14 *Other Aspects*

14.1 Bainite in Iron and its Substitutional Alloys

Bainite can be distinguished in high-purity iron or iron alloys which have an interstitial content that is less than 0.01 wt%. Kinetic experiments are difficult to conduct because the rate of reaction can be very large. The classification of microstructure therefore has to be based on limited evidence. The question arises as to whether there is any essential difference between martensite and bainite in interstitial-free alloys.

Many experiments have been conducted by monitoring the temperature during rapid cooling. The evolution of latent heat due to transformation can cause inflexions in the cooling curves, which in turn indicate the onset of transformation. Continuous cooling experiments on a Fe–14.43Ni–0.01C wt% alloy revealed two plateaux in a plot of the thermal arrest temperature versus the cooling rate (Wilson *et al.*, 1982). The plateau at the higher temperature was identified with bainite, the other with martensite but the microstructural evidence cited to support this interpretation was simply that the lath width was larger for the higher temperature arrest.

There does not seem to be satisfactory evidence to suggest that there is any difference in the mechanism of transformation between bainite and martensite in interstitial-free alloys. On the basis of the theory discussed in Section 6.1.2, the difference between bainite and martensite should vanish as the carbon concentration is reduced to zero.

14.2 The Weldability of Bainitic Steels

The region which is adjacent to the fusion zone of a weld is influenced by heat diffusion from the fusion zone. This region is the *heat-affected zone* (HAZ). Its boundaries need not be precisely defined because the definition depends on purpose. The heat dissipated into the HAZ can be detected as the temperature at any point rises to a maximum and then drops gently towards the far-field temperature. The severity of the heating or cooling cycles, and the peak temperature, depends on the location within the HAZ (e.g. Easterling, 1983;

Lancaster, 1986; Grong, 1997). For steels with a high hardenability, regions of the HAZ which have been austenitised by the heat pulse may transform during cooling into untempered martensite or to some other hard microstructure. These hard regions are susceptible to cold-cracking due to hydrogen embrittlement and other impurity effects.

This is the main reason why hardenable steels are difficult or impossible to weld (Fig. 14.1). The cooling rate can be reduced during welding, to avoid the formation of martensite in the HAZ. This can be done by heating the sample before welding begins. But this *preheating* adds to the cost of manufacture. It has been estimated, for example, that the cost of fabricating an aircraft carrier could be reduced by about £3 million if the number of panels requiring preheat can be reduced by 50% (Cullison, 1991).

The cold cracking susceptibility correlates empirically with a carbon equivalent (CE), which is a measure of hardenability. There are two popular formulae. The first one is a slightly modified version of the equation originally proposed by Dearden and O'Neill, and adopted by the International Institute for Welding. It is applied to steels containing less than 0.18 wt% of carbon:

$$\text{IIW} \quad > 0.18\,\text{wt\% C}$$

$$\text{CE} = \text{C} + \frac{\text{Mn} + \text{Si}}{6} + \frac{\text{Ni} + \text{Cu}}{15} + \frac{\text{Cr} + \text{Mo} + \text{V}}{5} \quad \text{wt\%} \tag{14.1}$$

where all the concentrations are in wt%. The other equation, due to Ito and Besseyo, has been adopted by the Japanese Welding Engineering Society:

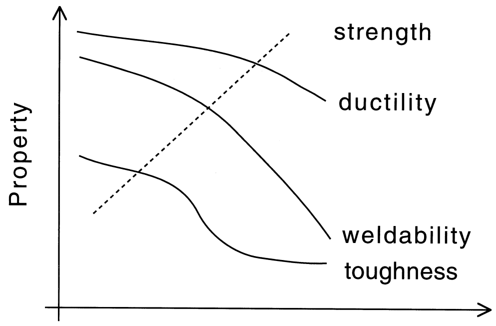

Fig. 14.1 Variation in mechanical properties of the heat-affected zone as a function of the carbon equivalent.

$$\text{Ito} - \text{Besseyo} \quad < 0.18\,\text{wt\% C}$$

$$\text{CE} = \text{C} + \frac{\text{Si}}{30} + \frac{\text{Mn} + \text{Cu} + \text{Cr}}{20} + \frac{\text{Ni}}{60} + \frac{\text{Mo}}{15} + \frac{\text{V}}{10} + 5\text{B} \quad \text{wt\%} \tag{14.2}$$

It is generally accepted that if the carbon equivalent is between 0.35 and 0.55 wt%, then the sample must be preheated prior to welding (the preheat temperature can be as high as 400 °C), and when CE $>$ 0.55, both preheating and postheating is considered essential to avoid cold-cracking and other difficulties.

The two equations give different values of CE, the Ito and Besseyo method taking a more conservative account of alloying additions. Equation 14.2 is appropriate for modern low-carbon, low-alloy steels such as the ultra-low-carbon bainitic steels (C $\simeq 0.01 \rightarrow 0.03$ wt%, e.g. Nakasugi *et al.*, 1980, Lorenz and Duren, 1983). For these alloys, the IIW CE gives a pessimistic assessment of weldability whereas the Ito and Besseyo equation works well. It has also been demonstrated (Lorenz and Duren, 1983) that for low carbon pipeline steels, the IIW CE overestimates the effects of alloying elements like manganese and molybdenum, a more realistic CE being given by:

$$\text{CE} = \text{C} + \frac{\text{Si}}{25} + \frac{\text{Mn} + \text{Cu}}{16} + \frac{\text{Cr}}{20} + \frac{\text{Ni}}{60} + \frac{\text{Mo}}{40} + \frac{\text{V}}{15} \quad \text{wt\%} \tag{14.3}$$

for weld cooling times of 2–3 seconds over the temperature range 800–500 °C (these conditions are typical for girth welds in pipelines).

There are good reasons for supposing that the same CE should not apply to medium-carbon and low-carbon steels. There is a disproportionate increase in the growth rates of both allotriomorphic and Widmanstätten ferrite as the carbon concentration drops below $\simeq$ 0.06 wt%, when compared with variations in carbon above this value (Bhadeshia *et al.*, 1985; Bhadeshia, 1990). This is because the average carbon concentration of the alloy approaches the equilibrium solubility of carbon in ferrite. The need to partition carbon into the austenite is thus reduced so that the diffusion-controlled velocity rises sharply.

14.3 Electrical Resistance

It follows from the Bloch theorem that anything which disrupts the periodic potential of the lattice causes an increase in the electrical resistance. Thermal vibrations, dislocations, solute atoms and other point defects therefore all contribute to electrical resistance.

The role of dissolved carbon has been modelled by Hoffman and Cohen (1973) using an analogy with the dynamic displacements associated with thermal vibrations. The analogy seems to work well even though the displacements due to carbon are static. Thermal vibrations involve dynamic displacements whose mean square value $\overline{u_T^2}$ is given by the Debye theory as:

$$\overline{u_T^2} - \overline{u_0^2} = \frac{145}{MT_D}\left(\frac{T}{T_D}\right)^2 \int_0^{T/T_D} \frac{vdv}{\exp\{v\} - 1} \tag{14.4}$$

where $\overline{u_0^2}$ represent the zero-point quantum vibrations (the units of the displacements are in Å), M is the atomic weight of iron in grams, v is a dummy integration variable. Using equation 14.4, $\overline{u_T^2} - \overline{u_0^2}$ is found to be $3.25 \times 10^{-5}\,\mathrm{nm}^2$ at 295 K at which temperature the measured resistivity $\rho_T = 9.8\,\mu\Omega\,\mathrm{cm}$. The ratio of the resistivity to mean square displacement is to a good approximation found to be constant. If the static displacements due to carbon are known, then this ratio can be used to estimate its contribution to resistivity. The ratio is found to be about $30.3\,\mu\Omega\,\mathrm{cm}$ per wt% C, in excellent agreement with a variety of experimental measurements on martensite.

Bainite is expected to have a lower electrical resistivity than martensite because it has less carbon in solid solution and a lower defect density. On the other hand, its dislocation density is larger than that of pearlite or allotriomorphic ferrite (Chapter 2). Therefore, the electrical resistivity of a specimen fully reacted to bainite is always found to be higher than that of pearlite at the same temperature (Radcliffe and Rollason, 1959). The resistivity decreases in the order austenite, martensite, bainite and pearlite for a given temperature; for a constant microstructure it decreases with temperature (Fig. 14.2).

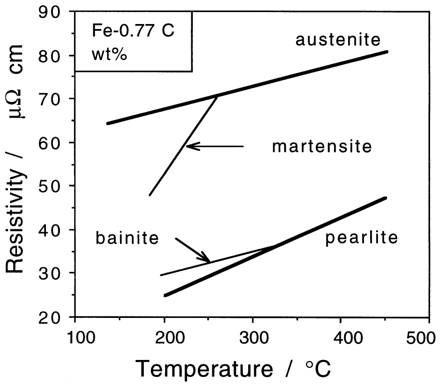

Fig. 14.2 Electrical resistivity of a variety of microstructures in steels (Radcliffe and Rollason, 1959).

14.4 Internal Friction

A material is said to be elastic when it exhibits a stress–strain curve which is fully reversible. The removal of stress eliminates the strain. The energy stored in the material when under stress is fully recovered. When such a material vibrates in a vacuum it may continue vibrating for an indefinite period of time.

Similar vibrations would decay naturally in an *anelastic* solid, since energy is dissipated by some process occurring within the sample during each vibration. The vibrations are said to be damped. An examination of the damping as a function of temperature and frequency can reveal information about the nature of the dissipative process. *Internal friction* measurements like these can be used to detect the onset of transformations, since moving interfaces can damp oscillations.

Internal friction measurements conducted during the continuous cooling transformation of a commercial steel to bainite have been interpreted to indicate a *prebainitic* microstructural change before the formation of bainite proper (Jihua *et al.*, 1989). The argument is based on an observed rise in damping during continuous cooling, at temperatures above B_S. These experiments are not supported by microstructural evidence nor is the association with bainite proven.

The same experiments have demonstrated that the degree of damping decreases monotonically as the transformation progresses, indicating that the concentration of dissipative units (whatever they may be) varies directly with the extent of reaction. The damping at any instant of time, increases as the temperature is reduced below B_S. This is expected because the total amount of bainite that can form increases with undercooling below B_S.

14.5 Internal Stress

It has long been recognised that the transformation of austenite to martensite causes the development of stresses which are retained in the transformed specimen. These *residual stresses* are usually attributed to the volume change due to transformation (Buhler *et al.*, 1932; Buhler and Scheil, 1933; Scheil, 1955; Buhler, 1955; Hildenwall, 1979).

The volume expansion is not unique to martensite; allotriomorphic ferrite, pearlite, Widmanstätten ferrite, and bainite all cause a decrease in density. There are no data for Widmanstätten ferrite, but the formation of bainite generates residual stresses (Radcliffe and Rollason, 1959; Diesburg *et al.*, 1981). As a general rule, X-ray diffraction peaks from transformations which are displacive (martensite, bainite, Widmanstätten ferrite) are found to be more diffuse than those from reconstructive reactions (allotriomorphic ferrite, pearlite). For example, Radcliffe and Rollason demonstrated a larger lattice strain with martensite and bainite than with pearlite. The diffusion that occurs during reconstructive transformation help accommodate the volume change, preventing the development of stresses.

The residual stresses develop mainly because transformation does not usually occur uniformly in all regions of the sample. This can be exploited for case-hardened components, where it is advantageous to have a compressive stress on the surface of the component. The compressive stress prolongs the fatigue life and makes the component more resistant to surface initiated fracture. In steels which are surface carburised and then quenched, the lower carbon core transforms at a higher temperature. The resulting core–volume expansion puts the still austenitic surface regions into tension, though the tensile stress is partly relaxed by plastic deformation. When the surface region eventually transforms to martensite on further cooling, its volume expansion causes stress reversal, so the surface ends up in compression relative to the core (Koistinen, 1958).

Because of the smaller volume expansion that accompanies the transformation to bainite (Goldak *et al.*, 1985), and since plastic relaxation eases at higher temperatures, a bainitic case is not as effective in introducing a compressive stress at the surface when compared with a martensitic case (Diesburg *et al.*, 1981). Samples containing bainite in the case have lower levels of compressive residual surface stresses. Thus, the performance of case-hardened samples can be improved by adding elements such as molybdenum which encourage martensite to form at the expense of bainite.

14.6 Bainite in Iron–Nitrogen Alloys

Both nitrogen and carbon exist in interstitial sites in iron and their respective binary phase diagrams with iron show eutectoid reactions in which austenite decomposes into a mixture of ferrite and carbide or ferrite and nitride (Fe_4N). It is therefore reasonable to expect similar sorts of phase transformations to occur in both alloy systems. It is well established that martensite can form in both Fe–C and Fe–N alloys, but the first report of bainite in an Fe–N alloy was by Bell and Farnell (1969). A Fe–1.8N wt% alloy when transformed isothermally at 350 °C was observed using light microscopy to contain ferrite and Fe_4N with an appearance similar to that of upper bainite in Fe–C alloys. The transformation products were stifled in their growth by austenite twin boundaries, consistent with growth in which there is a co-ordinated movement of atoms.

Foct *et al.* (1988) showed that in a Fe–9N at.% alloy, the transformation to bainite is sometimes preceded by the precipitation of Fe_4N. The resulting localised depletion of nitrogen then stimulates the austenite to transform into bainitic ferrite and more Fe_4N, without any change in the lattice parameter of the residual austenite.

Whilst there is evidence for the existence of a bainite reaction in Fe–N alloys, it would be useful to conduct a detailed microstructural characterisation, thermodynamic analysis and crystallographic experiments including the study of surface relief.

14.7 Effect of Hydrogen on Bainite Formation

Hydrogen has a bad reputation in the context of steels because when in solution, it undoubtedly embrittles ferrite. However, there are examples in titanium metallurgy where it is introduced temporarily to enable processing, after which it is removed by heat treatment. With this in mind, Yalci and Edmonds (1999) conducted what is probably the first study on the influence of hydrogen on the microstructure and properties of upper bainite.

The studies were conducted on silicon-rich steels to avoid the formation of cementite. The hydrogen was introduced at a pressure of two atmospheres, whilst the alloys were in the austenite phase field. This was followed immediately by isothermal transformation in the bainite temperature range.

The introduction of hydrogen apparently led to a greater amount of upper bainite and the thickness of bainite plates was reduced from 0.31 ± 0.06 μm to 0.21 ± 0.09 μm in the hydrogenated alloy. The hardness of the hydrogenated samples was measured to be greater than those which were simply heat treated in helium (Table 14.1). The increase in hardness in the hydrogenated samples is consistent with an increase in the fraction of bainite, since ferrite is at low temperatures harder than austenite.

Table 14.1 Hardness of Fe–0.2C–3Mn–2.1Si wt% and Fe–0.4C–4.09Ni–1.99Si wt% alloy transformed isothermally at 390 °C after austenitisation at 920 °C for 30 min. The samples were sealed in chambers containing either helium or hydrogen at 2 atmospheres pressure throughout these heat-treatments. After Yalci and Edmonds (1999). HV_{30} refers to the Vicker's hardness measured using a 30 kg load.

Alloy	Mn-containing alloy		Ni-containing alloy	
Environment	Helium	Hydrogen	Helium	Hydrogen
HV_{30}	367	409	364	383

14.8 Magnetically-Induced Bainite

It has long been known the magnetic fields must influence the transformation from austenite to martensite (Krivoglaz and Sadovskii, 1964; Kekeshita *et al.*, 1985). The two phases have different magnetic properties so the application of a magnetic field encourages the formation of the ferromagnetic martensite.

Ohtsuka and coworkers (2000) have recently verified the same effect of an externally applied magnetic field on the bainite transformation. The major effect of the field is to accelerate transformation (Fig. 14.3).

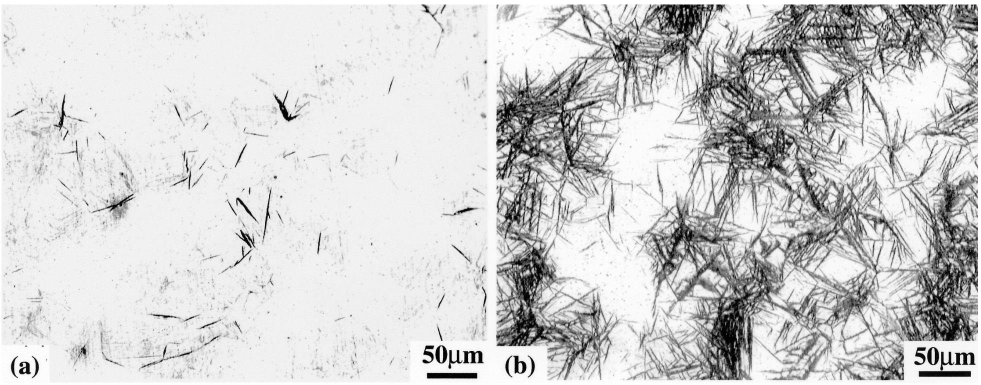

Fig. 14.3 A Fe–0.52C–0.24Si–0.84Mn–1.76Ni–1.27Cr–0.35Mo–0.13V wt% steel austenitised at 1273 K for 600 s and transformed isothermally to bainite at 573 K for 480 s, followed by helium quenching to ambient temperature: (a) zero magnetic field; (b) sample under the influence of a 10 Tesla magnetic field during transformation.