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Novel Steels for Rails

Modern railway systems are subject to intense use, with fast trains and large axle loads. There are many criteria which determine the suitability of a steel for rail track applications. The primary requirement is structural integrity, which can be compromised by a variety of fatigue mechanisms, by a lack of resistance to brittle failure, by localised plasticity and by excessive wear. All of these depend on interactions between engineering parameters, material properties and the environment. The track material must obviously be capable of being manufactured into rails with a high standard of straightness and flatness in order to avoid surface and internal defects which may cause failure. Track installation requires that the steel should be weldable and that procedures be developed to enable its maintenance and repair. Commercial success depends also on material and life-time costs.

1. Track Materials

There are many kinds of loadings which can adversely affect the life of rails; amongst these, wear and plastic deformation induced by contact stresses can combine to cause unacceptable changes in the rail head profile. The system of contact stresses is dependent upon the relative motions of the 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 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 1). Pearlite consists of a mixture of soft ferrite and a hard, relatively brittle iron carbide called cementite. Each colony of pearlite is a bicrystal of cementite and ferrite which in two-dimensional sections appears to consist of alternating lamellae of ferrite and cementite (Hillert, 1962). The hardness increases as the interlamellar spacing decreases or as the fraction of cementite within the pearlite is made larger. Pearlite presumably achieves a high resistance to wear because of the hard cementite and its containment by the more plastic ferrite.

The very cementite particles that confer hardness are also brittle. The situation is made worse by the fact that each pearlite colony is a bicrystal. It is the size of the colony, rather than the interlamellar spacing, which defines the length scale of fracture. Pearlitic steels are

not therefore tough. Fracture can occur from relatively minor stress-concentrating features inside the rail, or on its surface, as a result of manufacture or subsequent handling damage. At least one European railway network suffers almost 4000 rail fractures every year (Yates, 1996). Although such fractures are rarely dangerous when actively managed, they entail a high replacement cost and can be disruptive to the network (Yates, 1996).

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

Table 1: Compositions (wt%) of typical bainitic and pearlitic rail steels (Bhadeshia, 2001).

Wear is a system rather than a material property, but it is nevertheless useful to identify material factors which are important. With eutectoid steels, the wear rate decreases as the hardness increases (Fig. 1), although exceptions have been reported (Kalousek *et al.*, 1985b). Fig. 1 reveals that the microstructure influences wear; refining the microstructure prolongs the wear-limited rail life (Kalousek and Beynon, 1975).

The size of a pearlite colony and its interlamellar spacing can be refined by transformation at lower temperatures. But a greater degree of refinement can be achieved by transforming instead to bainite.

2. Bainite

Bainite is a non-lamellar aggregate of carbides and plate-shaped ferrite (Fig. 2). Each ferrite plate is about $10\mu\text{m}$ long and about 0.2×10^{-6} m thick, making an individual plate invisible in the optical microscope. The fine scale of the microstructure is beneficial to both the strength and toughness. Within the broad classification of bainite, there are two particular forms.

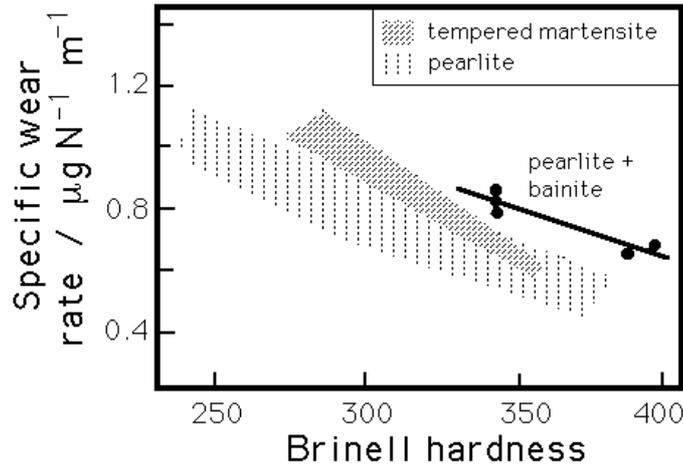


Fig. 1: The correlation of hardness with the wear rate for martensitic, bainitic and pearlitic microstructures (Hodgson and Preston, 1988).

Upper bainite consists of clusters of platelets of ferrite which are in identical crystallographic orientation and intimately connected to the austenite in which they grow. Elongated cementite particles decorate the boundaries of these platelets, the amount and continuity of the cementite layers depending on the carbon concentration of the steel.

As the transformation temperature is reduced, some of the carbon is encouraged to precipitate inside the ferrite plates, leading to the lower bainite microstructure (Fig. 2).

3. Bainitic Rails

There has been considerable effort devoted to finding alternatives to the pearlitic rails, but with confusing results. The wear resistance and toughness of pearlitic rails has been compared with a “bainitic” rail containing chromium and molybdenum (Ghonem *et al.*, 1982). The rail head microstructure in fact consisted of 70% bainite and 30% pearlite but exhibited unacceptable toughness; the wear on the gauge face where there is rolling–sliding contact, was also worse. Heller and Schweitzer (1980) compared the properties and service performance data from bainitic rail steels (alloys 8,11, Table 1) with those from a pearlitic rail steel (alloy 4, Table 1). The bainitic 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 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 indicated a wear rate some ten times faster than pearlite of the same hardness (Ichinose *et al.*, 1982). Even mixed microstructures of bainite and pearlite are found to be less wear resistant

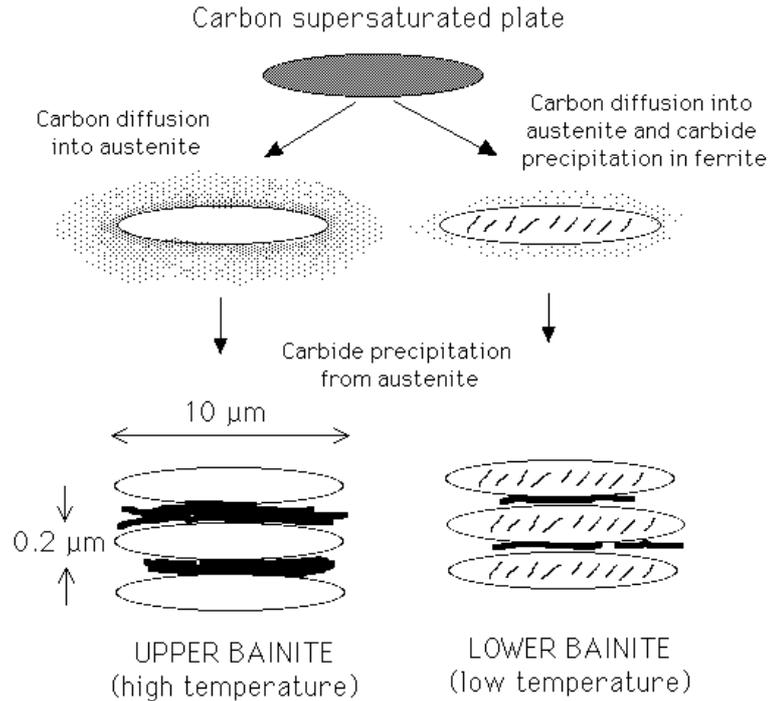


Fig. 2: Summary of the mechanism and microstructure of bainite in steels (Matas and Hehemann, 1961; Takahashi and Bhadeshia, 1990; Bhadeshia, 2001). A plate of bainite forms without diffusion, but any trapped carbon then partitions into the residual austenite where it eventually precipitates as cementite. If the transformation temperature is sufficiently small, some cementite may also precipitated within the supersaturated bainitic ferrite, giving rise to the lower bainitic microstructure.

when compared with fully pearlitic steels Fig. 3, (Kalousek *et al.*, 1985a,b).

By contrast, low-carbon bainitic steels (alloy 6, Table 1) 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 a low carbon concentration is that it enables the steels to be welded easily. Tests using pure sliding, cooled, pin-ring tests have demonstrated that low-carbon bainitic steels might have comparable or superior wear resistance to the pearlitic steels, Fig. 3. Even when the pearlitic and bainitic steels have similar wear characteristics, the lower carbon concentration of bainitic steels ensures better ductility, toughness and weldability.

Many of the pessimistic early results on bainitic steels can be challenged on the basis of inadequate characterisation and unrepresentative testing (Clayton *et al.*, 1987, 1990). Nevertheless, a useful study by Garnham (1989) proved that carbide-free bainite does not resist wear as effectively as pearlite. Although carbide containing bainitic steel performed better, it

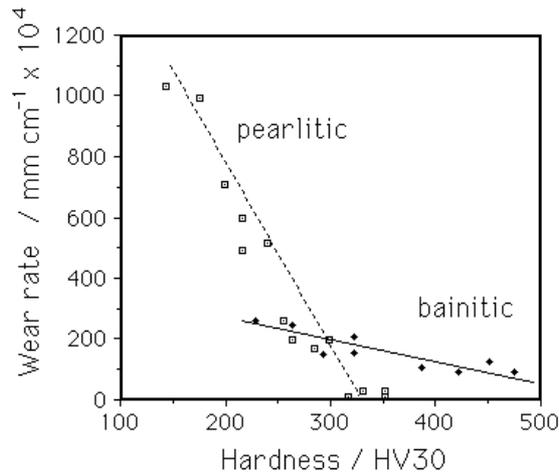


Fig. 3: Pin-ring wear rate versus hardness data for pearlitic and bainitic steels (Clayton *et al.*, 1987).

increased the wear of the mating pearlitic wheel. It may be concluded from Garnham's work that bainitic steel is not suitable for railway applications where rolling-sliding wear limits life.

The most recent work has been systematic in the assessment of wear and microstructure (Devanathan and Clayton; 1991). Three steels with carbon concentrations ranging from 0.04 to 0.54 wt% (alloys 8,9 & 12, Table 1) were examined in their bainitic condition. The lowest carbon steel, because of its greater ductility and work-hardening rate outperformed pearlitic steels at the same hardness. The medium carbon steel was similar in its wear performance to pearlite, whereas the higher carbon steel was found to be worse. Microstructural observations 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.

3. Silicon-Rich Carbide-free Bainitic Rail Steels

There clearly have been many attempts at improving rail steels using bainite, but with mixed outcomes. Low-carbon bainitic steels outperform pearlite but only in the low-hardness regime. High-carbon bainitic steels are harder but perform badly relative to pearlitic track. By analogy to high-strength steels (Bhadeshia and Edmonds, 1983; Caballero *et al.*, 2001), some of the problems of the higher strength bainitic rail steels must be attributable to the coarse carbides associated with the bainite microstructure (although this on its own does not explain all the problems encountered with bainite). The precipitation of cementite during the bainite transformation can be suppressed by alloying with about 1.5 wt.% of silicon, which has a very low solubility in cementite and greatly retards its growth (Bhadeshia, 2001).

An interesting microstructure results when this silicon-alloyed steel is transformed into

upper bainite. The carbon that is rejected into the residual austenite, instead of precipitating as cementite, remains in the austenite and stabilises it down to ambient temperature. The resulting microstructure consists of fine plates of bainitic ferrite separated by carbon-enriched regions of austenite (Fig. 4).

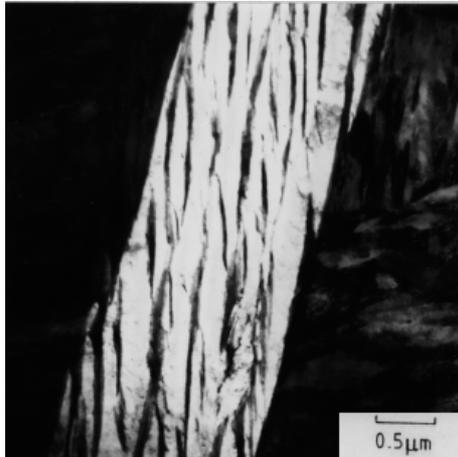


Fig. 4: Transmission electron micrograph of a sheaf of bainite, consisting of fine ferrite platelets separated by films of carbon-enriched retained austenite.

The potential advantages of the mixed microstructure of bainitic ferrite and austenite can be listed as follows:

- (a) Brittle cementite is eliminated, making the steel more resistant to cleavage fracture and void formation.
- (b) The bainitic ferrite is depleted in carbon; dissolved carbon is known to embrittle ferrite.
- (c) The microstructure, which is generated by a simple heat treatment, derives its strength from ultrafine ferrite plates, which are much less than one millionth of a meter in thickness. This cannot be achieved by any other commercially viable process. Furthermore, refinement is the only mechanism for simultaneously improving the strength and toughness of steels.
- (d) The ductile films of austenite which are intimately dispersed between the plates of ferrite have a crack blunting effect.
- (e) The diffusion of hydrogen in austenite is slower than in ferrite. Austenite can therefore lead to an improved stress-corrosion resistance.
- (f) Steels with this microstructure are cheap. All that is required is that the silicon concentration should be large enough to suppress cementite.

In spite of these appealing features, the bainitic ferrite/austenite microstructure does not always give the expected good combination of strength and toughness. Although the films of austenite enhance toughness, there remain relatively large “blocky” regions of austenite between the sheaves of bainite (Fig. 5). These blocks are less stable and transform into high-carbon martensite under the influence of a small applied stress. This untempered, hard martensite embrittles the steel. The blocks can in principle be eliminated by allowing more transformation, but as shown in the next section, there is a thermodynamic limit to the amount of bainite that can form in the absence of carbide precipitation. It is ironic that the prevention of cementite precipitation (in order to eliminate a brittle phase) leads to the retention of large blocks of unstable austenite which then embrittle the steel. To solve the problem, it is necessary to manipulate the thermodynamic limit to permit more bainitic ferrite to form and consume the large blocks of austenite whilst still retaining the films of stable austenite.

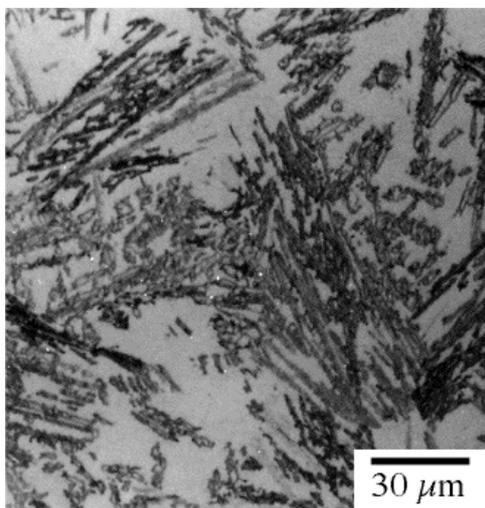


Fig. 5: Optical micrograph of upper bainite in an Fe-0.43C-3Mn-2.02Si wt.% showing the blocks of retained austenite between sheaves of bainite.

4. Mathematical Model

We have seen that the blocks of austenite are detrimental to toughness. Anything that can be done to reduce their size, or to increase their stability to avoid martensitic transformation, is expected to be beneficial.

Both of these aims can be achieved from an understanding of the mechanism of the bainite reaction; Fig. 2 shows that bainite grows like martensite, without diffusion, but that the excess carbon is soon afterwards partitioned into the residual austenite. Diffusionless growth requires that transformation occurs at a temperature below a temperature T_0 , when the free energy

of bainite becomes less than that of austenite of the same composition (Fig. 6). Fig. 6 also shows the locus of the T_0 temperature as a function of the carbon concentration, plotted on the Fe–C phase diagram. Growth without a composition change can only occur if the carbon concentration of the austenite lies to the left of the T_0 curve. It is necessary to account for the strain energy; this can be done by raising the free energy curve of the ferrite by a quantity equal to the strain energy, thus defining the T'_0 curve which occurs at slightly lower carbon concentrations with $T'_0 < T_0$.

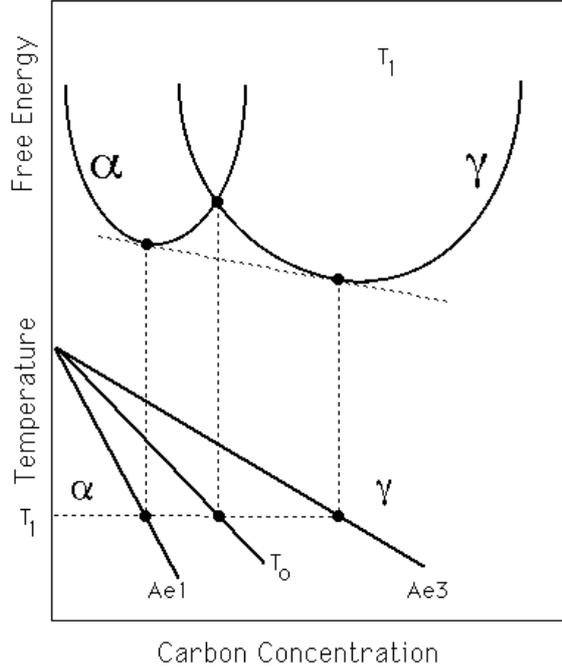


Fig. 6: Illustration of the origin of the T_0 construction on the Fe–C phase diagram. Diffusionless transformation is thermodynamically impossible if the carbon concentration of the austenite exceeds the T_0 curve.

Suppose that the plate of bainite forms without diffusion, but that any excess carbon is soon afterwards rejected into the residual austenite. The next plate of bainite then has to grow from carbon-enriched austenite (Fig. 7). This process must cease when the austenite carbon concentration reaches the T'_0 curve. It is this cessation of transformation which is responsible for the persistence of the large blocks of austenite in the microstructure illustrated in Fig. 5.

The maximum fraction of bainite that can form in a silicon-rich steel is given by the application of the lever rule to the T'_0 curve:

$$V_{\alpha_b}^{max} = \frac{x_{T'_0} - \bar{x}}{x_{T'_0} - x_{\alpha}} \quad (1)$$

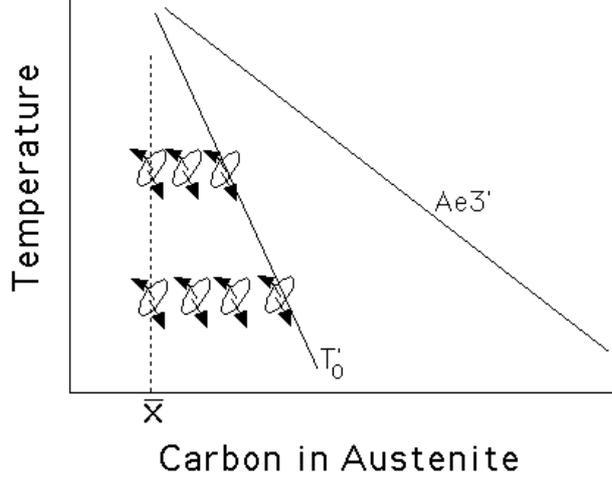


Fig. 7: The incomplete reaction phenomenon. During isothermal transformation, a plate of bainite grows without diffusion, then partitions excess carbon into the residual austenite. The next plate therefore has to grow from carbon-enriched austenite. This process continues until diffusionless transformation becomes impossible when the austenite composition eventually reaches the T'_0 boundary.

where $x_{T'_0}$ is the carbon concentration given by the T'_0 boundary, \bar{x} is the mean carbon concentration of the alloy, and $x_\alpha \simeq 0$ is the carbon concentration left in the ferrite.

An increase in $V_{\alpha_b}^{max}$ leads to a reduction in blocky austenite and an increase in the stability of the remaining austenite. Equation 1 shows that this can be achieved by modifying the substitutional solute concentration such that the T'_0 curve is shifted to higher carbon concentrations.

It can be demonstrated that the reduction ΔT_0 in the T_0 temperature caused by the addition of a substitutional element (concentration x in atomic %) is given approximately by (Honeycombe and Bhadeshia, 1995):

$$\Delta T_0 = \frac{x(b_{NM}\Delta T_{NM} + b_M\Delta T_M)}{b_{NM} + b_M} \quad (2)$$

where the values of b are listed in Table 2a. ΔT_{NM} and ΔT_M are temperature changes due to a unit change in the concentration of a substitutional solute. They come from the Zener factorisation of the magnetic and non-magnetic components of the free energy difference between austenite and ferrite. The effect of several alloying elements can be approximated by assuming additivity:

$$\Delta T_0 = \frac{\sum_i x_i (b_{NM}\Delta T_{NM_i} + b_M\Delta T_{M_i})}{b_{NM} + b_M} \quad (3)$$

Using these equations and the data from Table 2, we see that for the Fe–3Mn–2.02Si–0.43C wt% (2.97Mn, 3.87Si at.%) alloy, $\Delta T_0 = 116$.

The T'_0 curve for a plain carbon steel, accounting for 400 J mol^{-1} of strain energy, is over the temperature range of interest, given approximately by (Honeycombe and Bhadeshia, 1995)

$$T'_0(\text{ K}) \simeq 970 - 80x_C$$

so that for the alloy it is modified to,

$$T'_0(\text{ K}) \simeq 970 - 80x_C - 116 \quad (4)$$

Hence, manganese is seen to have a large effect in depressing the T'_0 temperature. An examination of Table 2b shows that one possibility is to replace all of the manganese with nickel. Thus, for a Fe–4Ni–2Si–0.4C wt% (3.69Ni, 3.85Si at.%) alloy, a similar calculation shows that $\Delta T_0 \simeq 72$ so that

$$T'_0(\text{ K}) \simeq 970 - 80x_C - 72 \quad (5)$$

Function	a	b	Temperature Range
$\Delta G_{NM}^{\gamma\alpha}$	–6660	7	$900 > T > 300 \text{ K}$
$\Delta G_M^{\gamma\alpha}$	650	–1	$900 > T > 620 \text{ K}$
$\Delta G_M^{\gamma\alpha}$	0	0	$T < 620 \text{ K}$

Table 2a: Approximate representations of the magnetic (ΔG_M) and non-magnetic (ΔG_{NM}) free energy components for the $\gamma \rightarrow \alpha$ transformation in pure iron. Each free energy term is given by $\Delta G = a + bT \text{ J mol}^{-1}$

This alteration in composition, which leads to an increase in the T'_0 curve, causes a remarkable improvement in toughness by eliminating the blocky austenite, without sacrificing strength (Bhadeshia and Edmonds, 1983). The same concept can therefore be used to design a carbide-free bainitic rail. The manufacture of rails involves continuous cooling transformation. Anisothermal transformation can be taken into account using a theory due Takahashi and Bhadeshia (1991). These theories were integrated into a mechanism-based model, leading to the design of a novel rail steel with a carbide-free bainitic microstructure (British patent, 1995). The approximate composition of one of the alloys thus designed is Fe–0.4C–1.5Si–2.0Mn–0.25Mo wt%, the molybdenum being added to prevent prior austenite grain boundary embrittlement due to phosphorus.

Solute	$\Delta T_M / \text{K per at.}\%$	$\Delta T_{NM} / \text{K per at.}\%$
Si	-3	0
Mn	-37.5	-39.5
Ni	-6	-18
Mo	-26	-17
Cr	-19	-18
V	-44	-32
Co	19.5	16
Al	8	15
Cu	4.5	-11.5

Table 2b: Values of ΔT_M and ΔT_{NM} for a variety of substitutional solutes (Aaronson *et al.*, 1966)

The new alloys designed in this manner have superior wear resistance (Fig. 8); uniquely, they also lead to reduced wheel wear (Yates, 1996). Tests have shown dramatic improvements in the rolling contact fatigue life. For example, whole wheel/rail tests showed crack initiation at two hundred and twenty thousand cycles on traditional pearlitic premium rail, whereas after one million cycles the new rail steel was crack free when the test was terminated to permit further trials to take place (Yates, 1996). The new steels are weldable using the thermit and flash butt welding processes.

There are a number of reasons why the new material exhibits these desirable properties. The high toughness contributes to wear resistance by reducing the rate of production of wear debris (Pitt, 1999). During wear, there is a transfer of material between the contact surfaces, followed by detachment of wear particles from the transfer layer. The bainitic steels have a softer transfer layer whilst retaining a high bulk hardness. The absence of brittle cementite particles must also contribute to the improved rolling-contact fatigue resistance.

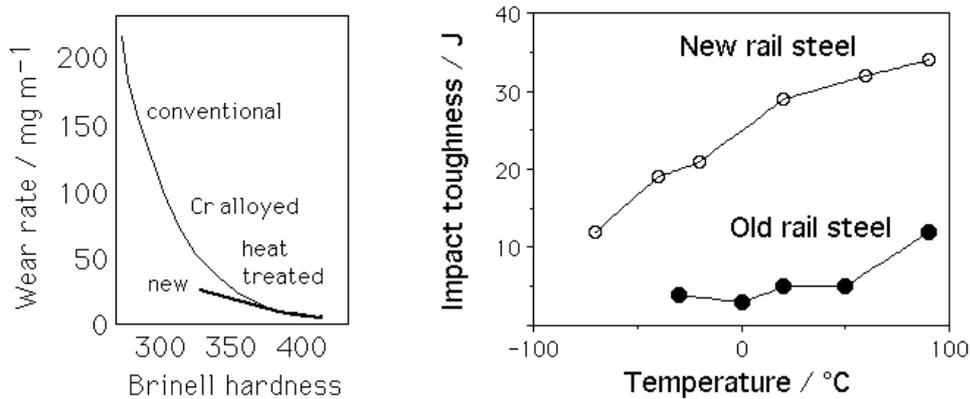


Fig. 8: (a) Comparison of the wear rates and hardness levels of a variety of conventional rail steels against the new alloys which have a carbide-free microstructure of bainitic ferrite and austenite. (b) Comparison of the Charpy toughness of new and conventional rail steel (Jerath *et al.*, 1991).

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