

## Bainite in Steels

Bainite forms by the decomposition of austenite at a temperature which is above  $M_S$  but below that at which fine pearlite forms. All bainite forms below the  $T_0$  temperature.

All time–temperature–transformation (TTT) diagrams consist essentially of two C–curves (Fig. 1). If we focus first on the Fe–Mn–C steel with the higher hardenability (slower rates of transformation) then the two curves are separated. The upper C–curve represents the time required to initiate reconstructive transformations such as ferrite or pearlite, whereas the lower C–curve represents displacive transformations such as bainite or Widmanstätten ferrite. Notice that as the hardenability of the steel decreases, the two curves tend to overlap so that in experiments it appears as if the TTT diagram contains just one curve with a complicated shape, describing all the reactions. This is not the case because it is possible to show that this is an experimental artifact caused by the inability to detect the two C–curves separately.

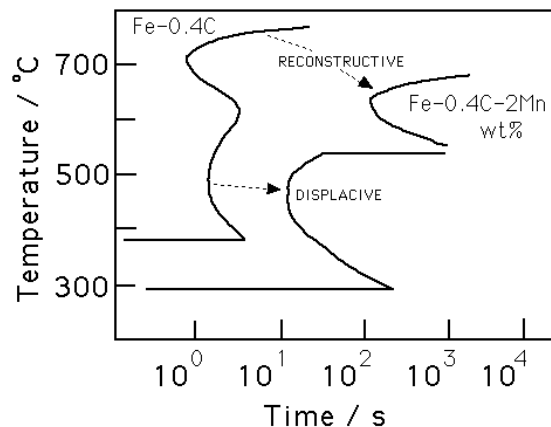


Fig. 1: TTT diagrams for two steels, one of which has a high hardenability.

A further feature to note (Fig. 1, Fe–Mn–C) is that the lower C–curve representing displacive transformations has a flat top. This represents the highest temperature  $T_h$  at which displacive transformations may occur. The temperature  $T_h$  may equal the bainite–start temperature  $B_S$  if the hardenability is high enough, but otherwise,  $T_h = W_S$  where  $W_S$  is the Widmanstätten ferrite start–temperature (Fig. 2). The latter does not form in high–hardenability

steels and we shall discuss in this lecture the detailed differences between bainite and Widmanstätten ferrite.

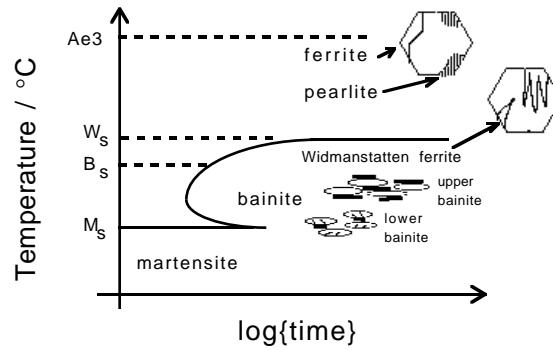


Fig. 2: TTT diagrams showing the different domains of transformation.

Bainite is a non-lamellar aggregate of carbides and plate-shaped ferrite (Fig. 3). As we shall see later, the carbide part of the microstructure is not essential; the carbides form as a secondary reaction, rather as in the tempering of martensite. The ferrite plates are each about  $10\mu\text{m}$  long and about  $0.2\mu\text{m}$  thick, making the individual plates invisible in the optical microscope.

**Upper bainite** consists of clusters of platelets of ferrite adjacent to each other and in almost identical crystallographic orientation, so that a low-angle boundary arises whenever the adjacent platelets touch. The ferrite always has a Kurdjumov-Sachs type orientation relationship with the austenite in which it grows.

Elongated cementite particles usually decorate the boundaries of these platelets, the amount and continuity of the cementite layers depending on the carbon concentration of the steel.

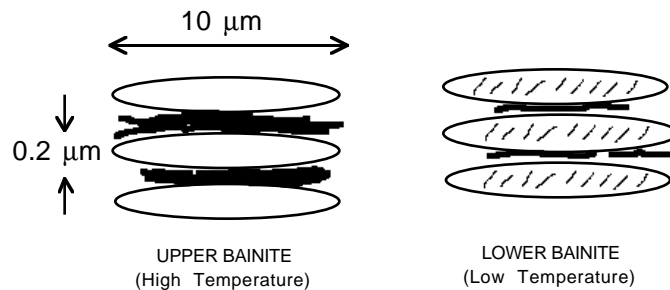


Fig. 3: Schematic illustration of the microstructure of upper and lower bainite.

The clusters of ferrite plates are known as ‘sheaves’ (Fig. 4); each sheaf is itself in the form of a wedge-shaped plate on a macroscopic scale. The sheaves inevitably nucleate heterogeneously at austenite grain surfaces. The cementite precipitates from the carbon-enriched austenite between the ferrite plates; the ferrite itself is free from carbides. Cementite precipitation from austenite can be prevented by increasing the silicon concentration to about 1.5 wt%; this works because silicon is insoluble in cementite. Silicon-rich bainitic steels can have very good toughness because of the absence of brittle cementite.

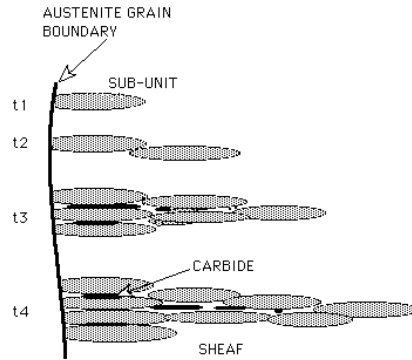


Fig. 4: Evolution of a bainite sheaf as a function of time.

### Shape Deformation

The formation of bainite causes a deformation (Fig. 5) which is an invariant-plane strain with a shear component of about 0.26 and a dilatational strain normal to the habit plane of about 0.03. This is consistent with a displacive mechanism of transformation.

Bainite forms at a relatively high temperature when compared with martensite. The parent austenite is weaker at high temperatures and cannot accommodate the large shape deformation elastically. It therefore relaxes by plastic deformation in the region adjacent to the bainite. This is evident in Fig. 5, but is also presented as a height scan in Fig. 6. The effect of this plastic deformation is to stifle the growth of bainite plates before they hit any obstacle. This is why each bainite plate grows to a size which is often smaller than the austenite grain size and then comes to a halt. Further transformation happens by the formation of a new plate and this is why the sheaf morphology arises.

### Substitutional Alloying Elements

These do not redistribute at all during transformation, even though equilibrium requires

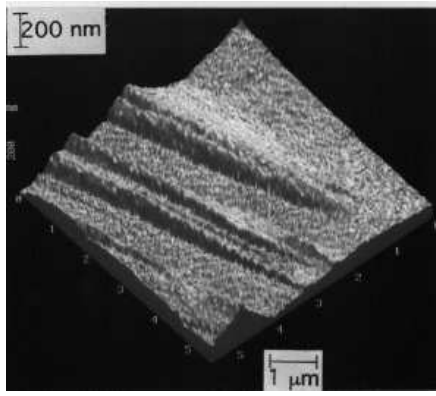


Fig. 5: Atomic force microscope image of the displacements caused on a polished surface of austenite by the growth of bainite. Notice the shear deformation (dark contrast) and indeed the plastic accommodation (light contrast tapering from the ridge of the region of dark contrast) of the shape change in the austenite adjacent to the bainite plates.

them to partition between the austenite and ferrite (Fig. 7). The ratio of substitutional to iron atoms remains constant everywhere including across the interface. This is consistent with a displacive mechanism of transformation and the existence of an atomic correspondence between the austenite and bainitic ferrite. The results exclude any mechanism which involves local equilibrium at the interface, or solute drag effects associated with interfacial motion.

### Interstitial Alloying Elements (C, N)

It appears from Fig. 7 that the carbon has partitioned into the austenite. It is simple to establish that martensitic transformation is diffusionless, by measuring the local compositions before and after transformation. Bainite forms at somewhat higher temperatures where the carbon can escape out of the plate within a fraction of a second. Its original composition cannot therefore be measured directly.

There are three possibilities. The carbon may partition during growth so that the ferrite may never contain any excess carbon. The growth may on the other hand be diffusionless with carbon being trapped by the advancing interface. Finally, there is an intermediate case in which some carbon may diffuse with the remainder being trapped to leave the ferrite partially supersaturated. It is therefore much more difficult to determine the precise role of carbon during the growth of bainitic ferrite than in martensite.

Diffusionless growth requires that transformation occurs at a temperature below  $T_0$ , when

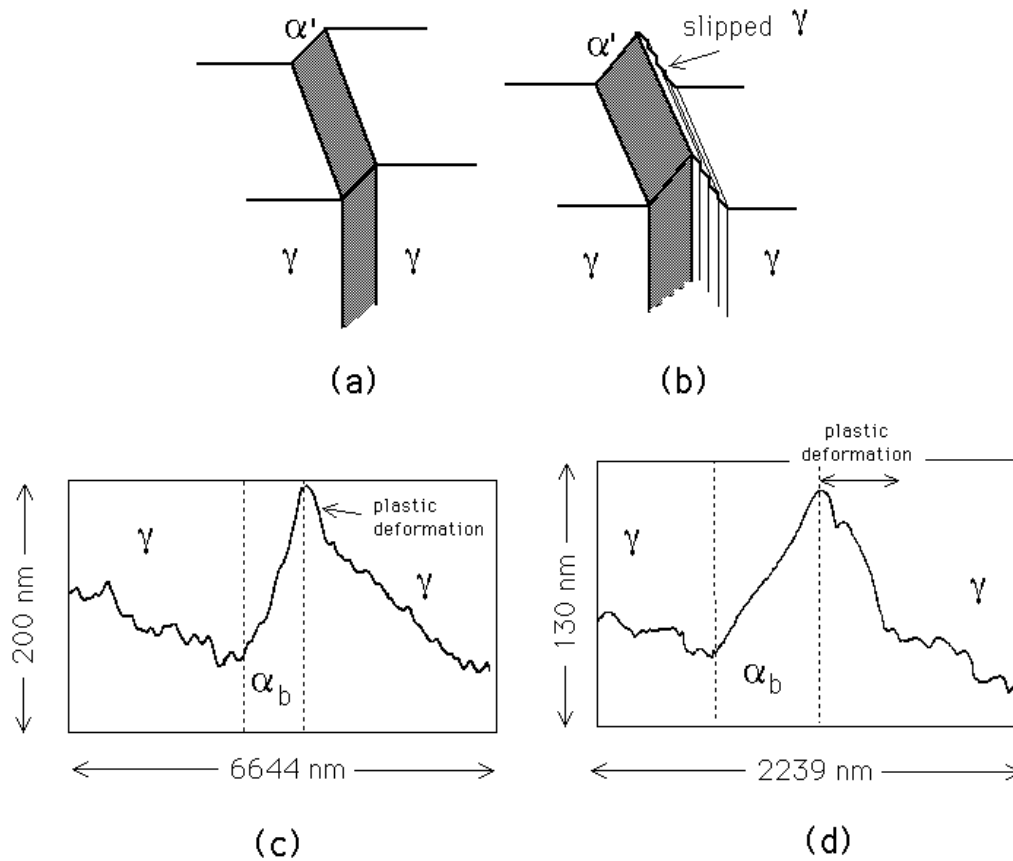


Fig. 6: (a) A perfect invariant–plane strain surface relief effect. (b) One where plastic relaxation of the shape change occurs in the adjacent matrix. (c,d) An actual atomic force microscope scan across the surface relief due to a bainite sub–unit (Swallow and Bhadeshia, 1996).

the free energy of bainite becomes less than that of austenite of the same composition. A locus of the  $T_0$  temperature as a function of the carbon concentration is called the  $T_0$  curve, an example of which is plotted on the Fe–C phase diagram in Fig. 8. Growth without diffusion can only occur if the carbon concentration of the austenite lies to the left of the  $T_0$  curve.

Suppose that the plate of bainite forms without diffusion, but that any excess carbon is soon afterwards rejected into the residual austenite. The next plate of bainite then has to grow from carbon–enriched austenite (Fig. 9a). This process must cease when the austenite carbon concentration reaches the  $T_0$  curve. The reaction is said to be incomplete, since the austenite has not achieved its equilibrium composition (given by the Ae3 curve) at the point the reaction stops. If on the other hand, the ferrite grows with an equilibrium carbon concentration then the transformation should cease when the austenite carbon concentration reaches the Ae3 curve.

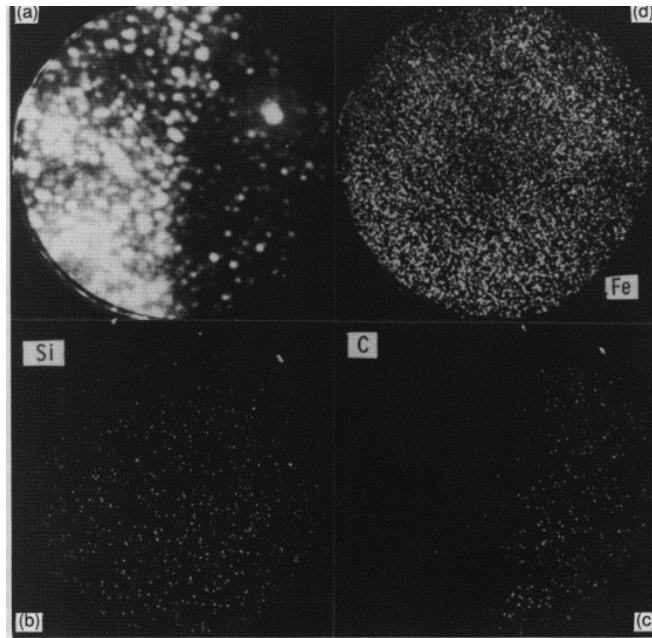


Fig. 7: Imaging atom-probe micrographs, taken across an austenite-bainitic ferrite interface in a Fe-C-Si-Mn alloy. The images confirm quantitative data (Bhadeshia and Waugh, 1982) showing the absence of any substitutional atom diffusion during transformation. (a) Field-ion image; (b) corresponding silicon map; (c) corresponding carbon map; (d) corresponding iron map.

It is found experimentally that the transformation to bainite does indeed stop at the  $T_0$  boundary (Fig. 9b). The balance of the evidence is that the growth of bainite below the  $B_S$  temperature involves the successive nucleation and martensitic growth of sub-units, followed in upper bainite by the diffusion of carbon into the surrounding austenite. The possibility that a small fraction of the carbon is nevertheless partitioned during growth cannot entirely be ruled out, but there is little doubt that the bainite is at first substantially supersaturated with carbon.

These conclusions are not significantly modified when the strain energy of transformation is included in the analysis.

There are two important features of bainite which can be shown by a variety of techniques, *e.g.* dilatometry, electrical resistivity, magnetic measurements and by metallography. Firstly, there is a well defined temperature  $B_S$  above which no bainite will form, which has been confirmed for a wide range of alloy steels. The amount of bainite that forms increases as the transformation temperature is reduced below the  $B_S$  temperature. The fraction increases during isothermal transformation as a sigmoidal function of time, reaching an asymptotic

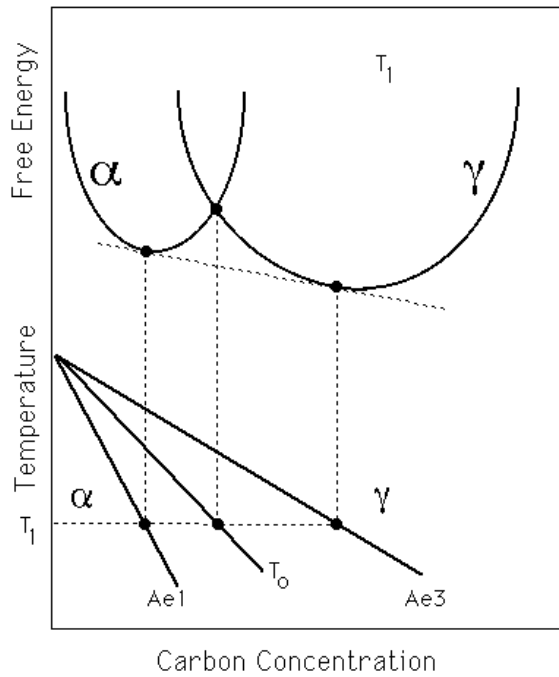


Fig. 8: Schematic illustration of the origin of the  $T_0$  construction on the Fe–C phase diagram. Austenite with a carbon concentration to the left of the  $T_0$  boundary can in principle transform without any diffusion. Diffusionless transformation is thermodynamically impossible if the carbon concentration of the austenite exceeds the  $T_0$  curve.

limit which does not change on prolonged heat treatment even when substantial quantities of austenite remain untransformed. Transformation in fact ceases before the austenite achieves its equilibrium composition, so that the effect is dubbed the “incomplete–reaction phenomenon”.

These observations are understood when it is realised that growth must cease if the carbon concentration in the austenite reaches the  $T_0$  curve of the phase diagram. Since this condition is met at ever increasing carbon concentrations when the transformation temperature is reduced, more bainite can form with greater undercoolings below  $B_S$ . But the  $T_0$  restriction means that equilibrium, when the austenite has a composition given by the Ae3 phase boundary, can never be reached, as observed experimentally. A bainite–finish temperature  $B_F$  is sometimes defined, but this clearly cannot have any fundamental significance.

### Summary

Bainite grows by displacive transformation; the growth is accompanied by a shape defor-

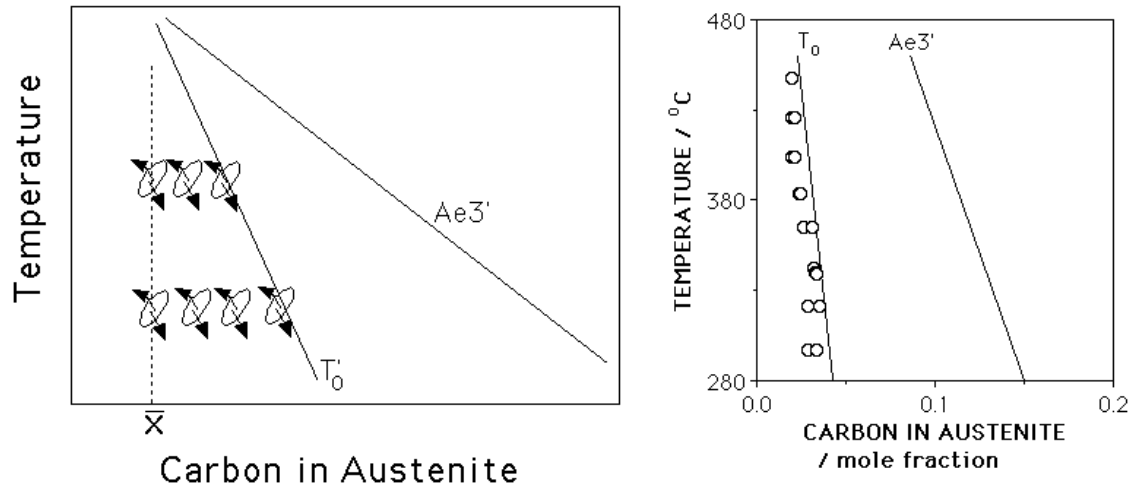


Fig. 9: (a) Illustration of the incomplete reaction phenomenon. During isothermal transformation, a plate of bainite grows without diffusion, then partitions its excess carbon into the residual austenite. The next plate therefore has to grow from carbon-enriched austenite. This process continues until diffusionless transformation becomes impossible when the austenite composition eventually reaches the  $T_0$  boundary. (b) Experimental data showing that the growth of bainite stops when the austenite carbon concentration reaches the  $T_0$  curve (Fe-0.43C-3Mn-2.12Si wt.% alloy).

mation which is an invariant-plane strain with a large shear component. The transformation is diffusionless but carbon atoms partition into the residual austenite (or precipitate as carbides), shortly after growth is arrested. The precipitation of carbides is therefore a secondary event.