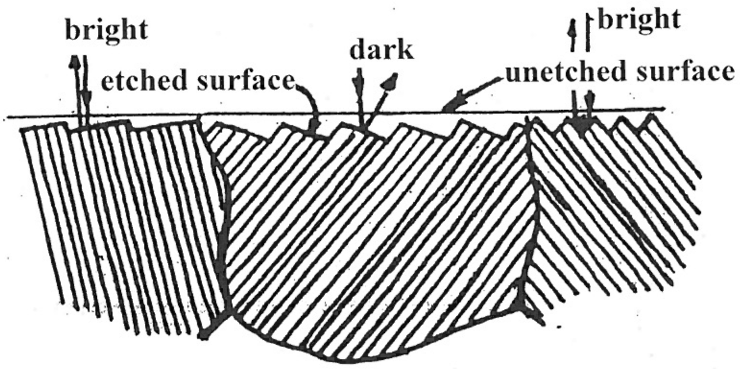


***Figure 15**: Reflection in the microscope at an etched surface (bright and dark contrast) in different grains.*

There are two causes to preferential etching at grain boundaries:

i) Due to different etch rates at various crystallographic orientations the grain surfaces are removed at different speed. Steps arise between grains with high etch rate and grains which are less affected. Small steps are seen in the microscope only as dark lines, whereas upper and lower edges are observed for higher steps, which are inclined. This is called double grain boundary (see Fig 16). For instance, pure iron is affected in this way by nitric acids.

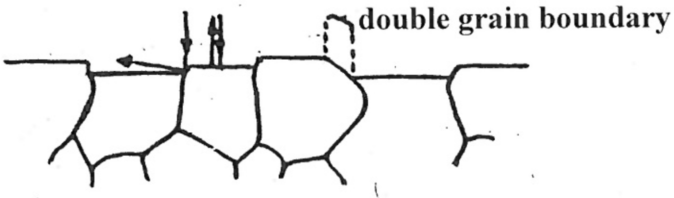


***Figure 16**: Schematics on the effect of different etching rates on different crystallographic orientations.*

ii) Grain boundaries, due to their high density of impurities and/or dislocations, may be etched away much faster than the grain surface. As a consequence grooves or trenches are formed in between the grains. These features are observed as dark lines as a result of their deficient reflection of the light (see Fig. 17).

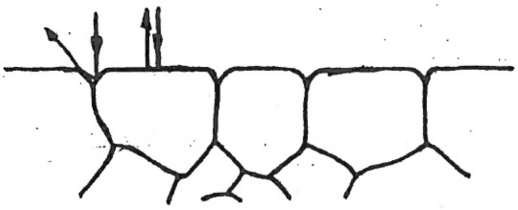


***Figure 17**: Schematics on the effect of different etching rates in the grain and grain boundaries. The kind of etching effect depends on the etchant as well as etching time.*

5 Microscopy

5.1 Practical rules

The microscope is a very expensive equipment. Even the easiest microscope for practise is a sophisticated optical apparatus and costs many ten thousands of SEK. Therefore, **handle the microscope with great care**.

Especially exposed to damage is the microscopes front lens or **objective**. This can be easily damaged by driving the sample into the lens during focusing. Always focus by lowering the objective lens and observing simultaneously from the side until the sample surface is close to the lens, after which the oculars should be used for the final focusing step and the sample should be moved upwards very carefully until the surface can be sharply seen. **Keep in mind in what direction the knob has to be turned to lower and raise the sample table.**

Before changing sample always lower the sample table to protect the objective lens from damage from the sample surface.

For certain highly magnifying objectives a drop of the so-called immersion oil has to be put between sample and lens. After usage, the oil has to be removed with dissolvent. Many objectives have been destroyed due to wrong cleaning agents. Use only benzene or xylene to clean the lens. Most of the other dissolvent damage the putty or glue, with which the objective lenses are mounted.

Be aware that an ordinary objective of a microscope costs easily as much as an expensive picture camera.

Please remember, the ocular does not become more transparent from the grease of the eyelashes. It is not necessary to pin the eye absolutely tight onto the ocular. Should the ocular be contaminated do not clean it with a handkerchief but only with pure lens tissue or Kleenex. Be aware that dust (or silicate particles in the paper) on the lens tissue scratches the anti-reflex film.

5.1 Theory

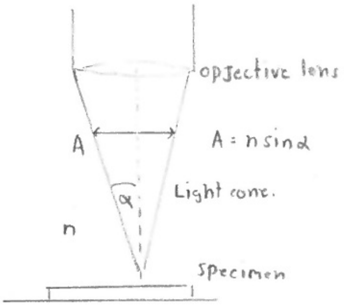
During the laboration a light microscope is used. The path of the light is shown in Fig. 18.

The **magnification of the microscope** is calculated by the product of the magnification of the objective and ocular lens and every other lens in the path of the rays. The magnification of the objective, however, depends on the distance to the ocular, the so-called tube length. Magnification and tube length for which it is valid are engraved at the side of the objective. An objective with a high magnification is corrected for a certain tube length and gives low quality pictures with the wrong tube length applied.

The **resolution of the microscope, d**, is defined as the smallest distance of two distinct points at which the two points are still resolvable. This depends on the **wavelength, λ**, used for illumination and the **numerical aperture, A**. The latter is defined as the product of **refractive index, n,** of the lens system and the **sinus of the opening angle** of the objective lens **sinα**, which leads to the following equation for the resolution:

$$d = \frac{1.22 \cdot \lambda}{2A} = \frac{1.22 \cdot \lambda}{2 \cdot n \cdot \sin\alpha}$$

This equation is the empirically found so-called Rayleigh criterion. The factor 1.22 in the equation is related to the ability of the human eye to distinguish two separate points. It becomes clear that resolution is very much limited by the applied wavelength. Therefore electron microscopes using much lower wavelengths (depends on the acceleration voltage of the electrons) have much higher resolution down to atomic resolution. Another way to improve the resolution is to apply immersion oils between the lens and the specimen. This increases n and thus enhances the resolution.



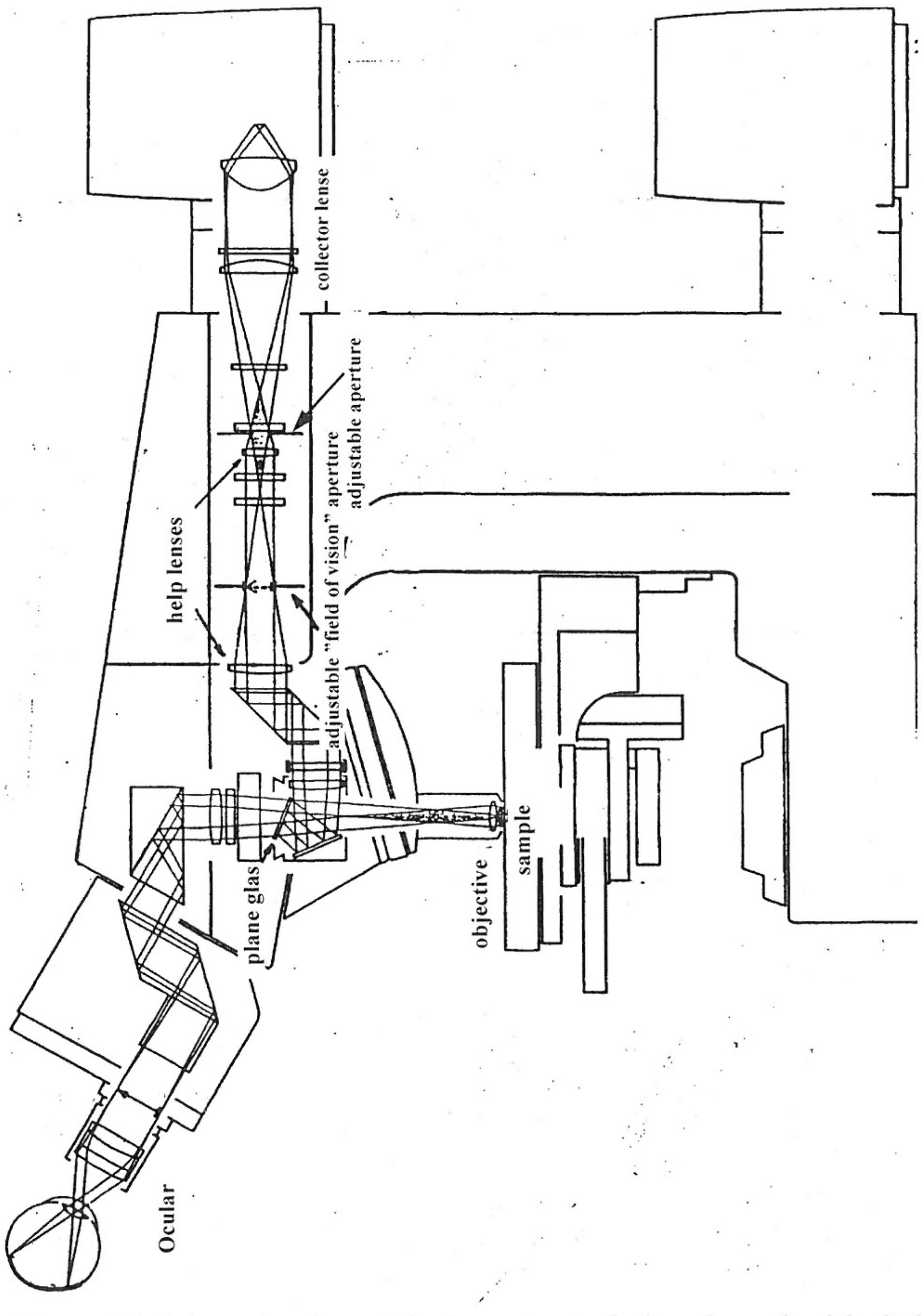
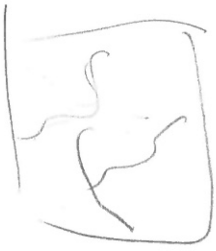


***Figure 18**: Schematic of a metal microscope including the path of the light.*

Laboration exercises

I. Solid solution structures



[X2]: Cu alloy with 4% Sn, sand-cast.

[X3]: As [X2], but annealed for 2 hours at 700°C

1) Draw the structural transformation in the phase diagram.
2) Although these specimens are examples of solid solution structures, a dendritic structure appears in the matrix of [X2]. Why?
3) How did the annealing treatment of [X3] affect the structure?
4) Dendritic shaped grains are very common due to one specific manufacturing process. What process?

II. Two-phase structures

[X8]: Cu alloy with 37% Zn. Sand cast.

1) Explain the basic difference between samples [X3] and [X8].
2) Compared to [X3], what will happen to the structure after heat treatment at a temperature below the solidus line?

III. Secondary phase precipitation: Widmanstätten structure

[X10]: $Cu_{58}Zn_{42}$. Sand cast.

[X11]: As [X10], but annealed for 1 hour at 800°C, followed by quenching in water.

[X12]: As [X10], but annealed for 1 hour at 800°C, cooled down in the oven until 600°C, and finally quenched in water.

[X13]: As [X12], but cooled in the oven until room temperature.

1) Explain how the heat treatment changes the microstructure's appearance.
2) How are the mechanical properties, such as hardness, ductility, brittleness, etc., affected by the different heat treatments?

IV. Eutectoid reactions

[X17]: Steel with 0.77% C. Annealed for 1 hour at 800°C. Cooled in furnace.

[X18]: As [X17], but cooled in air.

[X19]: Steel with 0.35% C. Furnace cooled from 870°C.

[X20]: Steel with 1.3% C. Furnace cooled from 970°C.

[X25]: Steel with 0.35% C. Quenched in water from 870°C.

[X26]: As [X25], followed by annealing for 2 hours at 600°C.

1) Identify the different structures in each sample.
2) Why are the structures different of samples [X17] and [18]?
3) Why can you see (with your naked eyes) two ring-shaped areas in sample [X25]?
4) Explain how the structure of sample [X25] transforms to [X26] during annealing. Why is annealing of martensite such an important procedure?

$E \propto |\vec{b}|^2$

V. Dislocation in LiF

- ❖ The LiF samples are very brittle, so handle them with care.
- ❖ Use tweezers to manipulate the samples.
- ❖ Place the samples on a Si wafer in order to get a better image.
- ❖ Focus on the specimen's edge, and on the upper surface.

Sample 1: Non-etched specimen

Identify the cleavage steps and explain how they originate

Sample 2: Etched specimen

Identify sub-grain boundaries and small etch pits on the surface, and explain how they originate.

Sample 3: Micro-indented and etched specimen

Identify different dislocation types in the <110> and <100> crystallographic orientation