Stabilisation of ferrite in hot rolled δ-TRIP steel

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A steel has recently been designed to benefit from the deformation induced transformation of retained austenite present in association with bainitic ferrite. It has as its major microstructural component, dendrites of δ-ferrite introduced during solidification. The δ-ferrite replaces the allotriomorphic ferrite present in conventional alloys of this kind. The authors examine here the stability of this δ-ferrite during heating into a temperature range typical of hot rolling conditions. It is found that contrary to expectations from calculated phase diagrams, the steel becomes fully austenitic under these conditions and that a better balance of ferrite promoting solutes is necessary in order to stabilise the dendritic structure. New alloys are designed for this purpose and are found suitable for hot rolling in the two-phase field over the temperature range 900–1200°C.

Keywords: TRIP steel, δ-ferrite, Alloy design, Phase transformation

Introduction

A delta transformation induced plasticity (δ-TRIP) steel has an unconventional microstructure in which the major constituent is δ-ferrite, which forms during solidification and is permanently retained after solidification.[1] The rest of the structure can be transformed into a mixture of bainitic ferrite and austenite, which is stabilised by the partitioning of carbon between these two allotropic forms of iron. It is this austenite, which behaves as in conventional TRIP assisted steels,[2–9] and undergoes deformation induced transformation to martensite and, as a result, enhances the ductility of the alloy. Hence, the acronym TRIP, which stands for transformation–induced plasticity.[10] The steel has interesting properties in its as cast state: an ultimate tensile strength of ~ 1000 MPa and a total elongation, almost all of which is uniform, of 23%.

There is a difficulty with the alloy design. In spite of cooling rates as slow as ~ 20 K s^{-1}, solid state transformation after solidification is completed exhibits large deviations from equilibrium conditions.[1] Compositions which should, according to phase diagram calculations, show large quantities of δ-ferrite dendrites, often display zero or much reduced fractions on casting. This is because the austenite that forms during cooling by the solid state transformation of cored δ-dendrites does so without the required partitioning of substitutional solutes, particularly aluminium.

It is important in the δ-TRIP concept to maintain, as far as is possible, the ferrite produced during solidification because its presence at all temperatures should improve its resistance spot weldability. If the material fails to become fully austenitic in the heat affected zone (HAZ), then it also becomes impossible to obtain a fully martensitic structure there. This should result in better welds because large gradients in properties in the affected region are known to dramatically reduce the shear resistance of spot welds.[11] Indeed, current TRIP assisted steels are known to be difficult to spot weld because of the production of fully martensitic regions in the HAZ.

The problem of retaining the δ-ferrite can, of course, be solved by adding aluminium in concentrations greater than required by equilibrium. This increases the fraction of δ-ferrite in the cast structure, even in the absence of equilibrium partitioning.[12] However, it is necessary to examine the stability of the δ-ferrite also in the reheated condition since the casting must ultimately be hot rolled in order to produce the form required for potential applications, such as in the car industry. The purpose of the present work is to assess the capability of the alloy system to sustain the ferrite during the reheating required for the hot rolling process, which is usually in the range 1200–900°C.

The purpose of the work presented here was to create a variant of the δ-TRIP steel in which the ferrite is retained during excursions to very high temperatures. It is hoped that the present work will form the basis of a future development programme, where the steel can be produced by hot rolling with the required quantity of ferrite, followed by the formation of the mixture of bainitic ferrite and retained austenite in a continuous process.

Experimental

Phase diagrams were estimated using MTDATA[13] in combination with the TCFE (version 1·21) database. The alloys were manufactured as 34 kg ingots of dimensions $100 \times 170 \times 230$ mm using a vacuum

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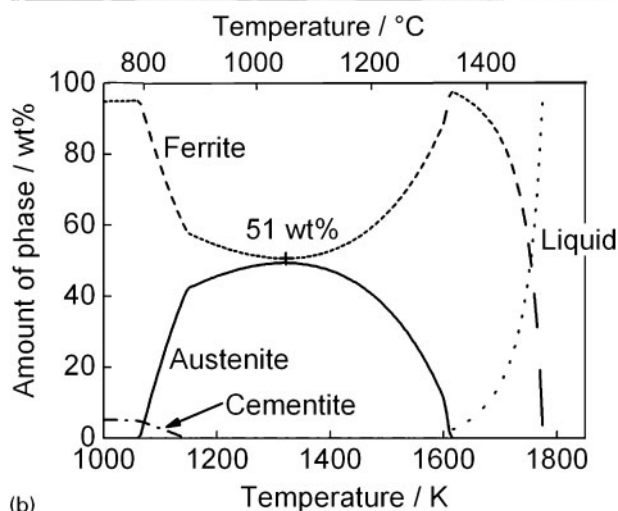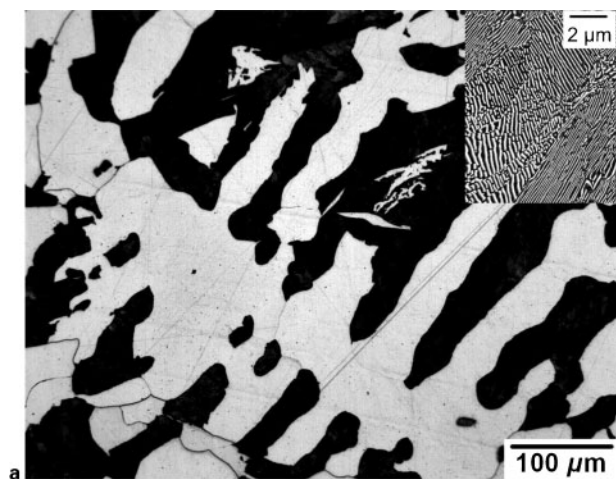
1 *a* as cast microstructure showing δ-ferrite dendrites: dark regions are pearlitic, as illustrated at high magnification in inset, and *b* calculated equilibrium phase diagram (Alloy 2)

furnace. In one case, representing the final alloy choice, the ingot was reheated to 1200°C for rough rolling to make 25–30 mm slabs followed by air cooling. These slabs were then reheated to 1200°C (heating rate to 1200°C not monitored) and hot rolled to 3 mm in thickness with the temperature always maintained above 900°C followed by cooling.

Optical microscopy samples were prepared using standard methods and etched in 2% nital. Heat treatments were conducted on cylindrical dilatometric samples of diameter 5 mm and length 10 mm using a

Table 1 Compositions achieved during manufacture, wt-%: alloys were manufactured as 34 kg ingots of 100 × 170 × 230 mm dimensions using vacuum furnace

	Alloy 2	Alloy 3	Alloy 4	Alloy 5	Alloy 6	Alloy 7
C	0·37	0·40	0·40	0·41	0·37	0·39
Si	0·23	0·26	0·74	0·26	0·76	0·77
Mn	1·99	2·02	1·99	1·53	1·53	1·50
Al	2·49	2·50	2·39	2·30	2·91	3·35
Cu	0·49	…	0·49	0·49	…	…
P	0·02	0·02	0·02	0·02	…	…
S	0·0036	0·0013	0·0015	0·0014	0·0042	0·0045
N	0·0048	0·0032	0·0024	0·0030	0·0020	0·0022

push rod 'BAHR DIL805' high speed dilatometer with radio frequency induction heating; the equipment has been described elsewhere.[14] The sample temperature is measured by a thermocouple welded to its surface using a precision welder and jig supplied by the dilatometer manufacturer.

The cast structure of the alloy identified as 'Alloy 2' in Table 1 has been investigated previously,[12] and hence, its designation is a maintained here in order to avoid confusion. The alloy is a new melt of the original δ-TRIP design,[1] needed to provide more material for experiments; the design emerged from the neural network and genetic algorithm analysis, and the details are described fully in Ref. 1.

Reheating experiments

The microstructure of the as cast state of Alloy 2 is illustrated in Fig. 1 along with the calculated equilibrium phase diagram, which indicated that a large fraction of the δ-dendrites generated during solidification should, under equilibrium conditions, persist during reheating at all temperatures.

Samples of the steel were heated at 20°C s^{-1} to peak temperatures of 800, 850, 900, 1000, 1100 and 1290°C for 5 min; further samples were similarly heated to 1360, 1380 and 1400°C for 1 min. After the small holding period, the samples were all quenched at −80°C s^{-1}. The heat treatments were conducted on the dilatometer. The heating rate and transformation times used here will be different from those practised in an industrial environment, but the difference is unlikely to be significant given the rapid rate of austenite formation at high temperatures.

Metallographic studies were conducted on all the samples, but only selected examples are presented here. Figure 2*a* shows the effect of heating to 800°C; there is clear evidence for the partial transformation of pearlite into austenite, so that quenching leads to martensite in the centres of the interdendritic regions, where austenite stabilising elements are partitioned.[12] The austenite that forms from pearlite then begins to penetrate the δ-dendrites. The amount of austenite obtained is inconsistent with the equilibrium phase diagram (Fig. 1), although these observations are not surprising given that the austenite grows without the required level of solute partitioning between the parent and product phases.[12]

By 950°C, the δ-dendrite arms are no longer visible as uniform features, but on a microscopic scale, consist of martensite and remnants of ferrite, which are in the form of a network, presumably reflecting segregation patterns (Fig. 2*b*). The sample heat treated to a peak temperature of 1100°C became almost fully austenitic, although the overall structure remained fine because of the pinning of austenite grain boundaries by remnants of ferrite (Fig. 3*a*). The ferrite content only recovered slightly when the peak temperature reached 1400°C, in a form reminiscent of the δ-ferrite, presumably following the solidification induced chemical segregation patterns existing in the sample.

It is evident the alloy would not be suitable for a process in which substantial δ-ferrite must be retained in the microstructure during the hot rolling process, even though the equilibrium phase diagram suggests otherwise. These large deviations from expectation make it difficult to design suitable alloys; in previous work,[12] the
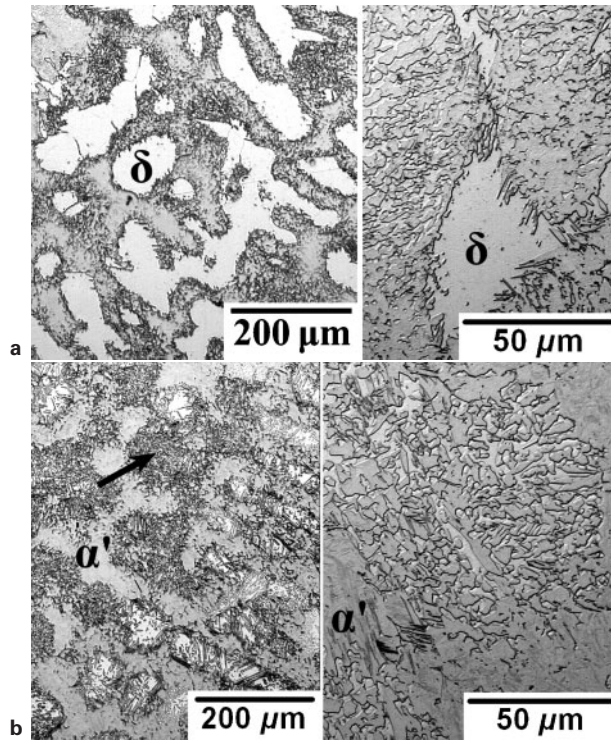
2 *a* Alloy 2 heated to 800°C and quenched: light etching regions represent δ-dendrites and darker regions are mixtures of martensite (austenite at 800°C) and ferrite, and *b* Alloy 2 heated to 900°C and quenched: arrow shows region which used to be δ-ferrite and which is now mixture of martensite and residual ferrite

authors used DICTRA, which allows growth to be modelled, to establish that the magnitudes of the kinetic effects are reasonable, but the method requires assumptions about shape and scale, which make its use for alloy design in the present context difficult. Therefore, a pragmatic approach was adopted in which new alloys were designed based essentially on experience, with the broad aim of stabilising the ferrite.
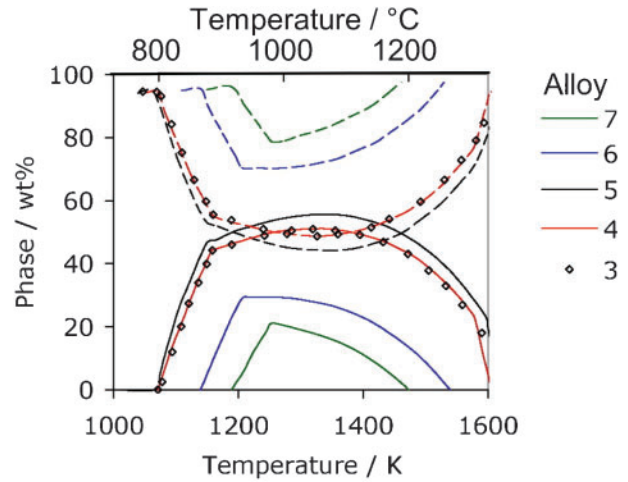

4 Calculated quantities of austenite (continuous lines) and ferrite (dashed lines): Alloy 3 is plotted as points in order to avoid confusion with Alloy 4; although only austenite and ferrite are illustrated for clarity, liquid and cementite phases were allowed to exist

New alloys

Referring to Table 1 and using Alloy 2 as a reference, Alloy 3 is based on the removal of copper, which is an austenite stabilising element. The silicon concentration is increased in Alloy 4 since both aluminium and silicon[15] have the same γ-loop forming tendency and hence favour the formation of δ-ferrite. Alloy 5 is based on a reduction in the manganese concentration and Alloy 6 is based on a simultaneous reduction in Mn and increase in the silicon and aluminium concentrations. Alloy 7 has a lower Mn, high Si and particularly high Al concentrations. Phase diagram calculations for the new alloys are illustrated in Fig. 4, which shows that the greatest potential for increasing the stability of δ-ferrite should be in Alloys 6 and 7.

Reheating experiments were conducted as described for Alloy 2, and quantitative data are presented in Fig. 6. δ-ferrite did not persist in Alloys 2–5 on
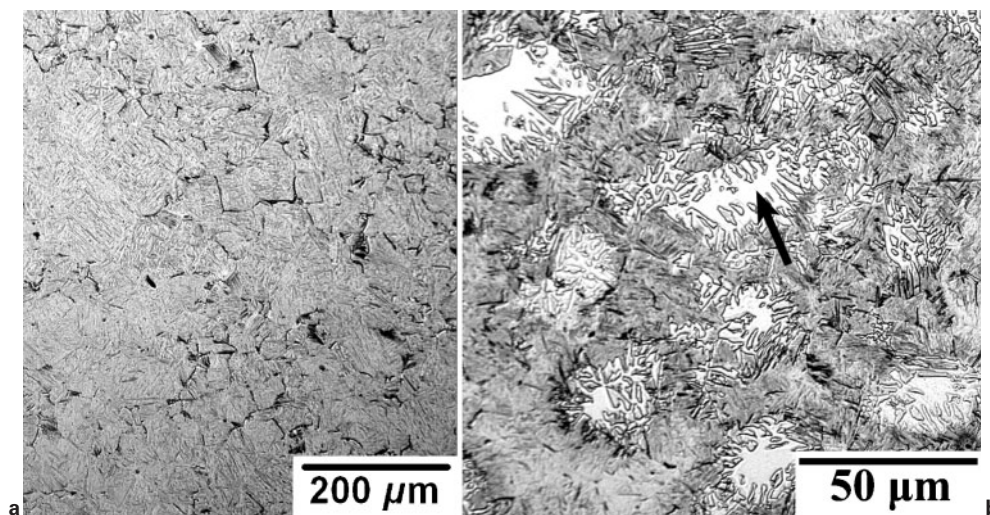
3 Alloy 2 *a* heated to 1100°C and quenched, and *b* heated to 1400°C and quenched: arrow indicates ferrite forming in segregated regions of sample, in shapes reminiscent of solidification structure
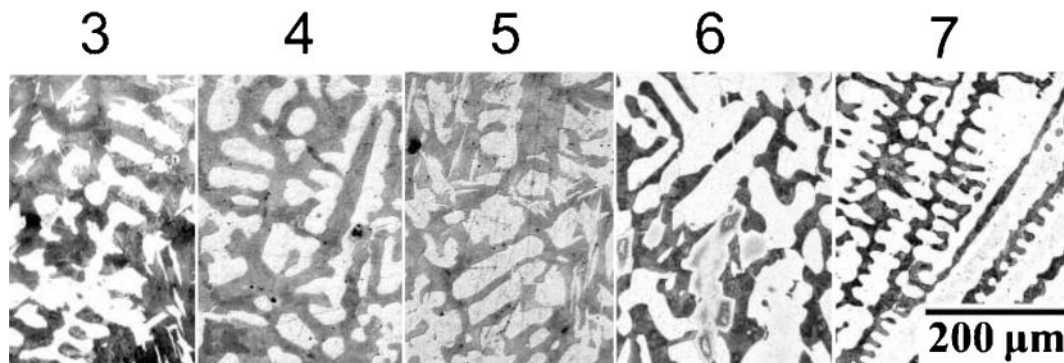
5 As cast microstructures of Alloys 3–7

reheating, but dendrites of this phase survived in Alloys 6 and 7, although the fraction of δ-ferrite is much less than expected from equilibrium (Fig. 7).

The microstructure after hot rolling is illustrated in Fig. 8, where it is evident that there are two kinds of ferrite present: the first, the δ-ferrite, which has been elongated by deformation and then the allotriomorphic ferrite, which precipitates as the temperature during rolling reduces towards 900°C. It is seen that the presence of ferrite at all temperatures maintains a fine austenite and ferrite grain structure.

Conclusions

The δ-TRIP alloy system is based on the use of a relatively large concentration of aluminium, which should lead to the formation of a substantial quantity of ferrite dendrites at equilibrium. Much of this ferrite, which forms during solidification, should, in principle, resist transformation into austenite, when the steel is heated to the high temperatures typical of hot rolling deformation. That it does not do so in practice has been shown in previous work[12] to be due to the fact that austenite is able to form by solid state transformation
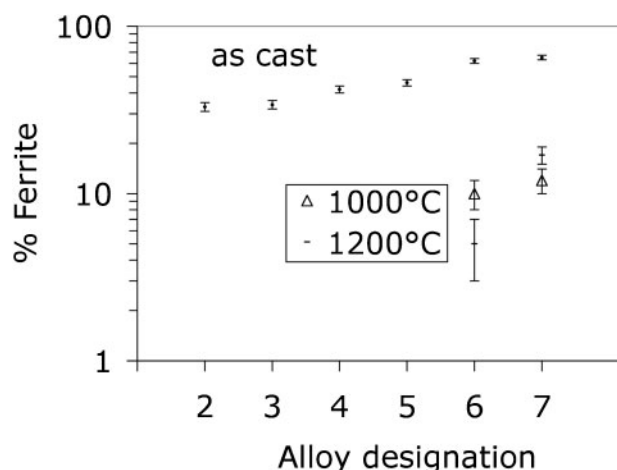
6 Volume percentages of optically resolvable ferrite for cast and reheated samples
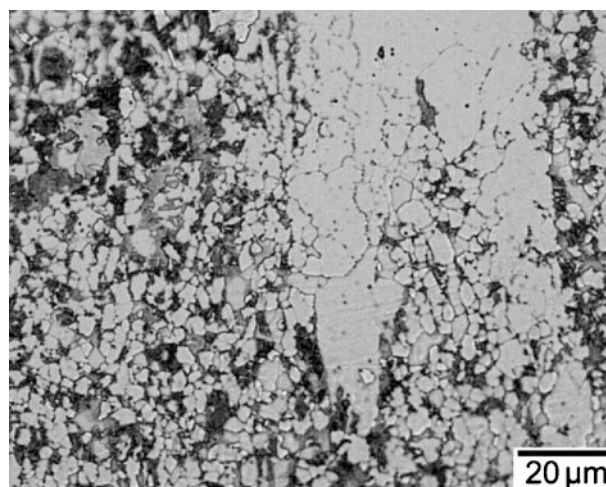
8 Alloy 7 after hot rolling
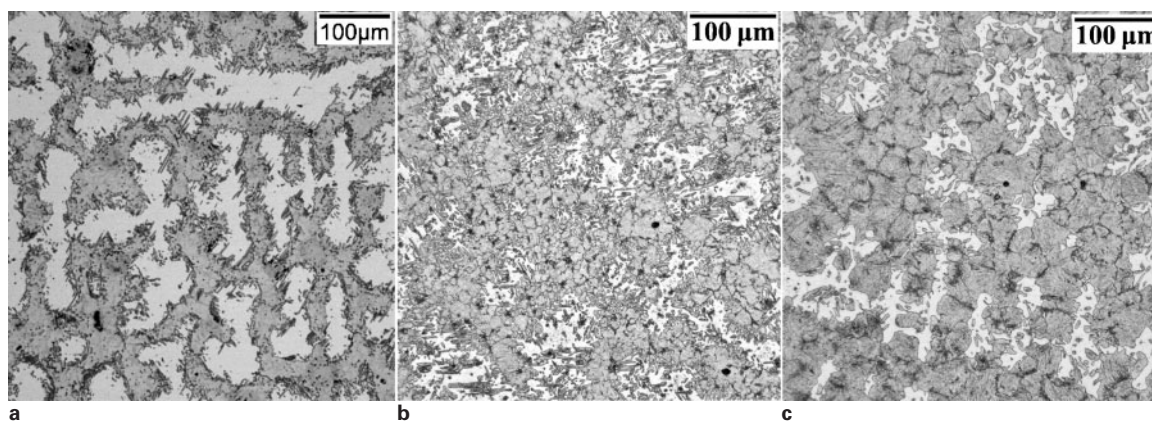
7 Alloy 7 following reheating to *a* 900°C, *b* 1000°C and *c* 1200°C

from *δ*-ferrite without the required level of partitioning of solutes, particularly aluminium.

In the present work, it has been demonstrated that suitable alloys, in which the *δ*-ferrite remains in the microstructure over the range 900–1200°C, have been designed by increasing the aluminium and silicon concentrations, accompanied by a reduction in the manganese and copper contents. It has been possible, therefore, to produce a rolled variant of the alloy, which will be used in further investigations towards the commercialisation of the steel on a continuous production line.

Acknowledgements

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