1.3. Displacive products of phase transformation

For displacive phase transformation disrupting of original arrangement in matrix and formation of a new crystallographic structure are typical. During this phase transformation diffusion processes are restricted extensively. Diffusion of Fe is impossible, but e.g. in case of bainite carbon is able of diffusion. In case of martensite neither Fe diffusion nor carbon diffusion is realized. By displacive (shear) mechanism coordinated atoms transfer on a short distance (shorter than lattice parameter) can be understand. a) in exactly defined directions leading to FCC austenite lattice to BCC ferrite. Among analysed products belong followed displacively formed:

a) Widmanstätten’s ferrite (WF)
b) Acicular ferrite (AF)
c) Bainite (B)
d) Martensite (M)

1.3.1. Widmannstätten’s ferrite (WF)

Widmanstätten’s ferrite (WF) is formed approximately at temperature by 80-100°C under idiomorphic ferrite (IDF) formation – Fig. 1.2. Its formation is only connected with carbon diffusion and no with substitution elements. Widmanstätten’s ferrite is always formed intergranularly, hence on the austenite grain size. To the WF formation comes in case when the cooling rate is higher and grain coarser (long path are for diffusion). Under those conditions ferrite nucleation is realised in cleavage facets of austenite grains and ferritic “arrows” are in those formed – see Fig.1.5. When the steel is strongly overheat and relatively fast cooled, those ferritic arrows are longer and also more frequent, go through all grain and forms characteristic figures. The WF microstructure occurrence (Fig. 1.5 and 1.6) is typical for materials after casting and faster cooling process, when structure coarse primary grain shows (long path for carbon diffusion) and in heat affective zone of weld joints. It is undesirable microstructure, because of Widmanstätten’s cleavage character. The WF microstructure can be eliminated by normalisation annealing. Without over recrystallize – grains interface shows low energy and is stable.
1.3.2. Bainite (B)

Under intermediate cooling rate the undercooled austenite is transformed into unbalanced heterogeneous structure – bainite (B). Bainite is mixture of oversaturated ferrite and carbides of un-laminated type. Diffusion of Fe atoms is practically un-possible, however, carbon diffusion is realised. The main base of bainite formation is austenite transformation into ferrite formed by shear mechanism. Arrangement of carbon is also changed (diffusion is possible) however none change of addition elements is supported, because their diffusion is practically unrealised. The bainite microstructure morphology is importantly depending on transformation temperature and on cooling rate through bainite region. Bainite can be divided to:

a) upper bainite (UB)

b) middle bainite (MB)

c) low bainite (LB)

Mechanism of B formation as well as its morphology is importantly changed with transformation temperature and chemical composition of steel, because the B can be divided into upper one (UB) and low one (LB). With decreasing transformation temperature B hardness and strength properties are increased. High B strength several kinds of strengthening cause. It is lath (plate) of bainite ferrite, which has higher dislocation density, further strengthening by carbon in solid solution and also dispersion strengthening by precipitated carbides. Between the UB and LB is situated so called middle bainite (MB) comes into existence, which includes some portion of upper- and low bainite.
**Upper bainite (UB)** is nucleated at temperatures of 350 – 500 °C, approximately 100 °C under WF formation temperature. It is formed on the austenite grain boundary (intergranularly). **Intragranularly** it is formed seldom. Bainite structure forms sheaves of coarser laths (plates) of bainite ferrite with longitudinally arranged cementite particles being first of all nucleated on the interface bainite ferrite and austenite. Its structure is plate-like (lath-like) and forms packets, which show high angle arrangement, while in frame of one packet the laths (plates) are parallel mutually, with low angle arrangement. Plates (laths) of bainite ferrite are formed by displacive mechanism (shear mechanism), “shoot” very fast, only into sure distance. Bainite ferrite can grow, however slowly into thickness, but more intensively into length. Carbon is divided through diffusion, it is able to get to interface bainite ferrite –austenite, where it precipitates and forms coarse carbides (Fe₃C – cementite). In case of cleavage crack initiation and its attack on carbide, that one is broken and the crack is propagated through matrix. Carbides are coarse, hard and usually are not able to stop the crack propagation. At the bainit-austenite interface crack is propagated by linking up of coarse carbides and simultaneously during deviation of secondary cracks through matrix of bainite ferrite, where are none obstacles for crack propagation. This is reason of other cracks dividing and other propagation of cracks, resp. generally the reason of unfavourable toughness of the upper bainite. The UB shows high strength, however low toughness with a tendency to high hydrogen embrittlement. **Figure 1.7a, b** shows the UB microstructure.

**Low bainite (LB)** is formed at low temperatures for B nucleation, is formed by thin plates of bainite ferrite oversaturated with carbon, with carbides precipitation, predominantly realised in within plates of ferrite matrix. Carbides precipitation is coming into existence on the stacking faults of austenite. Bainite ferrite is formed again by displacive (shear) mechanism. Temperature is already too low and this is reason why the bainite ferrite is not able to push carbon out, because the conditions for diffusion are not sufficient. Bainite ferrite grows very slowly and bainite ferrite plates are thin. Microstructure of LB is finer and properties are close to martensite. At cleavage crack propagation and by its attack on fine carbide the carbine will be not broken. Crack will deviated only and each deviation represents a loss of kinetic energy for cleavage crack propagation. After some repeating the kinetic energy is so low that it is not able to propagate through matrix and it will be stopped in short time. The LB shows high strength and simultaneously a good toughness. Microstructure of the LB is presented in **Fig. 1.8**.
Mechanical-metallurgical properties of bainite steels in comparison to tempered martensite fall behind. It is given by some difference in mechanism of formation and it is source of different microstructures of both products nucleated at low temperatures. Lower fracture toughness of bainite steels can be partially eliminated by higher carbon content (up 1.5 %). This is related to insignificant silicium solubility in cementite, leading to limited cementite precipitation at higher Si in steel.

Ductile of austenitic films leads to taking the edges off possible micro cracks and are connected with approximately 10-15 % of residual austenite.

1.3.3. Acicular ferrite (AF)

Acicular ferrite (AF) is formed intra-granularly. It is nucleated on specific precipitates or inclusions. These intra-granular un-ferrous particles are potential places for AF nucleation. The AF nucleation intensity is influenced by un-metallic particles size. AF has a form of laths (plates), shows favourable toughness and excellent hydrogen responsibility. Microstructure of
AF is depicted in Fig. 1.9. (Nital etching) and a coarser austenite matrix support the AF formation. AF shows “star” arrangement. Primarily formed laths (plates) can be secondary completed by so called “sympathetic” nucleation. Generally effective structure must show a chaotic, interwoven character.

Fig. 1.9a, b  Formation of intra-granular AF a) in volume of coarse grained austenite, b) Fig. 1.9a in detail

Conditions for AF formation, beside inclusion size (see Fig. 1.10a), also austenite grain size influences (see Fig. 1.10b).

Fig. 1.10a, b  Probability of AF formation in dependence on a) un-ferrous inclusions size, b) former austenite grains size

There is a question – are differences of final properties between AF and UB when both products are nucleated at the same temperature approximately, mechanism of bot his the same
(displacive one) and only differences are in nucleation positions? Both AF and UB are formed by displacive mechanism at the same temperature approximately, but the positions of nucleation are diametrically different. Laths (plates) of the UB are parallel in frame of individual packets and show identical crystallographic orientation. The UB is formed on the austenite grain boundaries, while the AF is nucleated in “points” in localised places of austenite grain volume. AF presence in matrix is of a great advantage, because it increases resistance against brittle rapture and strength level of steels. Conventional UB and AF are formed under conditions of identical isothermal phase transformation in the same steel type. Phase transformation to UB is preferentially realised in steels with finer austenitic grains, while the UB initiation comes into existence on the austenite grain boundaries, preferentially and further development is realised into the austenite grain volume. In case the austenite grain sizes are coarser (e.g. in continual casting process) the AF is preferentially nucleated. Consequently, the AF is not formed if the frequency of intra-granular nucleable potential positions is low and the austenite grains too fine.

![Graph showing angle of disorientation](image)

**Fig. 1.11  Results of disorientation of AF and UB plates (laths)**

Microstructure of AF is fine grained and leads to strength and toughness increase of given steel type. Laths boundaries of AF predominantly show high-angle phase arrangement representing obstacle against the cleavage crack propagation, while the UB represents the low-angle arrangement predominantly (in frame of one packet) – see **Fig. 1.11**. Presence of bainitic microstructure in frame of AF is un-favourable and makes the mechanical-
metallurgical properties worse. At matrix deformation, coarse austenite grains the AF formation to the exclusion of the UB contribute as well as narrow intrer-granular ATF bands. Could be characterized:

a) austenitization of steel including formed carbides solution and micro-alloyed additions in matrix;

b) “coarse” rolling process leading to a fine recrystallized austenite microstructure;

c) rolling in suppressed recrystallization region ensuring matrix strengthening and

d) final rolling process finished approximately 30-50 °C above the Ar₃ temperature followed by accelerated cooling process into approximately (according chemical composition) 500 °C, so that unfavourable segregation banding would be prevented. Final cooling can be carried out on the air. Of course, other controlled cooling is of more useful.

The length of cleavage crack path corresponds to 3 - 5μm, while in UB matrix the free path between packets reaches approximately 15-20 μm. In AF microstructure high angle miss-orientation predominates unlike the bainite, where the high angle miss-orientation between packets exists only, but predominantly low angle phase miss-orientations among laths and/or plates in frame of each packet assert being in majority. Consequently, the AF matrix represents higher resistance against cleavage crack propagation. This resistance can be also registered in case of hydrogen response. Schematically, transformation products of austenite into UB, AF and other possible products as well as crack propagation in the AF matrix Fig. 1.12 and 1.13 depict.

![Schematic illustration of austenite transformation products](image)

**Fig. 1.12** Schematic illustration of austenite transformation products
1.3.4. Martensite (M)

Martensite formation from austenite is technically the most important example of transformation without diffusion. M is formed at temperatures which are so low that none diffusion of substitution and interstitial elements is possible. $M_s$ (martensite start) represents temperature of starting martensite formation and $M_f$ martensite finishing temperature. Figures 1.14 - 1.16, dependence of martensite quantity formation (and/or residual austenite) on carbon and/or addition elements demonstrate.

Fig. 1.13 Schematic illustration of cleavage crack propagation in microstructure with AF

Fig. 1.14 Schematic illustration of martensite formation between $M_s$ and $M_f$

Fig. 1.15 Carbon influence on residual austenite portion

Fig. 1.16 Influence of addition elements on residual austenite portion
High alloyed steels with high carbon content show a great residual austenite portion. Extremal hardness and brittleness of quenched steel together with residual austenite can be partially removed by tempering. That operation should follow-up to quenching immediately. Minimal quenching rate \( (v_k) \) depends on chemical steel composition and microstructure type. In case of carbon steel it is approximately 1000 K.s\(^{-1}\). M formation is connected with undiffusion transformation (M is formed by displacive mechanism – by shear). Martensite transformation is realized after Ms reaching immediately. No incubation time is necessary for its realization. M is formed by arranged transports of atom groups from knot points of the original phase to knots of lattice of a new phase so, that relative changes of neighbouring atoms positions are shorter than distances among atoms.

Obr. 1.17. Martensite microstructure image a) light microscopy, b) TEM

Martensite (M) is formed by displacive mechanism (by shear) and laths and/or plates are formed as it Fig. 1.17a, b shows. These are semi-coherent with a great tendency to coherence with austenite. M in steel represents unbalanced solid solution of carbon in Fe-\(\alpha\), has tetragonal lattice centrally orientated (M = very often \(\alpha'\)). Carbon atoms are responsible for Fe-BCC lattice deformation. Steels with intermediate and higher carbon content (higher than approx. 0.45 wt %) show better detectable tetragonality corresponding the carbon content. Mechanism of lattice deformation carbon content influences, too, hence substructure of M. When carbon content is maximally 0.25 wt. %, deformation is realized by slip (dislocation martensite). Dislocation or lath martensite with dislocations is formed in case of low carbon steels and/or of high alloyed steels. With higher carbon content higher
deformation is realized in plates by twins formation (twin martensite) sometimes plate martensite called. It is often formed by plates which are transversely by fine twins divided.

It is formed in case of carbon or alloyed steels with medium and/or higher carbon content. The plate martensite with inner twins is harder (less tough) than lath martensite. The strengthened effect of twins is in given case compensated by dislocations effect. All M types, being in steels, have higher hardness than primary matrix represents. When the M from interstitial solid solution (e.g. Fe-C, Fe-N) is formed, hardness increase is the most obvious. High martensite hardness in steels is given by these increasing mechanisms:

- **grain size**
- **solid solution hardness** (especially due to interstitial carbon and to addition elements – influence is small)
- **strengthening awakened by martensite formations boundaries**
- **sub-structure strengthening** (dislocation or caused by twin M).

Martensite (M) is brittle. Its high brittleness is caused by above mentioned mechanisms together with inner tension. Specific volume during austenite to martensite transformation and thermal tensions formed during uneven cooling of quenched components are sources of inner tension. With carbon content tetragonality of lattice is on the increase, as well as volume changes during transformation and also twin martensite portion, which has lower ability the stress to relax by plastic deformation unlike the dislocation M. With higher carbon content dangerous of quenching cracks is higher (e.g. in places where martensite plate impacts on austenite grain boundaries and/or in place of martensite plates intersection).

1.4. **Physical engineering characteristics of granular bainite and conditions of M/A constituent formation**

Granular bainite (GB) consists of ferrite and some portion of the M/A constituent. It can be observed in low carbon micro-alloyed steels, resp. low alloyed steels. Granular bainite (GB) is formed during continual austenite decomposition, close under upper bainite formation. Firstly, austenite transforms to ferrite. With temperature decrease by diffusion mechanism carbon is driven out from ferrite matrix into austenite which is in this way enriched and stabilised. When to balance between ferrite and untransformed austenite comes,
Austenite is not able to transform into ferrite already. After crossing the $M_s$-temperature residual austenite transforms into M/A constituent. Strictly spoken, it is marstensite (M) and residual austenite (A). The higher is the M/A portion in matrix, the more strength is final matrix, however also more brittle. By conventional etching in Nital, M/A constituent can be not revealed, as it also the 1st right photos of Fig. 1.18 demonstrate, while after special etching the M/A constituent is quite good detectable, as it the middle photos of Fig. 1.18 show.

**Fig. 1.18** GB image after different etching, together with fracture character

The first row of photos (Fig. 1.18) shows material with high GB portion, resp. with higher volume fraction of M/A constituent), while the second row represents matrix with low
portion of GB, resp. with low volume fraction of M/A constituent. Nital etching shows practically none differences unlike etching in La Perra reagent. Influence of the higher portion of M/A presence cleavage trans-crystalline fracture demonstrates unlike the dimple morphology in case of low M/A constituent volume fraction in matrix, as it from Fig. 1.18 follows.

In praxis the GB formed e.g. in case of profiles showing different thicknesses (see Fig. 1.31, on the right down). Web and flange have different thickness leading to different cooling rates after rolling process. In the thicker parts higher M/A constituent portion can be detected as well as trans-crystalline fracture on the fracture surfaces unlike the areas with much lower M/A constituent volume fraction having not negative influence on fracture character showing dimple morphology, usually as from right down photo of the Fig. 1.31 follows.

**Fig. 1.32**  **Formed Fe₃C particles in places, where M/A constituent after rolling state was observed**

When the GB in higher portion is developed, its negative influence can be not eliminated by annealing. The areas with the M/A constituent that shows higher portion of carbon, those are transformed into Fe₃C particles during annealing process. New formed cementite particles degrade toughness of matrix as well as the M/A constituent particles. An example of the microstructure image with original M/A occurrence after its annealing at 600 °C/2h/air shows Fig. 1.32.
1.4.1. Control of the GB formation

We have three possibilities that can be applied independently and/or in combinations. In any case the aim is to reach austenite transformation into ferrite to the maximal extent, so that un-transformed austenite would be minimum, which will be consequently transformed into minimal portion of the M/A constituent.

a) the higher deformation intensity – it leads to higher number of potential nucleation positions for austenite transformation into ferrite. Through that we support its transformation into ferrite to higher extent. Given process is connected with increase of total free enthalpy of system – see eq. (1.1):

\[-\Delta G = - (\Delta G_{ch} - \Delta G_{A \rightarrow F})VF + E_{def}.\]  

where \( \Delta G_{ch} \) represents chemical part of the free enthalpy, \( \Delta G_{A \rightarrow F} \) a free enthalpy change as a logical consequence of the austenite transformation into ferrite and \( E_{def} \) is deformation degree during forming process.

b) more intensive material cooling, especially in the final stadium of rolling process and after finishing of the rolling, so that more intensive transformation of austenite into ferrite would be supported – under-cooling gradient will be increased and also total free enthalpy.

![Influence of chemical composition on free enthalpy change](image)

**Fig. 1.33** Influence of chemical composition on free enthalpy change
c) **Chemical way** – lower wt. % of elements making austenite stable (e.g. C, Mn) as it Fig. 1.33 shows. The curve \( f \) represents free enthalpy for ferrite and the \( r_1, r_2 \) and \( r_3 \) are curves for austenite. Higher portion of transformed ferrite is every connected with higher level of free enthalpy, on the contrary its lower level characterises higher austenite stabilisation – see eq. 1.1.

**Summarisation of capture terms**

In the end of capture the main terms are repeated, which you should master.

Allotriomorphic ferrite, idiomorphic ferrite, acicular ferrite, massive ferrite, Widmanstätten’s ferrite, bainites, martensite, granular bainite

**Questions**

- What the reconstructive and displacive formed transformation products represent?
- What the differences are between upper and low bainite?
- Could you express the differences between acicular ferrite (AF) and upper bainite?
- Which particles are nucleable for the AF and IDF formation?
- Have austenite grain size and size of nucleable particles impacts on the AF and UB (HB) formation?
- Explain the crack propagation principle in the UB and AF.
- Has residual austenite an impact on final properties?
- Explain the martensite formation.
- Describe granular bainite (GB) formation, influence of M/A constituent portion on steel properties. Is there a possibility to detect the M/A constituent by metallographic method?
- Describe possibilities of the GB control.
Literature:


