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Alternatives to the Ferrite–Pearlite Microstructures

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Abstract

Structural steels based on a mixed microstructure of allotriomorphic ferrite and pearlite have a well–established history of cost–effectiveness and reliability. The purpose of this paper is to review the possibility of alternatives to this microstructure for large scale applications. Controlled–rolled bainitic steels, accelerated cooled steels, ultra–low carbon bainitic steels and inoculated steels are discussed in this context.

Introduction

I had the privilege to work with the late Professor Urkola Galarza for about two years; we even shared in the teaching of steels for three wonderful days set in his Basque Country. He had a passion for steels – with Rodriguez Ibabe he wrote what must be the only book on the metallurgy of steels in the Basque language [1].

Professor Urkola's dedication to steel is easy to understand; the material is remarkable in its complexity and use. There are three stable allotropes that occur in nature, with a body– centered cubic, hexagonal close–packed and cubic–close packed crystal structure. Two further allotropes (face–centered tetragonal and trigonal) can be created artificially, and there are at least six different magnetic transitions in solid iron. Just the addition of carbon then leads to a large variety of new possibilities including the classical ferrite, Widmanstätten ferrite, pearlite, bainite and martensite microstructures.

The most popular microstructure in the context of structural steels has undoubtedly been a mixture of ferrite and pearlite. Typical chemical compositions are given in Table 1 together with an indication of the usual mechanical properties. Fig. 1 shows that the main effect of microalloying (and the associated thermomechanical processing) is to refine the microstructure. The niobium carbonitrides that form also strengthen the ferrite by interphase precipitation or by strain–induced precipitation [2–5].

Pickering [4] has listed some of the major applications of these steels: oil and gas pipelines, drilling rigs, production platforms, ships, pressure vessels and tubing, earth–moving equipment, heavy goods vehicles and automobile components, high– rise buildings, bridges, transmission towers, fuel and other storage tanks and reinforcement bars for concrete. 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 majeures*).

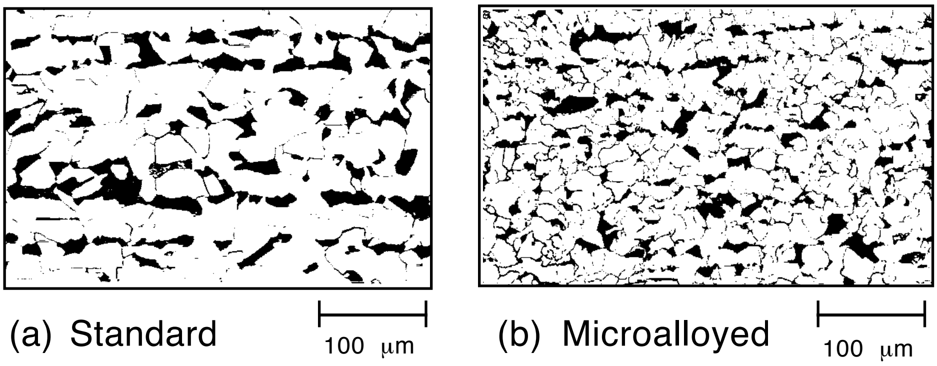

Fig. 1: Optical micrographs of banded ferrite–pearlite microstructures in (a) standard, (b) niobium microalloyed structural steel.

Type	Composition / wt.% C	Si	Mn	Nb	Stress / MPa σ_y^l	σ_y^u	UTS	Elongation %	C_V (−20 °C) / 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 interlamellar spacing
Standard	76	24	15 μm	0.22 μm
Microalloyed	78	22	9 μm	0.22 μm

Table 1: Typical ferrite–pearlite structural steels in both the standard and niobium microalloyed conditions.

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

Strength

The most elementary engineering specification is strength, which is easily obtained by transforming the austenite at ever decreasing temperatures (Fig. 2) and by increasing hardenability. And there are many other strengthening mechanisms available. Unfortunately, toughness does not necessarily increase with strength. The weldability may also deteriorate because the heat that flows into the steel causes the region adjacent to the fusion boundary to austenitise and more remote regions to be tempered. Two potential problems arise in this heat–affected zone (HAZ): (i) the excessive use of alloying elements can cause the formation of untempered hard phases in the fully austenitised region; (ii) there may be excessive softening in the tempered regions.

The tendency to soften depends on how far the microstructure is from equilibrium (Table 2). 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.

Steels that are stronger than 1000 MPa yield are important in certain applications, but the biggest markets are for lower strength varieties where the total alloy content rarely exceeds 2 wt.%. The

Phase Mixture in Fe–0.2C–1.5Mn wt.% at 300 K	Stored Energy / J mol^{-1}
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

Table 2: The stored energy as a function of microstructure, relative to the equilibrium state defined as a mixture of ferrite, cementite and graphite. 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 solutes are configurationally frozen (for martensite even the interstitial elements are frozen). Case 6 refers to an iron–base mechanically alloyed oxide–dispersion strengthened sample which to my knowledge is the highest reported stored energy prior to recrystallisation [6].

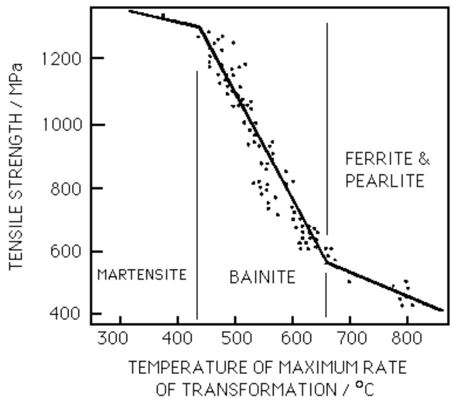

Fig. 2: Variation in the tensile strength of structural steels as a function of the temperature at which the rate of transformation is greatest during continuous cooling heat treatment [7].

alloy design then has to ensure sufficient hardenability to achieve the desired microstructure whilst at the same time avoiding the formation of hard martensite. Irvine and Pickering therefore developed a series of low–alloy, low–carbon steels, containing small amounts of boron and molybdenum to suppress allotriomorphic ferrite formation. Boron increases the bainitic hardenability and hence reduces the need for other solute elements which promote the formation of martensite. Steels like these (Alloy 1, Table 3) are found to transform into bainitic microstructures with very little martensite using simple, normalising heat treatments.

The other alloys listed in Table 3 are examples of some recent commercial bainitic steels. Although the steels listed appear to be similar, this observation is misleading because they have quite different mechanical properties because of the prominent role of the trace element concentrations, and because there are significant differences in the processing routes.

The steels developed by Irvine and Pickering exhibited reasonable combinations of toughness and

No.	C	Si	Mn	Ni	Mo	Cr	Nb	Ti	B	Al	N	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	0.020	0.20	2.00	0.30	0.30	–	0.050	0.020	0.0010	–	0.0025	ULCB
6	0.028	0.25	1.75	0.20	–	0.30	0.100	0.015	–	0.030	0.0035	ULCB (Cu 0.3, Ca 0.0004)
7	0.080	0.20	1.40	–	–	–	–	0.012	–	0.002	0.0020	acicular ferrite, TiO_x (Oxygen 0.0017)
8	0.080	0.20	1.40	–	–	–	–	0.008	0.0015	0.038	0.0028	acicular ferrite, TiB
9	0.080	0.20	1.40	–	–	–	–	0.019	–	0.018	0.0050	acicular ferrite, TiN

Table 3: Typical compositions (wt.%)of bainitic steels. The alloys appear similar in composition but the differences in carbon and trace element concentrations, and processing, are important.

strength, but in time proved to be unexciting when compared with the best of quenched and tempered martensitic steels. Nevertheless, the physical metallurgy principles established during their development are now being applied towards a new generation of bainitic steels, in which the emphasis is on further reductions in carbon and other solutes, together with advanced processing in order to refine the microstructure. The range of bainitic alloys which might reasonably be regarded as alternatives to the ferrite–pearlite microstructures are now discussed in greater depth.

Control–Rolled Bainitic Steels

The strengthening of iron by reducing its grain size is attractive because it gives a simultaneous improvement in the strength and toughness. This has led to the development of an impressive technology for reducing the austenite grain size and hence the scale of any transformation microstructure. In *controlled–rolling*, the ingots are reduced by hot–rolling at high temperatures making the austenite recrystallise many times before the finish–rolling temperature is reached, giving a fine austenite grain structure prior to transformation. Any grain growth during the hot–rolling process is hindered by the use of *microalloying* additions such as niobium, vanadium or titanium. These elements have a low solubility in austenite and are added in small concentrations ($\simeq$ 0.03-0.06 wt.%) to form stable carbides or carbonitrides which impede grain growth during hot deformation and subsequent cooling. It is also possible to continue deformation so that the austenite grains are flattened (pancaked) just before transformation because this leads to an even finer final microstructures.. A typical chemical composition of a steel suitable for control–rolling to a ferrite and pearlite microstructure is Fe–0.08C–0.3Si–1Mn–0.03Nb–0.004N wt.%.

The controlled–rolling process described above has now been adapted for bainitic alloys. There are two ways in which a bainitic microstructure can be obtained:

(i) The cooling rate can be increased to allow the austenite to supercool into the bainite transformation range.

(ii) Hardenability can be modified, thereby avoiding process changes.

This second option is discussed first; rapid cooling will be described in the next section. Alloying elements such as manganese are boosted in order to increase the possibility of forming bainite at the expense of allotriomorphic ferrite. Unlike conventional steels for control–rolling, TiN particles (of size $\simeq 0.02\mu\,\mathrm{m}$) 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 so that the alloys are best produced by continuous casting. Slabs of the material are then reheated to a relatively low reaustenitisation temperature of 1150 °C, the particles inhibiting austenite grain growth. Reductions in grain size are obtained by repeated recrystallisation during control–rolling. The particles also help to produce a more uniform grain structure than is obtained in conventional control–rolling treatments. Finish rolling is carried out at a temperature where recrystallisation does not occur, resulting in deformed and pancaked austenite grains which then transform to a fine bainitic microstructure on further cooling.

The details of the steelmaking process are important in determining the final properties of control–rolled bainitic 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.

Rapidly Cooled Control–Rolled Bainitic Steels

Bainitic microstructures can be generated in control–rolled steel by changing the cooling rate during processing, rather than by increasing hardenability. This is desirable when weldability is important, in order to avoid the formation of martensite in the heat affected zone of the welded plate. The technology of rapid cooling during controlled–rolling is not as trivial as it sounds, given the speed of production, the kinetics of transformation, the need to avoid shape distortion and the need to achieve uniform cooling rates. Considerable 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".

Pipeline and Plate Steels

There is a general 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, thereby reducing costs. The thickness of the steel used can be reduced by increasing the strength but without sacrificing toughness or weldability. 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 ($\simeq$ 10 MPa).

It is found that if, after controlled–rolling, the steel is cooled from the austenite phase field at a rate which prevents the formation of large quantities of allotriomorphic ferrite, but which is low enough to avoid martensitic transformation, then a fine–grained microstructure which is a mixture of allotriomorphic ferrite and bainite is formed. Such a microstructure has an increased yield strength and good toughness. The cooling rates involved are higher (10–40 $^\circ\mathrm{C\,s^{-1}}$ over the temperature range 800–500 °C) than those appropriate for normal control-rolled steel processes (Fig. 3). The accelerated cooling is achieved in industrial practice by the use of water spray curtains directed on either side of the hot plate (which could be some 15 mm in thickness) in a manner designed to ensure uniform cooling and to minimise distortion. The rapid cooling of thick plate requires different technology with much more careful control of water pouring in order to cope with the large thickness and width of the plates and the relatively slow rate at which the steel moves through the mill.

The process leads to a more refined microstructure in which the bainite platelets contribute to strength and toughness via their fine grain size. For a given strength level, the properties can therefore be achieved using lower alloying concentrations, with the concomitant advantages of reduced cost and in some cases, increased weldability.

Although most investigations of accelerated cooled steels have relied on light microscopy which lacks

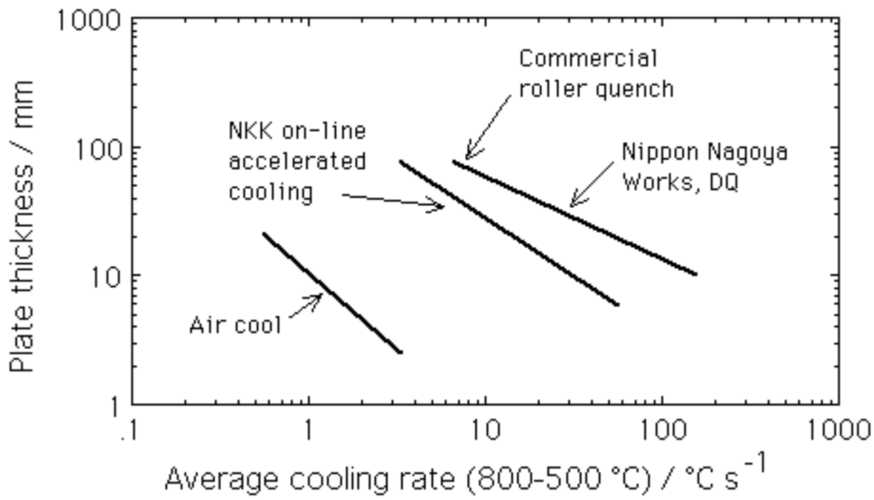

Fig. 3: An illustration of the average cooling rates associated with the manufacture of steels for structural applications [8].

the necessary resolution, a detailed investigation [9] has demonstrated that the microstructure of these rapidly cooled steels does indeed consist of a mixture of ferrite and bainite. The bainite consists of sheaves of platelets of submicron thickness, as compared with the equiaxed grains of allotriomorphic ferrite of size approximately 5 μm. It is also found to have a much higher dislocation density of $1.7 \times 10^{14}\,\mathrm{m}^{-2}$, compared with $0.4 \times 10^{14}\,\mathrm{m}^{-2}$ of the 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 by the bainitic transformation which occurs after the allotriomorphic ferrite [10]. It is as a consequence of the lower transformation temperatures involved, that the overall microstructure of these steels is found to be more refined relative to the conventional control–rolled steels. The volume fraction of bainite can vary from about $0.2 \rightarrow 1.0$ depending on the alloy chemistry and cooling conditions. Typical compositions for accelerated–cooled alloys are given as Alloys 2-4 in Table 3. Of these, Alloy 2 is the leanest and can be expected to contain the least quantity of bainite.

The production of the steels is not a continuous process of casting and controlled–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 very coarse austenite grain structure which evolves during ingot cooling is disrupted by transformation to ferrite. Hence, 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 $20\,^{\circ}\mathrm{C\,s}^{-1}$ to around 450 °C before allowing natural cooling. This treatment changes the microstructure of the plate from the normal mixture of ferrite and pearlite to a ferrite–bainite microstructure giving 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 and 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.* [11] 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 also do not exhibit sudden yielding, although the relevance of this to pipeline applications is not clear [12].

Process Parameters

There are a number of controllable factors which can significantly influence the properties of the final product [9, 12, 13]. For example, a high ingot reheating temperature allows more of the niobium carbonitrides to dissolve in the austenite; the niobium carbonitrides may subsequently precipitate during the $\gamma \rightarrow \alpha$ reaction to give fine dispersions within the ferrite ("interphase precipitation"), leading to an enhancement of its strength.

The temperature at which the rolling operation finishes is critical in the sense that it should leave the final austenite grains in an unrecrystallised, pancake shape. This ensures a further degree of refinement of the microstructure obtained after transformation and avoids the undesirable recrystallisation texture of austenite. If the finish rolling temperature is too low (below $Ar3$), then the allotriomorphic ferrite will deform, leading to an increase in strength but at the expense of toughness. For this reason, the deformation should be restricted to the austenite phase. The finish rolling temperature (T_R) also influences the variation in mechanical properties through the thickness of heavy gauge plates (Fig. 4). The surfaces, where the cooling rates are the largest, tend to be harder when 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 consequently tend to transform more rapidly; this counteracts the effect of the higher surface cooling rates [14].

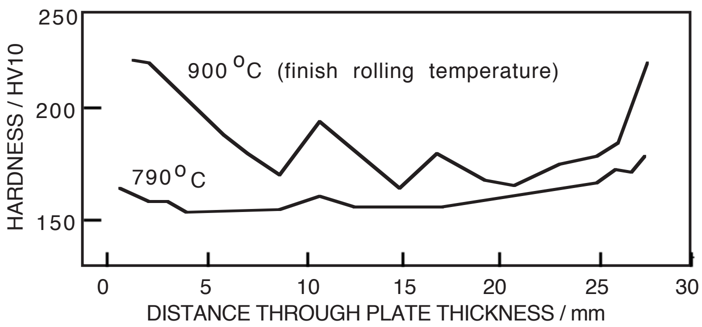

Fig. 4: The effect of the temperature at which rolling is completed, on the variation in hardness of a Fe–0.16C–0.63Mn wt% accelerated cooled steel containing bainite [15].

The steels used in the accelerated cooling operations have a hardenability which is so high as to prevent the completion of the bainite reaction during continuous cooling to ambient temperature. Large amounts of martensite are found in the microstructure under some circumstances, leading to a drop in toughness and a significant distortion of the plate product. As a consequence, the cooling has to be arrested by cutting off the water sprays at temperatures ranging between $600 \rightarrow 450\,^\circ\mathrm{C}$, depending on the steel chemistry and the exact cooling conditions. The subsequent slower cooling rate allows the bainite transformation to proceed to a larger extent, leaving only very small amounts of residual austenite which may transform to martensite on further cooling. Another related problem has been found in alloys with a relatively high hardenability, typically those low–alloy steels with more than about 1.4 wt.% manganese. When produced using controlled–rolling and accelerated cooling, the yield strength decreases even though the tensile strength does not [Fig. 5a, 16]. This is because martensite replaces bainite as the dominant hard phase, and as in dual–phase steels, yielding becomes a gradual process as the stress is raised. 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.

A two–stage accelerated cooling process has therefore been developed to enhance the chances of forming bainite instead of martensite, while at the same time retaining the high cooling rate required to refine the allotriomorphic ferrite that forms first (Fig. 5b). After thermomechanical processing while the steel is in the austenitic condition, it is cooled rapidly ($25\,^\circ\mathrm{C\,s^{-1}}$) through the ferrite temperature range in order to obtain the fine ferrite grain size, but the cooling rate is then reduced to about $3\,^\circ\mathrm{C\,s^{-1}}$ over the temperature range where bainite forms, giving a greater opportunity for transformation to proceed before the martensite–start temperature is reached. 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 also important. The

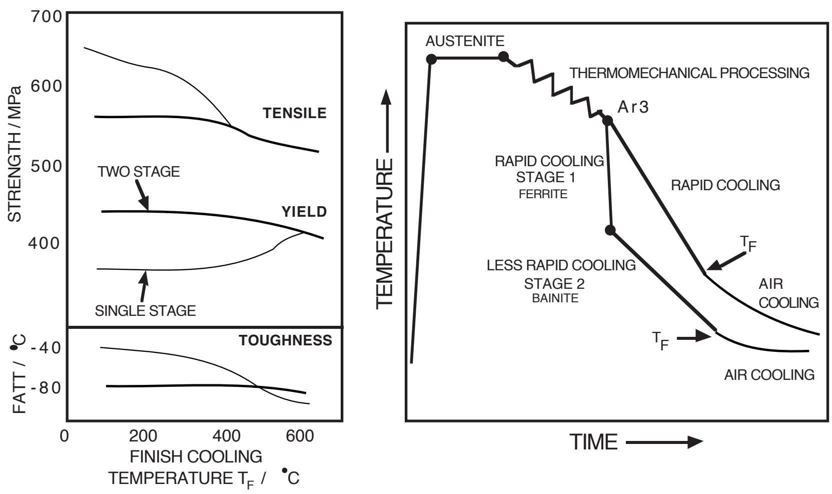

Fig. 5: (a) The relationship between the tensile and yield strength, and toughness of accelerated cooled steels as a function of the temperature at which the forced cooled is stopped, for the single and double stage processes. (b) Schematic illustration of the thermomechanical cycles associated with the two–stage accelerated cooling process [17].

mechanical properties are found to be much 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. 5a). The process is found to be successful in raising the yield stress of the steel when compared with the conventional accelerated cooling procedure.

Nishioka and Tamehiro [18] have emphasised that a general problem with the accelerated cooled steels is that the toughness and microstructure are not maintained in the heat– affected zones created after welding. On the other hand, the strength of these rapidly cooled steels is some 50 MPa greater than that of conventional control–rolled plates. For those applications where this excess strength is not needed, the carbon–equivalent of the steel could in principle be reduced further by the removal of alloying elements, thereby improving the weldability (Fig. 6). A further difficulty is that the process is not suitable for heavy gauge plates (20–30 mm thick) since it is then impossible to ensure uniform cooling throughout the depth of the samples [12]; it is often the case that the central regions of such samples transform instead to a ferrite and pearlite microstructure.

In control–rolled steels, heavy gauge plates which are cooled slowly after rolling, have a tendency to develop a coarse ferrite grain structure at the surfaces. This appears to be a consequence of the recrystallisation of ferrite grains deformed by rolling in the $(\alpha + \gamma)$ phase field, although the fact that recrystallisation only happens at the surface implies that the deformation must have been inhomogeneous. Accelerated cooling has the added advantage that it inhibits this recrystallisation at the surface [13].

Segregation

Control–rolled steels are made using the continuous casting process and therefore suffer from pronounced chemical segregation approximately along the mid–thickness of the plate. The resulting microstructure can be related directly to variations in the manganese concentration, which can reach twice the average value along the centre of the plate. Ferrite forms first in the manganese–depleted regions causing the carbon concentration of the manganese–enriched regions of austenite to increase. This is turn exaggerates the hardenability of the manganese–enriched regions, which then transform into bands of hard microstructure.

These hard bands are susceptible to hydrogen cracking. An advantage of the accelerated cooled

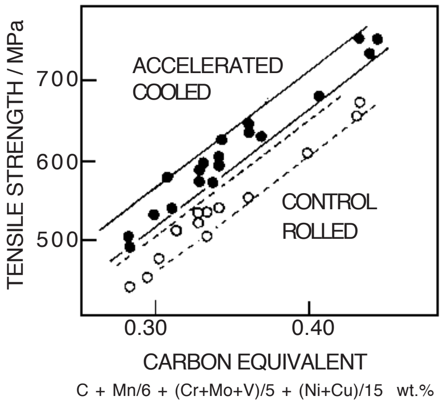

Fig. 6: Relationship between the carbon equivalent and tensile strength for conventionally produced control–rolled steels (CR) and accelerated cooled steels (AC); the latter have a mixed microstructure of ferrite and bainite [13].

steels is that the resulting microstructure of bainite and ferrite is experimentally found to be less sensitive to solute segregation, when compared with the gross banding effect observed with ferrite–pearlite microstructures [9,13]. The transformation in rapidly cooled steels is suppressed to lower temperatures, where nucleation is possible in all regions including the solute enriched zones, giving a more uniform distribution of carbon (Fig. 7). The resulting lower hardness in the segregated zone makes the steel less susceptible to hydrogen–induced cracking.

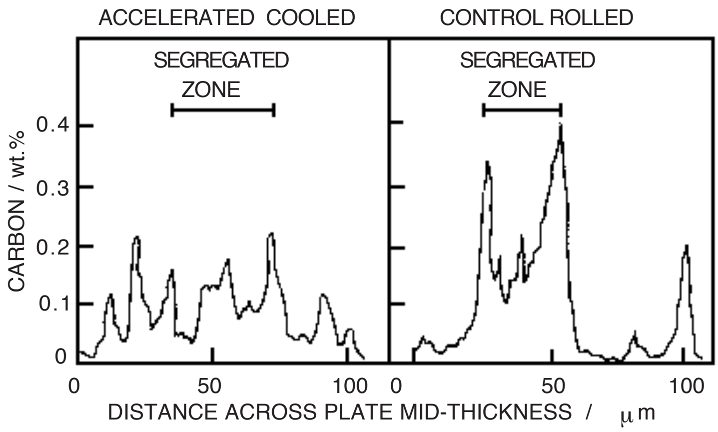

Fig. 7: Distribution of carbon concentration in the segregated zone for conventional control–rolled and rapidly cooled steel plates [19].

Ultra–Low–Carbon Bainitic Steels

Improvements in weldability brought about by reductions in carbon concentration can lead to major economies in fabrication, especially if the welding can be carried out without any preheat. Hence the development of the *ultra–low carbon bainitic* (ULCB) steels (Alloy 5 and 6, Table 3) with carbon concentrations in the range 0.01–0.03 wt.% [20, 21]. This level of carbon is sufficient to react with microalloying additions but not sufficient to allow the formation of martensite. The reduction in martensite at low carbon concentrations leads to an improvement in toughness without any undue loss of strength (Fig. 8). The carbon concentration should not fall below 0.01 wt.% since niobium carbide (and TiN) is necessary to ensure the development of a fine austenite grain structure during controlled–rolling. The niobium and titanium additions are also known to be effective in suppressing the precipitation of $Fe_{23}(CB)_6$ during control–rolling [22]. Only free boron is useful in enhancing hardenability. In other respects the ULCB steels are similar in concept to the Irvine and Pickering alloys; they contain boron and molybdenum or chromium to enhance bainitic hardenability, and titanium to getter nitrogen.

Boron can cause the hardness of the heat affected zone associated with welding to increase, but fortunately, its effect in ULCB steels is small because of very low carbon concentration, which also permits the greater use of substitutional solutes (Mn, Ni) compared with the Irvine and Pickering steels.

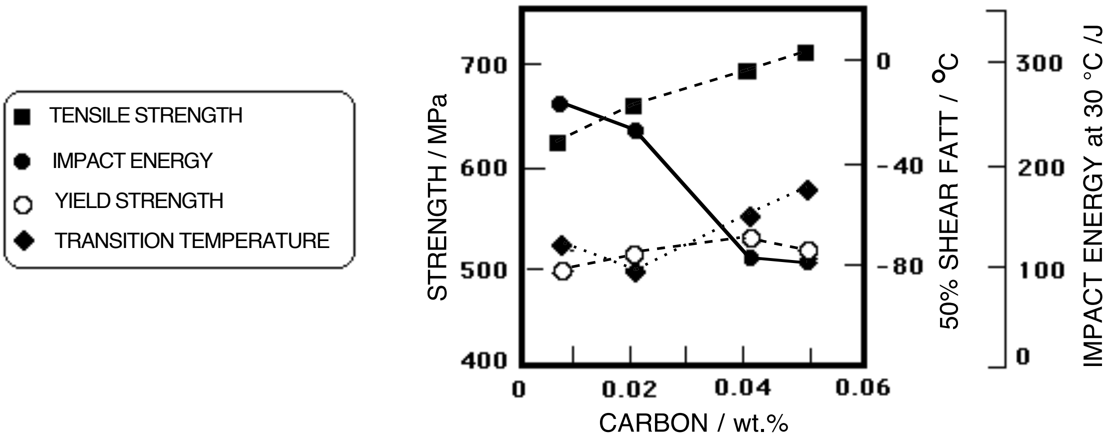

Fig. 8: Variations in mechanical properties as a function of carbon concentration in thermomechanically processed ultra–low–carbon steels [21]. The fracture assessed impact transition temperature (FATT) begins to increase as the carbon concentration falls below about 0.02 wt.%.

The solubility of niobium in ordinary steels is limited by their relatively high carbon concentrations, most of the niobium being present as carbonitrides. By contrast, a larger fraction of the niobium remains dissolved in ULCB steels, which also contain Ti to combine with free nitrogen. Their response to niobium additions is therefore different from that of conventional microalloyed steel. The free niobium can significantly influence the transformation behaviour [23,24]. The finish rolling temperature for ULCB steels has a large influence on the soluble niobium in the final steel (Fig. 9a). Less niobium remains in solution as the finish rolling temperature is reduced because there is then a greater opportunity for strain–induced precipitation.

The variations in dissolved niobium affect the microstructure in air–cooled ($1\,^{\circ}\mathrm{C\,s^{-1}}$) ULCB steels (Fig. 9b), perhaps because niobium strongly retards the formation of allotriomorphic ferrite, leading to an increase in the fraction of bainite. The effect is masked during accelerated cooling (15– $20\,^{\circ}\mathrm{C\,s^{-1}}$) because the transformation to allotriomorphic ferrite is avoided (Fig. 9b).

ULCB steels are strong (500–620 MPa), tough and weldable. They have been designated for use in high strength line pipe in Arctic or submarine environments.

Inoculated Acicular Ferrite Steels

Acicular ferrite is intragranularly nucleated bainite. Whereas bainite contains packets of parallel ferrite plates, the acicular ferrite microstructure is much more chaotic with plates lying on many different planes. Such clusters of nonparallel platelets lead to an enhancement of toughness since, unlike bainite, any propagating crack frequently encounters plates in different crystallographic orientations.

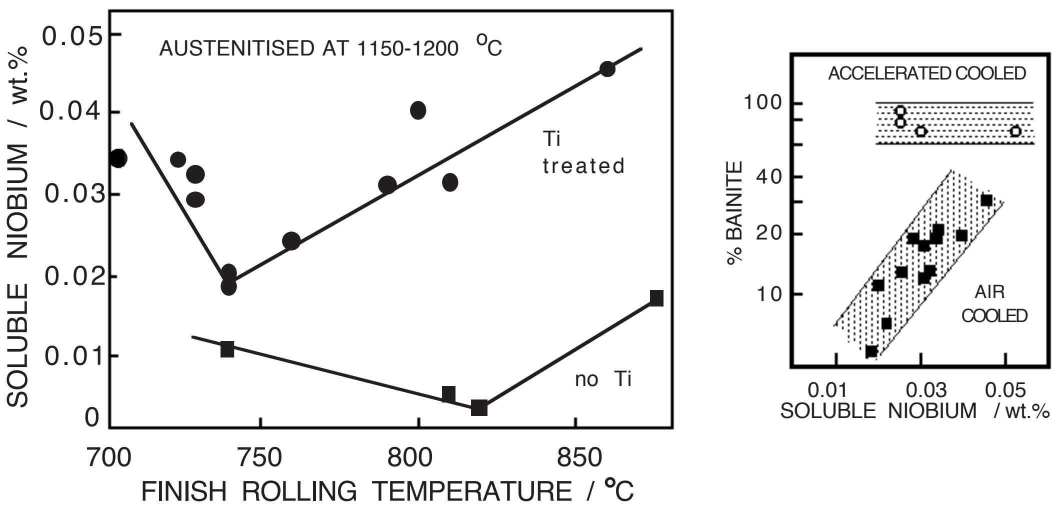

Fig. 9: Soluble niobium in ULCB steel [21]. (a) Variation in the soluble niobium concentration as a function of the finish rolling temperature; (b) variation in microstructure as a function of the finish rolling temperature.

The difference in microstructure arises because bainite nucleates from austenite grain *surfaces* thus allowing the development of packets, whereas acicular ferrite plates nucleate from *point* sites which are usually non–metallic inclusions.

Given the more desirable microstructure of acicular ferrite, it is now possible to inoculate steels with selected oxide particles in order to induce the formation of acicular ferrite for improved toughness [18,25,26]. Some examples of titanium oxides and nitride inoculated acicular ferrite steels are given in Table 3 (Alloys 7–9).

There appear to be considerable difficulties in the production of such steels so it seems unlikely that they will see many applications.

CONCLUSIONS

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

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