

## 13 *Modern Bainitic Steels*

Steels with yield strengths in excess of 1000 MPa are important in certain applications, but the biggest commercial markets are for lower strength varieties, where the total alloy content rarely exceeds 2 wt%. The alloy design therefore has to be careful with a proper balance of hardenability in the context of large scale steel production technologies. Lean steels tend to transform into mixtures of allotriomorphic ferrite and bainite, whereas any attempt to improve hardenability usually leads to partly martensitic microstructures. The solution therefore lies in low-alloy, low-carbon steels, containing small amounts of boron and molybdenum to suppress allotriomorphic ferrite formation, principles established originally by Irvine and Pickering. Boron increases the bainitic hardenability. Other solute additions can, in the presence of boron, be kept at sufficiently low concentrations to avoid the formation of martensite. Steels like these (Alloy 1, Table 13.1), when normalised, are found to transform almost completely into bainite with only small fractions of martensite or other residual phases.

The other alloys listed in Table 13.1 are examples of some of the latest commercial bainitic steels. A striking feature of the list is that they don't look particularly different in chemical composition. This impression is misleading because each steel has distinct mechanical properties. Trace element concentrations are vital in determining the details of the microstructure as are the thermomechanical processing routes used in their manufacture.

Any alloy development or 'advanced material' needs to be considered in the context of what already exists. For this reason, we first discuss the incredibly successful ferrite–pearlite microalloyed steels which set the standard by which any new alloy system must be assessed.

### 13.1 **Alternatives to the Ferrite–Pearlite Microstructure**

The most popular microstructure in the context of structural steels has undoubtedly been a mixture of ferrite and pearlite. There are hundreds of millions of tonnes of this manufactured annually. Typical chemical compositions are given in Table 13.2 together with an indication of the usual mechanical properties. Figure 13.1 shows that the main effect of microalloying and the associated thermomechanical processing is to refine the microstructure. The

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**Table 13.1** Typical compositions of advanced bainitic steels (wt%). Notice that the steels are fairly similar in composition with respect to the major alloying additions, although there are significant differences with respect to carbon and trace element concentrations. The fabrication procedure is also of vital importance in giving a large variation in properties in spite of the similarities of chemical composition. In the inoculated steels, the titanium oxide-bearing steel is the most effective.

No.	C	Si	Mn	Ni	Mo	Cr	Nb	Ti	B	Al	N	Others	Steel Type
1	0.100	0.25	0.50	–	0.55	–	–	–	0.0030	–	–	–	Early bainitic
2	0.039	0.20	1.55	0.20	–	–	0.042	0.015	0.0013	0.024	0.0030	–	Rapidly cooled bainitic
3	0.081	0.25	1.86	0.20	0.09	–	0.045	0.016	–	0.025	0.0028	–	Rapidly cooled bainitic
4	0.110	0.34	1.51	–	–	–	0.029	–	–	–	–	–	Rapidly cooled bainitic
5	0.100	0.25	1.00	–	–	–	–	–	–	–	–	–	Bainitic dual phase
6	0.040	–	0.40	–	–	–	–	–	–	0.05	–	–	Triple phase
7	0.150	0.35	1.40	–	–	–	0.022	0.011	–	0.035	–	–	Bainitic dual phase
8	0.120	1.50	1.50	–	–	–	–	–	–	0.045	0.0035	–	TRIP-assisted
9	0.020	0.20	2.00	0.30	0.30	–	0.050	0.020	0.0010	–	0.0025	–	ULCB
10	0.028	0.25	1.75	0.20	–	0.30	0.100	0.015	–	0.030	0.0035	Cu 0.3, Ca 0.004	ULCB
11	0.080	0.20	1.40	–	–	–	–	0.012	–	0.002	0.0020	O 0.0017	acicular ferrite, TiO <sub>x</sub>
12	0.080	0.20	1.40	–	–	–	–	0.008	0.0015	0.038	0.0028	–	acicular ferrite, TiB
13	0.080	0.20	1.40	–	–	–	–	0.019	–	0.018	0.0050	–	acicular ferrite, TiN
14	0.15	0.80	1.40	–	0.20	–	–	–	–	–	–	V 0.15	Forging (high strength)
15	0.09	0.25	1.00	0.50	1.00	–	0.10	0.02	0.002	0.04	0.006	–	Forging (100% bainitic)
16	0.09	0.40	1.40	–	–	–	0.07	–	–	0.04	0.010	V 0.06	Forging (Nb + V)
17	0.09	0.25	1.40	–	–	–	0.07	0.02	0.002	0.04	0.006	–	Forging (Nb + B)
18	0.012	–	1.60	–	–	–	0.08	–	0.004	–	–	–	Cold-heading

**Table 13.2** Typical ferrite–pearlite structural steels in both the standard and niobium microalloyed conditions.

Type	Composition, wt.%				Stress, MPa				
	C	Si	Mn	Nb	$\sigma_y^l$	$\sigma_y^u$	UTS	Elongation %	$C_V (-20^\circ\text{C}), \text{J}$
Standard	0.11	0.21	1.24		285	320	480	36	55–90
Microalloyed	0.11	0.30	1.40	0.034	395	430	525	32	100–190

Type	Ferrite %	Pearlite %	Ferrite grain size	Pearlite interlammellar spacing
Standard	76	24	15 $\mu\text{m}$	0.22 $\mu\text{m}$
Microalloyed	78	22	9 $\mu\text{m}$	0.22 $\mu\text{m}$

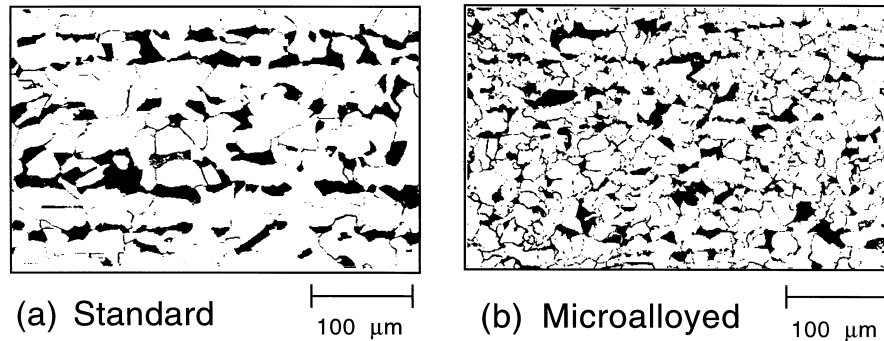


Fig. 13.1 Optical micrographs of banded ferrite–pearlite microstructures in (a) standard, (b) niobium microalloyed structural steel.

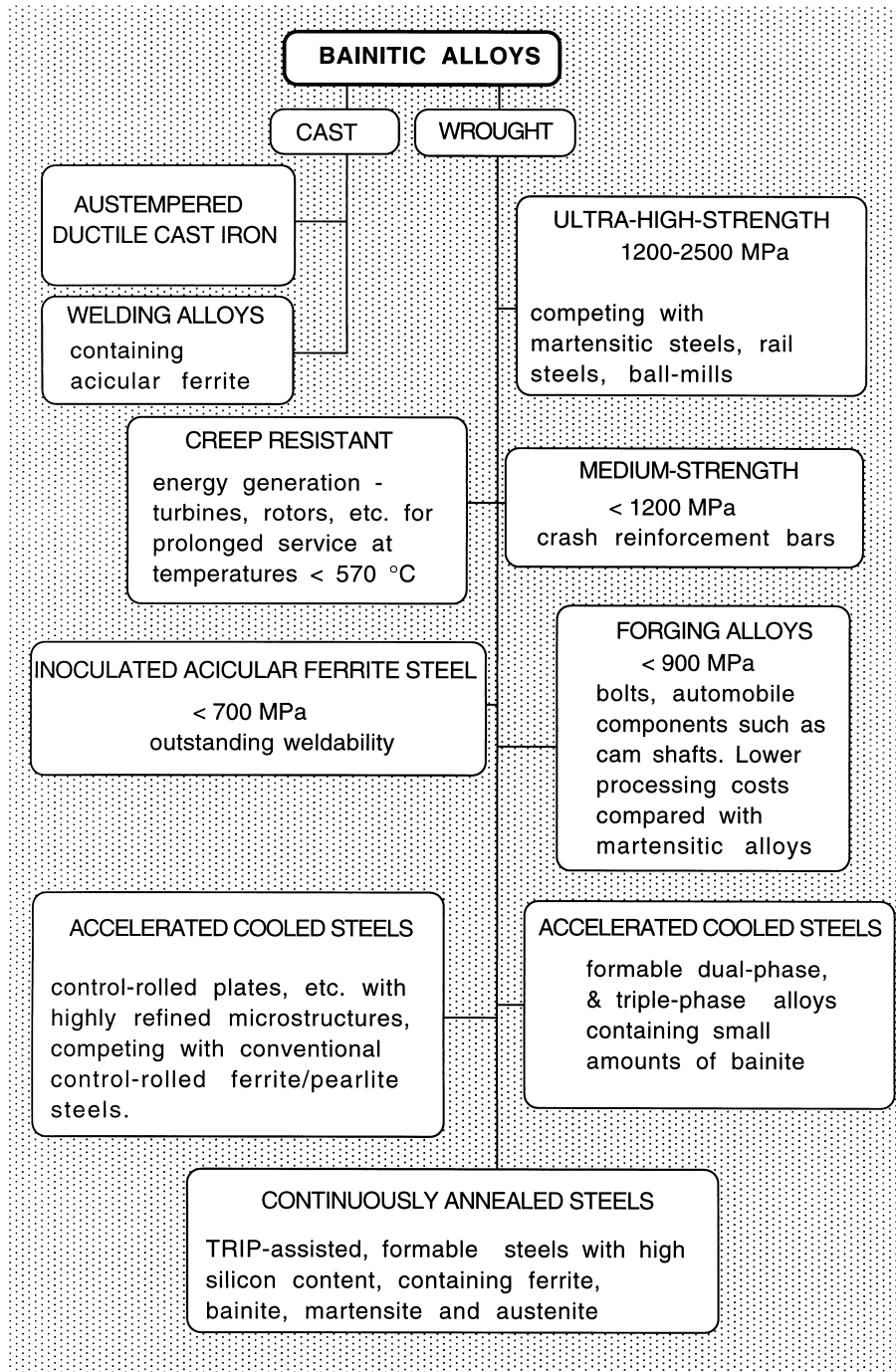
niobium carbonitrides that form also strengthen the ferrite by interphase precipitation or by strain-induced precipitation (Pickering, 1978, 1992; Honeycombe and Bhadeshia, 1995; Gladman, 1997).

Some of the major applications of structural steels include: oil and gas pipelines ( $\sigma_y = 350\text{--}485$  MPa), drilling rigs and production platforms (345–415 MPa), ships (315–450 MPa), pressure vessels and tubing, earth-moving equipment (345–550 MPa), heavy goods vehicles and automobile components, high-rise buildings (415–530 MPa), bridges (345–485 MPa), transmission towers, fuel and other storage tanks and reinforcement bars for concrete (Pickering, 1992; Bodnar *et al.*, 1997). A few of the attributes that are consistent with these applications include the low cost, ability to produce a variety of forms, weldability, fabricability, reliability under extreme conditions (such as fire and earthquakes and other *force majeure*s).

It seems unlikely in this context that the dominant position of the ferrite–pearlite steels will ever be challenged by alternative steel microstructures, let alone other materials! Nevertheless, we shall describe rivals based on bainite because such steels can be produced using a similar production route and without any additional heat treatments.

## 13.2 Strength

Strength is a basic engineering specification; steels can be made stronger by transforming the austenite at ever decreasing temperatures (Fig. 12.4) or by increasing hardenability. There are many other strengthening mechanisms available. Unfortunately, toughness does not necessarily increase with strength. The weldability may also deteriorate because the excessive use of alloying elements can cause the formation of untempered hard-phases in the austenitised regions of the heat-affected zone or there may be excessive softening in the tempered regions of the HAZ.



**Fig. 13.2** The range of bainitic steels currently available on a commercial basis. *TRIP* stands for transformation-induced plasticity.

The tendency to soften depends on how far the original microstructure deviates from equilibrium (Table 13.3). Large deviations naturally lead to greater rates of softening because the driving force for tempering is the energy stored within the material; the heat simply provides the thermal activation needed to bump the structure to lower energies. Higher strength steels are therefore more likely to be difficult to weld.

### 13.3 Bainitic Steels

It is evident that ferrite/pearlite steels are outstanding on any performance criterion which includes cost and reliability. They will certainly not be threatened by the alternative microstructures to be discussed here. On the other hand, the novel microstructures offer a better set of properties for high performance applications where their additional cost is not an issue.

The steels developed by Irvine and Pickering exhibited quite reasonable combinations of toughness and strength, but in time proved to fall short of the best quenched and tempered martensitic steels. Nevertheless, the physical metallurgy principles established during their development have been applied in the metallurgical design of a new generation of bainitic steels, in which the emphasis is on reductions in carbon and other alloying element concentrations, and on processing designed to refine the microstructure. We now proceed to describe developments in these high-technology steels and cast irons based on bainitic microstructures. Many of the lower alloy content steels are thermomechanically processed prior to their transformation so it is appropriate to begin

**Table 13.3** The stored energy as a function of microstructure, relative to the equilibrium state defined as a mixture of ferrite, cementite and graphite (Bhadeshia, 1998). The phases in cases 1 and 2 involve a partitioning of all elements so as to minimise free energy. In cases 3–5 the iron and substitutional solute are configurationally frozen (for martensite even the interstitial elements are frozen). Case 6 refers to an iron-base mechanically alloyed oxide-dispersion strengthened (ODS) sample which has the highest reported stored energy prior to recrystallisation (Bhadeshia, 1997)

Phase Mixture in Fe–0.2C–1.5Mn wt% at 300 K	Stored Energy, J mol <sup>-1</sup>
1. Ferrite, graphite & cementite	0
2. Ferrite & cementite	70
3. Paraequilibrium ferrite & paraequilibrium cementite	385
4. Bainite and paraequilibrium cementite	785
5. Martensite	1214
6. Mechanically alloyed ODS metal	55

with a description of the key industrial processes in the context of bainite. The range of bainitic alloys currently in use is summarised in Fig. 13.2; the inoculated acicular ferrite steels have already been dealt with in Chapter 10.

### 13.4 Controlled-Rolling of Bainitic Steels

The strengthening of iron via a reduction in grain size is an attractive option because a small grain size leads also to an improvement in toughness. This simple fact has led to the development of impressive thermomechanical processing technology capable of refining the austenite grain structure prior to its transformation to ferrite (Fig. 13.3). A fine austenite grain size leads to a correspondingly refined ferrite grain structure. The *controlled-rolling* process

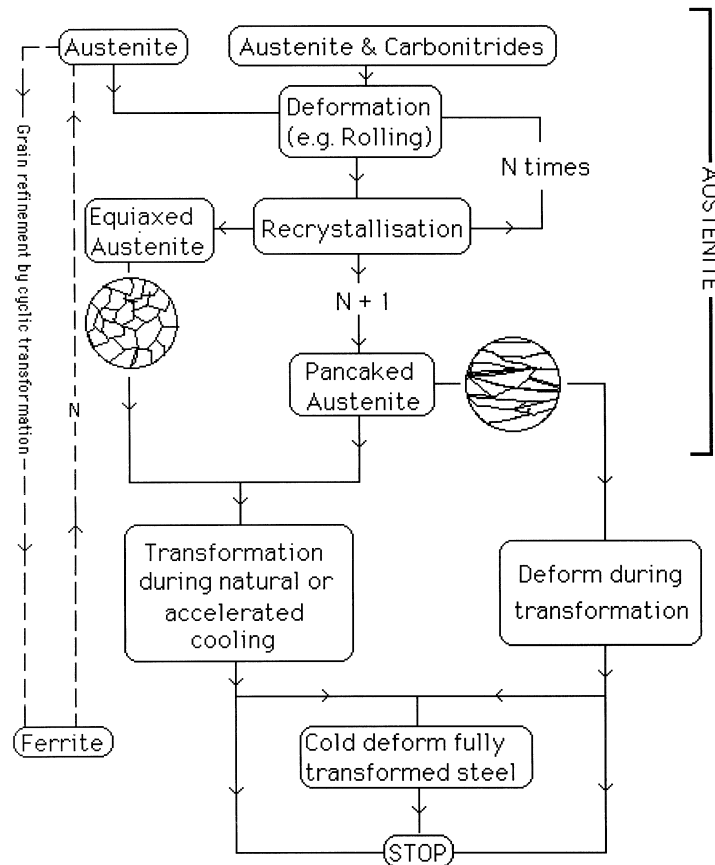


Fig. 13.3 Flow chart illustrating some of the thermomechanical processing routes available for the manufacture of steels.

involves a sophisticated rolling deformation of austenite. The subject has been reviewed by Speich *et al.* (1984), where the complex procedures and the variety of rolling and alloying practices are discussed in detail. The essence of the process is the reduction of ingots by hot-rolling in the austenite phase field, such that the austenite is induced to recrystallise many times before the finish-rolling temperature is reached. This gives a fine recrystallised austenite grain structure. The growth of these grains during the hot-rolling process is hindered by the use of microalloying additions such as niobium or titanium. These elements are added in small concentrations ( $\approx 0.01$ – $0.03$  wt%) to form stable carbides or carbonitrides which impede austenite grain growth. As will be seen later, it can be an advantage if the austenite grains are pancaked (flattened) because such grains transform into finer ferrite. A typical chemical composition of a steel suitable for control-rolling into a ferritic microstructure is Fe–0.08C–0.3Si–1Mn–0.03Nb–0.004N wt%.

Controlled-rolling has been used successfully for over 40 years to produce steels with a ferrite and pearlite microstructure; it has since been adapted for bainitic alloys (Nakasugi *et al.*, 1980, 1983). There are two ways in which a bainitic microstructure can be obtained: the first involves an increase in the cooling rate in order to allow the austenite to supercool into the bainite transformation range. The second is to modify the steel hardenability without substantially changing the processing conditions. Alloying elements such as manganese are boosted in order to retard the formation of allotriomorphic ferrite relative to bainite. Unlike conventional steels for control-rolling, TiN particles (of size  $\approx 0.02$   $\mu\text{m}$ ) are induced to precipitate during solidification and subsequent cooling to ambient temperature. The precipitation of fine TiN is stimulated by increasing the cooling rate of the molten steel during continuous casting. Slabs of the material are then reheated to a low austenitisation temperature of 1150 °C, the TiN particles inhibiting austenite grain growth. The grain size is reduced during repeated recrystallisation due to control-rolling. By inhibiting grain growth, the particles also help to produce a uniform grain structure than is obtained in conventional processes. Finish rolling is carried out at a temperature where recrystallisation does not occur within the time scale of the rolling sequence; the austenite grains are therefore pancaked and have a deformed microstructure just before they start to transform into bainite.

The details of the steelmaking process are important in determining the final properties of control-rolled steels. The higher quality steels are dephosphorised, desulphurised and vacuum degassed prior to casting. Typical concentrations of phosphorous and sulphur after these treatments are 0.015 and 0.0015 wt% respectively. In circumstances where formability and uniform ductility are important, the steel is usually treated with calcium which has the effect of fixing sulphur and of modifying the shape of the sulphide inclusions.

### 13.4.1 Crystallographic Texture

Polycrystalline materials do not in general contain grains which are randomly oriented. The crystals tend to align along particular orientations determined by the thermomechanical history of the material. This alignment is called *crystallographic texture* and its most important manifestation is in the development of anisotropic mechanical properties. The anisotropy may be exploited, as in deep-drawing steels where the texture is optimised to reduce plastic strain in the thickness direction. It can, on the other hand, be detrimental if the cleavage planes show a tendency to align because the fracture path becomes continuous across many grains.

In wrought steels, it is deformation, recrystallisation or phase transformation which can cause crystallographic texture. All of these processes occur during control-rolling. There are excellent reviews on the subject (Tanaka, 1981; Ray and Jonas, 1990), but the purpose here is to highlight the differences of crystallographic texture which arise between conventional and bainitic control-rolled steels.

A convenient (though incomplete) way of representing the texture in rolled sheet is by stating the set  $\{hkl\} \langle uvw \rangle$ , where  $\{hkl\}$  are the Miller indices of the planes which lie roughly parallel to the rolling plane, and  $\langle uvw \rangle$  the Miller indices of the direction in  $\{hkl\}$ , which tends to be parallel to the rolling direction. The texture can sometimes be deconvoluted into components:

$$\text{texture} = \sum_i \lambda_i \{hkl\}_i \langle uvw \rangle_i \quad (13.1)$$

where  $\lambda$  represents the weighting given to a particular type of texture. The major components of the deformation texture of austenite are  $\{110\} \langle 1\bar{1}2 \rangle$  and  $\{112\} \langle 11\bar{1} \rangle$ , the so-called *brass* and *copper* textures respectively.

Because ferrite has an orientation relationship with the austenite, the brass texture gives rise to a  $\{332\} \langle 11\bar{3} \rangle$  ferrite texture, and the copper texture to a  $\{113\} \langle 1\bar{1}0 \rangle$  ferrite component (Fig. 13.4). It is found experimentally that the  $\{332\} \langle 11\bar{3} \rangle$  ferrite texture is beneficial to the deep drawing qualities of steels and for strength and toughness. Control-rolling should therefore aim to maximise the brass texture component of austenite.

The texture of steel obtained from recrystallised austenite is made up predominantly of  $\{100\} \langle 011 \rangle$  ferrite, originating from the  $\{100\} \langle 001 \rangle$  *cube* recrystallisation texture of the parent austenite. This variety of ferrite texture can in principle be detrimental to the through thickness mechanical properties of rolled steels, since the  $\{100\}$  ferrite cleavage planes tend to align with the rolling plane. The fraction of  $\{100\}$  normals within  $10^\circ$  of the plate normal tends to increase as the finish rolling temperature decreases, but there does not appear to be any systematic correlation with measures of toughness



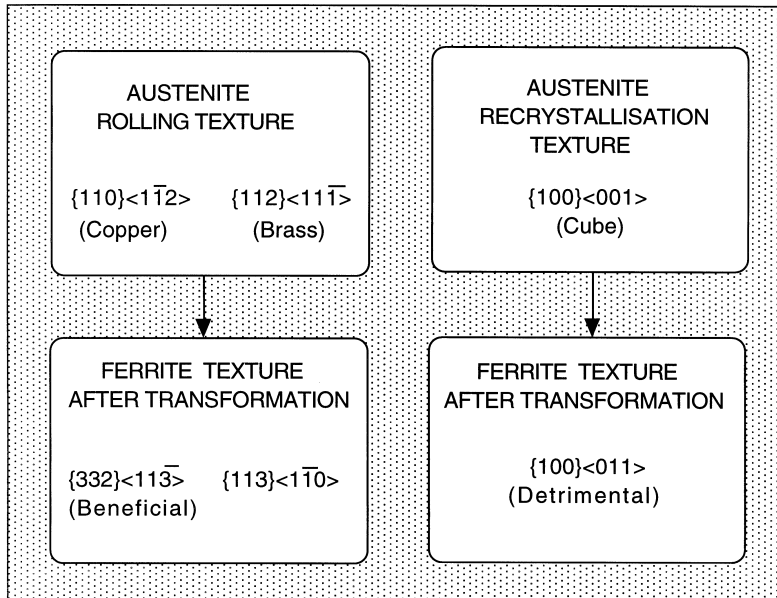
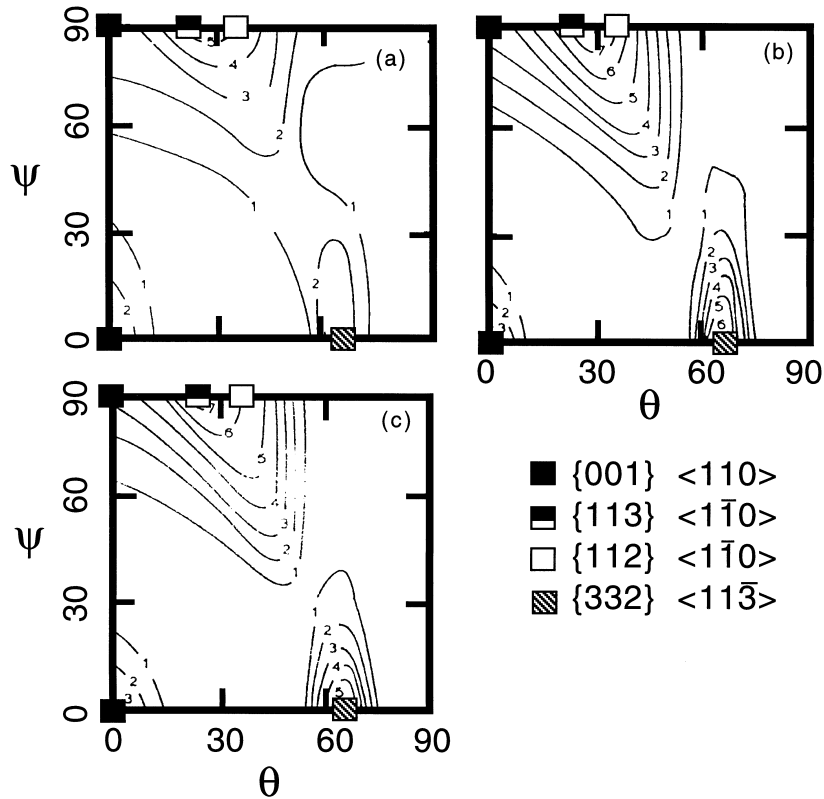


Fig. 13.4 Summary of crystallographic texture phenomena in wrought steels (adapted from Ray and Jonas, 1990).

(Davies *et al.*, 1977). It is possible that the texture is not uniform on a local scale. In fact, the cleavage process itself is not well defined for fine microstructures such as bainite, where the mode of brittle fracture is more accurately described as *quasi-cleavage*. In this, the cleavage planes are frequently disrupted by fibrous fracture at finely spaced plate boundaries. The complete story is therefore likely to be more complex given that inclusions such as MnS also tend to align parallel to the rolling plane. Whatever the detailed explanation, both the texture and inclusions make it easier for the steel to delaminate parallel to the rolling plane. The texture enhances the toughness along the length and width directions at the expense of the through-thickness properties. It is important therefore to ensure that the austenite is in a deformed, rather than in its recrystallised state just before it begins to transform.

Transformation textures associated with martensitic or bainitic microstructures (or acicular ferrite) are much more pronounced when compared with the case where the austenite transforms into allotriomorphic ferrite (Fig. 13.5). It is however, difficult to gauge the significance of the observations as far as delamination is concerned. Although the texture increases in intensity when the austenite transforms by a displacive mechanism, all of the components are strengthened. Any deleterious influence of the  $\{100\}\langle 011\rangle$  ferrite component may therefore be masked by stronger components, and



**Fig. 13.5** Sections along  $\phi = 45^\circ$  of the crystal orientation distribution functions showing transformation textures of control-rolled steels with a variety of microstructures (Yutori and Ogawa, 1979). For steels the most important features of the orientation distribution function are in the section of Euler space at  $\phi = 45^\circ$  because it contains orientations of the form  $\{100\} \langle uvw \rangle$  and the fibre textures  $\{hkl\} \langle 110 \rangle$  and  $\{111\} \langle uvw \rangle$ . (a) Allotriomorphic ferrite/pearlite; (b) acicular ferrite; (c) martensite. The points represent the exact textures indicated on the diagrams.

there do not seem to be any mechanical property data to indicate that bainitic steels which have been control-rolled have a greater propensity for delamination. Indeed, results to the contrary have been reported by Tamehiro *et al.* (1985b), who demonstrated that the delamination tendency is greater for conventional control-rolled steels, when compared with the more modern accelerated-cooled ferrite/bainite steels.

The observed differences in the strength of transformation textures are due to the displacive growth mechanisms of martensite, bainite and acicular ferrite;

the experimental data have been critically assessed by Ray and Jonas (1990). It appears to be a general result that when the transformation mechanism is reconstructive, all possible variants of the austenite/ferrite orientation relationship occur during transformation. There is therefore no *variant selection* during reconstructive transformations. On the other hand, attempts at predicting the product texture during displacive transformation, without assuming variant selection, have been unsuccessful in explaining experimental observations. The stresses and strains due to deformation favour the formation of particular variants in individual grains.

### 13.5 Rapidly Cooled Control-Rolled Steels

It was suggested in the previous section that bainitic microstructures can be generated in control-rolled steel either by increasing the hardenability or by changing the cooling rate during processing. The latter is the preferred route because the weldability deteriorates as the hardenability increases. The technology of rapid cooling during controlled-rolling is not trivial given the speed of production, the kinetics of transformation, the need to avoid distortion and nonuniform cooling. Progress has nevertheless been made and the rapidly cooled steels described here are commercially available. It is worth noting that rapidly cooled steels are often referred to as *accelerated cooled steels*.

Accelerated cooling plate mills are not essentially different from conventional rolling mills (Bodnar *et al.*, 1997). The distance from the finishing mill to the point where the plate can be removed from the mill can vary from 12–116 m. The cooling unit uses up to about 45 m of this distance with a variety of water and air systems to achieve a controlled cooling rate which can be as high as  $100 \text{ K s}^{-1}$  for a plate which is 10 mm thick.

#### 13.5.1 Pipeline and Plate Steels

There is a demand for a reduction in the wall-thickness and an increase in the diameter of pipelines for gas transmission. Thinner walls permit faster and less troublesome girth-welding operations. Thinner sections can be achieved by increasing the strength of the steels used, as long as toughness and weldability are not sacrificed in the process. When thickness considerations are not paramount, an increase in strength has the further advantage that the gas can be transmitted more efficiently under increased pressure ( $\approx 10 \text{ MPa}$ ).

It is found that if, after thermomechanical processing, the steel is cooled at a rate which is high enough to reduce the amount of allotriomorphic ferrite, but

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low enough to avoid martensite, then a fine-grained microstructure which is a mixture of allotriomorphic ferrite and bainite is obtained. Such a microstructure has the required higher yield strength and toughness. The cooling rates involved are larger ( $10\text{--}40\text{ }^{\circ}\text{C s}^{-1}$  over the temperature range  $800\text{--}500\text{ }^{\circ}\text{C}$ ) than those appropriate for normal control-rolling (Fig. 13.6). The accelerated cooling is achieved using water spray curtains directed on either side of the hot plate such that distortion is avoided; the plates can be as thick as 15 mm. The rapid cooling of thicker plate requires different technology with careful control of water pouring and to cope with the slow rate at which the plate moves through the mill.

It has been demonstrated that the microstructure of these rapidly cooled steels consists of a mixture of ferrite and bainite (Graf *et al.*, 1985). The bainite consists of sheaves of platelets of submicron thickness with a large dislocation density of  $1.7 \times 10^{14}\text{ m}^{-2}$ , compared with  $0.4 \times 10^{14}\text{ m}^{-2}$  in allotriomorphic ferrite. In fact, the dislocation density of the allotriomorphic ferrite in rapidly cooled steels is known to be about four times larger than in other steels containing ferrite, possibly because of plastic deformation caused by the formation of bainite (DeArdo, 1988). The volume fraction of bainite can vary from about  $0.2 \rightarrow 1.0$  depending on the steel composition and cooling conditions. Typical compositions for accelerated-cooled alloys are given as Alloys 2–4 in Table 13.3. Of these, Alloy 2 is the leanest and can be expected to contain the smallest amount of bainite.

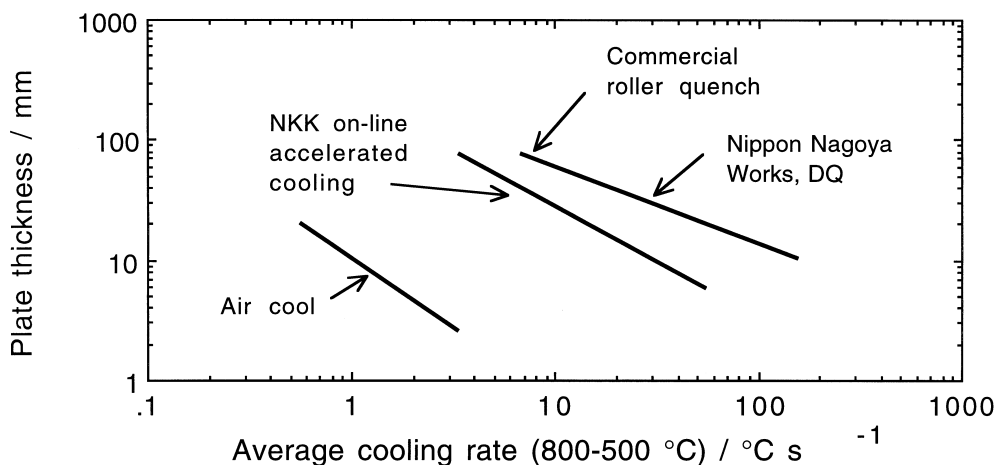


Fig. 13.6 An illustration of the average cooling rates associated with the manufacture of steels for structural applications (Gross *et al.*, 1995).

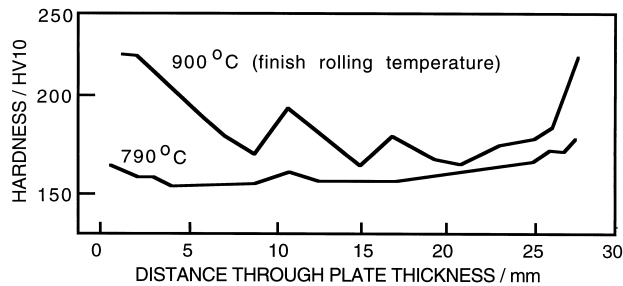
The production of the steels is not a continuous process of casting and control-rolling followed directly by accelerated cooling. Instead, cast ingots are first allowed to cool to ambient temperature and then reheated for the thermomechanical treatment. This ensures that the coarse austenite grain structure which evolves during ingot cooling is disrupted by transformation to ferrite. The processing involves the reheating of thick ingots to 1150 °C, followed by rolling during cooling of the ingot to 740 °C, with the total reduction in thickness being more than 600%, followed by accelerated cooling at 20 °C s<sup>-1</sup> to around 450 °C before allowing natural cooling. This treatment alters the normal microstructure, which is a mixture of ferrite and pearlite, to one which consists of ferrite and bainite, resulting in a better combination of mechanical properties. The tensile strength achieved is typically 700 MPa which is about 50–70 MPa higher than that of conventional control-rolled steels; the Charpy impact toughness can be an impressive 160–200 J at –20 °C. The extra strength is attributed to the fine size of bainite plates, although Morikawa *et al.* (1985) have demonstrated that the strength of the allotriomorphic ferrite also increases with the accelerated cooling, probably because of the dislocation density increase described above. The steels show gradual yielding, although the relevance of this to pipeline applications is not clear (Collins *et al.*, 1985).

### 13.5.2 Process Parameters

There are many processing variables which influence the properties of steel (Graf *et al.*, 1985; Tamehiro *et al.*, 1985a; Collins *et al.*, 1985). For example, a high ingot reheating temperature allows more of the niobium carbonitrides to dissolve in the austenite; the niobium may subsequently precipitate during the  $\gamma \rightarrow \alpha$  reaction to give fine dispersions of interphase precipitates within the ferrite, thereby increasing its strength.

The temperature at which the rolling operation finishes is critical because it should leave the austenite grains in an unrecrystallised, pancake shape. This not only helps refine the microstructure but also helps avoid the undesirable recrystallisation texture of austenite. If the finish rolling temperature is too low then transformation happens during deformation; the deformed ferrite then is stronger but less tough. The finish rolling temperature ( $T_R$ ) also influences the variation in mechanical properties through the thickness of heavy gauge plates (Fig. 13.7). The surfaces, where the cooling rates are greatest, are harder compared with the central regions of the plates. The differences diminish as  $T_R$  is reduced because rolling deformation becomes focused at the plate surfaces, which then transform more rapidly, counteracting the effect of the higher surface cooling rates (Tanaka, 1988).

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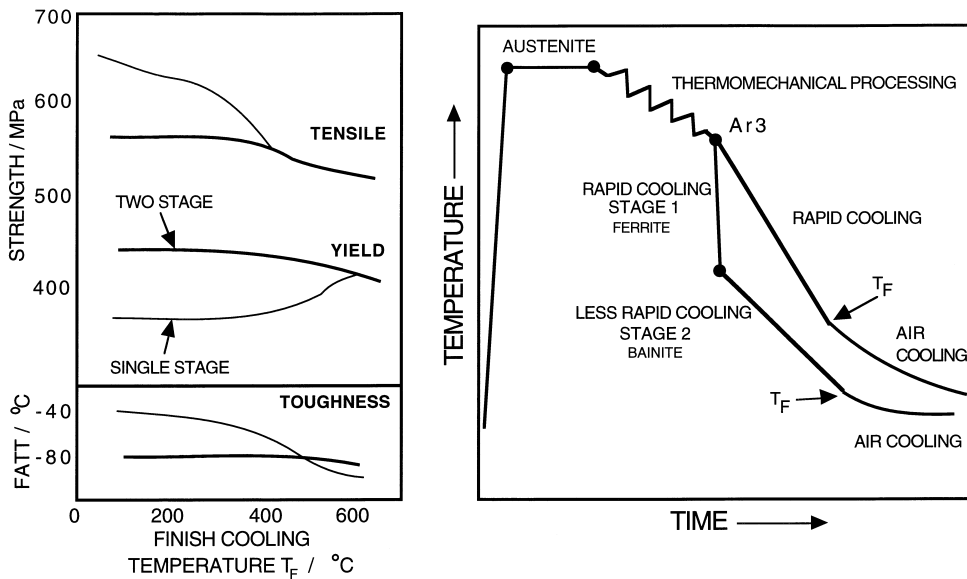


**Fig. 13.7** The effect of the temperature at which rolling is completed, on the variation in hardness of a Fe-0.16C-0.63Mn wt% accelerated cooled steel (Tamukai *et al.*, 1981).

Some of the steels processed using accelerated cooling have a high hardenability which introduces martensite into the microstructure with an accompanying reduction in toughness and increase in plate distortion. To avoid this, the cooling is arrested by cutting off the water sprays at temperatures between  $600 \rightarrow 450$  °C. This has the effect of permitting more bainite to form thus reducing the amount of martensite that can form.

Another related problem occurs in alloys with high hardenability, for example, those containing more than about 1.4 wt% manganese. Thermomechanical processing actually reduces the yield strength even though the tensile strength increases (Fig. 13.8a, Shiga *et al.*, 1983; Amano *et al.*, 1988). This is because martensite replaces bainite as the dominant hard phase so yielding becomes a gradual process, giving the reduction in the proof strength. Although this has clear advantages for applications involving forming operations, the lowering of yield strength is a disadvantage for pipeline and heavy plate fabrications where the design thickness is calculated using yield criteria.

Assuming that the difficulty can be avoided by limiting martensite formation, a two-stage accelerated cooling process has been developed to promote bainite over martensite, while at the same time retaining the high cooling rate required to refine the allotriomorphic ferrite that forms first (Fig. 13.8b). After thermomechanical processing in the austenitic condition, the steel is cooled rapidly ( $25 \text{ K s}^{-1}$ ) through the ferrite temperature range in order to obtain the fine ferrite grain size, but the cooling rate is then reduced to about  $3 \text{ K s}^{-1}$  over the temperature range where bainite forms, thereby reducing martensite. The temperature  $T_F$  at which the forced cooling is stopped to allow the steel to air cool in the second stage of the process is important. The mechanical properties are less sensitive to  $T_F$  for the two-stage process presumably because much of the bainitic transformation is completed at a relatively high temperature during the second stage (Fig. 13.8). The process



**Fig. 13.8** (a) The relationship between the tensile and yield strength, and toughness of accelerated cooled steels as a function of the temperature at which the forced cooled is stopped, for the single and double stage processes. (b) Schematic illustration of the thermomechanical cycles associated with the two-stage accelerated cooling process (Amano *et al.*, 1988).

succeeds in raising the yield stress of the steel when compared with the conventional accelerated cooling procedure.

A general problem with accelerated cooled steels is that the toughness and microstructure are not maintained in the heat-affected zones due to welding (Nishioka and Tamehiro, 1988). The steels are nevertheless about 50 MPa stronger than conventional control-rolled plates. When this additional strength is not required the carbon-equivalent of the steel can be reduced to improve weldability (Fig. 13.9). Accelerated cooling is not appropriate for heavy gauge plates (20–30 mm thick) because it is not possible to ensure uniform cooling; the central regions of thick plates transform into a ferrite and pearlite microstructure rather than the desired bainite (Collins *et al.*, 1985).

In control-rolled steels, thick plates which are cooled slowly after rolling develop a coarse ferrite grain structure at the surfaces. This is a consequence of the recrystallisation of ferrite grains deformed by rolling in the  $(\alpha + \gamma)$  phase field, due to heterogeneous deformation. Accelerated cooling avoids this difficulty because it inhibits recrystallisation, particularly at the surface (Tamehiro *et al.*, 1985a).

### Bainite in Steels

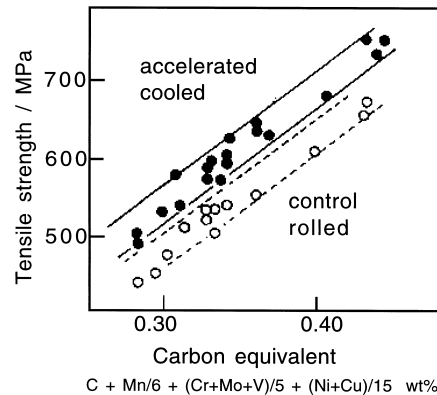


Fig. 13.9 Relationship between the carbon equivalent and tensile strength for conventionally produced control-rolled steels and accelerated cooled steels; the latter have a mixed microstructure of ferrite and bainite (Tamehiro *et al.*, 1985a).

### 13.5.3 Chemical Segregation

Control-rolled steels are cast continuously so they contain pronounced chemical segregation along the mid-thickness of the plate. For example, the manganese concentration at the centre can reach twice the average value. Ferrite naturally forms first in the manganese-depleted regions; the carbon partitioned as the ferrite grows ends up in the manganese-rich regions of austenite. This exaggerates the hardenability of the manganese-rich regions which transform into bands of hard microstructure.

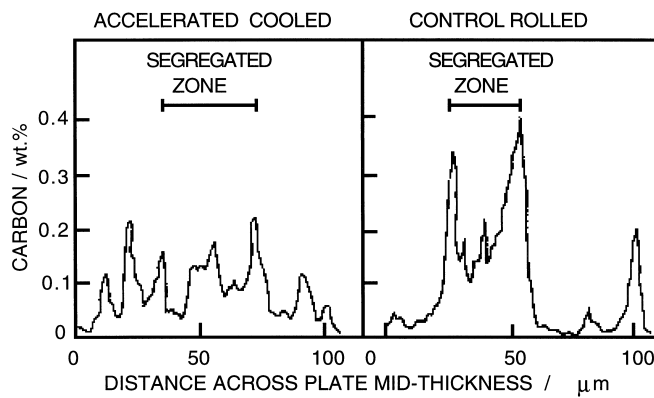
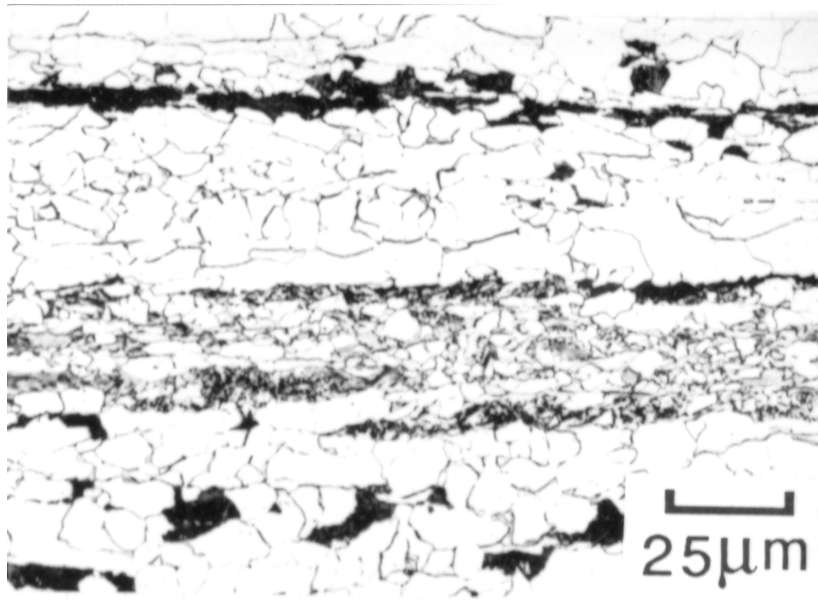
These bands are susceptible to hydrogen cracking. Hydrogen can be infused into the steel through corrosion reactions or other phenomena. An advantage of the accelerated cooled steels is that they are more microstructurally homogeneous (Fig. 13.10); this is because the ferrite and bainite form at a larger undercooling during accelerated cooling, so transformation occurs everywhere, even in the manganese-rich regions. The gross banding characteristic of ferrite-pearlite microstructures is therefore minimised or avoided altogether (Graf *et al.*, 1985; Tamehiro *et al.*, 1985a). The resulting lower hardness in the segregated zone makes the steel less susceptible to hydrogen-induced cracking. Cracking ceases to be a problem because the hardness in all regions becomes less than about 250 HV (Tamehiro *et al.*, 1985).

## 13.6 Steels with High Formability

The oil crisis of the seventies led to the development of the *dual phase* steels in an effort to reduce the weight of cars and make them fuel efficient.



Modern Bainitic Steels



**Fig. 13.10** (a) A light micrograph illustrating the effect of chemical segregation along the mid-thickness of heavy gauge plate. (b) Distribution of carbon concentration in the segregated zone for conventional control-rolled and rapidly cooled steel plates (Tamehiro *et al.*, 1985c).

Conventional automobile steels consist either of mixtures of ferrite and small amounts of pearlite, or wholly of ferrite, with a tensile strength of  $\simeq 420$  MPa and sufficient formability. In the context of automobile manufacture, formability includes the effects of operations like deep drawing, bending and stretching.<sup>†</sup>

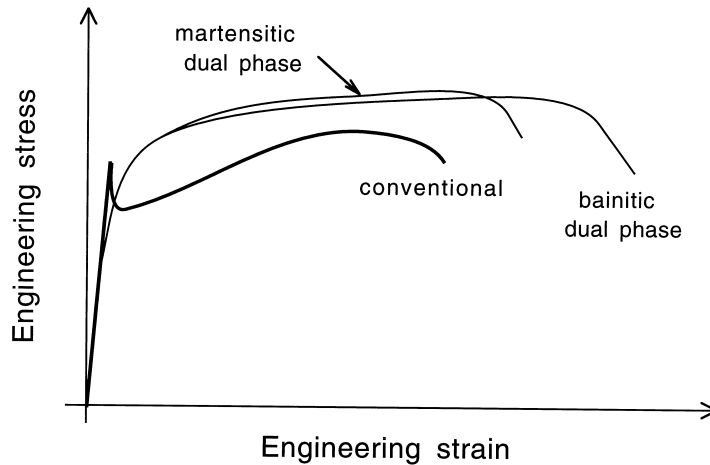
Ordinary steels are troublesome in forming operations because they exhibit discontinuous yielding behaviour, which causes the formation of unsightly stretcher strains on the finished products. Dual-phase steels are similar in composition to conventional alloys, but are heat-treated to generate a mixed microstructure of ferrite and martensite. Their mechanical properties are characterised by continuous yielding, a small proof to tensile strength ratio and a high uniform elongation (Fig. 13.11). The free dislocations in martensite, the strain induced in the ferrite due to martensitic transformation, and the large difference in hardness between the phases all contribute to yielding behaviour. Although the 0.2% proof stress can be small, the dual phase steels strain harden rapidly so the ultimate tensile strength can be quite large at 500–650 MPa, without loss of formability. More details on dual phase steels can be found in a review by Owen (1980). A typical dual-phase steel would have a chemical composition Fe–0.09C–0.6Si–1.0Mn wt% and would be heat-treated to give a mixed microstructure containing about 20% martensite and 80% ferrite. The high silicon concentration is known to enhance formability; alloying elements such as chromium are sometimes added in small concentrations ( $\simeq 0.5$  wt%) for hardenability and solid solution strengthening. The carbon concentration is normally less than 0.2 wt% to avoid brittle spot-welds.

One method of heat treatment involves annealing of strip at a temperature in the two phase ( $\alpha + \gamma$ ) field, followed by cooling at a rate which ensures that the austenite transforms to martensite. However, it is less expensive to integrate the heat treatment into the steel production process, by adjusting the conditions to allow the required microstructure to develop immediately after the hot rolling and coiling process.

There are difficulties associated with the ferrite–martensite dual phase steels. They do not have sufficient stretch flangeability and can suffer from localised necking in the heat-affected zones of flash butt welds. The final coiling

<sup>†</sup>A common test involves the measurement of a *hole expansion limit* which is the ratio of the hole radius before and after expansion. The test can be carried out using a cone or a flat punch. Hole expandability increases with the difference between the total elongation and uniform elongation, and depends on the cleanliness of the steel. Since local necking and local-fracture limit the ability of the hole to expand, it is the freedom of the steel from inclusions which often determines acceptability. Microstructures without cementite are, for the same reason, favoured. High strength fully bainitic or martensitic steels which have more homogeneous microstructures, show better hole expandability than, for example, dual-phase steels.

### Modern Bainitic Steels



**Fig. 13.11** Schematic stress–strain curves comparing the deformation behaviour of conventional high-strength, low-alloy automobile steel and that of dual phase steels. The bainitic dual-phase steel fares better because of its high retained austenite content.

temperature has to be below  $M_s$  with a sufficiently large cooling rate; this is technically difficult to attain at high mill speeds. Also, the liberal use of alloying elements is costly and the silicon causes scale formation on the steel surface during the manufacturing stage.

Attempts have therefore been made to create dual phase steels in which the hard phase is bainite instead of martensite (Esaka *et al.*, 1985). These new dual phase steels are simple in composition (Alloy 5, Table 13.1). They are hot-rolled in the austenite phase field to give a fine austenite grain size, cooled rapidly into the bainite transformation region (typically  $400^\circ\text{C}$ ) and coiled in that temperature range to promote bainite. The ferrite forms during the cooling operation, and the cooling rate ( $40 \rightarrow 100 \text{ K s}^{-1}$ ) determines the fraction of bainite, which may vary from 0.2–1.0 depending on the cooling conditions and composition. A coiling temperature below  $M_s$  can induce martensite as well, giving a triple phase steel.

The new steels have better weldability because the bainite which forms in the heat-affected zone is softer than martensite. The martensite does not form in the HAZ because of the lower carbon equivalent of the new steels.

Although conventional dual phase steels have been used in automobile manufacture, but not for body panels where their formability is inadequate. Bainitic dual phase steels have superior formability, but because the difference between the hardness of bainite and martensite is relatively small, they sometimes exhibit discontinuous yielding. A compromise solution would be

a triple phase steel containing some martensite in addition to bainite and ferrite, to promote gradual yielding. Sudo and coworkers (1981, 1982) have demonstrated that triple phase steels (Alloy 6, Table 13.3) do have the best combination of mechanical properties for press forming applications. Notice that their triple phase steel is leaner in alloy content than the dual phase alloys; the development of the required microstructure therefore depends on more difficult heat treatment practice.

Bainitic dual phase steels have recently been developed for the building industry, specifically to exploit their low yield to tensile strength ratio (Terada *et al.*, 1990). High-rise buildings require heavy gauge, strong and weldable steels. It is also necessary for earthquake-resistant design, to ensure that the steel has enough ductility after yielding to sustain localised deformation without collapse. Recent earthquakes have established that steel is the only cost-efficient construction material which imparts reliability to constructed structures (Williams, 1991). During an earthquake, the beams and columns of the building framework, which support axial loads, experience bending moments. The maximum bending moment that the beam can support before plastic collapse is reduced when the yield ratio is large. Conventional steels for these applications are quenched to a mixed microstructure of bainite and martensite, which is then tempered; any attempt to increase the yield strength unfortunately also increases the yield ratio. For reasons already discussed, this is not the case for dual-phase steels where the heterogeneity of the microstructure lowers the yield strength but increases the tensile strength. The bainitic dual phase steel developed for building purposes (Alloy 7, Table 13.1) has a yield ratio of about 0.7 compared with the quenched and tempered martensitic steel for which the ratio is 0.9. The steel is produced by controlled rolling, and the temperature where the forced cooling cycle commences determines the exact mix of phases and the precise mechanical properties. Typical properties of plates ranging in thickness from 12–80 mm, are 460 MPa yield strength, 600 MPa tensile strength, 35% elongation and > 200 J of Charpy impact energy at ambient temperature. These properties are not degraded by welding.

### **13.6.1 TRIP-Assisted Steels**

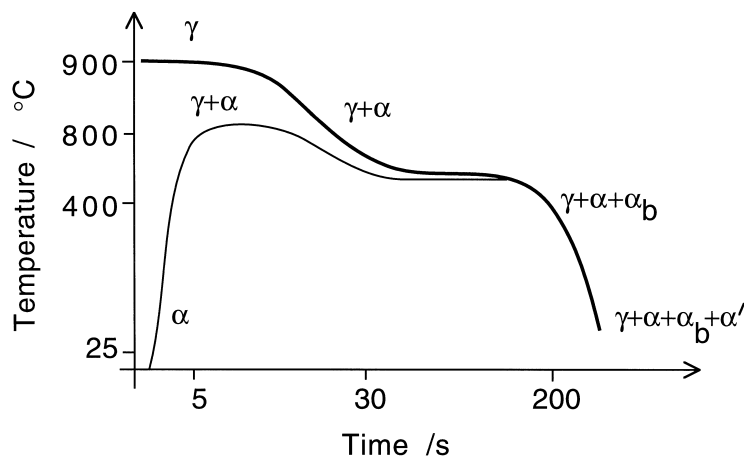
These alloys are also referred to as ‘continuously annealed steels’ because the required microstructure can be generated using a complex heat treatment within a matter of seconds during the processing of steel strip. The ultimate microstructure consists of allotriomorphic ferrite as the major phase mixed with a total of 30–40% of bainite, martensite and carbon-enriched retained austenite. The chemical composition is typically Fe–0.12C–1.5Si–1.5Mn wt%

(Alloy 8, Table 13.3); the silicon prevents cementite precipitation so the carbon rejected by bainite enriches the austenite which is retained.

The alloys are used to improve the safety of automobiles without adding to their weight. The strong steel which connects the front bumper to the frame was in the past made of high-strength low-alloy steels with a UTS of about 400 MPa. These have now been replaced by the TRIP-assisted steels with an ultimate tensile strength of 600 MPa with consequent reductions in weight. Side-impact bars with similar microstructures are used with the UTS in the range 900–1500 MPa.

The large silicon concentration causes scale formation during hot-rolling, resulting in a poor surface finish. The steels are therefore confined to applications where the components are hidden from view.

There are two kinds of TRIP-assisted steels. In the first case a cold-rolled strip is heated rapidly from ambient temperature to form some austenite (Fig. 13.12). This stage is known as an ‘intercritical anneal’.<sup>†</sup> The strip is then cooled at a controlled rate which may lead to the transformation of the



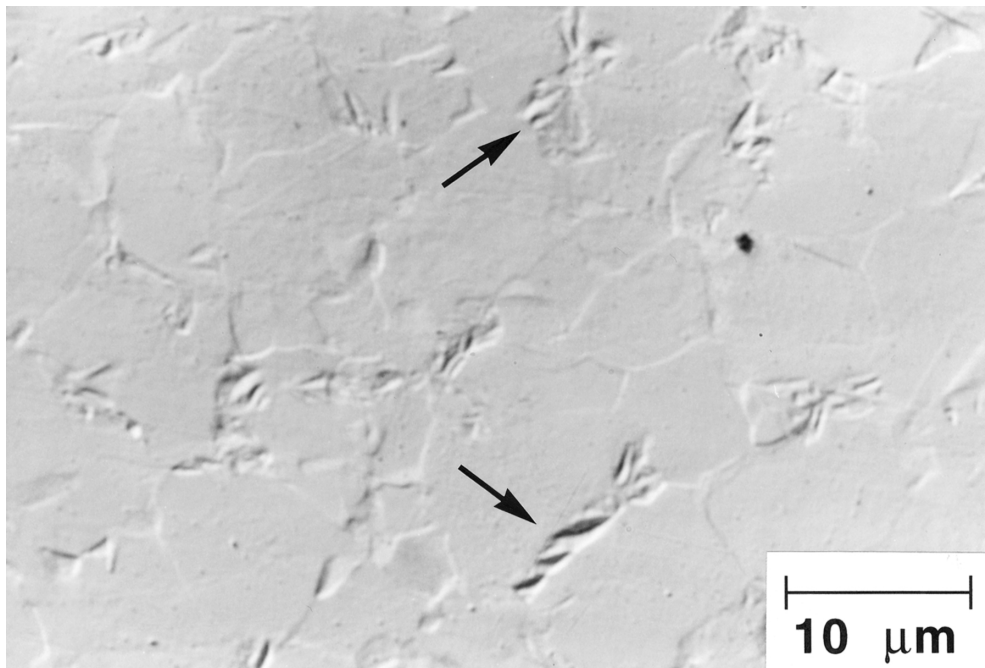
**Fig. 13.12** Schematic illustration of the two kinds of heat treatment used to generate the microstructures of TRIP-assisted steels. Some typical times and temperatures are indicated on the graph axes. The lighter curve represents an intercritical annealing heat treatment whereas the bold curve involves transformation from a fully austenitic sample. The terms  $\gamma$ ,  $\alpha$ ,  $\alpha_b$  and  $\alpha'$  represent austenite, allotriomorphic ferrite, bainite and martensite respectively.

<sup>†</sup>The  $Ae_1$  and  $Ae_3$  temperatures on the iron–carbon phase diagram are often identified as the ‘critical temperatures’. The term ‘intercritical annealing’ refers to heat treatment in the temperature range  $Ae_1 \rightarrow Ae_3$  when the alloy is in the  $\alpha + \gamma$  phase field.

### *Bainite in Steels*

austenite first into allotriomorphic ferrite and subsequently into bainitic ferrite. This latter reaction causes the austenite to become enriched in carbon, some of which may decompose to martensite on cooling, although a fraction ( $\approx 0.05$ ) may be retained (Fig. 13.13). The details of the microstructure and mechanical properties can be altered by changing the cooling rate or varying the rate as a function of the temperature. For example, it is common practice to allow more time in the bainite transformation range than at the higher temperatures where allotriomorphic ferrite grows.

The second kind of heat treatment starts from a hot-rolled strip which is fully austenitic (Fig. 13.11) and forms both allotriomorphic ferrite and bainite during the cooling part of the thermal cycle. This has the advantage that the microstructure can be produced directly from the hot strip which has been rolled to its final dimensions. The process is cheap since the strip does not have to be heated to the intercritical annealing temperature. However, hot-rolling mills are restricted by rolling loads to strips thicker than about 3 mm, although there



**Fig. 13.13** A Nomarski interference micrograph of a TRIP-assisted steel. The sample was metallographically polished, protected against oxidation and then subjected to an intercritical anneal and subsequent partial transformation of the austenite into bainite. The latter phase is here identified by the displacements due to its growth mechanism. Micrograph provided by Jacques, 1999.

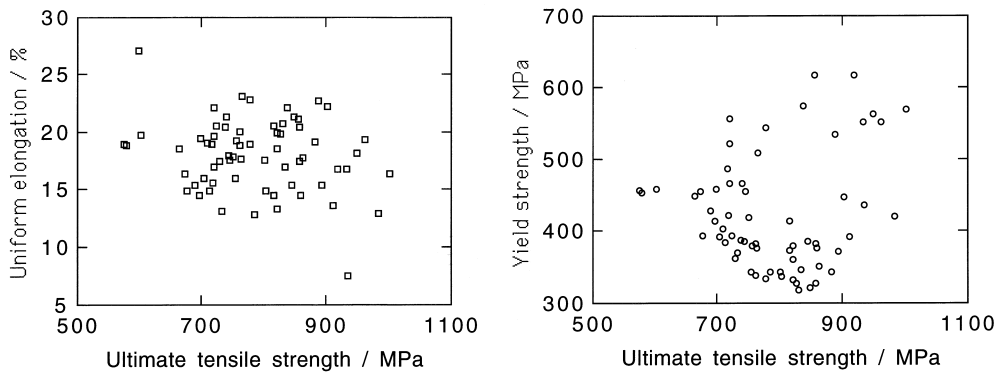
are modern mills which can cope with 1.4 mm thickness. Cold-rolled strips can, on the other hand, be made routinely into thinner gauges. Hot-rolled strips are preferred for automobile applications where cost is a prime factor in the choice of materials.

Virtually all such TRIP steels are based on the simple Fe–Si–Mn–C system with a narrow composition range. In spite of this, it is possible to obtain a great variety of combinations of the yield strength, ultimate tensile strength and uniform elongation by making subtle changes to the heat treatment (Fig. 13.14).

A common parameter used to assess formability in the context of TRIP-assisted steels is the product of the tensile strength and uniform elongation. The parameter is related in some way to retained austenite, but the formability must depend on many more variables including the stability of the austenite, the mechanical inhomogeneity of the microstructure, inclusions, carbide particles and the texture of the steel as a whole. The fact that some of the austenite transforms under stress must also be important. There is no satisfying quantitative interpretation of all these observations.

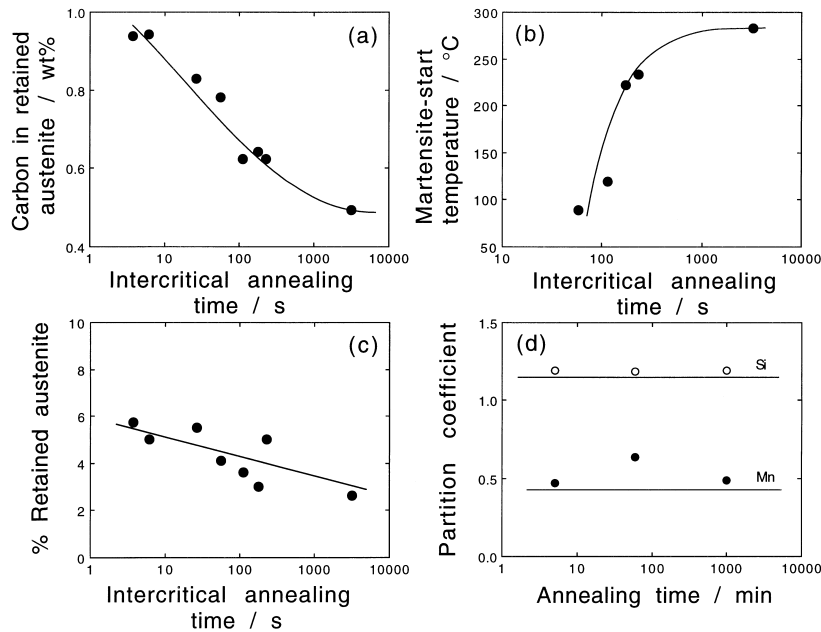
### 13.6.2 Transformations During Intercritical Annealing

The effect of intercritical annealing is to transform some of the initial microstructure into austenite. The part which does not transform is severely tempered, causing the ferrite to recrystallise. The quantity of austenite determines its chemical composition and stability (Fig. 13.15a–c). Austenite nucleates at cementite particles; it has a high carbon concentration to begin with but this becomes diluted as the austenite grows, raising its  $M_s$  temp-



**Fig. 13.14** The remarkable range of mechanical properties that can be obtained from TRIP-assisted steels containing allotriomorphic ferrite, bainite, martensite and retained austenite. The steels all are within the composition range 0.15–0.25C, 0.9–2.0Si and 1.4–1.8Mn wt% with a range of processing conditions.

### Bainite in Steels



**Fig. 13.15** (a–c) Some effects of intercritical annealing on a Fe–0.11C–1.5Si–1.53Mn wt% steel at 750°C. Data from Samajdar *et al.* (1998). (d) Ratio of concentration of element in ferrite to that in austenite as a function of the intercritical annealing time, for a steel of approximate composition Fe–0.2C–1.5Si–1.5Mn annealed at 800°C. The horizontal lines represent values calculated assuming equilibrium partitioning. Data from Minote *et al.* (1996).

erature and hence reducing the fraction that is retained on cooling to ambient temperature. The size of the austenite grains also influences their stability; larger grains transform more readily to martensite. Intercritical annealing can also lead to the redistribution of the substitutional solutes (Fig. 13.15d), the degree of partitioning depending on the annealing temperature and time.

The heat treatment influences also the crystallographic texture of both the austenite and the recrystallised ferrite (Samajdar *et al.*, 1998). Newly formed austenite grains which nucleate at cementite particles do not show preferred orientation. The texture changes during annealing but the mechanism is not understood.

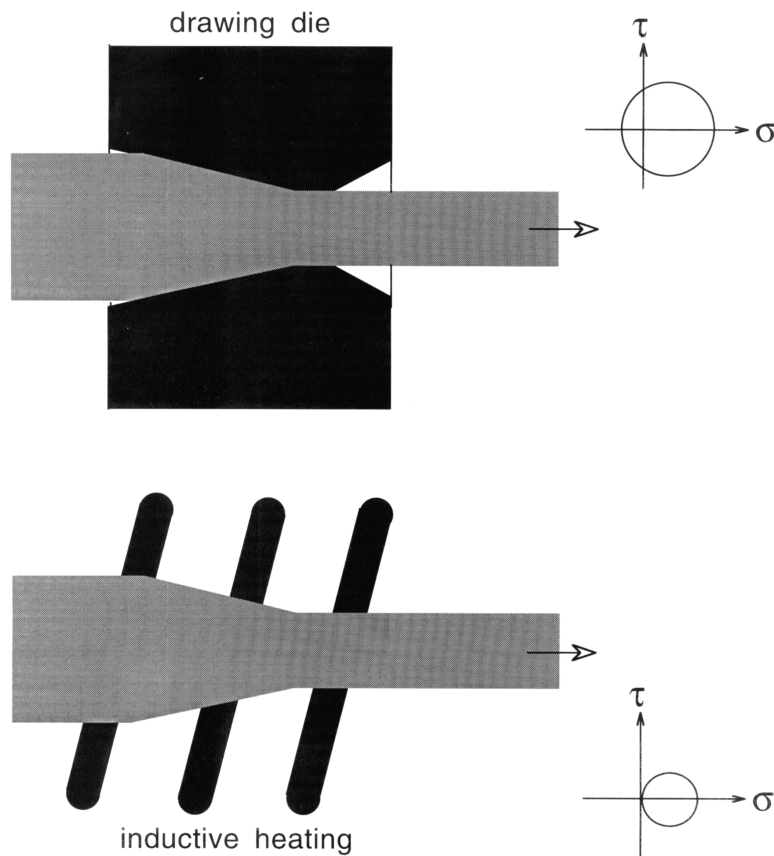
### 13.6.3 Dieless-Drawn Bainitic Steels

Wires or rods are conventionally made by drawing, in which a reduction in cross-sectional area is achieved by passing the stock through a die (Fig. 13.16).



The drawing reduction is determined by the ability of the thinner, work-hardened section which leaves the die, to sustain the drawing force without further deformation.

An alternative process achieves the reduction in section using inductive heating. The region of the rod which passes through the hot zone softens and is extended by the drawing force. It stops deforming on leaving the hot zone. The process avoids all the difficulties associated with die erosion and requires a smaller drawing force (Fig. 13.16) because there are no stresses due to die-constraint. Dieless drawing can be applied to a greater variety of shapes, for example, square rods or tubes. The thermomechanical processing combined with accelerated cooling can be used to selectively develop microstructures,



**Fig. 13.16** A comparison of conventional drawing with dieless drawing. The Mohr's circle representations of stress are also presented;  $\sigma$  and  $\tau$  represent the normal and shear stresses respectively (Weidig *et al.*, 1999).

including dual-phase mixtures of ferrite and bainite (Weidig *et al.*, 1999). The technology has yet to see major applications but the absence of discontinuous yielding in the dual phase microstructure is an advantage when drawn tubes subsequently have to be formed into complex shapes. Weidig *et al.* have managed to produce predominantly bainitic microstructures using dieless drawing and accelerated cooling in a low-hardenability steel of chemical composition Fe-0.1C-0.14Si-0.7Mn wt%.

### 13.7 Ultra-Low-Carbon Bainitic Steels

Ever since Irvine and Pickering (1957), it has been apparent that good mechanical properties can be achieved in bainitic steels by reducing their carbon concentrations. Lower concentrations not only reduce the coarse cementite but also regions of untempered martensite which originate from the incomplete transformation of austenite to bainite. For this latter reason, large concentrations of substitutional solutes are also detrimental if they limit the extent of transformation to bainite. And yet, the steel must have sufficient hardenability to avoid phases which occur before bainite during continuous cooling transformation.

Irvine and Pickering compromised by choosing a low concentration of carbon at about 0.1 wt% but ensuring hardenability using boron and molybdenum. They were therefore able to produce fully bainitic steels by continuous cooling transformation. Carbon is a potent strengthener of ferrite but for the same reason, it can lead to embrittlement following welding. A low carbon concentration can therefore provide major economies in fabrication, often allowing welding to be carried out without preheat.

The *ultra-low carbon bainitic* (ULCB) bainitic steels use this philosophy (Alloys 9 and 10, Table 13.1). The carbon concentration is limited to the range 0.01–0.03 wt% (Nakasugi *et al.*, 1980, 1983; Hulka *et al.*, 1988). The steels are microalloyed so some of the carbon precipitates in austenite as niobium carbide, so that the actual concentration in the austenite prior to its transformation is even smaller than the average value. The resulting reduction in the fraction of martensite improves toughness without undue loss of strength (Fig. 13.17). The carbon concentration should not be reduced to less than 0.01 wt% since niobium carbide precipitation then decreases, with an accompanying loss of toughness. The niobium carbide and TiN prevent austenite grain growth during control-rolling operations and suppress  $\text{Fe}_{23}\text{CB}_6$  and BN precipitation, thereby leaving the boron free to make its contribution to hardenability (Tamehiro *et al.*, 1987a,b). In fact,  $\text{Fe}_{23}\text{CB}_6$  particles should be avoided since they stimulate the nucleation of allotriomorphic ferrite.

Because there is insufficient carbon to combine with niobium in ULCB steels, a substantial proportion of the niobium remains dissolved in austenite. Since

Modern Bainitic Steels

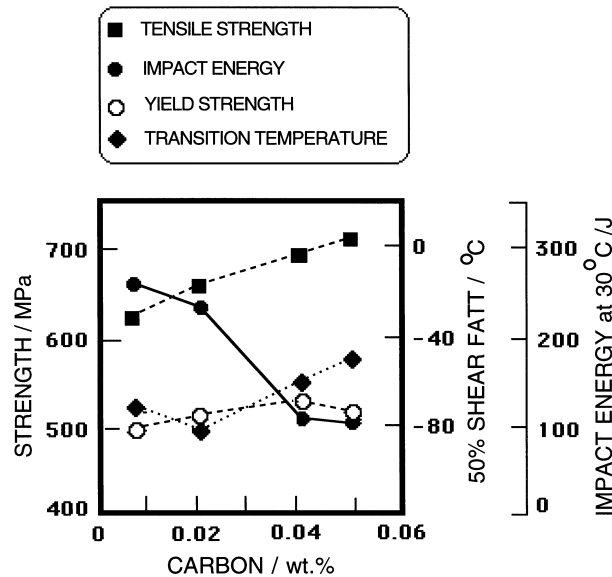


Fig. 13.17 Mechanical properties as a function of carbon concentration in thermo-mechanically processed ultra-low-carbon bainitic steels (Hulka *et al.*, 1988). The fracture assessed impact transition temperature (FATT), an indicator of the ductile-brittle transition temperature, begins to increase as the carbon concentration falls below about 0.02 wt%.

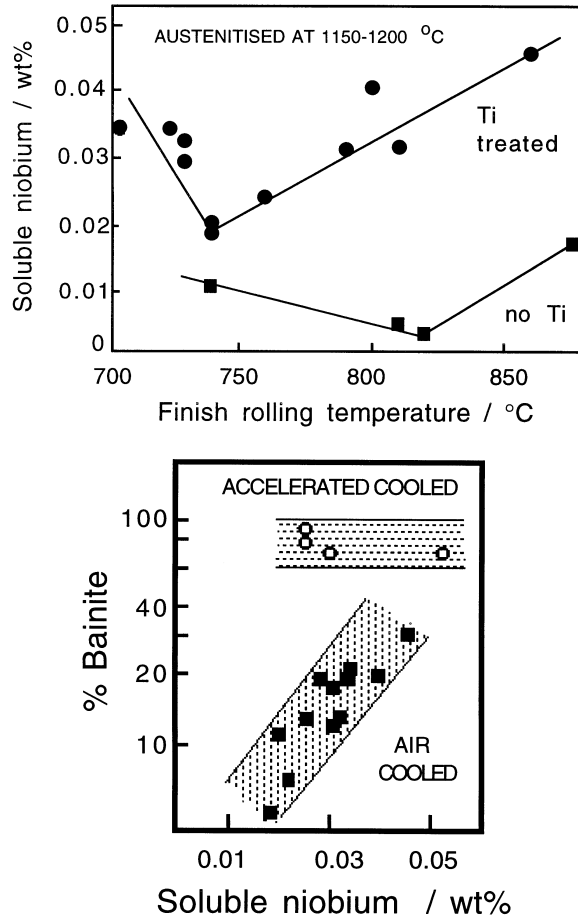
the equilibrium between niobium and carbon is temperature dependent, the finish rolling temperature  $T_R$  has a large influence on the state of the niobium (Fig. 13.18a). Less niobium remains in solution as  $T_R$  is reduced due to the greater propensity for strain-induced precipitation during deformation of the austenite at low temperatures.

Changes in the dissolved niobium concentration influence the evolution of microstructure in ULCB steels (Fig. 13.18b). The effect is unlikely to be purely thermodynamic given the small concentrations involved. Soluble niobium strongly retards allotriomorphic ferrite, permitting more bainite to be obtained in the final microstructure. The effect is not obvious in rapidly cooled steels because the ferrite is suppressed irrespective of the niobium concentration (Fig. 13.18b).

There is a synergistic effect between vanadium and niobium. Vanadium competes for carbon leaving more niobium in solution and thereby promoting a bainitic microstructure (Leber *et al.*, 1987).

ULCB steels have outstanding toughness, strength and weldability combinations so they find use in the construction of pipelines in Arctic or submarine environments. The highest strength values are obtained using low finishing

### Bainite in Steels



**Fig. 13.18** Soluble niobium in ULCB steel (Hulka *et al.*, 1988). (a) Variation in the soluble niobium concentration as a function of the finish rolling temperature; (b) variation in microstructure as a function of the finish rolling temperature.

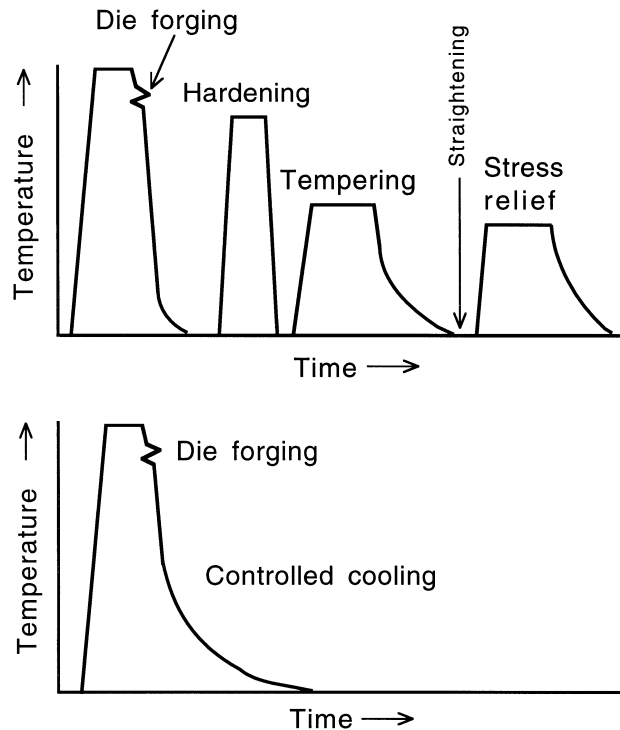
temperatures and accelerated cooling. The strength can be increased further by retarding cooling below 550 °C to allow NbC precipitation in the ferrite. This can be achieved by coiling or stacking the hot-steel product.

## 13.8 Bainitic Forging Steels

Forging is a method of working metal into the required shape by hot or cold plastic deformation. Large objects can be shaped using open forging-dies, whereas the mass production of precise components is done in closed dies

which are filled with solid metal using forge pressure. Forging is attractive as a manufacturing process because it reduces the machining costs and can enhance properties along directions consistent with the application. Typical components which are forged range from small scale items such as crankshafts, connecting rods, piston shafts, bolts, axles and fasteners to components which might weigh many tonnes, such as the rotor shafts in steam turbine generators.

Fe-C-Mn-Si mild steels have served the forging industry reliably for many decades. They have been available in two strength ranges, 350–450 MPa ferrite/pearlite alloys, and high strength (> 600 MPa) quench and tempered martensitic steels.<sup>†</sup> With the martensitic steels, the forged components have to be austenitised, quenched, tempered, manipulated to remove distortion due to heat treatment, and finally, stress-relieved (Fig. 13.19a). For large



**Fig. 13.19** An illustration of the heat-treatment procedures for forging steels (Wright *et al.*, 1987): (a) conventional quenched and tempered martensitic steels; (b) microalloyed forging steels.

<sup>†</sup>An excellent review on the subject has been published by Jones *et al.* (1985).

components, the steels has to be heavily alloyed to obtain martensite at all positions.

There have therefore been attempts to reduce costs by reducing the concentrations of alloying elements and by simplifying the heat treatments. Modern forging steels are microalloyed to produce either fine ferrite/pearlite microstructures, or are alloyed to give bainitic or predominantly bainitic microstructures. Unlike martensite, these microstructures can be produced by transformation during cooling from the forging temperature, with savings in heat treatment, handling and fabrication costs (Fig. 13.19b). Strength as high as 500–700 MPa are readily achieved without compromising toughness and with improved fatigue performance, machinability and weldability. The martensitic steels are still the toughest when it comes to medium carbon steels with a strength of 1200 MPa.

There are special considerations necessary when using directly transformed microalloyed steels; rapid induction heating is often used to heat the stock to forging temperature. The time at the austenitisation temperature has to be long enough to permit microalloying elements such as niobium to dissolve (Wright *et al.*, 1987). The way in which the finished components are stacked after the final forging operation can determine their cooling rates.

The new alloys open up the possibility of *controlled forging*, which by analogy with controlled rolling, aims to refine the austenite grain structure prior to transformation (Jones *et al.*, 1987). The final stages of forging are carried out at a temperature where the austenite does not recrystallise or alternatively, it recrystallises to a fine grain size. The disadvantage is that reductions in forging temperature cause an increase in the forging force necessitating more powerful production equipment. The increased stress also causes more rapid die wear.

Some of the new forging steels contain carbide-forming elements such as Nb, V, Mo (Alloys 14–18, Table 13.3). Tempering causes fine carbide precipitation which strengthens the final product. The response to tempering is related to the amount and type of bainite in the mixed microstructures of  $\alpha$ ,  $\alpha_b$  and pearlite (Leber *et al.*, 1988). The high dislocation density of bainite provides nucleation sites for the carbides, leading to rapid hardening. Predominantly bainitic microstructures are therefore preferred over those containing large fractions of allotriomorphic ferrite. Where niobium and boron additions are used to develop low carbon bainitic microstructures with high work hardening rates (Alloy 18, Table 13.3), cold deformation can be used to increase the strength of the final product. A good example is the series of steels developed for the production of high-strength bolts by cold heading operations (Heritier *et al.*, 1984).

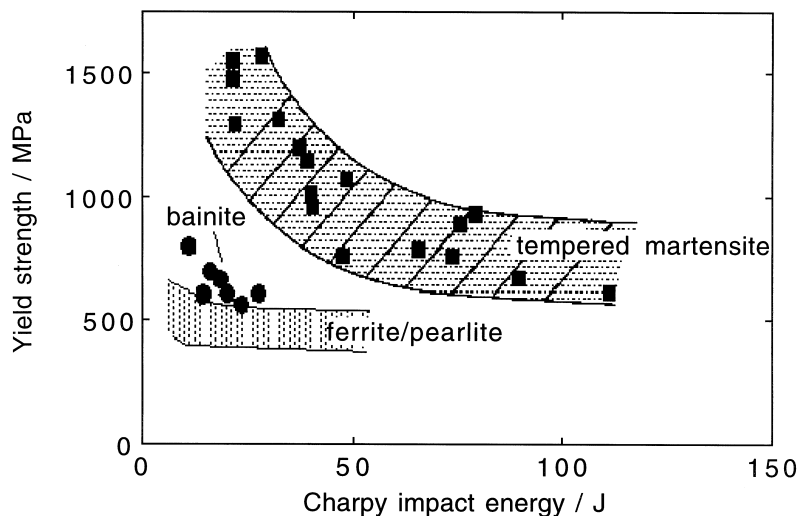
Strong, bainitic-forging steels containing high silicon and molybdenum concentrations have been developed for automobile applications, with a typical composition: Fe–1.4Mn–0.8Si–0.15V–0.2Mo wt% with 0.1–0.40 wt% C

(Heitmann and Babu, 1987; Grassl *et al.*, 1989). Such steels are intended to replace quenched and tempered martensitic alloys. The compositions are chosen to avoid the formation of carbides during bainitic transformation, carbides which can be detrimental to toughness. The toughness turns out to be better than that of ferrite–pearlite microstructures at comparable strength levels. However, at higher carbon concentrations the bainitic steels compare unfavourably with martensitic steels which have superior toughness (Fig. 13.20). Although this might militate against the use of bainitic-forging steels, their toughness is still more than adequate for many applications, where their cost advantage may be usefully exploited.

### 13.9 High Strength Bainitic Steels without Carbides

We have seen in Chapters 2 and 3 that an interesting microstructure results when a silicon or aluminium-alloyed steel is transformed into upper bainite. The carbon that is partitioned into the residual austenite does not precipitate as cementite, but remains there to make the austenite stable at ambient temperature. The microstructure obtained consists of fine plates of bainitic ferrite separated by carbon-enriched regions of austenite (Fig. 13.21).

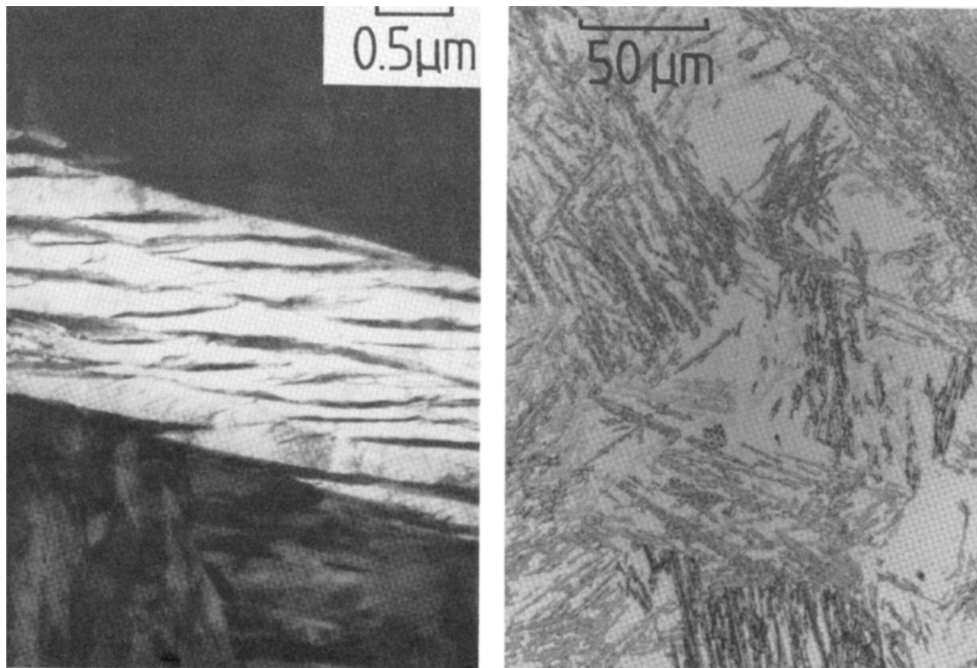
The potential advantages of this mixed microstructure can be listed as follows:



**Fig. 13.20** A comparison of the properties of ferrite/pearlite, bainitic- and martensitic-forging steels (Grassl *et al.*, 1989).

### *Bainite in Steels*

- (i) Cementite is responsible for initiating fracture in high-strength steels. Its absence is expected to make the microstructure more resistant to cleavage failure and void formation.
- (ii) The bainitic ferrite is almost free of carbon, which intensely strengthens ferrite and hence embrittles it.
- (iii) The microstructure derives its strength from the fine grain size of the ferrite plates, which are less than  $1\ \mu\text{m}$  in thickness. It is the thickness of these plates which determines the mean free slip distance, so that the effective grain size is less than a micrometer. This cannot be achieved by any other commercially viable process. Grain refinement is the only method available for simultaneously improving the strength and toughness of steels.
- (iv) The ductile films of austenite which are intimately dispersed between the plates of ferrite have a crack blunting effect. They further add to toughness by increasing the work of fracture as the austenite is induced to transform to martensite under the influence of the stress field of a propagating crack. This is the TRIP, or transformation-induced plasticity effect (Chapter 12).



**Fig. 13.21** (a) Transmission electron micrograph of bainitic ferrite plates separated by films of stable austenite. (b) Optical micrograph showing the large blocks of austenite left untransformed.



- (v) The diffusion of hydrogen in austenite is slower than in ferrite. The presence of austenite can therefore improve the stress corrosion resistance of the microstructure.
- (vi) Steels with the bainitic ferrite and austenite microstructure can be obtained without the use of expensive alloying. All that is required is that the silicon concentration should be large enough to suppress cementite.

In spite of these appealing features, the microstructure does not always give the expected good combination of strength and toughness. This is because the relatively large *blocky* regions of austenite between the sheaves of bainite (Fig. 13.21b) readily transform into high-carbon martensite under the influence of stress. This untempered, hard and coarse martensite regions severely embrittle the steel.

If it is assumed that a fraction  $\phi$  of a sheaf consists of films of austenite, then the ratio of the fractions of film and blocky austenite (prior to any martensitic transformation) is given by:

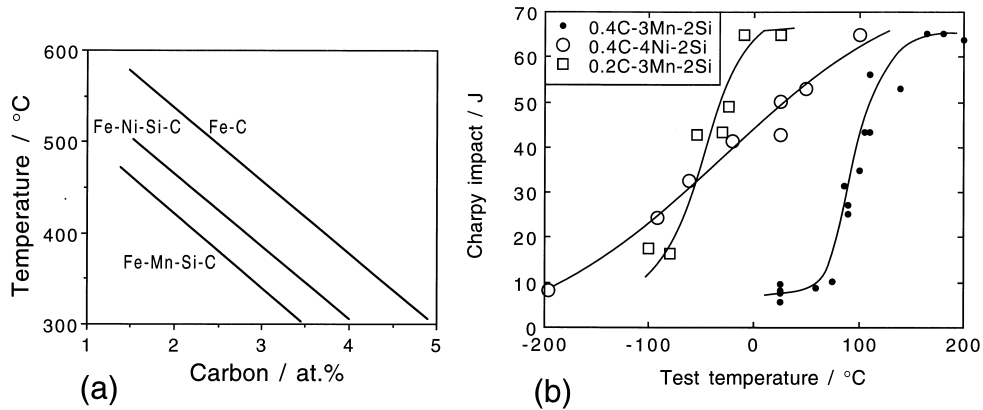
$$\frac{V^{\gamma-F}}{V^{\gamma-B}} \simeq \frac{\phi V^{\alpha}}{V^{\gamma} - \phi V^{\alpha}} \quad (13.2)$$

where  $V^{\gamma-F}$  and  $V^{\gamma-B}$  are the volume fractions of film and blocky type retained austenite respectively, and  $V^{\alpha}$  and  $V^{\gamma}$  the total volume fractions of bainitic ferrite and residual austenite respectively. It is found experimentally that both strength and toughness are optimised by achieving a ratio greater than 0.9. Any large blocks of austenite can be consumed by promoting the formation of bainite without precipitating carbides. The maximum fraction of bainite that can be obtained at any temperature is, from the lever rule,

$$V^{\alpha} \simeq \frac{x_{T_0} - \bar{x}}{x_{T_0}} \quad (13.3)$$

It follows that there are three ways of eliminating blocky austenite:

- (i) By reducing the isothermal transformation temperature to increase  $x_{T_0}$ . The lower limit is set by either the lower bainite or martensite-start temperature.
- (ii) By reducing the overall carbon concentration of the steel, so that the austenite reaches its limiting composition at a later stage of reaction.
- (iii) By moving the  $T_0$  curves of the phase diagram to larger carbon concentrations. This can be done by adjusting the concentration and type of substitutional solute (Fig. 13.22a).



**Fig. 13.22** (a)  $T_0$  curves for the alloy-steels listed in Table 13.3, and a corresponding curve for a plain carbon steel. (b) Impact transition curves showing improved toughness without loss of strength, obtained by reducing the amount of blocky austenite in a mixed microstructure of bainitic ferrite and austenite. The Fe-0.43C-2Si-3Mn wt% alloy has  $V^{\gamma-F}/V^{\gamma-B} = 0.5$ . The reduced carbon Fe-Mn-Si-C steel and the Fe-Ni-Si-C steel both have  $V^{\gamma-F}/V^{\gamma-B} > 1.5$ . After Bhadeshia and Edmonds (1983a,b).

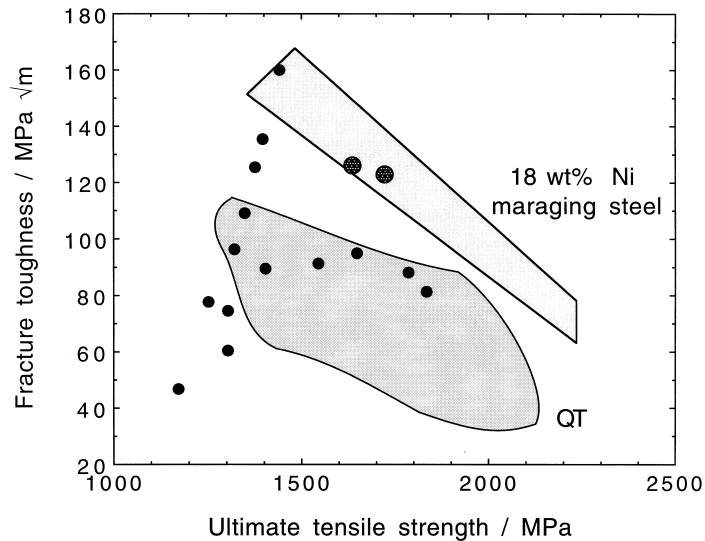
The effect on toughness in reducing the amount of blocky austenite is indicated in Fig. 13.22b, which shows the large changes in the impact transition temperatures as the ratio of film to blocky austenite is increased in the manner just described. Note that for a duplex  $\alpha + \gamma$  microstructure, the strength actually increases as the fraction of bainitic ferrite increases, so that the better toughness is obtained without sacrificing strength.

Typical compositions of high-strength steels which show good toughness are given in Table 13.4; Fig. 13.23 shows how the mechanical properties compare

**Table 13.4** Chemical compositions (wt%) of experimental high-strength steels with microstructures consisting of mixtures of bainitic ferrite and retained austenite. After Caballero *et al.* (2001).

C	Si	Mn	Ni	Cr	Mo	V	YS/MPa	UTS/MPa	$K_{IC}/\text{MPa m}^{1/2}$
0.3	1.5	2.0	—	1.3	0.25	0.1	1170	1800	
0.3	1.5	—	3.5	1.5	0.25	0.1	1150	1730	125
0.3	1.5	—	3.5	1.5	0.25	—	1100	1625	128

### Modern Bainitic Steels



**Fig. 13.23** Comparison of the mechanical properties of mixed microstructures of bainitic ferrite and austenite, versus those of quenched and tempered (QT) low-alloy martensitic alloys and maraging steels. The two large points refer to the latest bainitic steels, which match the maraging steel but are thirty times cheaper (about £900 per tonne compared with some £30,000 per tonne for maraging steels). Data from Miihkinen and Edmonds (1987c) and Caballero *et al.* (2001).

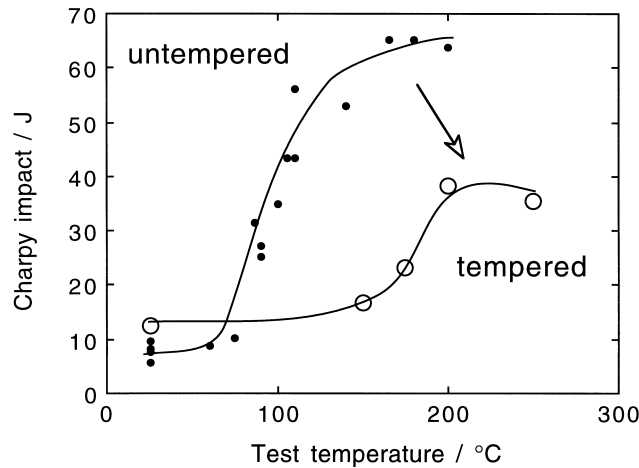
with quenched and tempered steels. It is evident that in some cases, the properties match those obtained from much more expensive maraging steels.

The properties of these steels do not change much when tempered at temperatures close to the transformation temperature at which the original bainite formed. However, annealing at elevated temperatures or prolonged periods at low temperatures can lead to the decomposition of the austenite into ferrite and carbides, with a simultaneous drop in strength and toughness, especially the upper shelf energy (Fig. 13.24). The latter effect can be attributed directly to the void nucleating propensity of carbide particles in the tempered microstructure, as illustrated by the much smaller void size evident in the fracture surface of the tempered sample (Fig. 12.11).

## 13.10 Thermomechanically Processed High-Strength Steels

The alloys described here are not terribly useful in practice, but the phenomena they reveal are interesting and add to the understanding of bainite.

### Bainite in Steels



**Fig. 13.24** Charpy impact toughness data for Fe-0.43C-2Mn-3Si wt% alloy transformed to a mixture of bainitic ferrite and carbon-enriched retained austenite. The data from the untempered microstructure are from Fig. 13.9. The other data were obtained after tempering the microstructure at 500 °C for 1 h to induce the decomposition of the austenite into a mixture of ferrite and carbides.

#### 13.10.1 Ausformed Bainitic Steels

Ausforming is the deformation of austenite at temperatures well below  $A_{e3}$ , followed by transformation to martensite or bainite. Its main effect is to increase the strength with a small loss in ductility. The deformation reduces the effective size of the austenite grains leading in turn to a more refined bainite (Kalish *et al.*, 1965, Duckworth, 1966; Edwards and Kennon, 1974, 1978; Umemoto *et al.*, 1986). The microstructure can become biased by the deformation which might favour particular crystallographic variants to form over others (Umemoto *et al.*, 1986).

The bainite that grows from deformed austenite has a greater density of dislocations (Irani, 1967; Edwards and Kennon, 1978). This is because it inherits the dislocations present in the parent austenite. The bainite sheaves adopt a smaller aspect ratio in ausformed samples (Tsuzaki *et al.*, 1989). Although there is an overall refinement of the microstructure, each grain of deformed austenite transforms into fewer variants of bainite. Indeed, an entire austenite grain can transform into a single packet of bainite, which may not be desirable from a toughness point of view (Tsuji *et al.*, 1999). On the other hand, the refinement of cementite particles in ausformed bainite improves the toughness. As a rough estimate, each one percent increment of plastic deformation leads to a 5 MPa increase in strength.

### Modern Bainitic Steels

Not all steels are suitable for ausforming because the austenite must remain stable during deformation at a low temperature. The steel must exhibit a deep bay in the region of the TTT diagram between the two C curves representing the reconstructive and displacive reactions (Table 13.5). Some of the alloying elements used to achieve the required hardenability in ausforming steels are also strong carbide formers. It is suspected that there is a precipitation of fine carbides in the austenite during its deformation; this must contribute further to the strength of ausformed samples (Duckworth *et al.*, 1964, Thomas *et al.*, 1965). When strong carbide-forming elements are absent, electrical resistivity measurements have shown that deformation does not affect the concentration of dissolved carbon in the austenite (Hoffman and Cohen, 1973).

**Table 13.5** Chemical compositions (wt%) of bainitic steels that have been studied with respect to the ausforming process. The steels are not in some cases custom made for ausforming operations.

C	Si	Mn	Ni	Mo	Cr	V	Cu	W	
0.39	1.00	0.25	–	1.39	5.25	0.54	–	–	Kalish <i>et al.</i> , 1965
0.48	0.25	0.86	0.18	0.04	0.98	–	0.09	–	Duckworth, 1966
0.85	–	1.39	–	–	0.59	–	–	0.53	Edwards & Kennon, 1978
0.59	2.01	1.02	–	–	–	–	–	–	Tsuzaki <i>et al.</i> , 1989

The plastic deformation of austenite at low temperatures is rarely homogeneous. Slip tends to concentrate into narrow bands (Schmatz and Zackay, 1959; Evans and O'Neill, 1959; Freiwillig *et al.*, 1976). The austenite in the bands then transforms into a peculiar narrow band of ferrite, whose transformation mechanism is not known, although the bands do act as nucleation sites for bainitic ferrite (Jepson and Thompson, 1949; Freiwillig *et al.*, 1976; Edwards and Kennon, 1978). Transmission electron microscopy has shown the bands to consist of single crystals of ferrite, or parallel laths of ferrite whose long axes lie in the plane of the band, together with some retained austenite. The bands also contain carbides which assist the nucleation of ferrite by locally reducing the carbon concentration. It is likely that the microstructure of the band is simply some highly refined form of bainite (Edwards and Kennon, 1978).

The tempering behaviour of ausformed bainite is somewhat different from that of ordinary bainite. The dislocations in the ausformed bainite rearrange into cells, whereas those in the lower dislocation density ordinary bainite remain dispersed (Edwards and Kennon, 1978).

Ausformed steels are not commercially successful because of the expense of the thermomechanical treatment and because of the limited variety of shapes that can be produced. With the exception of some low-carbon bainitic steels, ausformable alloys are costly because of the alloying elements they contain; a typical ausforming steel which is transformed to bainite might contain 5Cr–2Mo–0.5V–0.3Mn–1.0Si wt%. In general, ausformed martensitic steels exhibit better combinations of toughness and strength when compared with ausformed bainitic steels (Kalish *et al.*, 1965). This is because of the coarser carbides bainitic steels; it is not therefore surprising that ausformed low-carbon bainitic steels are tougher than martensite in the same alloy (Durbin and Krahe, 1973).

There have been developments in ausformed bainitic steels. Silicon-rich steels respond well to thermomechanical processing because of the absence of cementite (Tsusaki *et al.*, 1989). A particularly promising application is in the manufacture of wire-springs; the shape of this product is ideal for thermomechanical processing. Ausformed lower bainite wires have a hardness of 650 HV which is higher than the same spring in its tempered martensitic condition (Tsuji *et al.*, 1999).

### **13.10.2 Strain-Tempered Bainitic Steels**

Strain tempering involves the deformation of martensitic or bainitic microstructures, followed by an ordinary isothermal tempering heat-treatment. The steel compositions are similar to those of ausformed steel. The process leads to large increases in strength, with some loss in ductility and toughness. The strengthening that is obtained increases with the level of prior deformation, but is not just a reflection of the effects of deformation on the microstructure. The tempering causes an increase in strength as fine carbides precipitate on deformation-defects (Kalish *et al.*, 1965). The effect is greater for bainite than for martensite, because the retained austenite in the former microstructure undergo stress-induced transformation to harder martensite. Strain tempering is more effective as a method for increasing the yield strength than ausforming, but does reduce ductility.

### **13.10.3 Creep Tempering of Bainite**

Both recovery and recrystallisation processes are accelerated when bainite is tempered during creep deformation (Ridal and Quarell, 1962; Murphy and Branch, 1971). The unstressed regions of creep test samples retain a higher hardness than those in the gauge length. In molybdenum-containing bainitic steels, the transition from  $\text{Mo}_2\text{C} \rightarrow \text{M}_6\text{C}$  is accelerated by creep testing, the

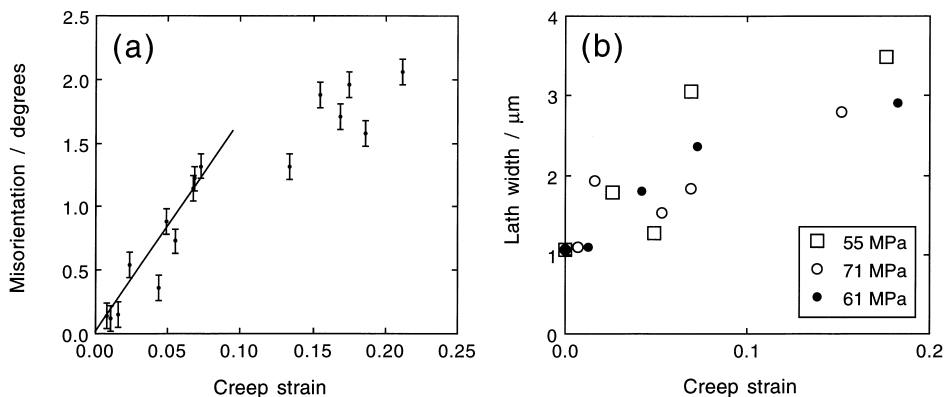
change being most noticeable at low temperatures or high strain rates, i.e. when reaction kinetics for the  $\text{Mo}_2\text{C} \rightarrow \text{M}_{23}\text{C}_6$  transition are slow.

It may be that the precipitation rate is increased by the greater number density of nucleation sites in a deformed sample, and by defect-assisted diffusion. However, there are complications in that the  $\text{Mo}_2\text{C} \rightarrow \text{M}_{23}\text{C}_6$  reaction has been reported to be retarded when the microstructure is tempered martensite (Ridal and Quarell).

### 13.10.3.1 Orientation changes during creep tempering

Cold-deformation experiments show that once a dislocation cell structure is established, further strain leads to a decrease in the cell size and a corresponding increase in the crystallographic misorientation between adjacent cells (Langford and Cohen, 1975). The dislocations move and their density and distribution also changes during the course of creep deformation. These changes are more complex than those associated with straight annealing, since there is in fact a coarsening of the microstructure. There is certainly an increase in the misorientation between adjacent cells in bainite as shown in Fig. 13.25a but the bainite lath width actually increases with the creep strain (Fig. 13.25b).

The measurements are for a  $2\frac{1}{4}\text{Cr}1\text{Mo}$  steel with a bainitic microstructure. The cell structure is found to be much coarser in the same steel when the microstructure is a mixture of ferrite and pearlite, with smaller misorientations between cells. Minute misorientations develop between dislocation cells within individual ferrite grains over a scale of about  $10\ \mu\text{m}$ . This contrasts with the



**Fig. 13.25** (a) Crystallographic misorientation between adjacent laths of bainite in  $2\frac{1}{4}\text{Cr}1\text{Mo}$  steel as a function of the secondary-creep strain. The samples were tested at a variety of stresses and temperatures. (b) Corresponding variations in lath size (Lonsdale and Flewitt, 1979).

orientation differences between adjacent laths of bainite, which are over distances less than 1  $\mu\text{m}$ .

## **13.11 Bainite in Rail Steels**

Modern railway systems are subjected to intense use, with fast trains and increasing axle loads. Rails have to be more wear resistant and achieve higher standards of straightness and flatness in order to avoid the surface and internal defects which may lead eventually to failure. The shape of the manufactured rail depends to a large extent on the uniformity of thermomechanical processing; the most advanced mills are computer controlled with continuous feedback from the product during manufacture (Fitzgerald, 1991). In this section we consider advances in the steels used for making rails, particularly the bainitic steels.

### **13.11.1 Track Materials**

Although a variety of loadings can adversely affect the life of rails, wear and plastic deformation can lead to unacceptable changes in the rail head profile, changes controlled primarily by a system of rolling contact stresses encountered during service. This system is dependent upon the relative motions of wheel and rail within a small contact zone of about one square centimeter. The motions lead to lateral and longitudinal surface tractions and a spin moment. The rate of rail degradation depends also on the exact location; rail head erosion is at a maximum in regions where the track curves.

Ordinary rail steels contain about 0.7 wt% of carbon and are pearlitic (Table 13.6). However, special grade steels have been considered for tracks carrying high axle loads or fast trains. Since the steels may be used in continuous welded track, they must be amenable to welding to other kinds of low-grade rail steels. Conventional welding processes include flash butt welding and the thermite process. The Hadfield cast austenitic manganese steels are excellent for wear resistance but are not weldable to ordinary rail steels (Sawley *et al.*, 1985).

Wear is a system property more than a material property, but it is nevertheless possible to identify material factors which are important. With eutectoid steels, the wear rate decreases as the hardness increases (Fig. 13.26), although exceptions have been reported (Kalousek *et al.*, 1985b). Figure 13.26 shows also that the microstructure influences wear; refining the microstructure prolongs the wear-limited rail life (Kalousek and Beynon, 1975).

The majority of rail steels are of eutectoid composition with a pearlitic microstructure. The pearlite colony size and interlamellar spacing can be refined by transformation at lower temperatures. But a higher level of refinement can be

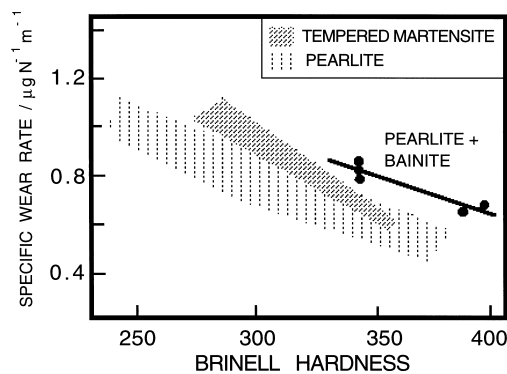


Modern Bainitic Steels

**Table 13.6** Compositions (wt%) of typical bainitic and pearlitic rail steels.

No	C	Si	Mn	Ni	Mo	Cr	V	Nb	B	Al	Ti	Other
1	0.55	0.25	1.00	-	-	-	-	-	-	-	-	Pearlitic rail steel
2	0.80	0.30	1.00	-	-	-	-	-	-	-	-	Pearlitic rail steel
3	0.70	1.90	1.50	-	-	-	-	-	-	-	-	Pearlitic rail steel
4	0.75	0.70	1.00	-	-	1.0	0.1	-	-	-	-	Special grade pearlitic rail steel
5	0.65	0.25	0.70	-	-	-	-	-	-	-	-	Pearlitic tyre steel
6	0.04	0.20	0.75	2.0	0.25	2.8	-	-	0.01	0.03	0.03	Bainitic rail steel
7	0.09	0.20	1.00	-	0.50	-	-	-	0.003	0.03	0.03	Experimental bainitic rail steel
8	0.07	0.30	4.50	-	0.50	-	-	0.1	-	-	-	Experimental bainitic rail steel
9	0.10	0.30	0.60	4.0	0.60	1.7	-	-	<0.01	0.03	0.03	Experimental bainitic rail steel
10	0.30	0.20	2.00	-	0.50	1.0	-	-	0.003	0.03	0.03	Experimental bainitic rail steel
11	0.30	1.00	0.70	-	0.20	2.7	-	0.1	-	-	-	Experimental bainitic rail steel
12	0.52	0.25	0.35	1.5	0.25	1.7	-	0.1	<0.01	-	-	Experimental bainitic rail steel
13	1.00	0.25	0.25	-	-	1.50	-	-	-	-	-	Roller bearing alloy

achieved by transforming instead to bainite. We shall consider first the conventional bainitic steels in which the microstructure contains carbides. These are in general found to perform worse than the pearlitic steels. Ghonem *et al.* (1982) compared the wear and toughness of pearlitic rails with that of an experimental bainitic rail containing chromium and molybdenum. The rail head microstructure consisted of 70% bainite and 30% pearlite but not only was the toughness unacceptable, so was the wear on the gauge face where there is rolling–sliding contact. Heller and Schweitzer (1980) compared the properties and service performance data from bainitic rail steels (alloys 8, 11, Table 13.6) with those from a pearlitic rail steel (alloy 4, Table 13.6). The bainitic

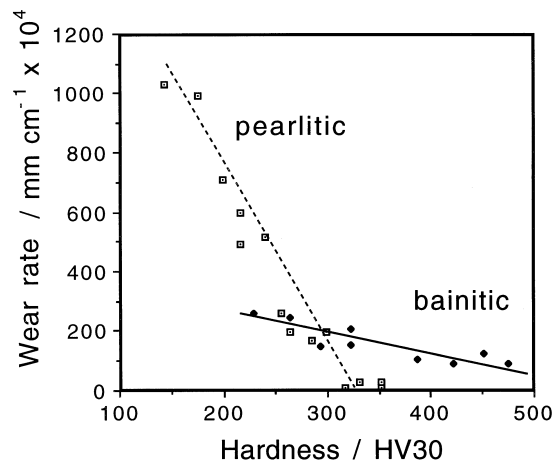


**Fig. 13.26** The correlation of hardness with the wear rate for martensitic, bainitic and pearlitic microstructures (Hodgson and Preston, 1988).

steels achieved higher tensile and fatigue strengths and performed well in service. There were, however, unspecified problems during welding. Service trials subsequently revealed that the bainitic steels wore faster than conventional pearlitic steel rails, when comparisons were made at the same hardness (Heller and Schweitzer, 1982). Work on a low-carbon bainitic steel using laboratory tests involving rolling/sliding contact has indicated a wear rate some ten times faster than a pearlitic steel of the same hardness (Ichinose *et al.*, 1982). Other results confirm these general trends and indicate further that mixed microstructures consisting of bainite and pearlite are less wear resistant when compared with fully pearlitic steels Fig. 13.26 (Kalousek *et al.*, 1985a,b; Mutton, 1985).

By contrast, low-carbon bainitic steels (alloy 6, Table 13.6) have been tested successfully for railway crossing applications where impact erosion and fatigue of the crossing nose were the major wear problems with conventional pearlitic rail steels (Callender, 1983; Garnham, 1989). An advantage of low carbon concentrations is that it enables the steels to be welded easily. Tests using pure sliding, cooled, pin-ring tests has demonstrated that low-carbon bainitic steels might have comparable or superior wear resistance to the pearlitic steels, Fig. 13.27. Even when the pearlitic and bainitic steels have similar wear characteristics, the lower carbon concentration of bainitic steels ensures better ductility, toughness and weldability.

It has been argued that most of the early pessimistic results on bainitic steels could be challenged because they were not systematic investigations and the



**Fig. 13.27** Pin-ring wear rate versus hardness data for pearlitic and bainitic steels (Clayton *et al.*, 1987).

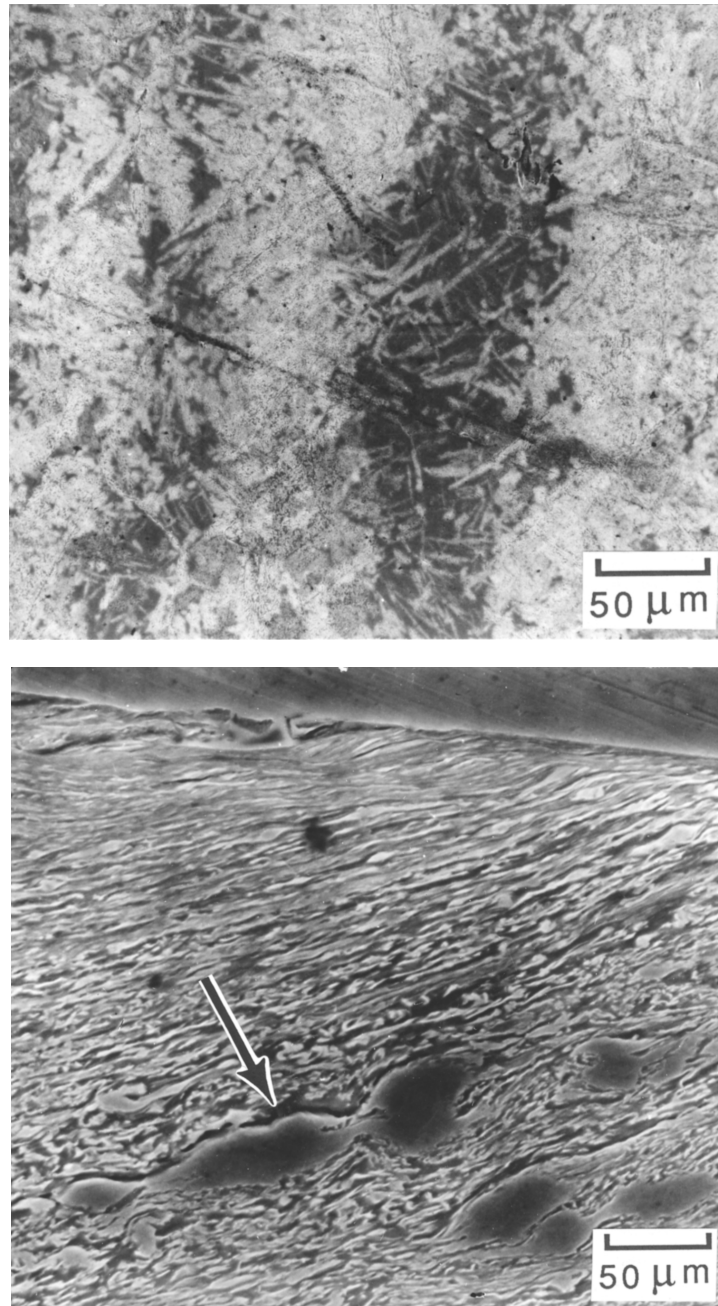
microstructures had been inadequately characterised (Clayton *et al.*, 1987, 1990). Contradictory results can also be attributed to the different ways in which wear resistance was measured. The pin-ring test is not representative of service conditions where there is rolling–sliding wear. Using tests designed to simulate rolling–sliding wear, Garnham (1989) has demonstrated, using a variety of steels that carbide-free bainite has poor wear resistance relative to pearlite. Although carbide-containing bainitic steel performed better, it increased the wear of the mating pearlitic railway wheel steel so that the combined wear rates were generally no better than for pearlite–pearlite combinations. It may be concluded from Garnham’s work that bainitic steel is not suitable for railway applications where rolling–sliding wear is the major cause of rail or wheel replacement.

The most recent work has been more systematic in the assessment of wear resistance (Devanathan and Clayton; 1990). Three steels with carbon concentrations ranging from 0.04 to 0.54 wt% (alloys 8, 9 & 12, Table 13.6) were examined in their bainitic condition. The lowest carbon steel, which also had the lowest starting hardness, outperformed pearlitic steels at the same hardness level. This is because the bainitic steel work hardened rapidly and had a greater ductility. The medium carbon steel was similar in its wear performance to pearlite, whereas the higher carbon steel was found to be worse. Microstructural observations (Fig. 13.28) revealed that the high carbon steel was chemically segregated, leading to bands of high carbon martensite separated by bainite; cracking initiated at the interfaces between these bands.

### 13.11.2 Silicon-rich Carbide-free Bainitic Rail Steels

We have seen that rail steels are largely based on high-carbon steels of near eutectoid composition and pearlitic microstructures. It is the carbide phase which is crucial in providing the necessary hardness. A different approach has been based on medium carbon bainitic alloys without any carbides, using precisely the same design philosophy as described for high-strength steels in section 13.9; much of the work is commercially sensitive and the subject of many patents, but there is a brief review article published by Yates (1996) and a paper by Jin and Clayton (1997). The steels contain a large silicon concentration and hence have a microstructure which is a mixture of bainitic ferrite, carbon-enriched retained austenite and some martensite. They have already been demonstrated to have a much improved wear resistance (Fig. 13.29) and rolling-contact fatigue resistance, together with a high toughness even at sub-zero temperatures. A typical composition for the new steel is Fe–0.4C–1.5Si–2.0Mn–0.25Mo wt%, the composition being decided using thermodynamic and kinetic theory as described in section 13.9. The hardness throughout the rail section is about 400 HV.

*Bainite in Steels*



**Fig. 13.28** (a) Banding apparent in a high carbon bainitic rail steel (Alloy 12, Table 13.6); (b) cracking at the interface between a pool of martensite and bainite (after Devanathan and Clayton, 1990).

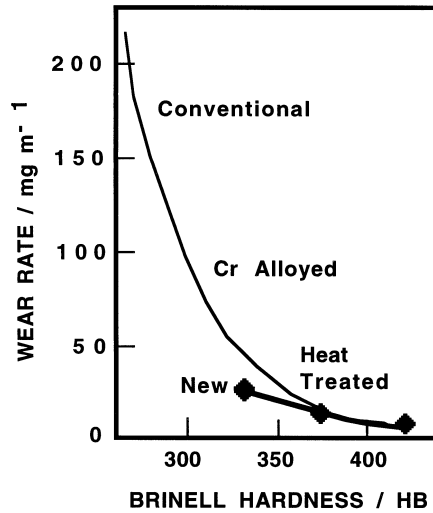


Fig. 13.29 Comparison of the wear rates and hardness levels of conventional rail steels against the new alloys which have a carbide-free microstructure of bainitic ferrite and austenite (Jerath *et al.*, 1991).

### 13.11.3 Wheels

It is possible that railway wheels will also in the future be based on a bainitic microstructure. A particular problem with wheels is that the alloys must be resistant to microstructural change due to *wheel spin burning*. This can lead to the formation of brittle martensite when material at the surface of a pearlitic steel is momentarily heated to a temperature high enough to cause austenite to form. The same material is then cooled rapidly as the heat is dissipated into the underlying bulk of the wheel, causing the austenite to transform into brittle martensite. This effect can be of greater importance than the rolling-sliding wear of the tread (Sawley *et al.*, 1988). Steels containing less carbon and a bainitic microstructure have been shown to outperform the pearlitic steels because their lower hardenability makes martensite formation less likely; if martensite does form, it is not as brittle given the low carbon concentration and high  $M_s$  temperature.

### 13.11.4 Bearing Alloys

Roller bearings represent another wear-related application in which bainitic microstructures might have an advantage over the traditional alloys used in the quenched and tempered condition (Akbasoglu and Edmonds, 1990). The

studies to date have focused on altering the microstructure of available alloys rather than starting from scratch. The standard bearing alloy is the Fe–1C–1.5Cr steel (Alloy 13, Table 13.5), austenitised at 850 °C for 20 min, oil quenched and then tempered at 175 °C for 1 h. The tempering temperature can be increased to 250 °C for better toughness, but at the expense of hardness. A lower bainitic microstructure can be produced by isothermal transformation below the  $B_S$  temperature (a typical heat-treatment might consist of 250 °C for 40 min).

For roller bearings, the main life-limiting factor is rolling contact fatigue, a parameter which can be sensitive to the environment (oil, water) in which the bearing operates. Maximum shear stresses arise under the outer surface of the bearing and can lead to severe spalling. When operating in water-based lubricants, hydrogen evolving from electrochemical reactions can cause embrittlement and can accelerate the fatigue crack growth. In these circumstances, bainitic microstructures fare better than the more hydrogen sensitive martensitic microstructures.

### 13.12 Bainitic Cast Irons

Cast irons typically contain 2–4 wt% of carbon with a high silicon concentration and a greater concentration of impurities than steels. The *carbon equivalent* (CE) of a cast iron helps to distinguish the grey irons which cool into a microstructure containing graphite and the white irons where the carbon is present mainly as cementite. The carbon equivalent is defined as:

$$CE(\text{wt}\%) = C + \frac{Si + P}{3} \quad (13.4)$$

A high cooling rate and a low carbon equivalent favours the formation of white cast iron whereas a low cooling rate or a high carbon equivalent promotes grey cast iron.

During solidification, the major proportion of the carbon precipitates in the form of graphite or cementite. When solidification is just complete, the precipitated phase is embedded in a matrix of austenite which has an equilibrium carbon concentration of about 2 wt%. On further cooling, the carbon concentration of the austenite decreases as more cementite or graphite precipitates from solid solution. For conventional cast irons, the austenite then decomposes into pearlite at the eutectoid temperature. However, in grey cast irons, if the cooling rate through the eutectoid temperature is sufficiently slow, then a completely ferritic matrix is obtained with the excess carbon being deposited on the already existing graphite.

White cast irons are hard and brittle; they cannot easily be machined. When they are hypoeutectic (Fig. 13.30)

Modern Bainitic Steels

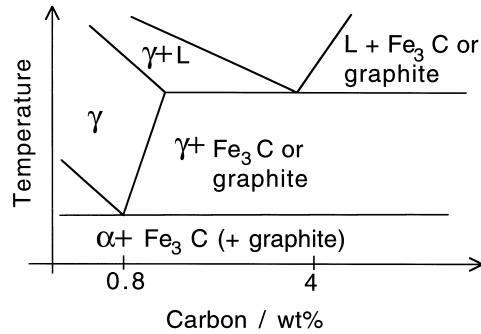
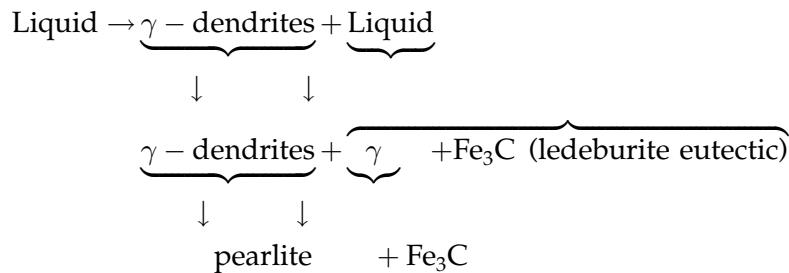


Fig. 13.30 Schematic representation of the iron-carbon phase diagram showing the eutectic and eutectoid reactions.



Grey cast irons are softer with a microstructure of graphite in transformed-austenite and cementite matrix. The graphite flakes, which are rosettes in three dimensions, have a low density and hence compensate for the freezing contraction, thus giving good castings free from porosity.

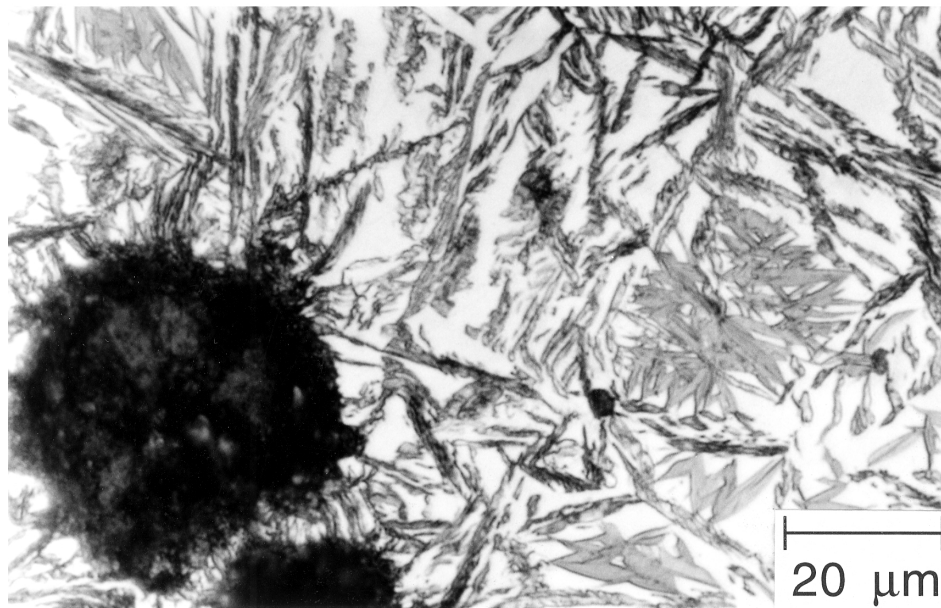
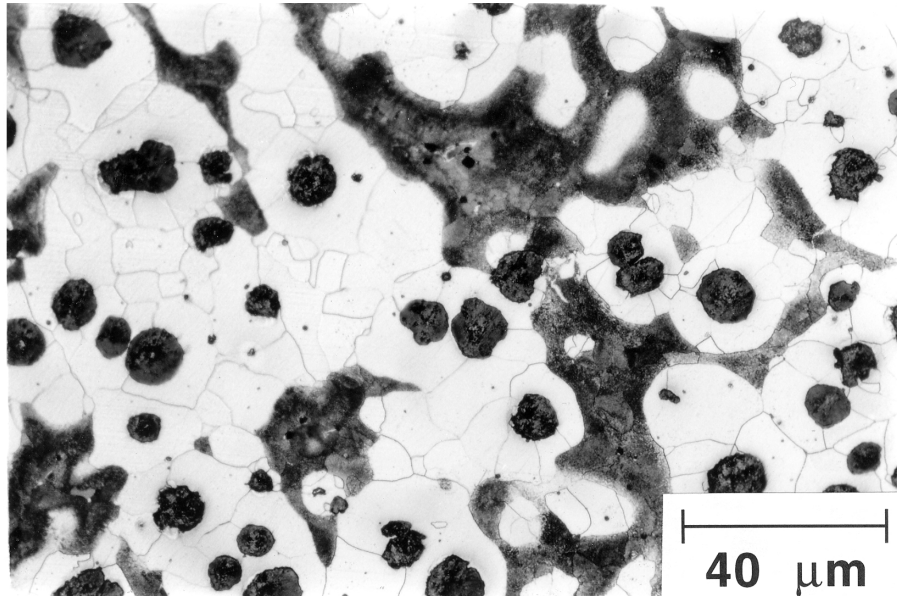
The flakes of graphite have good damping characteristics, good machinability but are stress concentrators, leading to poor toughness.

The addition of minute quantities of magnesium or cerium poisons preferred growth directions and leads to isotropic growth resulting in spheroids of graphite. This spheroidal graphite cast iron has excellent toughness and is used widely, for example in crankshafts.

### 13.12.1 Austempered Ductile Cast Irons

The latest breakthrough in cast irons is where the matrix of spheroidal graphite cast iron is not pearlite, but bainite (Fig. 13.31). This results in a major improvement in toughness and strength. The bainite is obtained by isothermal transformation of the austenite at temperatures below that at which pearlite forms.

*Bainite in Steels*



**Fig. 13.31** Optical micrographs: (a) the microstructure before austempering; (b) the microstructure after austempering, with graphite in a matrix which is a mixture of bainitic ferrite, retained austenite and some martensite (after Yescas, 2000)



### Modern Bainitic Steels

The isothermal heat treatment which leads to the formation of bainite is known as *austempering* and the resulting cast iron, which is known as *austempered ductile cast iron*, has much improved ductility and toughness (Dorazil *et al.*, 1962; Blackmore and Harding, 1984; Moore *et al.*, 1985a; Ueda and Takita, 1986; Shea and Ryntz, 1986; Franetovic *et al.*, 1987a,b; Rundman *et al.*, 1988), together with improved abrasion wear resistance (Shepperson and Allen, 1987; Lu and Zhang, 1989). The austempered ductile cast irons have a typical composition Fe–4.0C–2.0Si–0.5Mn–0.06Mg wt%, with a carbon equivalent of about 4.5 (Table 13.7). The nodular iron is prepared by treating the melt with magnesium which alters the flake graphite morphology associated with untreated melts to that of nodular graphite. The solidified iron is then held at 930 °C for 12 hours followed by slow cooling, to give a structure consisting of graphite nodules in a ferritic matrix.

**Table 13.7** Typical compositions of austempered ductile cast irons.

Type	Chemical composition/wt%								Reference
	C	Si	Mn	Mo	Cu	S	P	Mg	
Typical	3.6	2.0	0.5	–	–	0.02	0.02	0.06	
Reduced manganese	3.44	2.41	0.15	–	–	0.007	0.0015	0.064	Putatunda & Gadicherla, 1999

Heat-treatment consists of austenitisation at around 950 °C followed by isothermal transformation of the austenite at temperatures where bainite is expected. Because the cast irons have a large silicon concentration (1–2 wt%), carbide precipitation is prevented during the formation of upper bainite so that the microstructure at the isothermal transformation temperature consists of a mixture of bainitic ferrite and carbon-enriched residual austenite. Some of the latter phase then decomposes into untempered martensite on cooling to ambient temperature (Moore *et al.*, 1985a; Franetovic *et al.*, 1987a).

One important feature of the heat treatment is that bainite is allowed to form until it reaches a maximum in its volume fraction, but the heat treatment must not go on so long that carbides begin to precipitate from the carbon-enriched austenite. This is why the carbon concentration of the residual austenite at the isothermal transformation temperature can be calculated using the  $T_0'$  boundary (James and Thomson, 1999). This is useful because not only can the carbon concentration of the austenite be calculated using the  $T_0$  curve, but so can the maximum fraction of bainitic ferrite since:

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$$V_{\alpha_b}^{max} \simeq \frac{\bar{x} - x^\alpha}{x_{T_0} - x^\alpha} \quad (13.5)$$

where  $V_{\alpha_b}^{max}$  is the maximum fraction of bainitic ferrite,  $\bar{x}$  is the average carbon concentration in the austenite before the formation of bainite,  $x_{T_0}$  is the concentration of carbon given by the  $T_0$  curve of the phase diagram and  $x^\alpha$  is the solubility of carbon in ferrite. Furthermore, this theory also explains why the carbon concentration of the austenite is independent of the starting concentration  $\bar{x}$  (Moore *et al.*, 1985b). Naturally, the calculations must all allow for the presence of chemical segregation in the cast iron.

As emphasised earlier, prolonged heat-treatment at the isothermal transformation temperature leads eventually to the diffusional decomposition of the residual austenite into carbides and more ferrite. The period between the achievement of a maximum volume fraction of bainitic ferrite and the onset of carbide precipitation is known as the heat-treatment window.

Austempered ductile cast iron behaves in the same manner as silicon-rich steels transformed to bainite, except that the carbon concentration of the austenite depends both on the time and temperature to which the cast iron is heated. This is because the austenite is in contact with graphite and the equilibrium between these phases depends on temperature; the time required to achieve equilibrium also depends on the austenitising temperature. An increase in austenitising time leads to a greater dissolution of the graphite into the austenite, thereby increasing the austenite carbon concentration to a maximum level consistent with the ( $\gamma/\gamma + \text{graphite}$ ) phase boundary (Moore *et al.*, 1987). Consequently, a cast iron isothermally transformed to bainite after holding at the austenitising temperature for a long period gives lower bainite (with  $\eta$  carbides within the bainitic ferrite) whereas the bainite obtained after a short austenitising treatment is upper bainite, free of carbides. This is consistent with the theory for the transition from upper to lower bainite (Chapter 7).

Cast irons need a long time (typically 2 h) at the austenitising temperature to reach equilibrium, when compared with steels (Rouns and Rundman, 1987). This is because the dissolution of graphite produces carbon concentration gradients in the matrix, leading to a heterogeneous microstructure. For reasons which are not clear, the ferrite also seems to take longer to dissolve at the austenitising temperature.

Consistent with the lower carbon level expected in the austenite following a low temperature austenitising treatment, the rate of formation of bainite is found to increase with a decrease in the austenitising temperature (Moore *et al.*, 1985a,b). Of course, the austenite grain size also decreases with the austenitising temperature and may contribute to the acceleration of transformation, but the effect of carbon is expected to be the main factor responsible for the increased reaction rate.

A further effect of austenitising temperature that is peculiar to the cast irons is the mechanism by which phosphorous embrittles the austenite grain boundaries (Klug *et al.*, 1985). The phosphorous combines with magnesium to form particles (possibly  $Mg_3P_2$ ) at the boundaries. The phase begins to redissolve at the grain boundaries if the austenitisation temperature is too high: for a Fe-3.74C-2.4Si-0.19Mn-0.15Mo-0.02P-0.045Mg wt% alloy, the redissolution becomes important at temperatures exceeding  $\sim 1050^\circ\text{C}$ . The dissolving particles act as a source for free phosphorous which then spreads along the austenite grain. There is then a drop in toughness as fracture occurs preferentially at the embrittled austenite grain boundaries.

As is the case for bainitic steels containing large concentrations of silicon, austempered nodular graphite cast irons have the highest toughness when carbides are absent and when the retained austenite is mechanically stable (Moore *et al.*, 1985a; Franetovic *et al.*, 1986). The time at austenitising temperature therefore is important because it determines the matrix austenite carbon concentration, and hence its stability with respect to both carbide precipitation and martensitic decomposition.

The role of carbide precipitation in causing a reduction in the ductility of bainitic cast irons is not as clear cut as in the wrought alloys. The austempered cast irons contain numerous large graphite particles which should be the main sites for failure initiation. It seems unreasonable to suggest that the much smaller carbides associated with bainite can control failure initiation. It is probable that the reduction in austenite volume fraction and stability caused by carbide precipitation causes the reduced ductility. Fractography could establish this hypothesis.

Common defects in cast irons, such as shrinkage, slag inclusions and films, segregation, and certain eutectic products can negate the benefits achieved by the presence of austenite in the austempered ductile iron (Moore *et al.*, 1987). For example, the brittle interdendritic carbides which form when the molybdenum concentration is greater than 0.5 wt%, makes the iron brittle in spite of the austenite. The problems become more severe for larger castings where cooling rate variations might necessitate larger alloy concentrations at the risk of exaggerating segregation. There is a growing trend towards alloys with reduced concentrations of manganese and molybdenum (Table 13.7) because these have a greater tendency to segregate into the liquid phase when compared with solutes such as Si, Ni, and Cu which partition preferentially into the solid phase (Hayrynen *et al.*, 1988; Putatunda and Gadicherla, 1999).

It is a well-known result for bainitic steels that the retained austenite occurs in two forms. There are the films trapped between the platelets of bainitic ferrite, and the coarser blocky regions of austenite trapped between different bainite sheaves. The blocks tend to transform more easily to untempered

high-carbon martensite which causes embrittlement. Their volume fraction therefore needs to be minimised. Similar results have been obtained for the austempered ductile iron (Moore *et al.*, 1987, Rouns and Rundman, 1987).<sup>†</sup> The ductility decreases as the amount of blocky austenite increases. A smaller fraction of this austenite can be achieved by allowing a longer time  $t_1$  at the austempering temperature assuming that the reaction has not saturated. The difficulty is to ensure that the time  $t_2$ , when the residual austenite begins to decompose to carbides is longer than  $t_1$ . For example, it has been demonstrated that beyond a certain manganese concentration,  $t_2$  is always found to be smaller than  $t_1$  (Moore *et al.*, 1986, 1987; Rouns and Rundman, 1987). This limiting manganese concentration is found to depend on the austenitising temperature which determines the carbon concentration in the austenite (Moore *et al.*, 1987). Hence, a higher austenitising temperature should lead to more of the blocky austenite. With lower bainite, the volume fraction of blocky austenite is reduced because some of the carbon is then tied up as carbides in the bainitic ferrite (Moore *et al.*, 1987). However, the higher strength causes a reduction in toughness. Cast irons containing upper bainite generally have a tensile strength in the range 960–1150 MPa with a tensile elongation up to 13%, whereas the corresponding data for lower bainite are 1310–1495 MPa and 5% elongation (Moore *et al.*, 1987).

There are other special effects concerning the bainite transformation in cast irons. In wrought steels, bainite inevitably nucleates at the austenite grain boundaries. In cast irons, the nucleation of bainite is also found to occur at the austenite/graphite interfaces (Moore *et al.*, 1985a). The interface between graphite and iron is weak, and it may in fact be the case that nucleation occurs at the free surface, produced by detachment of the graphite from the matrix. The regions where the nodules of graphite form are in general poorer in alloy concentration, and this might explain their ability to preferentially stimulate the nucleation of bainite (Rouns and Rundman, 1987). Cast irons usually contain chemical segregation which is more pronounced than in steels. Solute concentrations tend to be highest in the interdendritic and intercellular regions,

<sup>†</sup>The terminology used to identify the blocky regions is confusing. The regions are designated *untransformed austenite volumes (UAV)*, their volume fraction being measured using point counting on a light microscope. However, the films of austenite within the bainite sheaves are not included in this analysis, even though they are also untransformed. Hence, the volume fractions of austenite reported using X-ray diffraction analysis (which includes both films and blocks) are higher than the values reported for the UAV regions. A further difficulty is that the carbon concentration of the UAV regions is assumed to be unchanged by the formation of bainite (Moore *et al.*, 1986); this is incorrect and must lead to an overestimation of carbon in the remaining microstructure, since some of the microstructural parameters are derived using mass conservation conditions.

which are the last to solidify (Moore *et al.*, 1985a,b). The problem is made worse by the rather high mean alloy concentration of bainitic cast irons, necessitated by the need for the austenite to have sufficient hardenability to avoid the formation of pearlite, especially for applications where heavy section castings are required (Rundman *et al.*, 1988). The presence of pearlite leads to poor mechanical properties when compared with bainite (Dorazil *et al.*, 1962; Shiokawa, 1985). The effect of chemical segregation is to give a non-uniform distribution of bainite. Solute-rich regions which are unable to transform to bainite during austempering decompose into untempered, high-carbon martensite on cooling to ambient temperature and cause a marked decrease in ductility (Moore *et al.*, 1985a; Rundman *et al.*, 1988).

A more uniform distribution of bainite can be obtained in spite of the chemical segregation, by holding for a longer time at the isothermal transformation temperature. The regions rich in austenite-stabilising elements then have an opportunity to transform. A difficulty is that carbides might form in the regions which transform first to bainite (Moore *et al.*, 1985a,b). Another procedure involves ausforming prior to austempering, the rate of transformation to bainite being higher everywhere, in lightly deformed austenite (Moore, 1985a). Ausforming is a specialised treatment involving austenite deformation before transformation and does not seem practicable for cast irons.

There is an effect of segregation which cannot be eliminated by any of the treatments discussed above. The good properties of austempered ductile irons are due to the stable austenite. A short austempering time leads to less carbon enrichment in the residual austenite, which consequently suffers from mechanical instability. Longer times, on the other hand, induce carbide precipitation. There is therefore an optimum austempering time which depends on alloy chemistry. Segregation causes this optimum time period to be different in different regions of the sample, making it impossible to obtain stable austenite throughout the sample (Rundman *et al.*, 1988).

### **13.12.2 Wear of Bainitic Cast Irons**

Many cast irons are used in situations where resistance to abrasive wear is an important requirement. The mechanical properties (high strength and ductility) of austempered cast irons, together with their cheapness, make them potential candidates for applications involving abrasive wear. Many investigations have now indicated that the austempered ductile cast irons have better abrasive wear resistance when compared with normalised cast irons of the same chemical composition (Lu and Zhang, 1989) or abrasion-resistant steels, including the Hadfield manganese steel, of equivalent hardness (Shepperson and Allen, 1987). It appears that the improved wear resistance is

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at least partly a consequence of the stress-induced transformation of austenite into hard, high carbon martensite, together with the inherent ductility of the microstructure (Shepperson and Allen, 1987). The high work hardening rate usually associated with the transformation induced plasticity may also make a contribution.