H. K. D. H. Bhadeshia

Case Study: Weld Microstructures

The microstructure obtained as the weld cools from the liquid phase to ambient temperature is called the *as–deposited* or *primary* microstructure. Its major components include allotriomorphic ferrite, Widmanstätten ferrite, and acicular ferrite (Fig. 1). There may also be some martensite, retained austenite or degenerate pearlite. These latter phases occur in very small fractions, and are known by the collective term *microphases.* Bainite, consisting of sheaves of parallel platelets, is not generally found in well–designed welding alloys. Instead, acicular ferrite is induced to nucleate heterogeneously on nonmetallic inclusions.

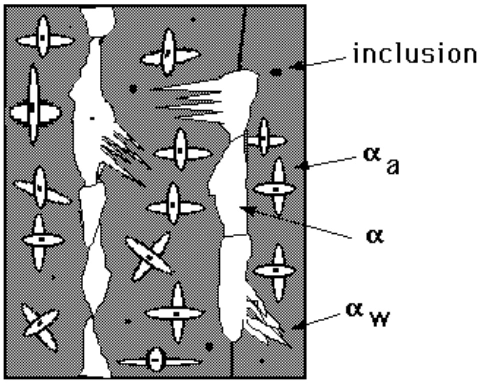

Fig. 1: Schematic illustration of the essential constituents of the primary microstructure in the columnar austenite grains of a steel weld deposit.

Allotriomorphic ferrite

Allotriomorphic ferrite (α) is the first phase to form on cooling the austenite grains below the Ae_3 temperature. It nucleates at the columnar austenite grain boundaries. Because these boundaries are easy diffusion paths, they become decorated with thin, continuous layers of ferrite. The layers then thicken at a rate which is controlled by the diffusion of carbon in the austenite ahead of the transformation interface. Under isothermal conditions, the ferrite thickness q changes parabolically with time t:

$$q = \alpha_1 t^{\frac{1}{2}} \tag{1}$$

where α_1 is called the parabolic rate constant. This is illustrated in Fig. 2 for alloys with different carbon concentrations; note that the growth kinetics become sensitive to the carbon concentration as the latter approaches the solubility of carbon in the ferrite. The kinetic equations all contain a supersaturation term

$$\Omega = \frac{c^{\gamma\alpha} - \overline{c}}{c^{\gamma\alpha} - c^{\alpha\gamma}} \tag{2}$$

where $c^{\gamma\alpha}$ is the concentration in austenite in equilibrium with ferrite and $\overline{c}$ is the average concentration. Thus, the supersaturation tends to unity as $\overline{c} \rightarrow c^{\alpha\gamma}$.

The theory so far has dealt with isothermal growth; the transformation in welds occurs as it cools. Therefore, the allotriomorph thickness has to be integrated over temperature:

$$q = \int_{t=0}^{t_1} 0.5\alpha_1 t^{-0.5} + \frac{\partial \alpha_1}{\partial t} t^{0.5} \quad dt \tag{3}$$

Once the thickness of the ferrite has been obtained, it can be converted into a volume fraction by taking into account the geometry of the austenite grains.

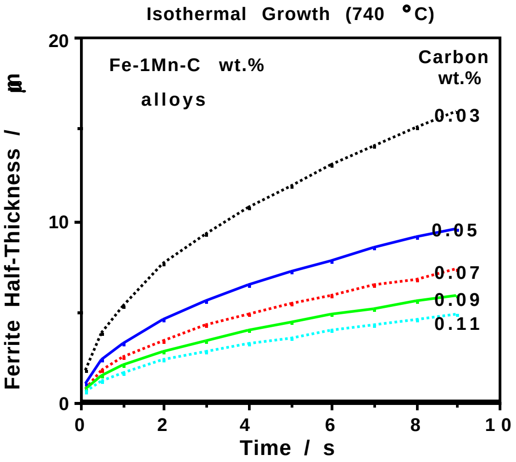

Fig. 2: An illustration of the parabolic thickening of ferrite during isothermal transformation. Each curve represents a Fe-1Mn–C wt.% steel with the carbon concentration as indicated on the diagram.

The magnitude of the parabolic rate constant depends on the equilibrium compositions of the austenite and ferrite, and on the diffusivity of carbon in austenite. Alloying elements such as manganese, which stabilise austenite, are associated with smaller values of α_1. In welding, transformations are not isothermal, but nevertheless, because nucleation is not rate limiting, the fraction of allotriomorphic ferrite obtained correlates directly with the parabolic rate constant (Fig. 3a).

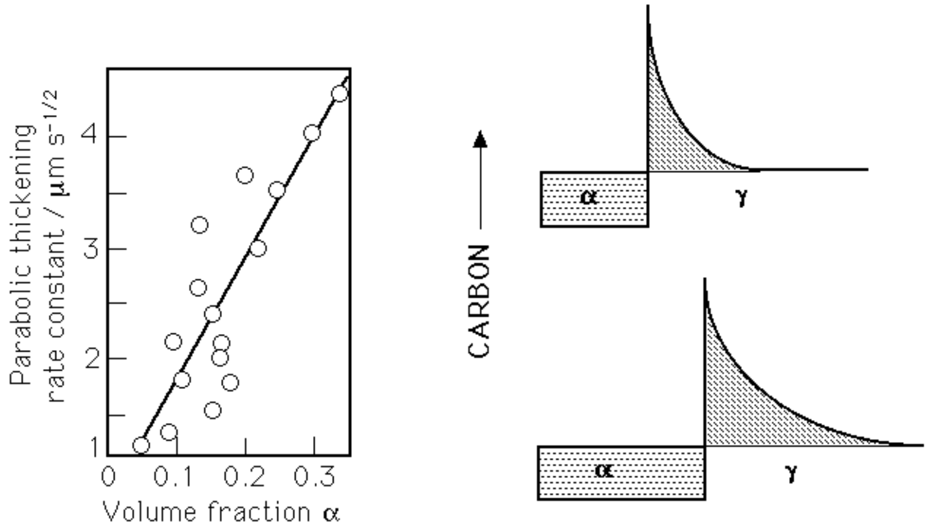

Fig. 3: (a) The correlation between the calculated parabolic thickening rate constant (a variable related to the growth rate) and the volume fraction of allotriomorphic ferrite obtained in a series of manual metal arc weld deposits, fabricated using similar welding parameters but with different chemical compositions. The rate constant is calculated for transformation at 700°C. (b) The diffusion distance increases as the ferrite layer thickens, slowing down the rate of growth.

The fact that the thickness of the ferrite varies with the square root of time, means that the rate of growth decreases as the ferrite layer gets thicker. This is because the distance over which carbon has to diffuse increases with time (Fig. 3b). The growth rate for a given alloy goes through a maximum as a function of temperature, because the driving force for transformation increases with undercooling whereas the diffusivity decreases. Consequently, as the weld cools to temperatures less than about 600 °C, the diffusional growth of ferrite slows down so much, that the layers of allotriomorphic ferrite reach a limiting thickness. Widmanstätten ferrite formation does not involve the diffusion of substitutional solutes, and therefore its growth is not sluggish at low temperatures. The remaining austenite therefore begins to transform to Widmanstätten ferrite (Figure 1).

Although substitutional solutes and iron atoms do not diffuse during the growth of Widmanstätten ferrite, carbon does partition during transformation. Because of its plate shape, much of the carbon can be accommodated at the sides of the growing plate, so that the plate tip always encounters fresh austenite. This is unlike the case for allotriomorphic ferrite, where the partitioned carbon builds up ahead of the interface and progressively slows down the rate of growth. Widmanstätten ferrite plates therefore lengthen at a constant rate.

The growth rates are found to be so large for typical weld compositions, that the formation of Widmanstätten ferrite is usually completed within a fraction of a second. Hence, for all practical purposes, the transformation can be regarded as being isothermal (Fig. 4a).

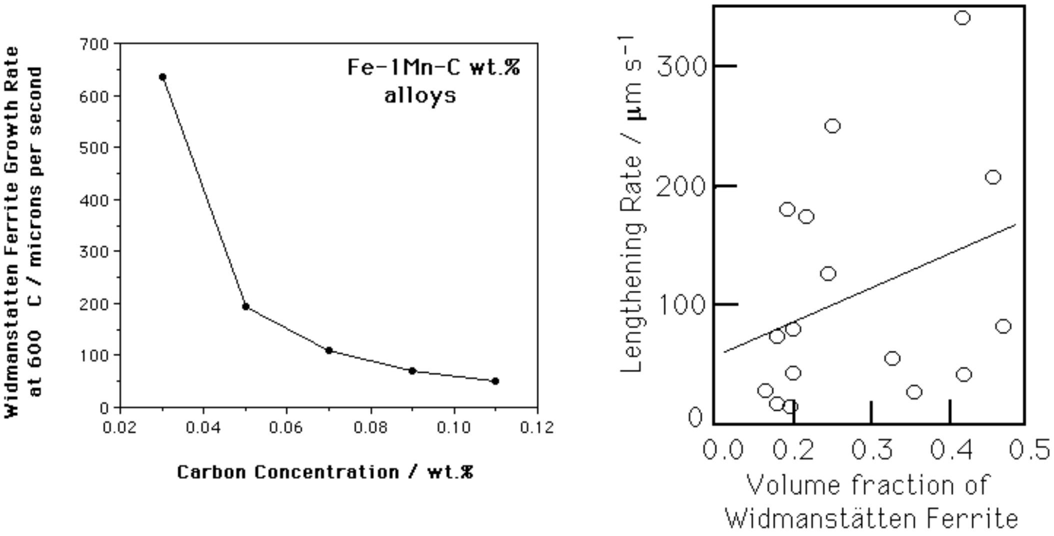

Fig. 4: (a) The isothermal growth rate of Widmanstätten ferrite in a series of Fe–1Mn–C wt% alloys as a function of carbon concentration. Notice that the growth rates are so large, that the plates could grow right across typical austenite grains within a fraction of a second. (b) Poor correlation of the volume fraction of Widmanstätten ferrite against the calculated growth rate.

Unfortunately, the fraction of Widmanstätten ferrite that forms in weld deposits correlates badly with the plate lengthening rate, as illustrated in Fig. 4b. This is because there is an interference between the plates of Widmanstätten ferrite that grow from the austenite grain boundaries, and acicular ferrite plates which nucleate at nonmetallic particles dispersed throughout the weld (Fig. 5). The formation of Widmanstätten ferrite and acicular ferrite is therefore competitive. Anything that increases the number density of inclusion nucleation sites relative to austenite grain nucleation sites, favours the formation of acicular ferrite at the expense of Widmanstätten ferrite. Hence, the refinement of austenite grain size, or a reduction in the oxide content of the weld below a limiting value, both lead to a decrease in the acicular ferrite content.

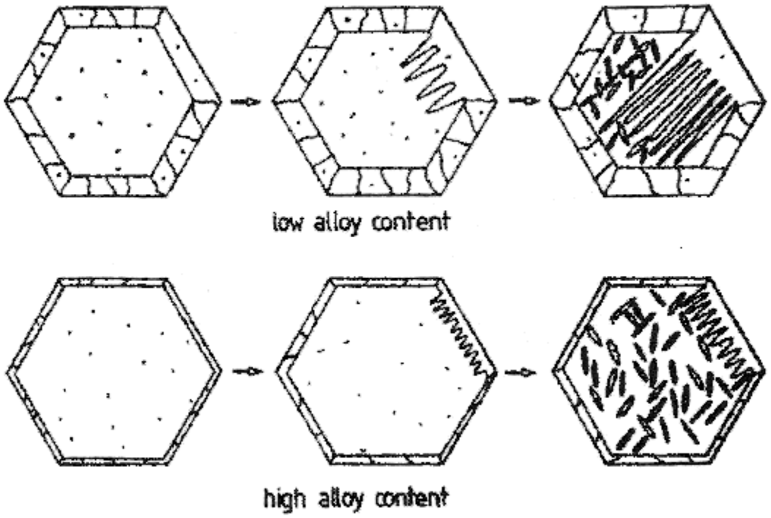

Fig. 5: Diagrams illustrating the development of microstructure in two weld deposits with different chemical compositions. The hexagons represent cross–sections of columnar austenite grains whose boundaries first become decorated with uniform, polycrystalline layers of allotriomorphic ferrite, followed by the formation of Widmanstätten ferrite. Depending on the relative transformation rates of Widmanstätten ferrite and acicular ferrite, the former can grow entirely across the austenite grains or become stifled by the intragranularly nucleated plates of acicular ferrite.

By the time the weld deposit cools to about 500 °C, most of the austenite has been consumed. The small quantity of remaining austenite (about 5%) is enriched in carbon and either transforms to martensite, or into pearlite which is degenerate because it does not have the opportunity to establish a lamellar structure. Slower cooling rates favour the formation of pearlite relative to martensite. Some austenite may also be retained to ambient temperature. Because of their small volume fractions in the overall microstructure, these phases are, in welding terminology, called "microphases". The microphases are relatively hard and behave in many respects like brittle inclusions. They are therefore of importance n determining the toughness of weld deposits.

Sensitivity to carbon

It is striking that small variations in carbon concentration can have a major influence on the microstructure of welds, especially since the average carbon concentration of a weld is usually kept very small. It is apparent from the previous discussions of the growth rates of allotriomorphic and Widmanstätten ferrite, that the sensitivity of growth kinetics to carbon becomes larger as the concentration of carbon decreases.

These are important observations given that the general trend in the steel industry is to reduce the carbon concentration, sometimes to levels approaching the maximum solubility of carbon in ferrite. The rate at which ferrite grows increases sharply as the carbon concentration of the steel approaches its solubility in ferrite. This is because there is no need for the carbon to diffuse ahead of the γ/α interface, since it can all be accommodated in the ferrite.

Hence, the effect of carbon is seen to be larger (Figs. 2,4) when its concentration changes from $0.03 \rightarrow 0.05\,\mathrm{wt\%}$, when compared with the change from $0.09 \rightarrow 0.11\,\mathrm{wt\%}$. Changes in mechanical properties are found to reflect this behaviour, the strength of low–carbon steels being particularly sensitive to the carbon concentration. This increased sensitivity of the γ/α transformation to carbon at low concentrations, leads to a corresponding decreased sensitivity to substitutional alloying elements. Carbon is in effect controlling the kinetics of transformation.

In welding, the hardenability of the steel is often expressed as a carbon equivalent (CE). The concentration of each solute is scaled by a coefficient which expresses its ability, relative to carbon, to retard the γ/α transformation. Steels with a carbon equivalent in excess of about 0.4 wt% cannot easily be welded because of their increased tendency to form martensite. There are in fact two popular expressions for the carbon equivalent, one due to the International Institute for Welding (I.I.W.), and the other due to Ito and Besseyo, covering the high and low ranges of carbon respectively:

$$\mathrm{I.I.W.} \qquad > 0.18\,\mathrm{wt\%\ C}$$

$$\mathrm{CE} = \mathrm{C} + \frac{\mathrm{Mn}+\mathrm{Si}}{6} + \frac{\mathrm{Ni}+\mathrm{Cu}}{15} + \frac{\mathrm{Cr}+\mathrm{Mo}+\mathrm{V}}{5} \qquad \mathrm{wt\%} \tag{4}$$

$$\mathrm{Ito-Besseyo}, \qquad < 0.18\,\mathrm{wt\%\ C}$$

$$\mathrm{CE} = \mathrm{C} + \frac{\mathrm{Si}}{30} + \frac{\mathrm{Mn} + \mathrm{Cu} + \mathrm{Cr}}{20} + \frac{\mathrm{Ni}}{60} + \frac{\mathrm{Mo}}{15} + \frac{\mathrm{V}}{10} + 5\mathrm{B} \qquad \mathrm{wt\%} \tag{5}$$

The Ito–Besseyo carbon equivalent formula has smaller coefficients for the substitutional solutes when compared with the IIW formula. It is believed to be more reliable for low carbon steels. The IIW formula shows much smaller tolerance to substitutional alloying elements than the Ito–Besseyo equation. At already discussed, with low carbon concentrations, the kinetics of transformation are more sensitive to carbon than to substitutional solutes. Hence, it is logical that there should be two different empirical expressions for the carbon equivalent, for the low and high carbon weldable steels. Fig. 6 illustrates that as expected, both the microstructure and mechanical properties change more rapidly at low carbon concentrations.

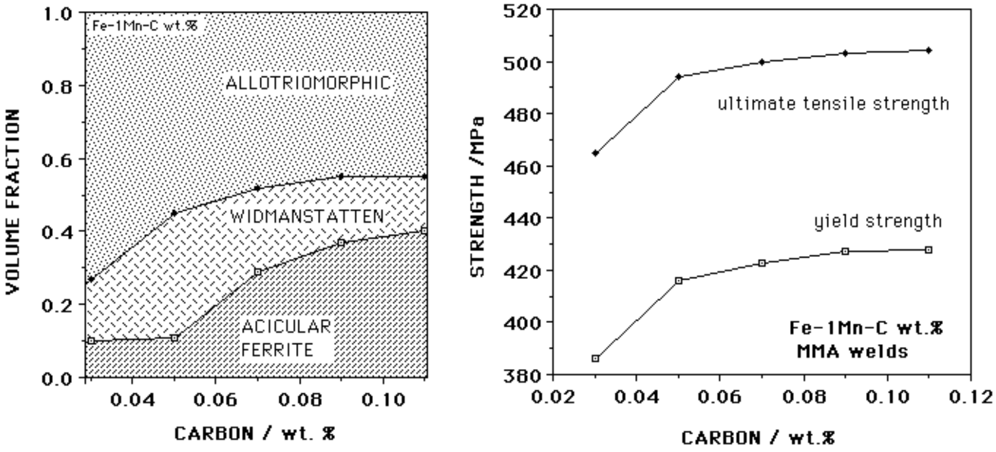

Fig. 6: Variations in microstructure and mechanical properties as a function of carbon concentration in Fe–1Mn–C wt% steel weld deposit using manual metal arc welding (1 kJ mm^{-1}).