**The influence of a high aluminium addition on the strength and impact behaviour of hot rolled Nb containing steels**

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## Abstract
An addition of 0.3 wt.%Al has been made to a Nb containing 0.1 wt.%C, 1.4 wt.%Mn steel in an attempt to improve its impact behaviour on hot rolling to 15mm thick plate; the intention being to replace some of the specifications that presently can only be achieved by the more expensive methods of control rolling or normalising. Previous work had shown that a 0.2 wt.%Al addition to a hot rolled plain C-Mn steel resulted in the impact transition temperature decreasing by 40$^o$C with no significant change in strength making such a proposition possible. Unfortunately, the coarse austenite grain size of the hot rolled Nb containing steel resulted on transformation in colonies of lower transformation products and Widmanstätten ferrite forming in the ferrite/pearlite microstructure resulting in poor impact behaviour. Reducing the cooling rate by working with thicker plate, 30mm, led to little change in behaviour on hot rolling. In contrast, on control rolling, the finer grained steels which were virtually free of these lower transformation products showed a small benefit from the high Al addition even when Nb was present, particularly for the slower cooling rate, the 27J impact transition temperature decreasing by ~20$^o$C together with a small increase in yield strength. Further work is suggested to decrease the hardenability so that the original aim might be achieved.

**Key Words**: Hot rolling, Steel, Nb, high Al, MA, Lower transformation products, Impact, Strength.

## Introduction
High strength and good impact behaviour are the primary requirements for the increased industrial demands of the construction, automotive and pipeline industries [1]. These requirements can be met by control rolled steels. Control rolling although giving very good properties is more expensive than hot rolling and not only this, many smaller steel companies do not have the control rolling facilities and the powerful rolling mills needed to work at the low temperatures required. (Finish rolling temperature is 900- 800$^o$C for control rolling). However, there is always the possibility that better compositional design may allow the cheaper process of hot rolling to be used. Adding a higher than normal Al addition has been found as a promising means of achieving this [2,3,4].

Generally, only a small amount of Al is added to steel, 0.02-0.04 wt.% and this addition has no influence on the strength and impact behaviour of hot rolled steel plate; the addition being used as a de-oxidant or as a grain refiner on normalising.

In the hot rolling process, the billets are soaked at 1200-1250$^{o}$C and rolled without holds, finish rolling in the temperature range 950-1000$^{o}$C [4]. With hot rolling, the steel is rolled as fast as possible, finish rolling at a high temperature giving high productivity.
Steels with a normal Al addition of 0.02-0.04 wt.% do not usually precipitate out AlN on hot rolling, since AlN precipitates very sluggishly [5]. No grain refinement can then take place and the Al and N remain in solution. The grain size is therefore normally very coarse, 6-8mm$^{-1/2}$ (30-15µm) making it difficult to achieve either a high strength or good impact behaviour. This is in contrast to control rolling when rolling to lower temperatures not only refines the austenite grain size but by deforming belreow the non-recrystallisation temperature creates many dislocations, which supply extra sites for the nucleation of ferrite on transformation so that an even finer ferrite grain size can be achieved (8-5µm). In addition for heavily control rolled steels there is strengthening from dislocation hardening and texture development [6].

Unfortunately, because of the coarse grain size in hot rolled structural steels, they are more susceptible to Widmänstatten ferrite forming [7], as well as the lower transformation products (LTPs) like bainite and M-A (martensite/retained austenite), which can all very adversely affect the impact performance of air cooled plate.

For optimum properties on hot rolling, the composition and processing route must always ensure that these structures are solely ferrite/pearlite. For example, manganese additions by decreasing the transformation temperature so refining both the grain boundary carbides and the ferrite grain size, improves the impact performance of hot rolled steels. However, the maximum amount that can be added in steels with the present composition (~0.1 wt.%C) on air cooling 15mm plate is ~1.5 wt.%Mn. Above this amount of Mn, martensite forms making the steel brittle [8].

Similarly, as long as lower transformation products are not formed, increasing the cooling rate through the transformation encourages finer grain boundary carbides as well as a finer ferrite grain size, thus improving the toughness. However, above a certain cooling rate, Widmänstatten ferrite and martensite/bainite form and the benefit to impact behaviour is lost [9,10].

Al has also been shown to encourage martensite formation when added at the very high 1 wt.% level [11]. Although in this case, Al raises the transformation temperatures making it in theory harder for martensite to form, it also delays carbide precipitation so that there is a build-up of carbon in the austenite when transformation occurs, encouraging retained austenite to form which can then transform to martensite under deformation [12,13].

However, as long as the microstructure remains almost totally ferrite/pearlite, it has been found that by adding ~0.2 wt.%Al, the properties of hot rolled plain C-Mn steels can be significantly improved [2,3,4].

Previous work on hot rolled steel plate had examined the influence of Al additions of 0.2, 1 and 2 wt.%Al on the impact behaviour of low Si and Si-killed 0.1 wt.%C, 1.4 wt.%Mn, 0.005 wt.%N steels and the results are summarised in Fig.1, [2]. Increasing the Al content from 0.02 to 0.2 wt.% caused the impact transition temperatures to decrease by 40$^{o}$C, Fig.1, without changing the strength level giving a 27J impact transition temperature (ITT) of -95°C and a lower yield strength (LYS) of ~300MPa in 13mm thick air-cooled plate [2].

However, raising the Al level from above 0.2 wt.% to the next level in that programme, 1 wt.%Al, caused the impact behaviour to deteriorate as martensite formed. This deterioration became most marked at the 1.9 wt.%Al level when martensite formation was fully established, Fig.1.
Up to the 1 wt.%Al level there was no significant influence of Si on the impact behaviour and therefore only one average curve for 0.02 and 0.3 wt.%Si levels is shown. However, the Si addition of 0.3 wt.% at the 0.2 wt.%Al level, although not influencing the impact behaviour did increase the yield strength by 30MPa [2, 11].

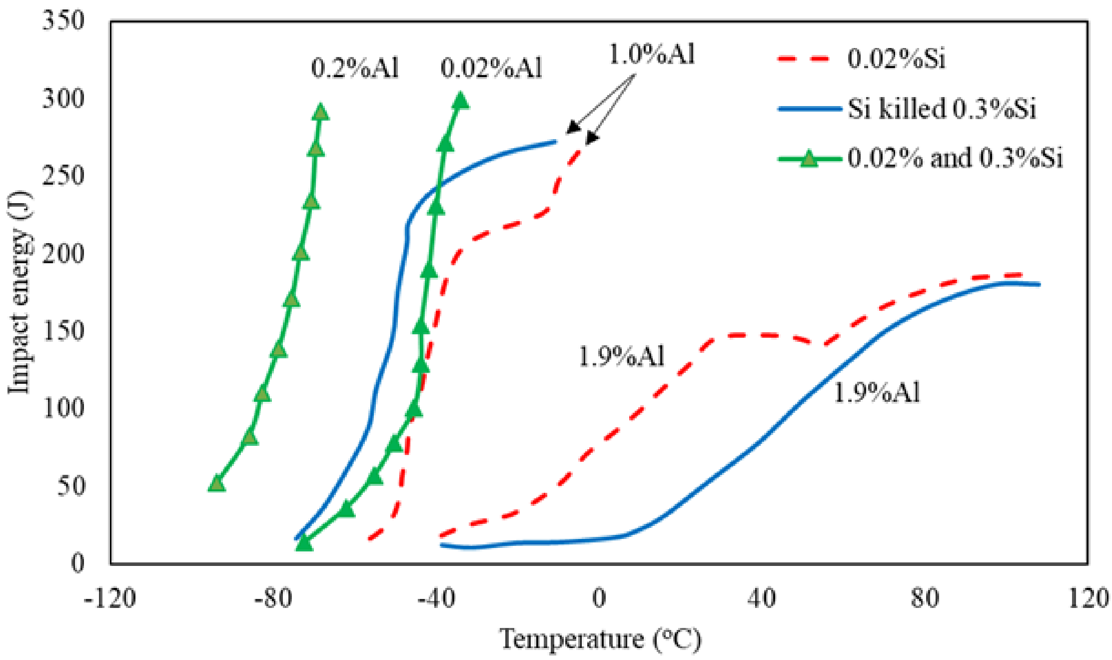



Fig.1 Schematic diagram showing the influence of Al on the ITT curves for hot rolled plain C-Mn steels. Steels had the base composition (wt.%): 0.10C, 1.4Mn, 0.01P, 0.002S and 0.004N and were examined at two Si levels, 0.02 and 0.3Si and Al levels of 0.02, 0.19, 1.0 and 2.0 [2].

Subsequent work [11] reduced the nitrogen level to a very low level ~0.001 wt.% to remove any effect of N on the impact and strength behaviour both from preventing any grain refinement by AlN and from the interstitial hardening by N. The carbon level was also reduced to 0.02 wt.% resulting in no lower transformation products being formed when air cooling the 13mm thick plates, the intention here being to isolate the effect of Al in solution on the yield strength. Al levels examined this time were 0.02 wt.%, 0.5 wt.% and 1 wt.% and increasing the Al level from 0.02 wt.% to 1 wt.% was found to increase the strength by ~55MPa due to the solid solution hardening by Al. The impact behaviour was not, however, influenced by this solid solution hardening possibly because the adverse influence on the impact behaviour of the solid solution hardening from Al was balanced by another benefit of Al, its ability to refine the grain boundary carbides [11].

Nevertheless, when N is present, previous examinations have consistently shown that adding 0.2 wt.%Al to a plain C-Mn steel improves the impact behaviour, the 27J ITT decreasing by ~40°C [2,3,4] with little change in strength. This indicates that removal of N from solution is important for achieving the benefits of the Al addition.

All the previous work [2,3,4] was carried out on simple plain C-Mn steels and although these properties are encouraging, the main intention has been to develop hot-rolled steels that can replace some of the lower strength control rolled steels (400MPa strength level).

The problem with the hot-rolled high Al (0.2 wt.%Al) steel, at this stage in the research programme, was that its strength, 300MPa, was not sufficiently high. An increase in strength, except in the case of grain refinement, is normally gained at the expense of impact toughness. Nevertheless, the good impact behaviour of this steel, when no precipitation hardeners are present, should allow an increase in strength by precipitation hardening even without any grain refinement and still meet the toughness requirements of some of the control rolled steels at the lower end of the control rolled strength spectrum. Increasing the yield strength to 400MPa from the presently achieved strength level of 300MPa and 27J ITT of -95$^{o}$C solely by precipitation hardening would result from previous work, in the ITT increasing to ~-50$^{o}$C (an increase in precipitation hardening by 10MPa corresponds to a ~4-5$^{o}$C rise in the ITT [8]). This is still a workable impact value to replace some of the control rolled steels.

To achieve the required strength in the current work, a Nb addition was chosen, Nb being both able to provide this precipitation hardener by forming Nb(CN) and also to refine grain. For hot rolled steels, the start rolling temperature is always sufficiently high to ensure all the Nb is in solution and will be able to precipitate out on cooling during or after rolling. The contribution to strength from precipitation hardening will then depend on the fineness of the precipitate and the volume fraction that is present.

**Experimental**

The steels were cast as 50kg vacuum melts. The dimensions of the ingot prior to rolling were 381x172 x108mm. Separate casts were made for the hot and control steels. After casting, the ingots were soaked at 1200oC and rolled to give 15mm thick air-cooled plate; finish rolling at 950$^{o}$C and 900$^{o}$C, for the hot and control rolled plates, respectively. For the control rolling process, after rolling the steel to 45mm thick plate, it was held to cool to 950$^{\circ}$C followed by further rolling to 15mm, the finish rolling temperature (FRT) being ~900$^{o}$C. The steel plates were then left to air-cool. The control rolled plates were examined so that the properties obtained could be compared to those obtained when hot rolled. The composition of the steels (weight percent) examined are given in Table 1.

An Al addition of 0.3 wt.%, higher than that used previously of ~0.2 wt.%Al was chosen [2,3,4]. Steels 2, 4, 5, 6, 8, 10, 11 and 12, Table 1, containing 0.3 wt.%Al were compared with similar steels having a normal Al addition of ~0.02 wt.%, steels 1, 3, 7 and 9. The Nb level was taken as a commonly used level for control rolled steels, 0.03 wt.%. The steels were chosen so that comparisons could be made between the influence of both Al and Nb on strength and impact behaviour of hot and control rolled steels and therefore for comparison purposes, steels without Nb were also examined, steels 1, 2, 7 and 8, Table 1.

Two of the 0.3 wt.%Al, Nb containing steels (steels 6 and 12) were produced with a relatively high S level (0.2 wt.% compared to ~0.002 wt.% for the other steels) and a relatively high Si level, (0.6 wt.% compared to 0.3 wt.%).

Table 1 Composition of steels (wt.%).

| Steel | Condition | C | Si | Mn | P | S | **Al** | **Nb** | N | $N_f$ |
|---|---|---|---|---|---|---|---|---|---|---|
| 1 | HR* | 0.09 | 0.26 | 1.37 | 0.013 | 0.0019 | **0.02** | **0.00** | 0.005 | 0.005 |
| 2 | HR | 0.11 | 0.28 | 1.39 | 0.016 | 0.001 | **0.29** | **0.00** | 0.007 | 0.001 |

| | | | | | | | | | | |
|---|---|---|---|---|---|---|---|---|---|---|
| 3 | HR | 0.10 | 0.27 | 1.37 | 0.016 | 0.0015 | **0.02** | **0.03** | 0.008 | 0.008 |
| 4 | HR | 0.11 | 0 .29 | 1.40 | 0.016 | 0.0016 | **0.29** | **0.03** | 0.008 | 0.001 |
| 5 | HR | 0.10 | 0.29 | 1.36 | 0.015 | 0.0015 | **0.31** | **0.03** | 0.007 | 0.001 |
| 6 | HR | 0.10 | 0.57 | 1.39 | 0.014 | 0.022 | **0.32** | **0.03** | 0.008 | 0.001 |
| | | | | | | | | | | |
| 7 | CR+ | 0.09 | 0.26 | 1.37 | 0.013 | 0.0019 | **0.02** | **0.00** | 0.005 | 0.005 |
| 8 | CR | 0.11 | 0.28 | 1.39 | 0.016 | 0.001 | **0.29** | **0.00** | 0.007 | 0.001 |
| 9 | CR | 0.10 | 0.27 | 1.37 | 0.016 | 0.0015 | **0.02** | **0.03** | 0.008 | 0.008 |
| 10 | CR | 0.11 | 0.29 | 1.40 | 0.016 | 0.0016 | **0.29** | **0.03** | 0.008 | 0.001 |
| 11 | CR | 0.10 | 0.29 | 1.36 | 0.015 | 0.0015 | **0.31** | **0.03** | 0.007 | 0.001 |
| 12 | CR | 0.10 | 0.57 | 1.39 | 0.014 | 0.022 | **0.32** | **0.03** | 0.008 | 0.001 |

HR* hot rolled, CR+ control rolled

The nitrogen in solution, $N_f$, in the steels prior to rolling is given in Table 1 and was obtained using Leslie's solubility equation [14]. For the 0.3 wt.%Al level, when ~0.008 wt.% total N is present in the steel, the amount of N in solution at $1200^{o}C$ is low and can be calculated from Leslie's solubility equation to be ~0.001 wt.% [14]. However, the driving force for AlN precipitation on cooling after rolling will increase with Al level, and with this high Al addition may remove the small amount of N remaining in solution. Nb also forms a carbo-nitride so will take some N out of solution even though it will be very small. The amount of free N in steels containing 0.3 wt.%Al has therefore been taken as zero or when the normal level of Al, ~0.02 wt.% Al is present; the total nitrogen level.

Standard, 10mm cross-section, Charpy V notch samples were machined parallel to the rolling direction of the plate while duplicate tensile specimens were machined transversely. The SEM (Scanning Electron Microscope) and optical microscope (OM) were used to examine the microstructural features associated with Widmanstätten ferrite and the LTPs in these steels. The etchant used was 2%Nital.

**Results**

A summary of all the microstructural measurements, tensile properties and impact behaviour is given in Table 2.

Table 2 Microstructural measurements, tensile and impact properties

| **Steel** | **Cond-ition** | **Al wt.%** | **Nb wt.%** | **Si wt.%** | **Pear-lite %** | **LTP** | **Grain Size $mm^{-1/2}$** | **Grain Size µm** | **LYS MPa** | **27J, ITT $^{o}C$** |
|---|---|---|---|---|---|---|---|---|---|---|
| 1 | HR* | 0.02 | - | 0.3 | 14.5 | x | 7.6 | 17.4 | 283 | **-65** |
| 2 | HR | 0.3 | - | 0.3 | 13.5 | - | 7.4 | 18.3 | 311 | **-50** |
| 3 | HR | 0.02 | 0.03 | 0.3 | 14 | √ | 8.7 | 13.3 | 381 | **-40** |
| 4 | HR | 0.3 | 0.03 | 0.3 | 16 | √ | 9.7 | 10.6 | 423 | **-10** |
| 5 | HR | 0.3 | 0.03 | 0.3 | 11 | √ | 8.6 | 13.5 | 407 | **-20** |
| 6 | HR | 0.3 | 0.03 | 0.6 | 12 | √ | 7.2 | 19.3 | 394 | **0** |
| | | | | | | | | | | |
| 7 | CR+ | 0.02 | - | 0.3 | 11 | x | 7.5 | 17.7 | 291 | **-65** |
| 8 | CR | 0.3 | - | 0.3 | 15 | x | 11.35 | 7.8 | 362 | **-85** |
| 9 | CR | 0.02 | 0.03 | 0.3 | 10 | x | 12.0 | 7.0 | 421 | **-100** |

| | | | | | | | | | | |
|---|---|---|---|---|---|---|---|---|---|---|
| 10 | CR | 0.3 | 0.03 | 0.3 | 15 | x | 12.3 | 6.6 | 473 | **-100** |
| 11 | CR | 0.3 | 0.03 | 0.3 | 15 | x | 10.9 | 8.4 | 430 | **-80** |
| 12 | CR | 0.3 | 0.03 | 0.6 | 12 | x | 11.5 | 7.6 | 443 | **-80** |

HR* hot rolled, CR+ control rolled. LTP- lower transformation products, x indicates Widmanstätten ferrite was not observed whereas √ indicates its presence. The pearlite % also includes WF.

The impact transition curves for the hot rolled plates are shown in Fig.2 and for the control rolled plates in Figs. 3a and b. In the case of the hot rolled steels, Fig.2, adding 0.3 wt.%Al (cf. curves 1 and 2) or 0.03 wt.%Nb (cf. curves 1 and 3) always made the impact behaviour worse and the deterioration is made far worse when the Al and Nb are present together (cf. curve 1 and curves 4,5,6).

The strength increases with these additions from ~280 to 400MPa but by then the 27J, ITT is as high as -10°C.

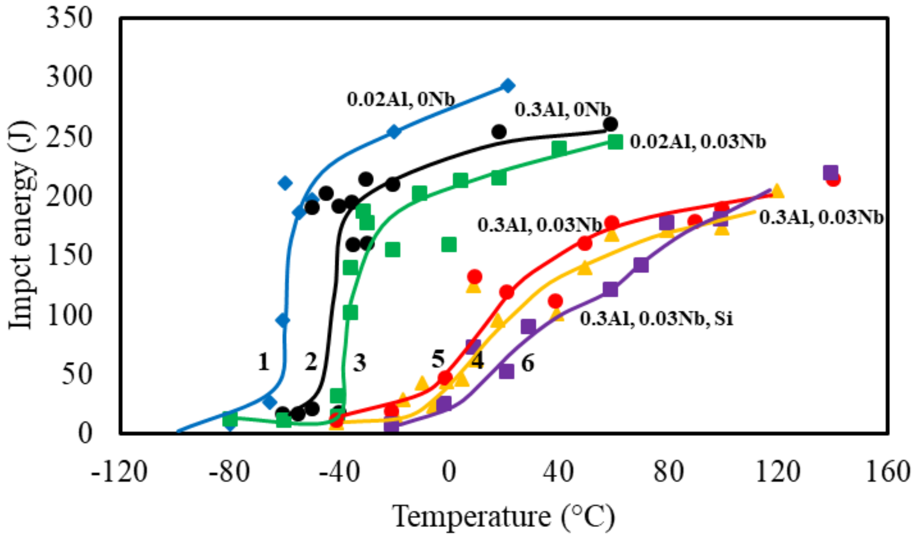
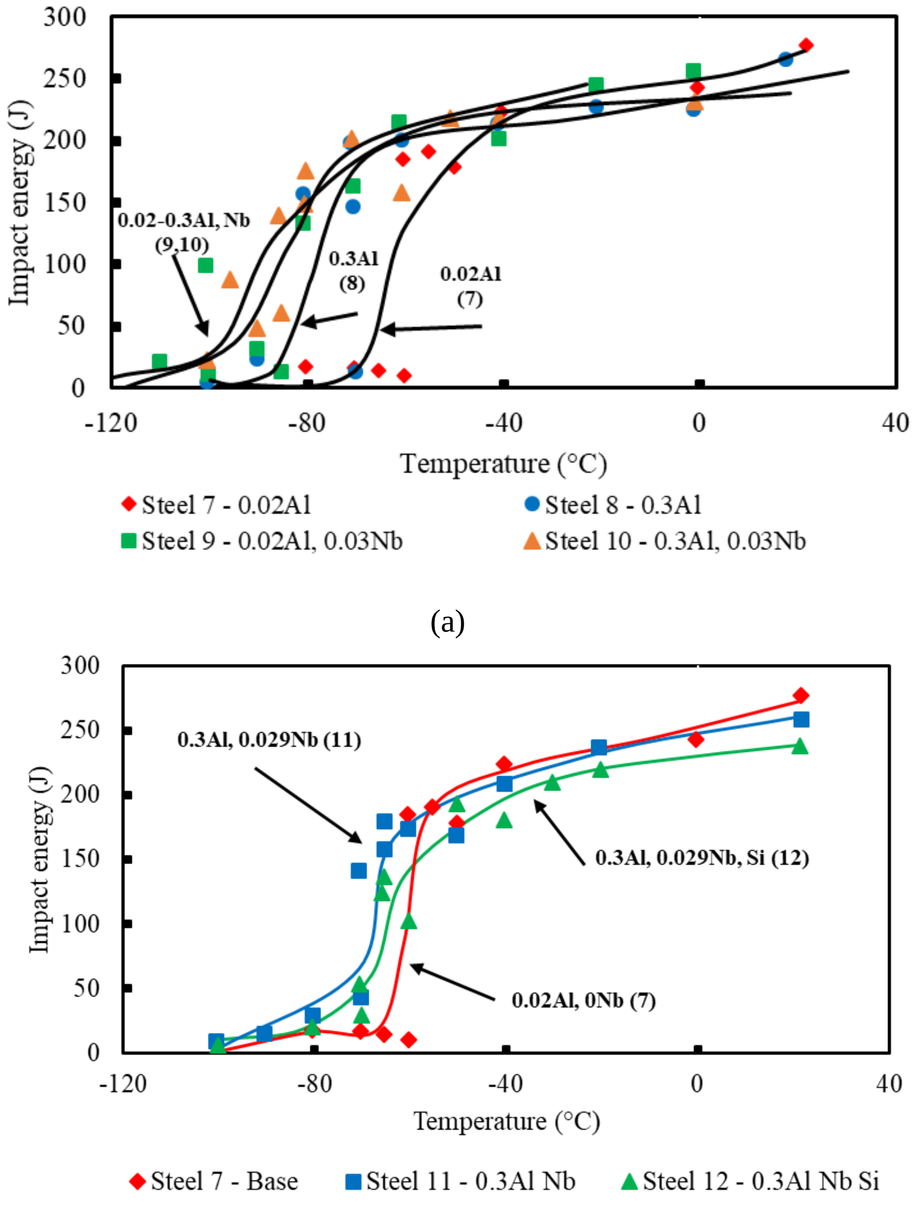



Fig.2 Impact transition curves of hot rolled steels, Table 1. Note the very poor impact behaviour when both Nb and the high Al addition (0.3 wt.%) are present together, steels 4, 5 and 6.

In contrast for the control rolled steels, because Nb has such a strong, grain refining ability, its addition was found to be beneficial, both for the impact behavior (cf. curves 7 and 8 with 9 and 10 in Fig.3a) and (cf. curve 7 with 11 and 12 in Fig.3b), but also for the strength, both improving considerably, Table 2.



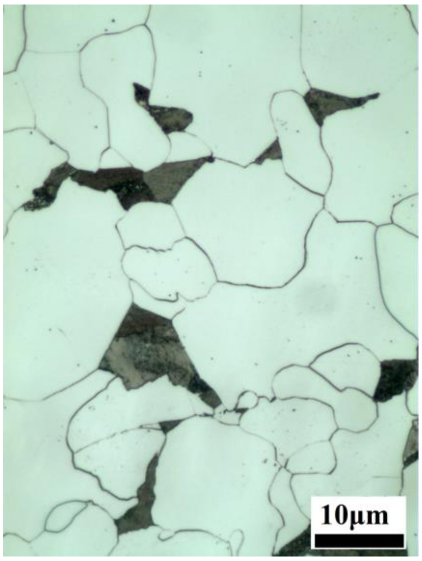

(a)

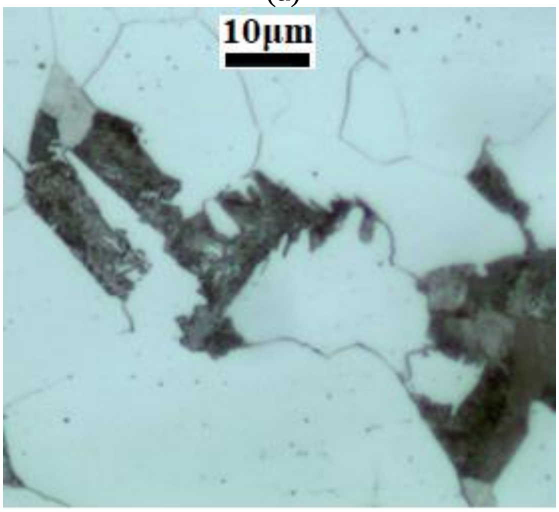


(b)

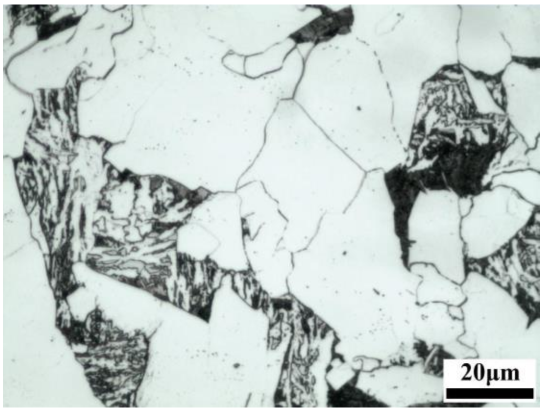
Fig.3 (a) Impact transition curves of control rolled steels, steels 7, 8, 9 and 10, (b) Impact transition curves of control rolled steels, steels 7, 11 and 12.

The microstructures for selected hot rolled steels are given in Figs.4a-4c and Fig.5 for the control rolled steels. For the hot rolled 0.02 wt.%Al, Nb free base steel, steel 1, a ferrite/pearlite structure was present, Fig.4a. Adding 0.3 wt.%Al introduced some Widmanstätten type structures, steel 2, Fig.4b. When the higher Al and Nb additions were present together, both higher volume fractions of Widmanstätten ferrite and lower transformation products were present, Fig.4c.



(a)


(b)



(c)

Fig.4 (a) Hot rolled 0.02 wt.%Al steel, steel 1, with normal ferrite/pearlite microstructure, (b) hot rolled, 0.3 wt.%Al steel, steel 2, in which small amounts of  Widmanstatten were found, (c) hot rolled steel with 0.3 wt.%Al and 0.03 wt.%Nb, steel 5 in which higher amounts of Widmanstätten ferrite were found. (OM)

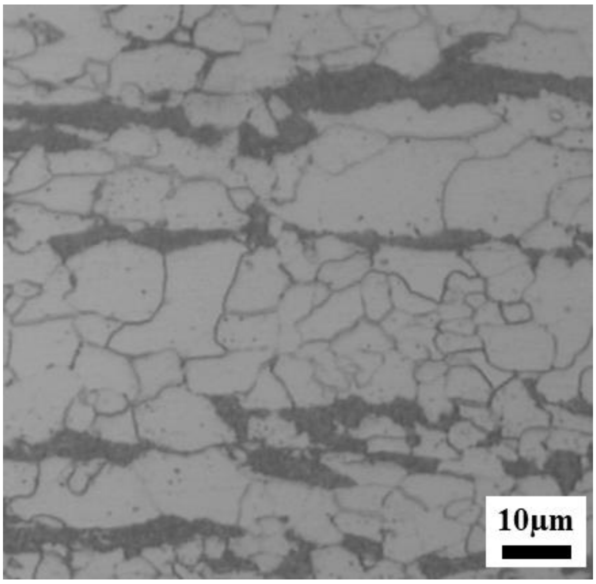



Fig.5 Control rolled steel with 0.3 wt.%Al and 0.03 wt.%Nb, steel 10. Essentially ferrite/pearlite micro-structure (OM)

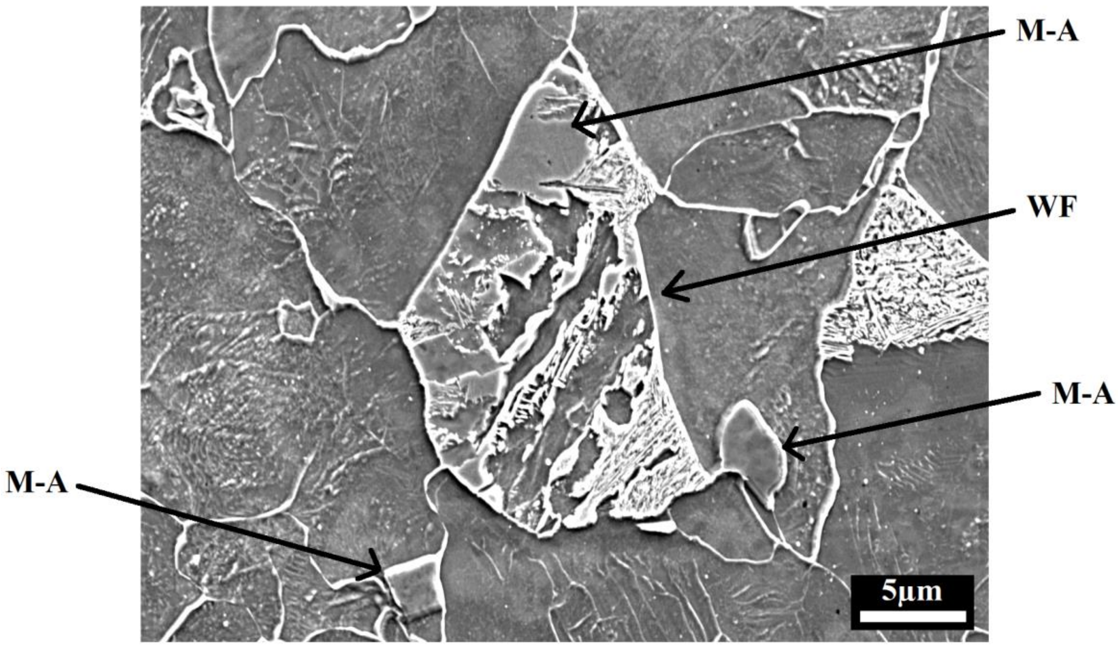



Fig.6 Widmanstätten ferrite (WF), ferrite, pearlite and martensite/retained austenite (M-A) in the hot rolled steel containing 0.03 wt.%Nb and 0.3 wt.%Al, steel 4. (SEM)

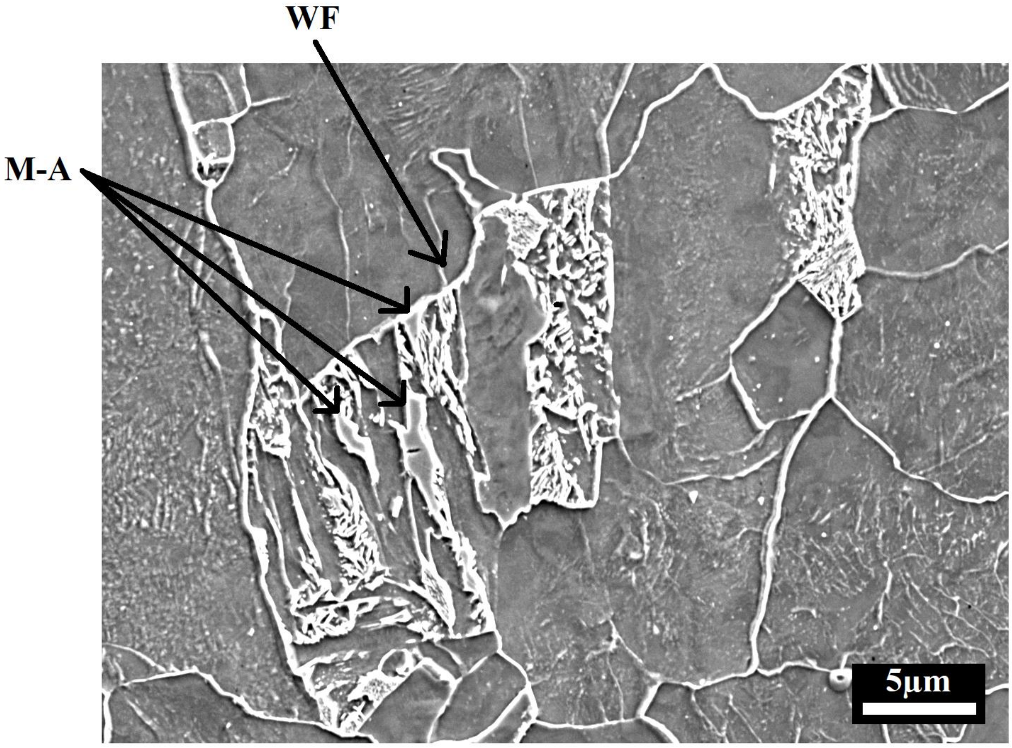



Fig.7 Widmanstatten Ferrite and MA constituents in the hot rolled 0.3 wt.%Al and 0.03 wt.%Nb steel, steel 5. (SEM)

For the control rolled steels, the finer grain size ensured essentially normal ferrite/pearlite structures with banding being evident and some elongation of the grains in the rolling direction as well as a mixed grain size structure, Fig.5.

SEM examination confirmed that M-A constituent or lower transformation products, bainite and Widmanstätten ferrite were present in all the hot rolled steels having 0.3 wt.%Al and Nb, Fig.6 and Fig.7 and this accounts for their poor impact behaviour, as shown in Fig.2.

Except for the low Al, Nb free steel, the coarse grained hot rolled steels were found to have mixed structures consisting of ferrite grains and colonies with varying partial transformation products, bainite and M-A as well as Widmanstätten ferrite. The presence of saw cut like structures was also indicative of the early stages of Widmanstätten ferrite formation, Fig.8.

In some cases partial transformation to pearlite had taken place at the periphery of these colonies leading to the formation of comb like teeth protruding into a Widmanstätten ferrite grain, Fig.9. It would be expected that these “brittle” teeth would easily crack and favour brittle behaviour.

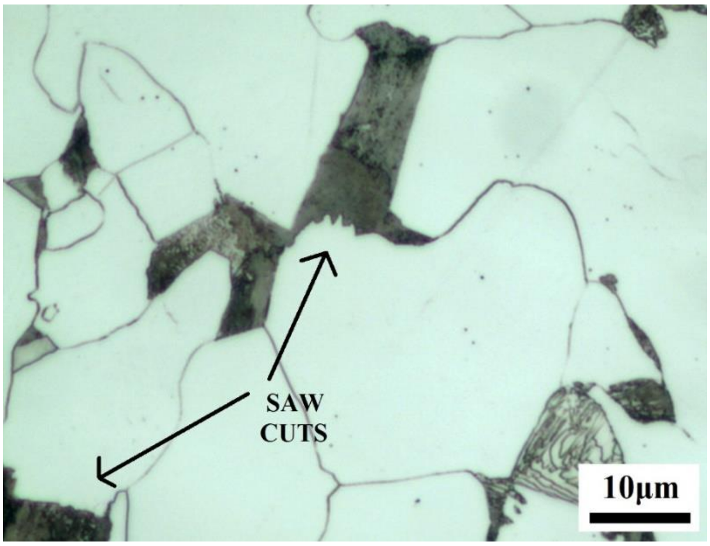



Fig.8 Saw-cut like structures which seem to be associated with the formation of Widmanstätten ferrite and poor impact properties, steel 4. (OM)

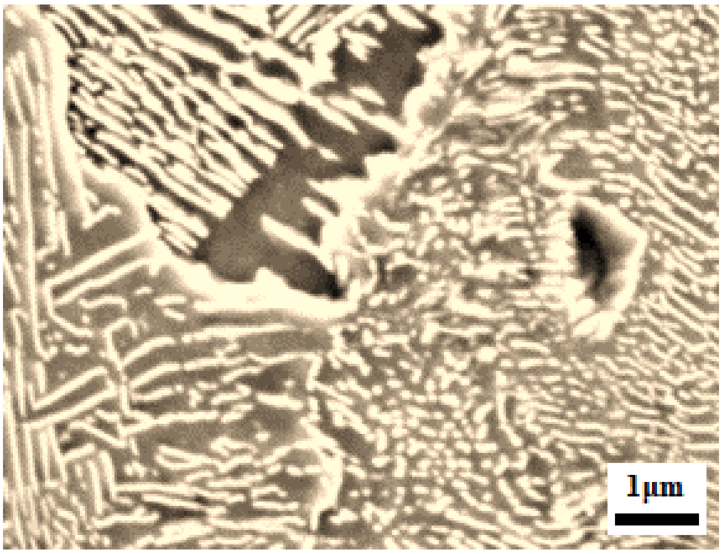



Fig.9 SEM micrograph of a Widmanstätten ferrite colony. The carbides are in the form of "teeth" protruding out into the ferrite grain, steel 4 (SEM)

Typical grain boundary carbides and pearlite colonies in a fully ferrite/pearlite microstructure are shown in Fig.10 for steel 1. Grain boundary carbides are not isolated but are normally attached to the pearlite colonies unless the carbon level is very low.

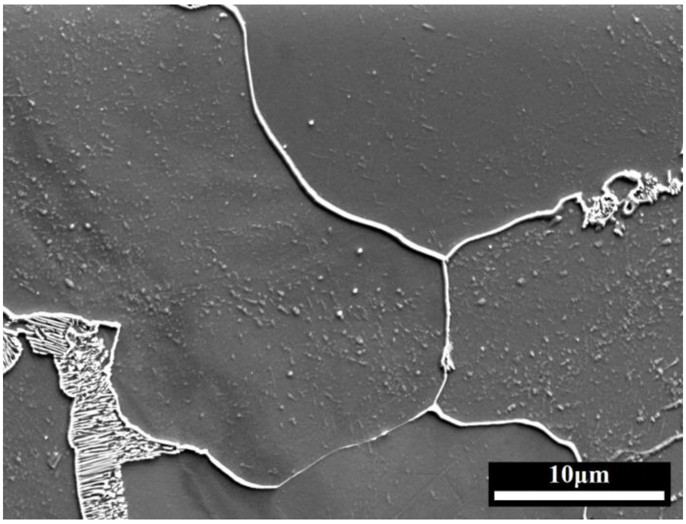



Fig.10 Steel 1 (0.02 wt.%Al) showing typical carbides (coloured white) at boundaries in a nominally ferrite/pearlite steel. Most of the boundaries are decorated with carbides. (SEM)

## Discussion

### Relationship between impact behaviour and lower transformation products including Widmanstätten ferrite

The microstructural examination has shown that in the hot rolled steels the presence of lower transformation products and Widmanstätten ferrite gives rise to poor impact behaviour, Fig.2 and Figs. 4b, 4c and 6. The presence of MA, Fig.6 and 7 is particularly deleterious to the impact behaviour.

### Hot rolled steels

In contrast to the previous work in which the addition of 0.2 wt.%Al improved the properties of plain C-Mn steels, increasing the Al to 0.3 wt.%Al caused the impact properties to deteriorate. For the hot rolled condition, increasing the Al level from 0.02 to 0.3 wt.% in the Nb free steels, 1 and 2, respectively, leads to the 27J, ITT increasing by 15$^{o}$C and the strength increasing by 28MPa, Table2. The microstructure for the 0.02 wt.%Al steel, steel 1 was essentially ferrite/pearlite, Fig.4a but a small amount of Widmanstätten ferrite was present in the high Al containing steel, steel 2 Fig.4b. When both Al (0.3 wt.%) and Nb (0.03 wt.%) are present together, steel 5, lower transformation products (LTPs) combined with Widmanstätten ferrite were there in larger amounts, Fig.4c and the ITT increased by ~45$^{o}$C over that for the base plain C-Mn steel, steel 1.

Because of the presence of LTPs and Widmanstätten ferrite influencing properties it was not possible to use normal Structure/Property relationships developed for ferrite/pearlite steels to interpret the influence of the microstructural data, the higher S level and Si content on the strength and impact behaviour.

### Control rolled steels

For the control rolled steels, the finer austenite grain size prevented MA forming on transformation and there was very little Widmanstätten ferrite resulting in the impact behaviour improving substantially. Wholly, ferrite/pearlite microstructures were present in the 0.02 wt.%Al containing steel, steel 7, the Nb free steel, steel 8 and the Nb containing, 0.02 wt.%Al steel, steel 9 and only occasional Widmanstätten ferrite colonies were observed in the high, 0.3 wt.%,Al and 0.03 wt.%Nb containing steels, 10, 11 and 12.

For Nb free control rolled steels, steels 7 and 8, respectively, Table 2 the addition of Al decreases the ITT by 20ºC and increases strength by 71MPa. However, the wide difference in grain size in these two steels, Table 2 makes it difficult to assess these results.

Again the benefit of adding 0.3 wt.%Al to a Nb containing steel is not too clear, steels 9 and 10, as the impact behaviour did not change even though these steels have similar grain sizes. Nevertheless, increasing the Al addition from 0.02 to 0.3 wt.%, (steels 9 and 10, respectively) increased strength by 52MPa, Table 2, giving an excellent strength/impact combination of 473MPa and a 27J, ITT of -100ºC. This increase in strength can be regarded as the equivalent to reducing the ITT by ~ 25ºC [8].

Steels 11 and 12 with similar fine grain sizes showed that increasing the Si from 0.3 to 0.6 wt.% increased the strength by 13MPa, also with no influence on the ITT, Fig.3b and Table 2.

Although ferrite/pearlite steels are the main requirement for achieving the benefits of a high Al addition, there are also other important variables that have to be met to satisfy achieving good properties and these will be discussed in turn.

**Grain boundary carbide thickness**

The difficulty in differentiating between carbides and M-A constituents at the grain boundaries made it impossible to do reliable grain boundary thickness measurements on the hot rolled steels in the present exercise.

However, previous work [11] has shown, Fig.11, that Al additions are able to refine the grain boundary carbides and so enhance impact performance.

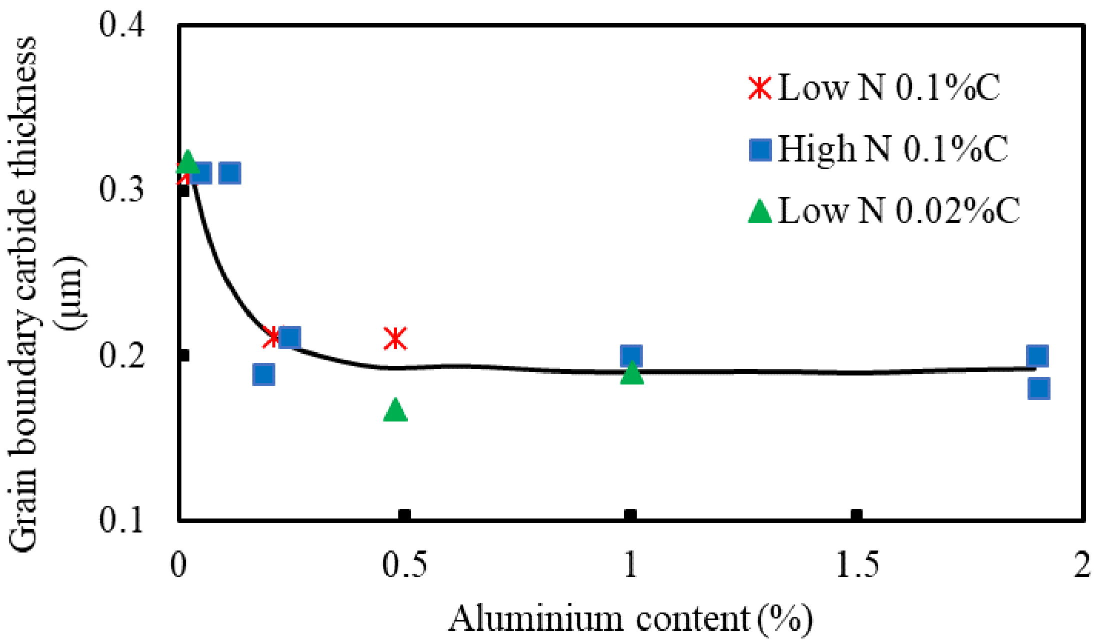



Fig.11 Influence of Al on the grain boundary carbide thickness [11].

Examination of Fig.11 shows there is little further refinement in the thickness of the grain boundary carbides once the Al level exceeds ~0.25 wt.%. The change in Al level from 0.02 wt.% to 0.3 wt.% as in the present work would lead to a change in carbide thickness from 0.3

to 0.2μm, Fig.11 and this would decrease the 27J, ITT by 18$^{o}$C according to the following multiple regression equation developed for hot rolled steels [8].

$$27\text{ J, ITT }^{o}\text{C} = 173t^{1/2} - 8.3d^{-1/2} + 0.37\Delta p - \text{Constant}$$

where t is the grain boundary carbide thickness in μm, d is the grain size in mm. and Δp is the precipitation-hardening component in MPa.

**Influence of S on the impact transition temperature**

Sulphur is generally important because it controls the volume fraction of MnS inclusions and so the ductile shelf energy, the lower the S content, the higher is the ductile shelf energy. The S level was not included in the original regression equations for relating the impact transition temperatures to composition and microstructure (1979) [8] because it was not believed then to have an influence on brittle fracture. Typical S levels were then 0.02 wt.%.

When S was included in a similar multiple linear regression exercise on normalised fine grained HSLA and plain C-Mn steels (1994), a multiplying factor of +1224 was found for sulphur for the 27J, ITT in $^{o}$C with a high statistical t ratio; i.e. a decrease in the S level by 0.02 wt.% results in a decrease in the ITT of 25$^{o}$C [15]. This showed that S as well as influencing the ductile shelf energy via combining with Mn to form MnS inclusions is also important in influencing the impact transition temperatures.

The effect of sulphur on ITT of these hot rolled steels can also be seen from some previous data on hot rolled steels in which processing conditions and the plate thickness were similar to those used in the present examination [3].

In this exercise, four steels, A-D, whose full compositions are given in Table 3 were examined after hot rolling [3]. Here, the grain size variations were small making it easier to compare the properties.

Table 3 Composition (wt.%) of steels with high and low S levels including grain size and 27J, ITT [3].

| Steel | C | Si | Mn | **S** | P | **Al** | N | 27J, ITT $^{o}$C | Grain Size $mm^{-1/2}$ | Grain Size μm |
|---|---|---|---|---|---|---|---|---|---|---|
| A | 0.08 | 0.26 | 1.41 | **0.022** | 0.008 | **0.01** | 0.0043 | -10 | 6.8 | 21.5 |
| B | 0.098 | 0.30 | 1.42 | **0.002** | 0.010 | **0.021** | 0.0030 | -62 | 6.5 | 23.8 |
| C | 0.08 | 0.28 | 1.46 | **0.021** | 0.007 | **0.09** | 0.0039 | -58 | 7.4 | 18.1 |
| D | 0.095 | 0.29 | 1.40 | **0.002** | 0.006 | **0.19** | 0.0040 | -105 | 6.6 | 22.7 |

Two of the steels (A and C) had a high sulphur level of ~0.02 wt.% and two a low S level of 0.002 wt.% (B and D). Each of the steels were examined at two Al levels, low Al (0.01-0.02 wt.%Al) and high Al (0.1and 0.2 wt.%Al). For both the low Al and high Al steel (A and C) the 27J, ITT decreased ~50$^{o}$C when the S level was reduced from 0.022 to 0.002 wt.%, Table 3 [3]. The effect of increasing the Al level from the low levels 0.01-0.02 wt.%Al to 0.1-0.2 wt.%Al can also be seen to be similar (~45$^{o}$C decrease in ITT).

Generally this effect of S on the impact energy ITT's can be explained by the impact energy criterion for ITT having both a ductile component to the fracture energy as well as a brittle a brittle component. The higher ductile shelf energy on decreasing the S leads to higher impact values at the lower temperatures. Decreasing the S content steepens the transition curve and thus moves the transition to lower temperatures. It is also possible that atomic S segregating to the grain boundaries, weakening them, might also be responsible [16]. Remarkably no influence of S on the ITT was obtained in the present examination but this may be because the beneficial influence of the higher Si level was obscuring the detrimental influence of a higher S level.

The reason Mn is added is more to prevent the formation of FeS which is a low melting point phase. When one is talking about S segregation to the boundaries this is possible even when there is adequate Mn present to make sure FeS does not form, which may be related to atomic sulphur segregating to the grain boundary and weakening them [16]. Combined effect of AIN and sulphur on hot ductility of high purity iron-base alloys. MnS inclusions will go partly back into solution on soaking at 1200-1250°C, granted very little but when it comes to segregation to the boundaries often very little is required to cause problems.

**Influence of Si on strength and impact behaviour**

Si is one of the few solid solution hardeners that can increase strength and lower the ITT [17,18,19]. There is evidence to suggest that the observed improvement in impact behaviour is due to Si reducing the $k_y$ value in the Hall-Petch relationship ($\sigma_y = \sigma_0 + k_y\, d^{-1/2}$), this being related to the ease of generating dislocations from the grain boundary. Si by segregating to the grain boundaries repels the interstitials there, thus making it easier to generate dislocations into adjacent grains [20]. Work by Mintz et al. [20] has shown that there is a relationship between the $k_y$ value and the amount of interstitials, C and N, at the ferrite grain boundaries; $k_y$ being proportional to the square root of this interstitial content. However, the benefit Si gives to yield strength is very dependent on the grain size [18]. At coarse grain sizes, Si additions give marked strengthening but there is evidence that as the grain size refines this strengthening becomes less and at ultra-fine grain sizes can cause softening [18]. Hence, adding Si to coarse grained hot rolled steels should be very beneficial but less so, in the fine grained control rolled steels.
It should be noted that increasing the Si level to ~0.6 wt.%, in the control rolled steels, (compare steels 11 and 12, respectively), Table 2 gives a modest increase in strength, 13MPa without again influencing impact behaviour. This may, as already be explained, be due to the higher Si steel (0.6 wt.%Si) also having a higher S level, 0.022 wt.%.; the beneficial influence of one on impact behaviour offsetting the detrimental influence of the other. Further work is required to substantiate these findings. Certainly increasing the Si level to 0.6 wt.% is beneficial.

**Influence of Al and N on the yield strength and impact behaviour**

In the case of Al, the $k_y$ value does not change with increase in Al [21] so this cannot explain its better impact behaviour but there is evidence, Fig.11 that Al additions refine the grain boundary carbides and in addition there is removal of N from solution as AlN [11]. The removal of N by Al as AlN would be expected to improve the impact behaviour but reduce the yield strength due to lack of interstitial hardening by N.
Mintz and Turner [17] have shown that an increase of 0.001 wt.%N in solution increases the ITT by 2.75°C so that an increase in N by 0.005 wt.% would increase the ITT by 14°C. A detailed examination of the literature by Morrison et al. [22] has concluded that 0.001 wt.%N increases the yield strength by ~5MPa. Using these multiplying factors, complete removal of the 0.005 wt.%N in the base steel 7, Table 2 should decrease the yield strength by 25MPa but improve the impact behaviour, by lowering the ITT by ~14°C. However, as noted previous work [21] has shown that Al is a powerful solid solution hardener similar to Si; an when all the data was analysed [21] it was concluded that a 1 wt.% increase in the Al level leads to a 70MPa increase in strength. A 0.3 wt.%Al addition would therefore cause a 21MPa increase in strength. If it assumed that 0.3 wt.%Al removes all the N from solution, then adding this amount will lead to a reduction in strength of only 4MPa and a decrease in ITT of ~14°C from the removal

of N. The 30-40$^o$C decrease in ITT which has been found in previous work on adding Al is likely therefore to be due to both refinement of the grain boundary carbides, a 14$^o$C reduction in the ITT, (18$^o$C according to Fig. 11) and removal of N from solution giving a similar 14$^o$C reduction. Unfortunately, it is difficult in the present work to isolate any benefit from adding the Al even in the control rolled steels because of the coarse grain size present in steel 7 compared to the other control rolled steels.
For hot rolling any benefit to impact behaviour on adding Al is lost when the Al level is increased from 0.2 to 0.3 wt.%Al as lower transformation products form, with consequent deterioration in impact behaviour, Fig.1 [2].

When considering the commercial relevance of the work it is important to realise that it is the combined effect of strength and impact behaviour that is important. Thus, a plot of ITT against LYS gives the most useful information and this is shown in Fig.12 for the 15mm thick plate. The results can be split into two curves, one for the control rolled plates and the other for hot rolled plates.

The top curve is the curve where LTP products dominate the properties and the lower curve is when the structure is essentially ferrite/pearlite.

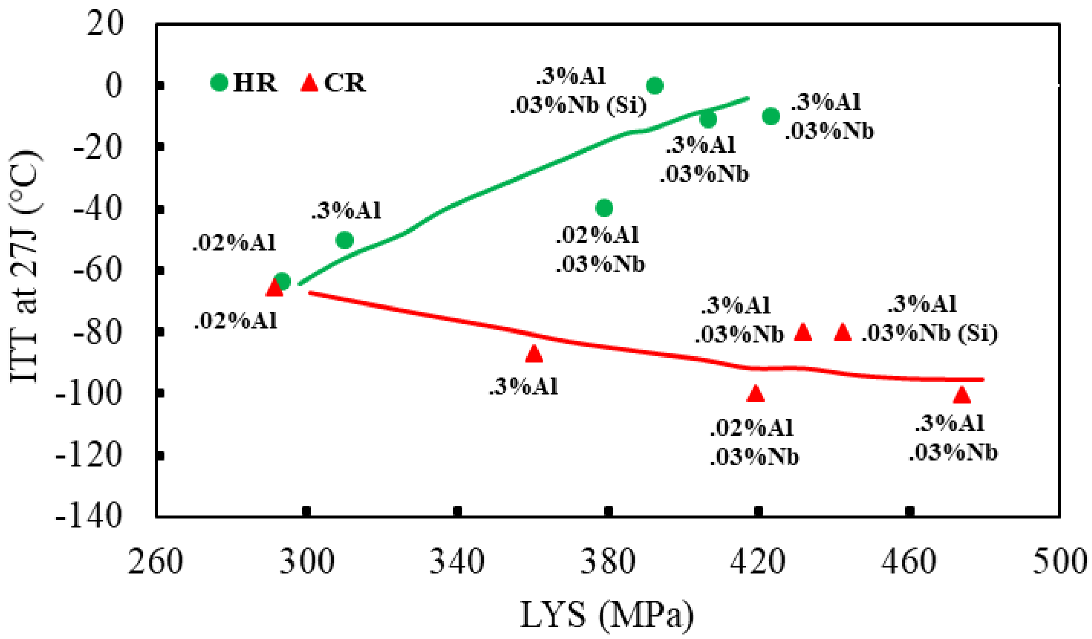



Fig.12 The combined influence of impact performance and yield strength on adding Al and Nb to hot rolled and control rolled steels for a plate thickness of 15mm.

For the control rolled steels, the impact behaviour improves when Al and Nb are added as the lower transformation products are not present. For the hot rolled steels the opposite occurs.
Following this work on 15mm thick plate a further four casts were made in which the final thickness was 30mm in the hope that the slower cooling rate would improve the hot rolled properties. The Al and Nb levels were also reduced from 0.3 to 0.25 wt.% for the Al and from 0.03 to 0.025 wt.% for the Nb.

Although properties were better, M-A and Widmänstatten ferrite islands were still present in the hot rolled plates, Fig.13 causing the impact properties to not reach their true potential.

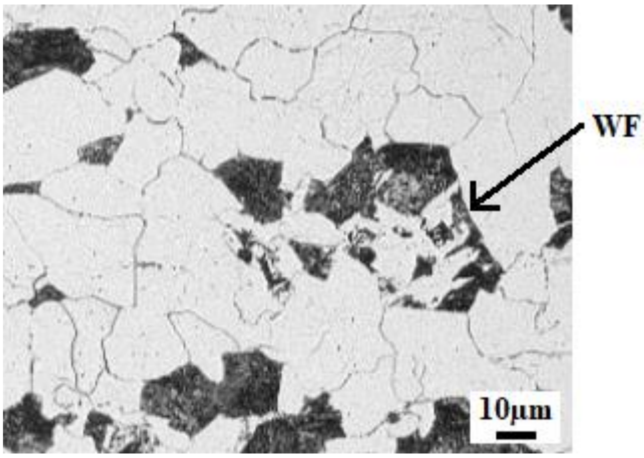
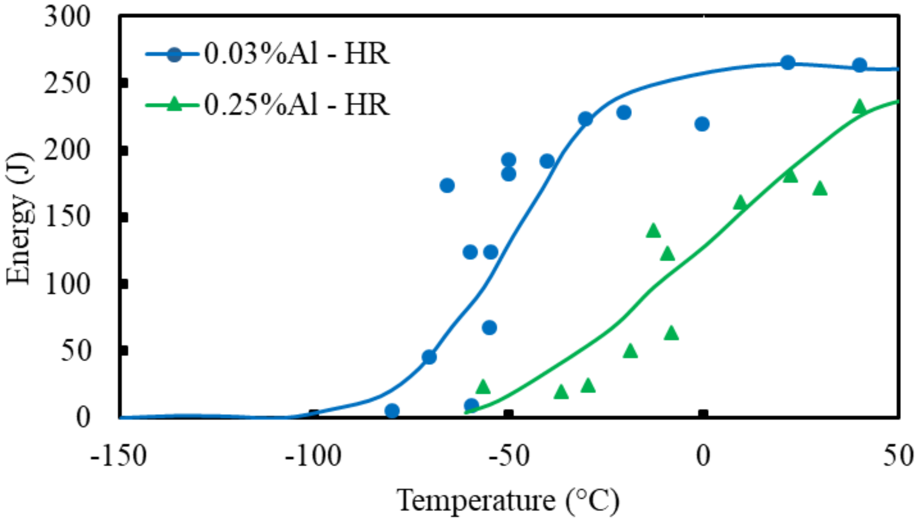
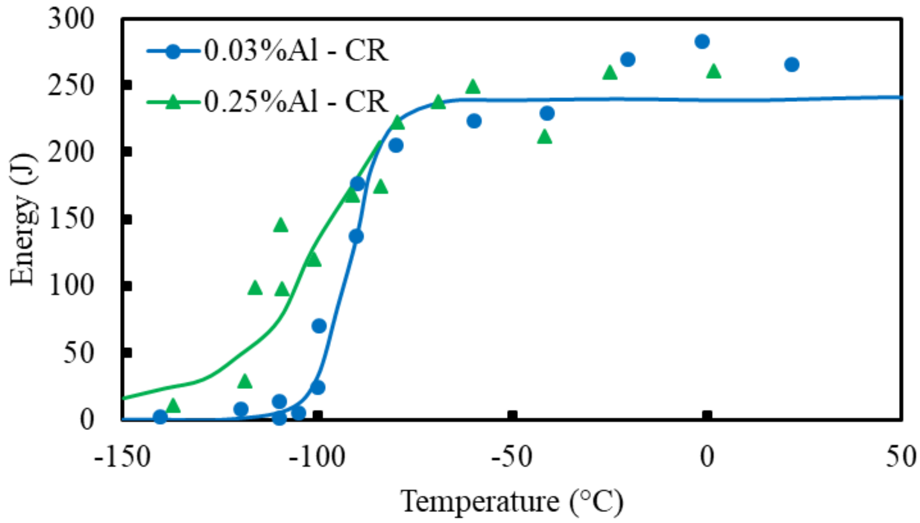



Fig.13 Widmanstätten ferrite in hot rolled steel containing 0.25 wt.%Al and 0.025 wt.%Nb hot rolled to 30mm thick plate.

The compositions of these steels are given in Table 4 with the mechanical properties in Table 5 and the ITT curves in Fig.14. The behaviour was similar to that shown by the thinner plate. Adding 0.25 wt.%Al to the Nb containing steels caused the ITT to rise 32°C on hot rolling,(steels 13 and 14) Fig.14a and decrease ~30°C when control rolled, Fig.14b (steels 15 and 16). However, because of the finer grain size, ~ 12µm in these hot rolled Nb containing steels, Table 5, steels 13 and 14, the ITT and yield strength were better than found in the 15mm thick plate, a yield strengths of 360MPa and ITT of -80°C being achieved in the steel with only 0.03 wt.%Al.




(a)




(b)

Fig.14 (a) ITT curves for hot rolled steels having 0.03 and 0.25 wt.%Al, (b) ITT curves for control rolled steels having 0.03 and 0.25 wt.%Al.

In contrast, the finer grain size of the control rolled steels prevented these deleterious phases from forming and showed clearly that adding Al was beneficial, decreasing the ITT by 20-30ºC, Fig.14b as well as giving a small increase in strength, 11MPa, Table 5, (cf. steels 15 and 16). The grain size for the two steels was similar making direct comparisons to be made more easily. Crowther has also shown for 30mm thick air cooled plates that adding ~0.2 wt.%Al to Nb containing control rolled steels will decrease the 27J ITT by 20ºC and increase the strength by 20MPa [23]. Hence it suggests that if the lower transformation products could be removed from the hot rolled plates the improvement that has been found in plain C-Mn steels on hot rolling could be achieved in coarse grained C-Mn-Nb steels.

Table 4 Composition of steels (wt.%) rolled to 30mm plate thickness.

| Steel | Condition | C | Si | Mn | P | S | **Al** | **Nb** | N | $N_f$ |
|---|---|---|---|---|---|---|---|---|---|---|
| 13 | HR* | 0.12 | 0.37 | 1.49 | 0.01 | 0.002 | **0.03** | **0.025** | 0.006 | 0.006 |
| 14 | HR | 0.11 | 0.39 | 1.49 | 0.01 | 0.002 | **0.25** | **0.025** | 0.005 | 0.001 |
| 15 | CR+ | 0.11 | 0.37 | 1.49 | 0.01 | 0.002 | **0.03** | **0.024** | 0.005 | 0.005 |
| 16 | CR | 0.11 | 0.38 | 1.49 | 0.01 | 0.002 | **0.25** | **0.024** | 0.006 | 0.001 |

HR* hot rolled, CR+ control rolled

Table 5 Microstructural measurements, LYS and 27J, ITT for the four steels rolled to 30mm thickness and air cooled.

| **Steel** | **Condition** | **Al** wt.% | **Nb** wt.% | **Pearlite wt.%** | **LTP*** | **Grain Size mm$^{-1/2}$** | **Grain Size µm** | **LYS MPa** | **27J, ITT ºC** |
|---|---|---|---|---|---|---|---|---|---|
| 13 | HR | 0.03 | 0.025 | 15 | 0 | 9.7 | 10.6 | 360 | -79 |
| 14 | HR | 0.25 | 0.025 | 13 | 5 | 9.0 | 12.4 | 372 | -47 |
| 15 | CR | 0.03 | 0.025 | 15 | 0 | 11.5 | 7.6 | 385 | -102 |
| 16 | CR | 0.25 | 0.025 | 15 | 0 | 11.9 | 7.1 | 396 | -130 |

LTP* lower transformation products including Widmanstätten ferrite

Al appears to benefit properties in a similar way to Mn but by different methods. In contrast to Mn it raises the transformation temperatures and is a ferrite former. Nevertheless, it is able to refine carbides probably by delaying the precipitation of iron carbide [13]. This delay although refining the carbides also leads to a build-up of carbon in the remaining austenite before transformation takes place, making martensite and retained austenite more likely to form. Al is also a solid solution hardener like Mn but its effect on strength is often obscured by the concomitant removal of nitrogen when Al combines with N to form AlN [21].

It is this removal of nitrogen by Al which probably also accounts partly for its influence in improving impact performance. There is also the possibility of some grain refinement taking place due to the precipitation of AlN pinning the grain boundaries but this has not been found in the present work.

It should also be noted that Mn tends to segregate during continuous casting to the centre line encouraging martensite formation on transformation and it is this centre line segregation, (which can lead to a 100% increase in the Mn level), which often limits the amount of Mn that can be added [24].

**Influence of rolling methods on texture**

The present paper has been aimed at improving the impact behaviour of hot rolled steels and the control rolling has only been included for comparison purposes. For control rolled steels because rolling is carried out at lower temperatures in the two phases $\gamma/\alpha$ range marked textures can develop which increase the strength when tested in transverse to the rolling direction and possibly the impact behaviour although the evidence does not really support this [25,26]. Dislocation hardening can also increase the strength of control rolled steels [6]. However, with hot rolling, the finish rolling temperature is always sufficiently high for recrystallisation to occur during rolling so that there is no dislocation hardening and the textures developed are too weak to influence the mechanical properties.

**Commercial implications and Future Work**

High Al additions are not favoured commercially because $Al_2O_3$ readily forms during casting and this tends to clog up the nozzles in the tundish [26] However, all continuous cast steels are Al killed and are cast without problems for the Al levels up to 0.04 wt.% and TRIP and TWIP steels are being cast successfully with 1-2 wt.%Al [27-29] using special mould fluxes and this level is very much higher than the level of 0.2%Al used in this work. Calcium is often added to change the oxide to a more fluid composition [30].

It is true that the improvement in properties that has been found by adding 0.2 wt.%Al in control rolled steels is relatively small but nevertheless if this could be replicated in hot rolled steels it would enable some of the hot rolled steel to replace the control rolled steels at the lower end of the control rolled spectrum. If this can be achieved then the extra cost of having a higher Al content might be justified.

Hardenability seems to be the major problem in any further development. This 30-40$^{o}$C lowering of the ITT only seems possible if Widmanstatten ferrite and LTPs are completely removed. Although the early work on adding Al to a plain C-Mn steel seemed very promising for a more complex steel, containing Nb, this has proved unsuccessful and even increasing the Al level from 0.2 to 0.3 wt.%, as in the present work, causes the impact behaviour to suffer.

Nb lowers the transformation temperatures and so will make the steel more susceptible to forming lower transformation products but there is also evidence [31] which suggests it raises the pearlite transformation temperature and because it is a carbide former encourages thicker and more numerous carbides. Its big advantage of course is its grain refining ability on rolling. Al in contrast raises the transformation temperatures but does not form a carbide itself and delays cementite precipitation in the bainite encouraging finer carbides, but only providing M-A does not form.

Unfortunately, only relatively small amounts of LTPs or Widmanstätten ferrite have to be present to impair the impact behaviour (>1 wt.% volume fraction, unpublished work by the authors) and when this happens, any improvement that Al additions might have made to the impact behaviour is removed.
Engineering the composition by reducing the Nb and Al levels, decreasing the carbon level and cooling at a slower rate during the austenite transformation may all help in solving this problem. For future work, it is suggested that in Nb containing steels, the Al level be reduced to 0.2 wt.%Al, the carbon level reduced to 0.06 wt.%C and the plate thickness increased to 30mm so that lower transformation products and Widmanstätten ferrite are avoided in the microstructure. For low C steels (up to at least 0.2%C), the yield strength is mainly controlled by the grain size and precipitation hardening. The C in solution is in general very low on air cooling [17] and does not vary much with the carbon level. The total carbon does control the amount of pearlite in the structure and this does have a small influence on the yield strength. An increase in pearlite by 1% pearlite raises the LYS by 1.5MPa [15]. Reducing the carbon from 0.1% to 0.06% will reduce the pearlite content from about 15% to 5% which would reduce the yield strength by 15MPa. Although reducing the carbon level from 0.1 wt.% to 0.06 wt.% will reduce the pearlite volume percentage by about 10 wt.% this will only lead to a small decrease in yield strength ~15MPa [15].

**Summary and Conclusions**

The work to date aimed at producing a hot rolled steel with a yield strength of 400MPa and an ITT of -50 to -60$^{o}$C in a high Al, Nb containing steel has not yet been achieved.
Unfortunately, the combination of a high Al level of 0.25 to 0.3 wt.% and 0.03 wt.%Nb has led to hot rolled structures having microstructures containing Widmanstätten ferrite as well as MA and bainite which have resulted in poorer than expected impact behaviour. Because Nb lowers the transformation temperatures, the Nb containing steels are more susceptible to martensite/bainite forming than the plain C-Mn steels and therefore a lower level of Al and Nb are required to prevent this from happening. Al does however improve the properties of control rolled steels as their finer grain size prevents the formation of lower transformation products.
It is clear from the fine grained control rolled steels that when the microstructure is solely ferrite/pearlite, high Al additions will also be beneficial to the properties even when Nb is present.
Sulphur level has again been shown to be important, as in addition to controlling the inclusion content which controls the ductile shelf energy, it has an equally important influence on brittle failure. Reducing the S level from 0.022 to 0.002 wt.% causes the 27J, ITT to decrease by ~50$^{o}$C.
The benefit that Al gives to a hot-rolled steel is due to a combination of refinement of the grain boundary carbides and to removal of N as long as structures remain ferrite/pearlite.

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