Chapter

VII

Types of diffusional transformations

VII-1. Introduction

One of the solid-state phase transformations that is of particular interest to researchers today is the precipitation reaction. This reaction occurs in certain materials during heat treatment (such as tempering or aging) and imparts the material with its working properties.

Precipitation processes from supersaturated solid solutions, which involve diffusion phenomena, are generally classified into two main categories: discontinuous and continuous precipitation. In practice, homogeneous precipitation is rare; nucleation almost always begins on heterogeneities in the initial phase, which lower the free enthalpy of nucleation by reducing either surface energy, strain energy, or both. These heterogeneities can include impurities, inclusions, dislocations, surfaces, or grain boundaries. It is well established that the presence of lattice defects promotes nucleation, even under conditions where it would be unlikely to occur without these defects. Consequently, precipitate crystallites often form on these defects before those resulting from homogeneous precipitation. During discontinuous precipitation, the initial supersaturated phase, which maintains a constant solute concentration, coexists with the precipitation products. These products correspond to equilibrium phases with compositions determined by equilibrium at the given temperature. Thus, a concentration discontinuity exists at the interfaces between regions that have undergone the decomposition reaction and those that have not yet undergone it.

VII-2- Definition of precipitation

The precipitation of a supersaturated solid solution is a solid-state transformation that results in an increase in the number of phases. It generally involves the decomposition of the supersaturated solid solution (the initial matrix) into two new phases:

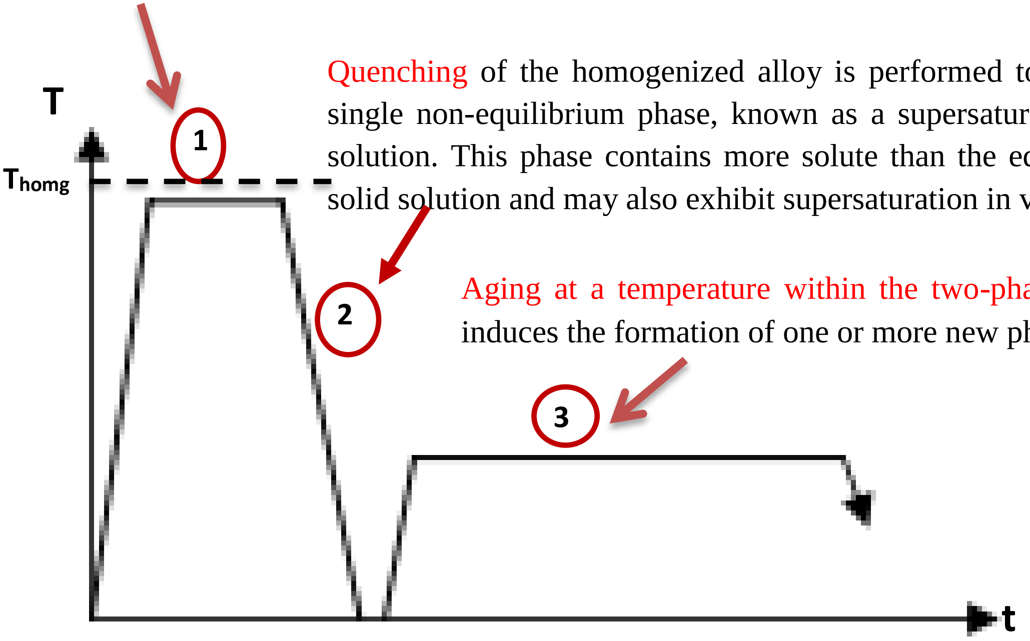
- The depleted matrix
- The precipitated phase

It follows that an alloy, which may be single-phase at a certain temperature, can become two-phase at lower temperatures. The supersaturated homogeneous alloy is relatively soft, with low hardness and elastic limit. However, the precipitation of a second phase induces structural hardening, which increases both the hardness and the elastic limit of the material.

To initiate a precipitation reaction, the following three steps must be carried out:

1. **Homogenization Treatment**: This step eliminates any segregation within the alloy and achieves a homogeneous single-phase system. The homogenization temperature is selected to be above the solubility limit, within the single-phase range.
2. **Quenching**: The homogenized alloy is rapidly cooled to obtain a single non-equilibrium phase known as a supersaturated solid solution. This solution contains more solute than the equilibrium solid solution at this temperature and may also experience additional supersaturation in vacancies.
3. **Aging**: The alloy is then heated to a temperature within the two-phase range, causing the formation of one or more new phases.

Homogenization treatment is used to eliminate any segregation in the alloy and achieve a homogeneous single-phase system. This is done at a temperature above the solubility limit, within the single-phase domain.



Quenching of the homogenized alloy is performed to obtain a single non-equilibrium phase, known as a supersaturated solid solution. This phase contains more solute than the equilibrium solid solution and may also exhibit supersaturation in vacancies.

Aging at a temperature within the two-phase range induces the formation of one or more new phases.

Fig. VII-1- : Schematic representation of the precipitation steps

There are generally two types of precipitation, as indicated in the introduction:

- **Continuous Precipitation**: In this process, the supersaturated solid solution depletes in a theoretically homogeneous and continuous manner throughout the matrix. The lattice parameter evolves continuously. Occasionally, continuous precipitation occurs predominantly at grain boundaries and slip planes, in which case it is referred to as localized or heterogeneous precipitation.

- **Discontinuous Precipitation**: This type of precipitation is characterized by the formation of two distinct regions during the transformation:
    - One region where the transformation is complete.
    - Another region where the solid solution remains supersaturated.

The transformation typically starts at the grain boundaries, leading to the development of cells—hence the term cellular precipitation. During this process, the lattice parameter changes discontinuously, taking on two distinct values.

VII-3. Characteristics of Precipitation

VII-3-1. Criteria for composition and sequence of heat treatments for alloys

Consider a binary aluminum-element (Al-M) alloy as illustrated in Figure VII-2-. For an alloy with a concentration X_1 (where X_1 is lower than the solubility limit X_S of the solute element M), and at a temperature T_1 that is higher than the solidus temperature T_S, the alloy should be maintained at this temperature for a sufficient duration to achieve thermodynamic equilibrium. This will ensure the complete dissolution of the solute atoms M.

When the alloy is cooled very slowly from temperature T_1 to a temperature below T_S, the solid solution of M in aluminum, which retains its face-centered cubic (FCC) structure (denoted as α), tends to decompose. This decomposition results in the formation of relatively coarse particles of the stable intermetallic compound Al_xM_y within a solid solution of aluminum that is now poor in M atoms. This equilibrium precipitation leads to a structure similar to that obtained during annealing. However, this structure is generally of limited interest for alloy hardening due to the large size of the precipitates (on the micrometer scale) and their relatively low number per unit volume of the alloy. The molar fraction f_m of the precipitates follows the lever rule or the rule of inverse segments.

$$f_m = \frac{BA}{BC} \qquad \text{(VII-1-)}$$

Where BA and BC are defined in figure VII-2-, at temperature T_2.

- The quenching operation involves rapidly cooling the alloy after dissolution to prevent the decomposition of the solid solution and the formation of coarse equilibrium precipitates. Quenching also traps the numerous vacancies that are stable at high temperatures. Upon quenching to room temperature T_2 (see Figure VII-2-), a supersaturated solid solution is obtained, where the solute atoms M and vacancies are randomly distributed within the face-centered cubic (FCC) lattice of aluminum. This results in a supersaturated solid solution, as illustrated in Figure VII-3-a-.

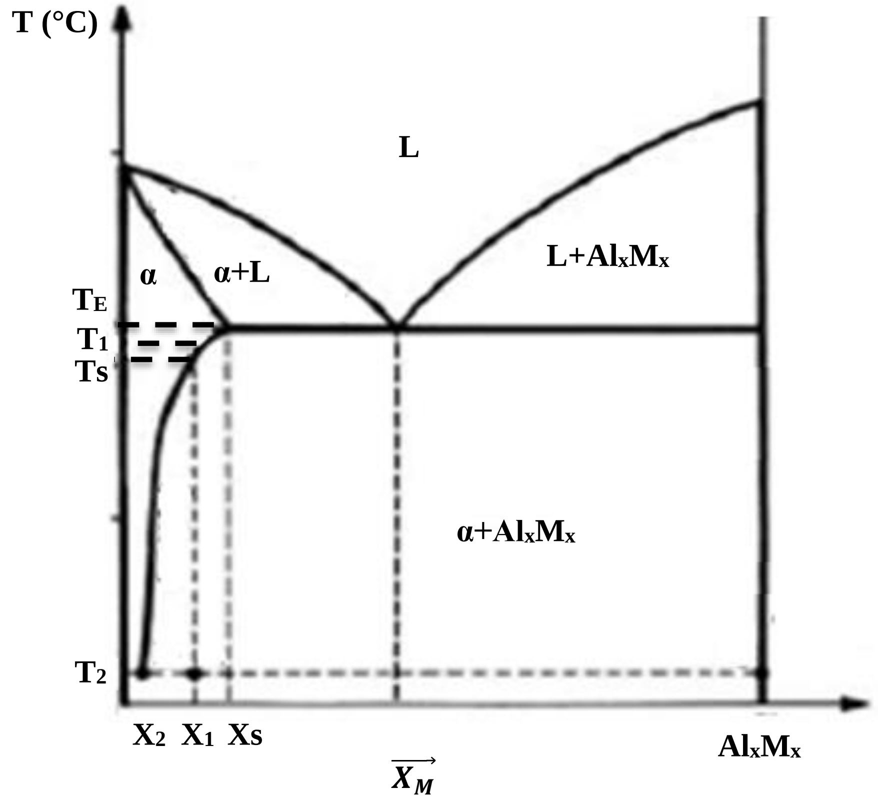
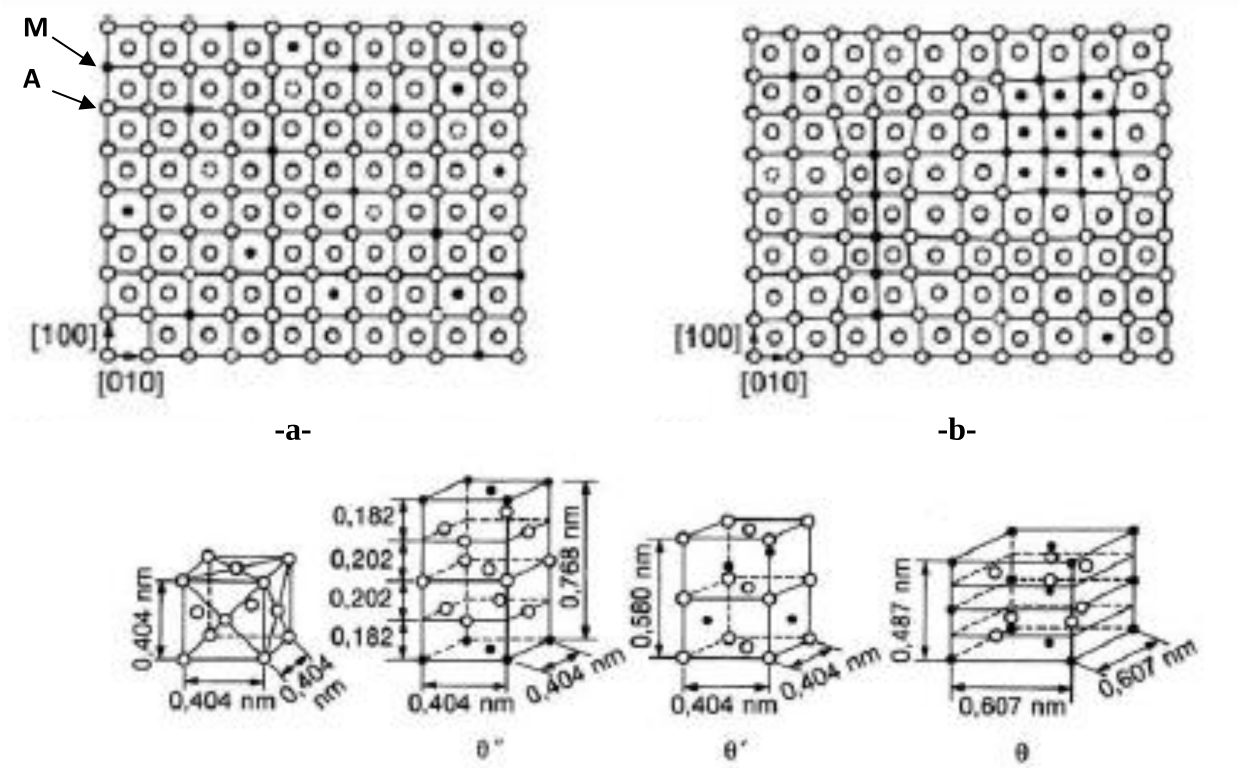


Fig. VII-2- : Principle of heat treatment for hardening aluminum alloys through precipitation in a binary phase diagram.

After dissolution and quenching, hardening through precipitation results from the decomposition of the supersaturated solid solution into its main solute elements. During aging at room temperature or at a low temperature following quenching, the supersaturated solid solution is metastable. This phase, known as natural aging or maturation, involves the gradual and spontaneous hardening of the alloy. The solid solution decomposes to form small clusters of solute atoms, known as *Guinier-Preston (GP) zones*, which occur at a very high density. This process represents the pre-precipitation stage.

These clusters of atoms are always located at the nodes of the aluminum matrix lattice and consist of small segments of crystallographic planes within the aluminum lattice, enriched in solute atoms (see Figure VII-3-b-). The presence of these Guinier-Preston (GP) zones introduces elastic distortions into the matrix lattice. The localized chemical enrichment and the resulting elastic deformations can enhance the precipitation process. In certain cases, structural defects such as dislocations can facilitate this precipitation by promoting higher diffusion rates of solute atoms or accommodating the associated elastic deformations.



-a-

-b-

-c-

Fig. VII-3- : Different steps of the precipitation sequence of the Al-Cu system (*B. DUBOST et al, Techniques de l'Ingénieur, traité Matériaux métallique, la fiche M240, 1991)*.
(a)- α-substitutional solid solution.n
(b) GP zones formation
(c) Lattice structures of the metastable intermetallic compounds θ'' and θ', and the equilibrium intermetallic compound θ (Al_2Cu).

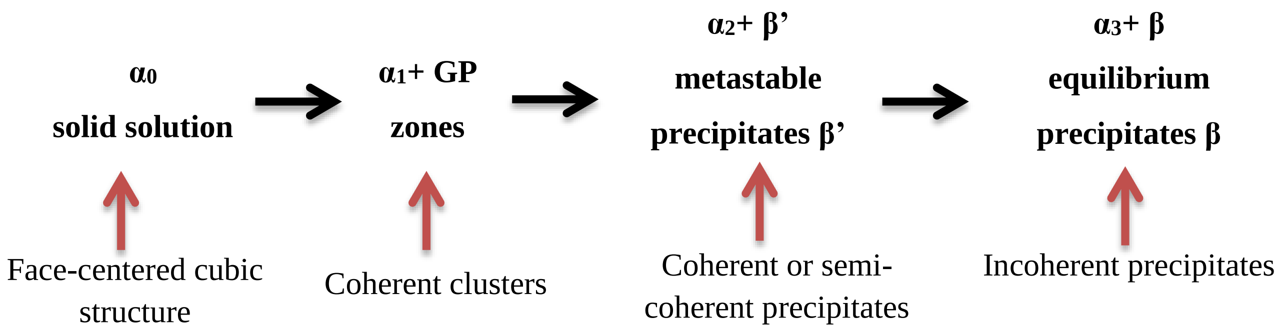
Subsequent heating at a moderate temperature (around 100°C), known as tempering, enhances the decomposition of the solid solution. This process involves the dissolution of the previously formed Guinier-Preston (GP) zones and the precipitation of a large number of very fine particles (ranging from 10^{16} to $10^{17}/cm^3$ with an average size between 2 and 50 nm). These precipitates are metastable intermetallic compounds, often referred to as transition precipitates.

They possess their own distinct crystal lattice and chemical composition, as illustrated in Figure VII-3-c-. This stage further hardens the alloy, reaching its maximum mechanical strength.

The precipitation of the equilibrium intermetallic compound in a matrix that is low in solute typically occurs only during an over-tempering treatment. This treatment is conducted at higher temperatures, generally above 160°C, and for a duration longer than that required to achieve the relative maximum hardening. During this stage, the alloy experiences relative softening due to the coalescence of precipitates. The largest of these precipitates, which can reach sizes on the micrometer scale, become visible under optical microscopy at high magnification. The microstructure thus tends to resemble that of the annealed state, which is the most thermodynamically stable. This annealed state is typically achieved after several hours at temperatures between 250 and 400°C.

VII-3-2. Precipitation sequence in quenched aluminum alloys

For thermodynamic and kinetic reasons, the quenching of a supersaturated solid solution almost never leads directly to the precipitation of the equilibrium intermetallic phase. Instead, the hardening of alloys at room temperature and during subsequent tempering generally results from a sequential decomposition of the α-solid solution and a progressive depletion of solute. This process typically follows the schematic steps:



For significant structural hardening, it is crucial to achieve a high density and a homogeneous distribution of clusters and/or metastable precipitates within the matrix.

VII-4. Types of precipitation in metal alloys

VII-4-1. Continuous precipitation

This reaction involves the diffusion of atoms over large distances from the initial phase to the precipitate nuclei, resulting in a gradual modification of the solute composition towards its equilibrium value. An experimental characteristic of this transformation is the continuous variation in the crystal lattice parameter of the matrix until it reaches a limiting value corresponding to the equilibrium state. Precipitation begins with the formation of nuclei at the expense of the Guinier-Preston (G.P) zones, which eventually disappear and give way to metastable or stable phases. The transition from the initial state to the final state typically involves a series of intermediate stages, as outlined in the previous discussion on aluminum alloys.

- **Guinier-Preston zones (G.P)**

In 1938, Guinier and Preston proposed the existence of zones enriched with dissolved atoms that form small domains coherent with the matrix. These zones, known as Guinier-Preston (G.P.) zones, are responsible for the initial stage of hardening.

In contrast to the precipitation mechanisms of metastable and stable phases, which typically occur on structural heterogeneities, the formation of Guinier-Preston (G.P.) zones happens homogeneously. This mechanism has been extensively studied in aluminum-based alloys, where it has been established that G.P. zones form at sufficiently low temperatures. Vacancies play a crucial role in this process. After quenching, excess vacancies, which are present at low temperatures and before reaching their equilibrium limit, facilitate the diffusion of solute atoms. These atoms then cluster together rapidly, forming coherent zones with a distinct shape and structure.

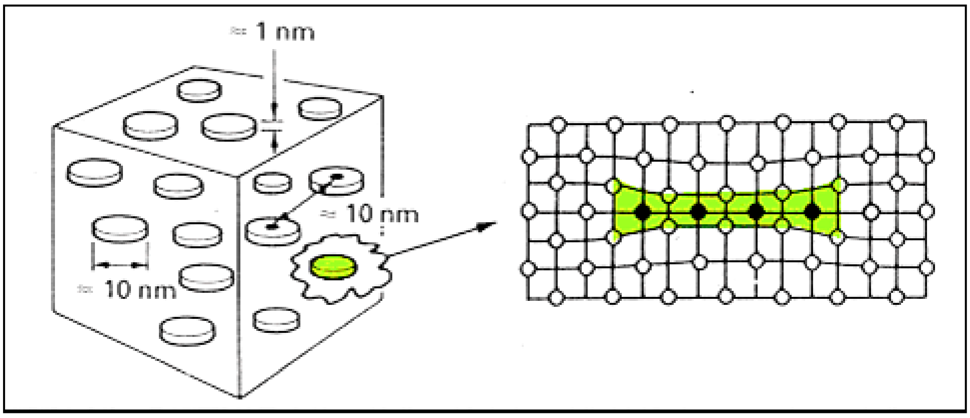


Fig. VII-4- : Guinier-Preston zones (G.P)

Due to their size, these precipitates create localized deformation fields (see Figure VII-5-). Dislocations are either attracted to or repelled by these particles, leading to their bending around the precipitates. To allow dislocation movement, a significant stress must be applied to overcome these elastic interactions. As the dislocation moves through or around a precipitate, it shears the particle, creating an interface of width b (see Figure VII-5-a-). The work required to move the dislocation is equivalent to the change in cohesive forces at this interface. This results in a high yield stress, which can increase with the particle diameter up to a critical size. Beyond this critical diameter, the Orowan mechanism becomes relevant, where circumventing the particle becomes easier than cutting through it (see Figure VII-5-b-).

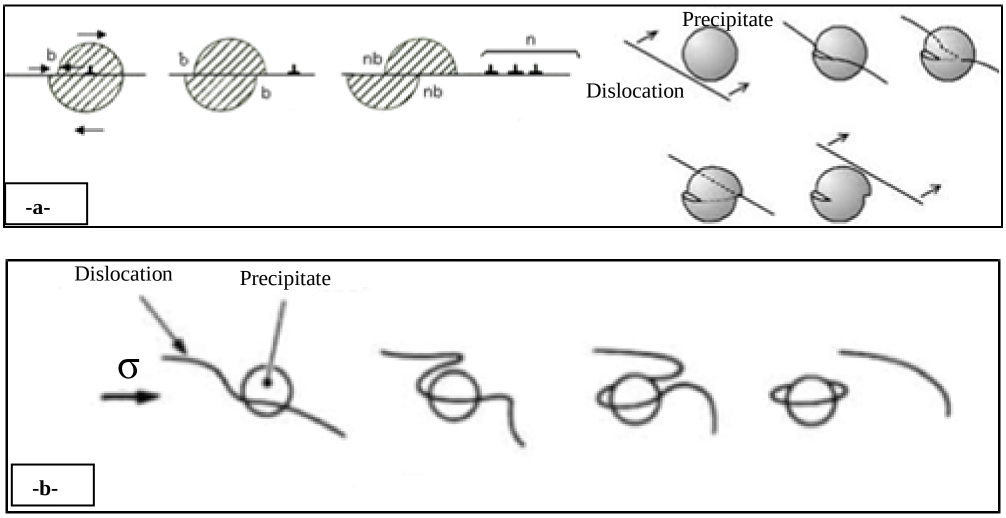


Fig. VII-5- : (a) Shearing of a coherent precipitate.
(b) Dislocation line bends around the precipitate: (Orowan mechanism)

VII-4-2. Discontinuous precipitation

Discontinuous precipitation is a type of discontinuous reaction involving two migrating phase boundaries. This heterogeneous solid-state reaction is diffusion-controlled at the boundaries, which serve as transport pathways for the solute. It typically occurs preferentially at regions with significant crystallographic misorientation. During this process, the supersaturated solid solution α' is replaced by a cellular structure composed of two phases, α and β.

$$\alpha' \longrightarrow \alpha + \beta$$

Where: α': The supersaturated matrix.

α: the same phase as α' but with a different concentration

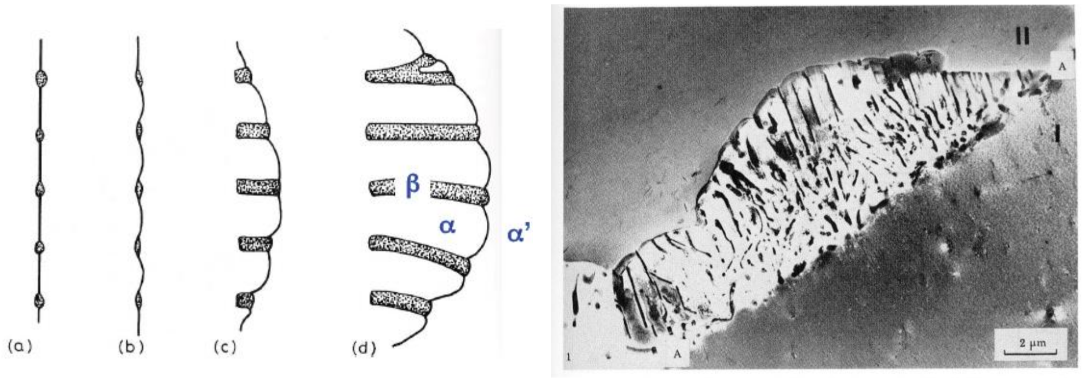
β: equilibrium precipitate (of the second phase).

Since Agreew et al. first discovered discontinuous precipitation in Ag-Cu alloys in 1930, our understanding of this reaction has significantly advanced. However, many important questions remain unresolved. It is still unclear which specific mechanism controls the process in a given alloy or why discontinuous precipitation occurs in some alloys and not in others. Additionally, establishing the criteria for this solid-state reaction remains challenging due to the multiple factors that can influence its occurrence.

✓ Different types of discontinuous precipitation

❖ Interfacial precipitation

The precipitation product features a cellular structure with alternating lamellae behind a reaction front that resembles an incoherent grain boundary (see Figure VII-6-). During isothermal annealing, the interlamellar distance generally remains constant. However, Gust et al. have demonstrated that in certain areas, the interlamellar distance of precipitates formed isothermally can vary and follows a Gaussian distribution. In this type of reaction, material transport is facilitated by interfacial diffusion along the reaction front, which is a grain boundary with significant misorientation.



-a-

-b-

Fig. VII-6-: (a)- Schematic representation of an interfacial type precipitation reaction

(b)- Cellular precipitation of $Mg_{17}Al_{12}$ in the Mg-9 at. % Al alloy annealed for 1 h at 220°C *(D. A. Porter, K. E. Eastorting. Phase transformation in metals end alloys, second edition, Chapman et Hall, London, (1992)).*

Dislocations play a significant role in the germination and growth of precipitates, as they create preferential regions for precipitation. This second type of precipitation differs from the first type in the following ways:

- The precipitated particles are not lamellar but consist of small platelets with a well-defined crystallographic structure.
- The volume diffusion of the solute element is involved.

Precipitation on dislocations occurs in systems where the specific volumes of the initial matrix and the reaction product are very different. It is assumed that, during the precipitation of the β-phase, the dislocations act as new nucleation sites. This type of precipitation on dislocations is not very common. However, in several solid solutions, both types of precipitation can be observed simultaneously.

❖ Precipitation at interphase boundaries

The issue of precipitation at interphase boundaries has been extensively studied theoretically. It is widely accepted that an immobile interphase boundary can provide favorable sites for the nucleation of a new phase, particularly at low temperatures.

In copper-based alloys, only two studies have focused on the investigation of discontinuous precipitation at interphase boundaries. Manna et al. were the first to explore the possibility of inducing discontinuous precipitation from interphase boundaries, similar to that observed at grain boundaries.

In their studies on a Cu-15% mass In alloy, Hamana and Boumerzoug demonstrated that the eutectoid phase can be obtained. Subsequent annealing at 400°C can lead to discontinuous precipitation at the eutectoid/matrix interphase boundaries, similar to what is observed at grain boundaries.

VII-5. Precipitation mechanisms

VII-5-1. Precipitation by germination and growth

According to the classical theory of Gibbs, extended to phase transformations in the solid state by Turnbull and Fischer, germination—the initial stage of precipitation—involves the formation of stable nuclei from the matrix. These nuclei are rich in solute, have a composition significantly different from that of the solid solution, and exhibit a defined interface with the matrix. Germs can arise from statistical composition fluctuations of sufficient magnitude (on the order of a nanometer) within the solid solution, with very rapid formation kinetics occurring in the presence of vacancies and supersaturation (homogeneous precipitation). Alternatively, they can preferentially form on lattice defects or pre-existing interfaces between different grains or particles (heterogeneous precipitation). The subsequent growth of these germs and precipitates is governed by the diffusion of solute atoms towards them. This diffusion is thermally activated at the tempering temperature, as long as the solid solution remains supersaturated. The proposed mechanisms for precipitation consider both thermodynamic factors and kinetic effects and apply to most precipitation-hardened aluminum alloys.

❖ Thermodynamics of germination

Thermodynamically, germination is only possible after dissolution and quenching if it results in a reduction in the overall free energy of the system, which initially consists of the homogeneous solid solution. This quantity, the Gibbs free energy, first includes the chemical term of the molar free energy associated with germ formation, followed by thermodynamic barriers that oppose the formation of these germs. Figure VII-7 illustrates the free energy curves for the formation of different phases as a function of the concentration X of the added element at the high temperature (T_H) and the low tempering temperature (T_B). In this figure, the phase equilibrium diagram of the system is derived from the minimization of its free energy across the entire range of considered temperatures.

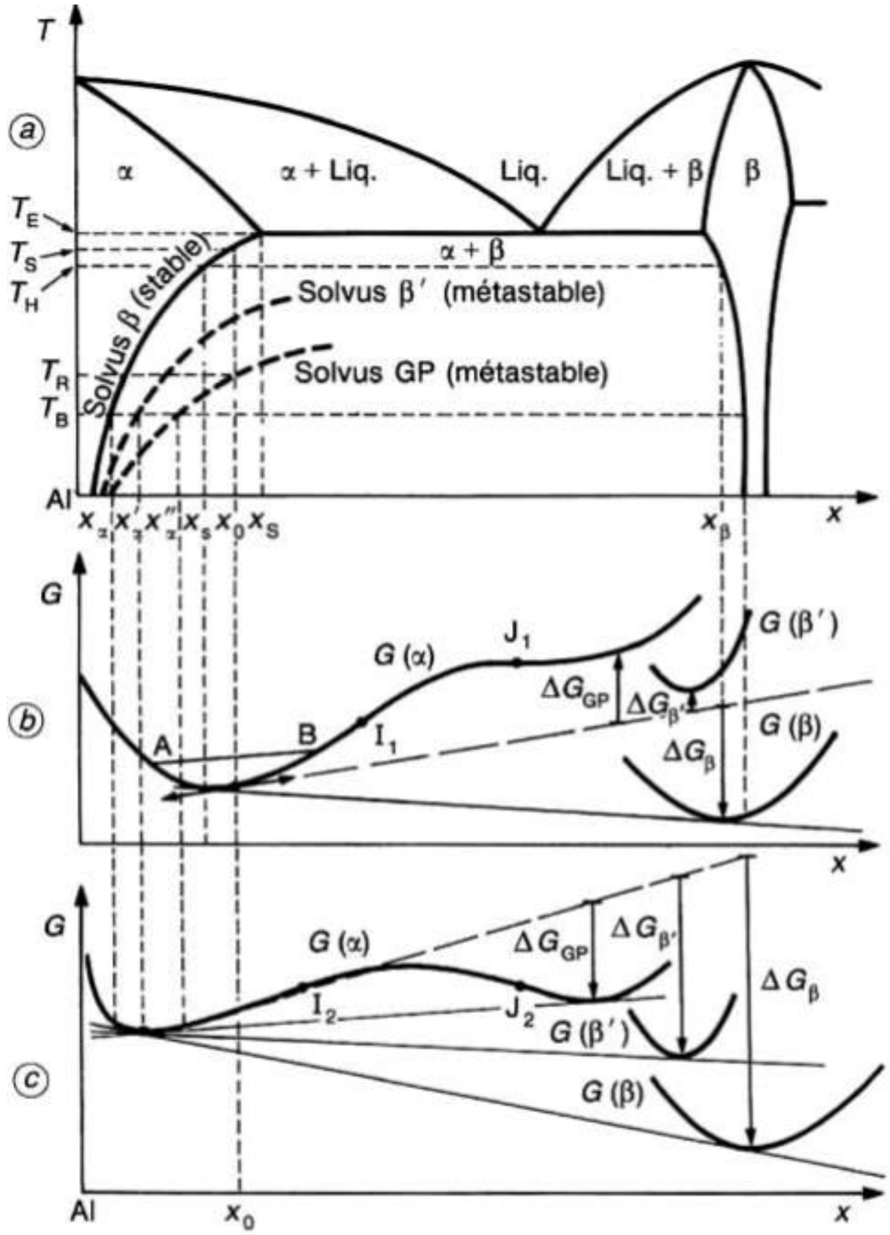
- The curve G(α) shows the variation in the molar Gibbs free energy of the solid solution α\alphaα as a function of solute concentration, as well as that of the GP zones, since the latter share the same crystalline structure as the matrix. This curve exhibits double concavity, with an absolute minimum near the solute concentration at equilibrium X_s (at low values of X) and a relative minimum close to the average solute concentration around the GP zones (at higher values of X).
- G(β′) and G(β) represent the respective free energy curves for the metastable phase β′ (transition precipitate) and the equilibrium phase β (stable phase) as a function of the concentration X of the added element M.

Any mode of decomposition of low amplitude into two phases represented by A and B (at temperature T_H) is thermodynamically impossible within the domain where the curvature of G(x) is positive (Fig. VII-7-b). This is because such decomposition would result in an increase in the system's free energy. The only thermodynamically favorable decomposition of the initial solid solution that decreases the system's free energy is demixing into a highly dilute solid solution (close to the equilibrium concentration) and clusters rich in solute atoms. These clusters can consist of GP zones (with a concentration beyond the second inflection point J_1 or J_2 on the G(α) curve, in the positively curved region) or precipitates of the metastable β′ phase or stable β phase with lower molar free energy. The compositions of the solid solution and the associated GP zones or precipitates are given by the abscissa of the contact points of the tangent common to the two positively curved regions of the G(x) curve (Figs. VII-7-b and VII-7-c).

VII-5-2. Precipitation by Spinodal decomposition

This mode of precipitation, with theory established by Cahn, typically affects alloys with a high volume fraction of precipitates (e.g., copper alloys) but can also explain the structural hardening of certain aluminum alloys in the Al-Zn and Al-Ag systems. On the Gibbs free energy molar curve shown at low temperature in Figure VII-8, the concentration lies between the inflection points I and J (with the curve connecting I to J being the Spinodal curve). Spinodal decomposition is then possible within the domain where the curve G(α) is convex (with negative curvature) and for any temperature where $\partial^2 G(\alpha)/\partial x^2 < 0$.

A continuous and progressive decrease in the chemical free energy of the system, which is thermodynamically favorable, is achieved by the demixing of the initial solid solution into two components: one increasingly poor in solute and the other increasingly rich in solute. The respective concentrations of these phases are given by the abscissas of segment MP, based on the portion of the G(α) curve with negative curvature. Precipitation, which initiates from random fluctuations in the concentration of the solid solution, occurs throughout the grain. Unlike precipitation by germination and growth, there is no energy barrier for nucleation. Initially, the differences in composition are very small, but as the decomposition progresses, these differences increase. The solute migrates towards the precipitates by short-distance diffusion, moving up the concentration gradient until the final composition of the precipitate X_p is reached (Fig. VII-9). This results in a microstructure characterized by an apparently periodic modulation of the distribution of precipitates within the matrix (homogeneous precipitation).



(a) Diagram of equilibrium phases and metastable phases
(b) Molar free energy G at a high temperature T_H
(c) Molar free energy G at low temperature T_B

Fig.VII -7- : Schema for Constructing the Phase Diagram of the Al-M System *(B. Dubost et al, Techniques de l'Ingénieur, traité Matériaux métallique, la fiche M240, 1991).*

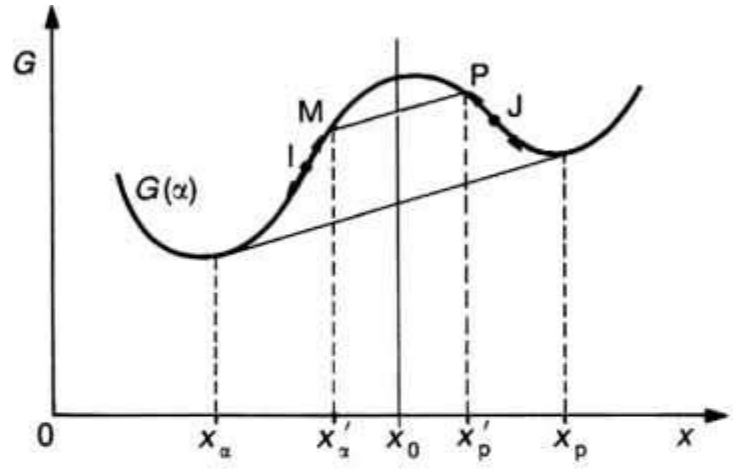


Fig. VII-8- : Free energy curve G at low temperature of an alloy with concentration x_0 *(B. Dubost et al, Techniques de l'Ingénieur, traité Matériaux métallique, la fiche M240, 1991).*

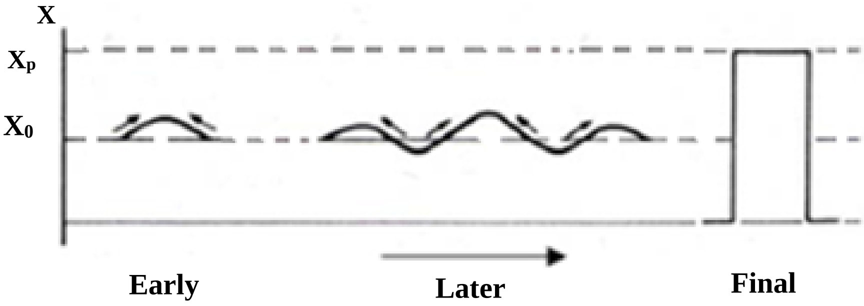


Fig. VII-9- : Evolution of the Solute Concentration Profile xxx in an Alloy Precipitated by Spinodal Decomposition, Depending on the Maturation Stage.

VII-6. Experimental methods for studying structural precipitation

VII-6-1. Metallography

Optical microscopy, due to its limited resolving power, is primarily used to visualize coarse precipitates on samples that have been polished and, if necessary, treated with specific solution reagents (Fig. VII-10). The images obtained reveal attack figures that provide valuable information about the distribution of the coarser phases, such as phases that are insoluble (i.e., phases that do not dissolve during homogenization treatments following alloy fabrication) and precipitates located at grain boundaries.

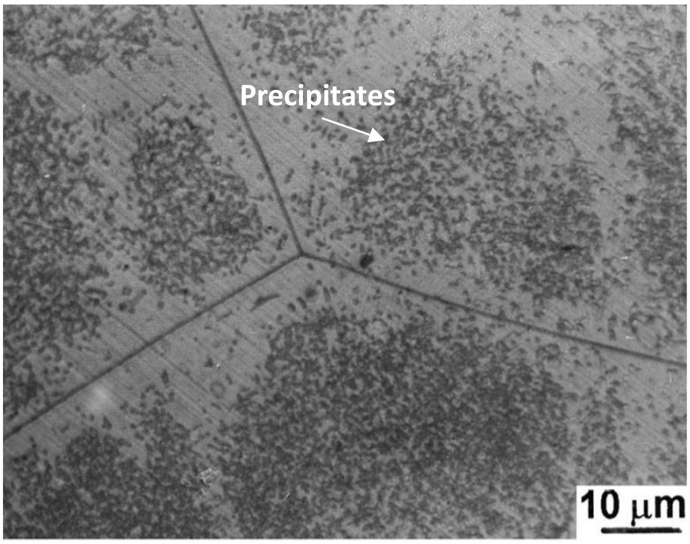


Fig. VII-10-: Microstructure of Al–12 wt.% Mg alloy homogenised 17 h at 430°C, quenched then aged during continuous heating at a rate of 2°C/min to 280°C *(D. Hamana et al., Journal of Alloys and Compounds 320 (2001) 93–102)*

Scanning electron microscopy on polished sections, utilizing the chemical contrast from differences in atomic number seen in backscattered electron imaging, provides similar information with higher spatial resolution while avoiding the need for chemical attack. These techniques can be effectively combined with devices for analyzing image contrasts and morphologies. This allows for quantitative and objective determination of the volume fractions of precipitates, as well as their geometric parameters such as size, shape, and distribution.

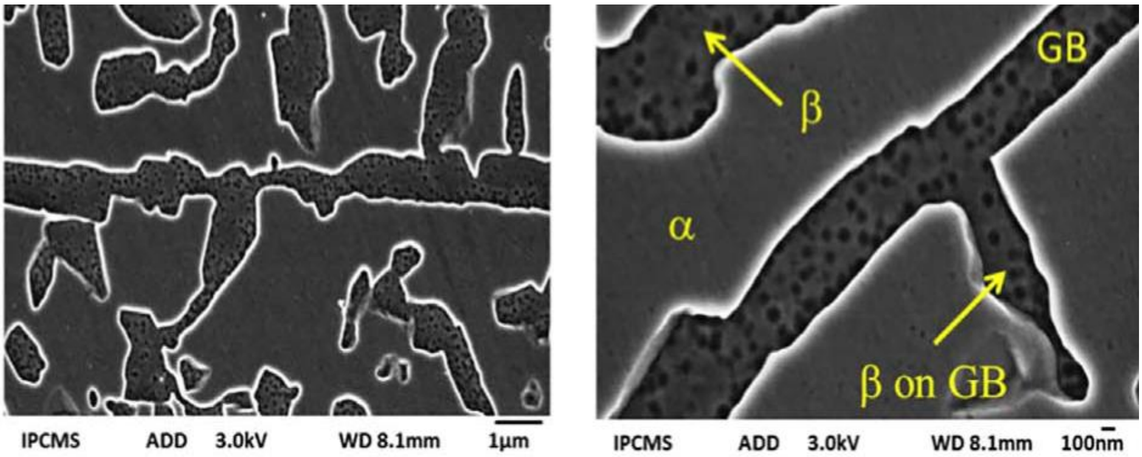


Fig. VII-11-: SEM images of Al-12 wt% Mg alloy homogenized for 17 h at 430 °C quenched and aged for 21 h at 250 °C *(C. Amrane et al., Materials Characterization 134 (2017) 49–54).*

Transmission electron microscopy (TEM) allows for precise characterization of phases in the alloy in terms of crystallographic structure (such as orientation relationships between phases), geometric features (including morphology, size, and distribution of phases at very high magnifications), and chemical composition (main elements). The spatial resolution is sufficient to visualize planes and columns of atoms constituting the microstructure through the interference of diffracted electron beams. Spatial resolution in crystallographic analysis (via electron diffraction) or chemical analysis (through energy-dispersive X-ray spectroscopy or energy analysis of transmitted electrons) can reach a few nanometers. This level of resolution is crucial for characterizing most precipitates in quenched aluminum alloys. However, it may be insufficient for certain cases, such as studying the initial stages of precipitation in Al-Mg-Si, Al-Zn-Mg, and Al-Cu-Mg alloys. TEM, based on electron diffraction, is particularly effective for investigating the orientation relationships of precipitates relative to the matrix and for analyzing structural defects (such as dislocations, lattice distortions, and grain boundaries).

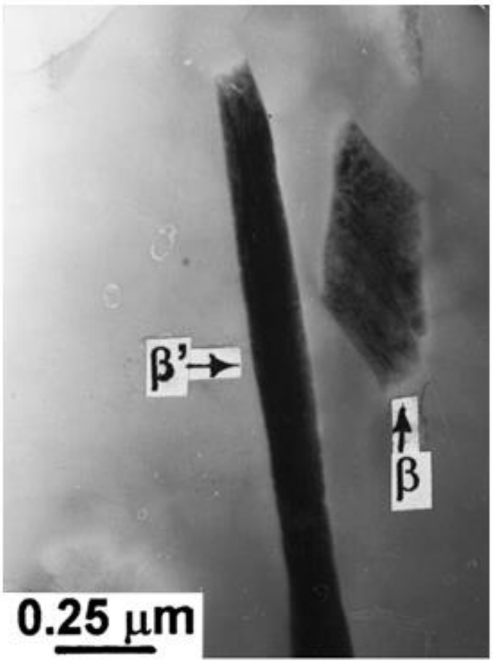


Fig. VII-12-: TEM micrographs of Al–12 wt.% Mg alloy homogenized 17 h at 430°C, quenched and aged 1 h at 250°C illustrating different particles of β' and β phases *(D. Hamana et al., Journal of Alloys and Compounds 320 (2001) 93–102).*

VII-6-2. X-ray and neutron diffraction

The extensive relationships between crystallographic structure and precipitation parameters make diffraction methods (including electron diffraction) essential tools for the microstructural study of metallic alloys. These techniques are employed for crystallographic characterization and, in some cases, for determining the morphology, volume fraction, and size of the precipitates present in the studied system.

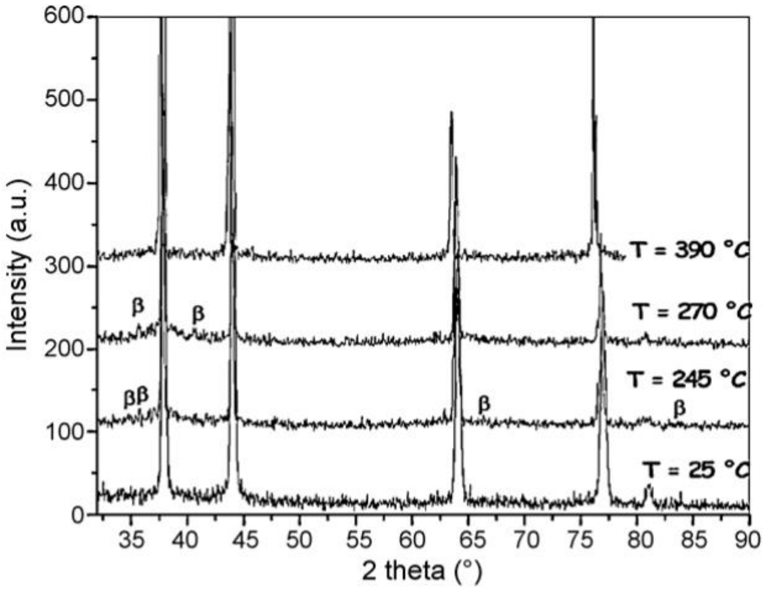


Fig. VII-13-: Sequence of X-ray diffraction patterns recorded during heating at a rate of 2°C/min for an Al–12 wt.% Mg alloy that was homogenized at 430°C for 17 hours, quenched in ice water, and then compressed to reduce its thickness to 16%. *(A. Hayoune and D. Hamana, Materials Science and Engineering A 527 (2010) 7261–7264).*

VII-6-3. Macroscopic physical methods

These methods exploit macroscopic variations in physical properties (such as electrical, thermal, and dimensional) of the material according to its state of structural separation.

Differential Scanning Calorimetry (DSC) is a highly sensitive calorimetric method that allows for the direct measurement of heat released or absorbed by an alloy sample during an exothermic process (such as precipitation) or an endothermic process (such as the dissolution of precipitates or GP zones) caused by a temperature increase at a controlled rate or by a controlled isothermal segment (Fig. VII-14). After calibration, this technique can quantitatively determine the mass fraction of phases formed in the matrix due to homogeneous or heterogeneous precipitation. DSC also enables accurate determination of phase transformation temperatures, which are also studied using Differential Thermal Analysis (DTA) and are crucial for defining the conditions for alloy dissolution.

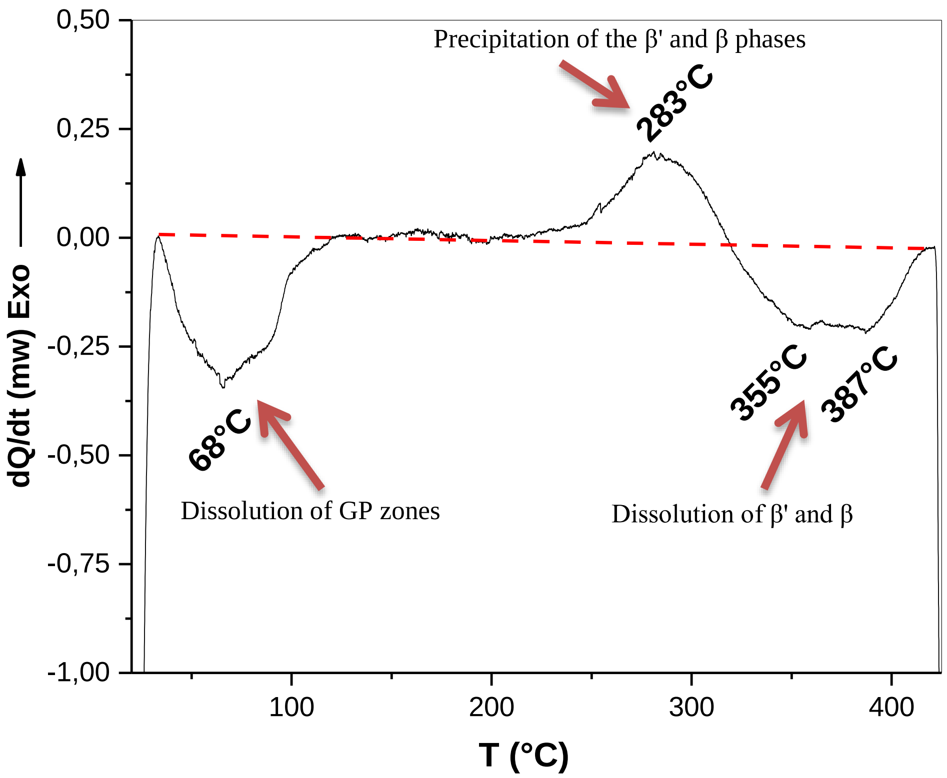


Fig. VII-14-: DSC curves of the Al-12wt. % Mg alloy homogenized then water quenched

Differential dilatometry allows for monitoring the kinetics and phase transformation ranges during heat treatments by measuring dimensional variations (contractions or expansions) induced by the aging or tempering of alloys (Fig. VII-15). However, the effects are generally weaker in industrial light alloys (particularly in ternary systems such as Al-Cu-Mg and Al-Mg-Si) compared to their constituent binary alloys (Al-Cu, Al-Mg, Al-Si). This is due to the compensation of individual effects caused by deviations in atomic radii of different solute atoms relative to the aluminum matrix and the precipitation of various metastable phases.

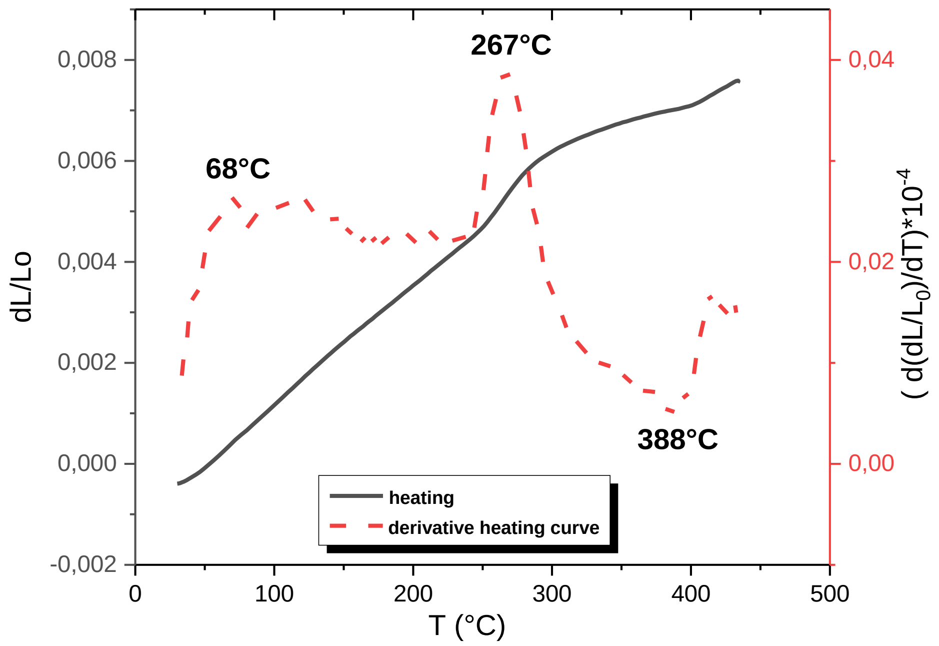


Fig. VII-15-: Derivative of the heating segment for an Al-12 wt.% Mg alloy that was homogenized for 17 hours at 430°C and then water-quenched.

The electrical conductivity and, conversely, the electrical resistivity are influenced by the progress of the solid solution's decomposition. Solute concentration has a significantly greater impact on these transport properties than the precipitate volume fraction. Measuring surface electrical conductivity and resistivity at room temperature is commonly used in industry as a non-destructive and rapid method for assessing the quenching and tempering state of semi-finished aluminum products, intended respectively for aeronautical and electrical applications. Electrical resistivity initially increases rapidly and then more slowly during maturation due to the formation of GP zones, likely because electron diffusion is affected by the clusters of atoms formed. It subsequently decreases progressively during tempering as the matrix solute is depleted, reaching a minimum value after extensive overaging or annealing. Finally, electrical resistivity is also influenced by the surface area of the precipitates and their shape factor (the ratio of the largest dimension to the smallest dimension).

Practice exercise

The heat treatment sequence used industrially for the Al-12 wt. % Cu alloy is given in Figure 1.

1- What happens during the high temperature plateau?
2- Why do we quench the alloy after the first stage?
3- Is the quenched alloy then in equilibrium?

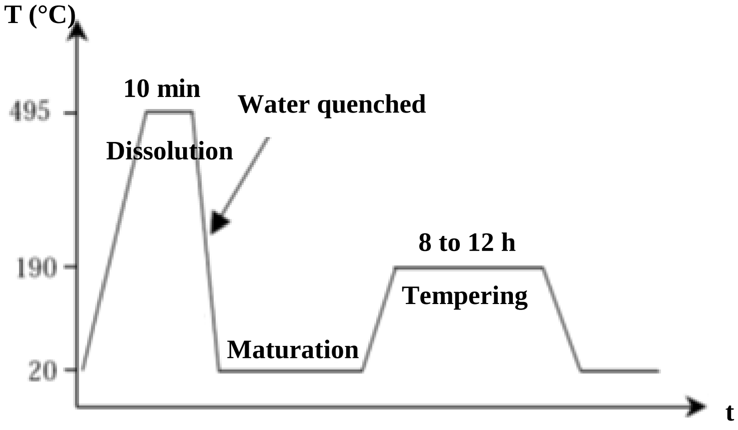


Figure 1

Answers

1. During the high-temperature plateau, the alloy undergoes solution treatment. This process involves heating the alloy to a high temperature where the solute elements (such as copper) dissolve completely into the aluminum matrix. This results in a single-phase solid solution. The purpose of this stage is to homogenize the alloy and dissolve any precipitates or phases present before quenching.

2. The alloy is quenched after the first stage to rapidly cool it and trap the solute elements in a supersaturated solid solution. Quenching prevents the precipitates from forming during cooling, which allows for the subsequent aging process to control the precipitation and enhance the alloy's strength and hardness. This rapid cooling helps to achieve a non-equilibrium state that can be further processed to develop the desired microstructure.

3. No, the quenched alloy is not in equilibrium. After quenching, the alloy is in a non-equilibrium state with a high concentration of solute elements trapped in a supersaturated solid solution. The alloy will undergo a subsequent aging process to allow controlled precipitation of phases and achieve the desired mechanical properties. During this aging process, the solute elements will diffuse and form precipitates, eventually leading to a more stable microstructure.