

THE MANUFACTURE, PROPERTIES AND WELDABILITY OF VANADIUM-CONTAINING STEELS

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ABSTRACT

This paper examines a range of aspects related to the production and use of vanadium-containing steels.

During continuous casting vanadium steels can exhibit high levels of hot ductility giving them low susceptibility to transverse cracking, thus offering potential yield improvements. Furthermore, the range of cooling rates obtained during casting can lead to a precipitate particle size and distribution which are almost ideal for the production of Ti-V-N steels.

In the course of rolling, vanadium steels tend to recrystallise over a wide temperature range, leading to lower rolling loads than observed with other micro-alloyed steels, particularly as the temperature is reduced. In addition, the recrystallised austenite grain size is reasonably constant over a wide range of finish rolling temperature, resulting in relative insensitivity of properties and this grain size can be refined by controlled additions of Ti and N. Furthermore, if rolled at low temperature, in the two-phase region, steels containing vanadium alone can develop recovered microstructures containing both a high dislocation density and a precipitate size and distribution which lead to a combination of high strength and good toughness. Thus, by application of appropriate processing procedures including hot rolling, temperature controlled rolling, accelerated cooling and two-phase rolling a family of steels with a wide and controllable range of properties can be obtained. The properties of some of these steels are considered.

Vanadium steels are weldable over a range of heat inputs and exhibit the additional benefit that the incorporation of vanadium can give rise to advantageous heat affected zone microstructures, particularly at slower HAZ cooling rates, and this may lead to increasing use of such steels as heat inputs are increased. Results which demonstrate the effect of vanadium on HAZ toughness are presented.

INTRODUCTION

Since vanadium was first knowingly used by Henry Ford, in the USA, to obtain superior properties in axle, crankshaft, gear and other steels for his early automobiles, its use has grown widely. By far the greatest use of vanadium remains in the steel industry, with approximately 35-45 tonnes being consumed per million tonnes of steel produced, world-wide. Within the steel industry, by far the greatest proportion of this vanadium is used in the production of high strength, low alloy, structural and linepipe steels. It is the intention, therefore, of the present paper to concentrate on the effect of vanadium on the production, properties and use of such steels.

PRODUCTION OF VANADIUM-CONTAINING STEELS

Most iron ores contain little or no vanadium and only in the titaniferous magnetites

of China, Russia and South Africa are higher levels found. Even in these cases the maximum vanadium pentoxide content of the ore is typically less than 2.0%. Additionally, the free energy of formation of V_2O_5 in liquid steel at 1500°C has been estimated to be⁽¹⁾ -47 170 k cal mol⁻¹, which is significantly more positive than that of Al_2O_3 at -102 263 k cal mol⁻¹. Thus, vanadium is unlikely to figure significantly in the normal oxidation/reduction reactions which accompany steel refining. However, it should be noted that some steel companies around the world, when manufacturing vanadium-containing steels, have used the relatively high free energy of formation of vanadium oxide to obtain in-situ reduction in combination with suitable reducing agents.

At the lower temperatures associated with casting, rolling and heat treatment, vanadium forms both a stable nitride and a stable carbide and it is from the solubility, precipitation and growth of these compounds that vanadium-containing steels obtain many of their properties.

a) Effect of Vanadium on Hot Ductility

Transverse cracking is a common defect particularly, though by no means exclusively, observed during continuous casting of microalloyed HSLA steels. It occurs as a result of longitudinal stresses being applied to the solidifying strand in regions where the steel ductility is low, e.g., the straightener where the temperature range may be 750-950°C.

As the strand cools into this temperature range the relatively coarse austenite grains become enveloped in a thin layer (<20µm) of ferrite, some of which may be deformation induced. As this ferrite has a significantly lower flow stress than the austenite from which it forms, the strain associated with the applied stress is accommodated by the ferrite. Anything which strengthens the austenite, such as precipitation, work hardening, solute drag etc, or weakens the grain boundary, e.g. second phase particles, or precipitate free zones, will reduce the ability of the ferrite to accommodate the strain and lead to a reduction in hot ductility and the appearance of transverse cracking. The effects of niobium on transverse cracking have been well documented over the years.

A number of workers^(2, 3, 4) have investigated the effects of vanadium on hot ductility and recent work by Mintz and Abushosha⁽⁵⁾, Figure 1, clearly indicates the relationship between vanadium, nitrogen and the level of hot ductility, in a tensile test, in a 0.1%C/1.4%Mn/0.03%Al steel. A comparison with a steel containing 0.03%Nb is also given. At lower levels of vanadium and nitrogen, i.e., <0.1%V, 0.005%N, the level of hot ductility at 850°C is less than that of the C/Mn/Al steel but well above that

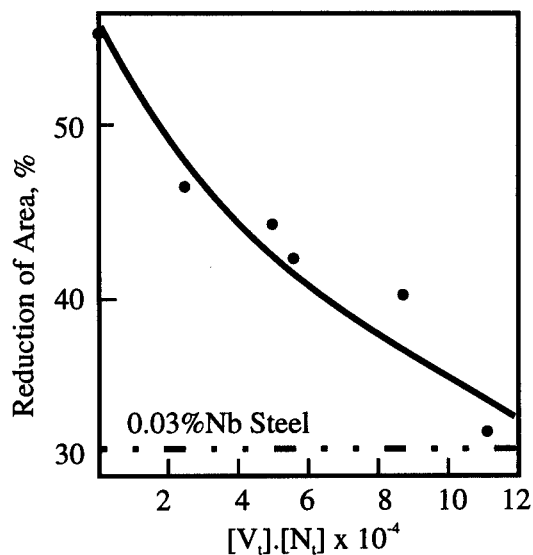


Figure 1. Influence of Product of Total V and N Contents on RA Value at 850°C (Ref. 5)

of the 0.03%Nb steel. It is only when vanadium levels approach 0.1%V, or greater, accompanied by nitrogen levels approaching 0.01%, or greater, that the level of hot ductility in the vanadium steels approaches that of the 0.03%Nb steel. The reason for this behaviour is thought to be associated with the fact that, under the conditions which exist on continuous casting machines, vanadium carbonitrides coarsen more rapidly than niobium carbonitrides and, thus, have a less deleterious effect on hot ductility than niobium. As coarsening under these conditions depends directly on the solubility of vanadium in austenite it would be expected that increasing the nitrogen content would give rise to a reduction in hot ductility, as observed. Any increase in the volume fraction of precipitate would also be expected to contribute to a reduction in hot ductility.

This suggests that, at the typical levels of vanadium and nitrogen found in many modern steels i.e., <0.1%V/<0.008%N, they should be able to be continuously cast without running into serious problems with transverse cracking. However, if high vanadium levels coincide with higher nitrogen levels, the same levels of care as currently exercised when casting niobium steels may be required.

b) Effect of Vanadium on Reheating Temperatures

Both vanadium carbide and vanadium nitride exhibit significant solubility in austenite. As can be deduced from the following equations:

$$\log VC = -\frac{9500}{T} + 6.72^{(6)}$$

$$\log VN = -\frac{7840}{T} + 3.02^{(7)}$$

the solubility of the carbide is much greater than that of the nitride. However, the above equilibrium relationship predicts that even with the high levels of 0.15%V and 0.02%N present, the equilibrium solution temperature of VN would be 1140°C, thus permitting the use of lower soaking temperatures. This relatively high solubility, especially of the carbides, is also important in heat treated steels as it permits vanadium to be taken into solution at the heat treatment temperature and made available for subsequent reprecipitation as fine strengthening precipitates during cooling.

A potential drawback of these relatively low temperatures for complete solution of vanadium, carbon and nitrogen in austenite is that grain coarsening may occur during reheating. It is now well established that this can be overcome by making an addition of 0.01-0.015%Ti, while ensuring that there is sufficient "free" nitrogen remaining for strengthening by precipitation of VN.

In this context Sage et al⁽⁸⁾ have investigated the effect of post solidification cooling rate (1470-1100°C) on the particle size of TiN particles and on the austenite grain coarsening temperature of vanadium containing steels. As shown in Figure 2(a) the particle size decreased by a factor of five as the cooling rate increased from 8 to 50°C/min. It should be noted that this also resulted in an increase in the average titanium content of the particles. The effect of PSCR on the austenite grain size of steels rolled with 50-60% total deformation from 1100, 1200 and 1300°C to finish rolling in the temperature range 1050-1120°C

is shown in Figure 2(b). It can be concluded from this figure that the optimum PSCR is 25-35°C/min and that the lower the reheating temperature the better. However, if the steels require a high level of heat affected zone toughness after welding then the titanium content of the particles needs to be greater than 80-90% and this can only be achieved at slab reheating temperatures of 1200°C or greater. Nevertheless, the potential to obtain a relatively fine as-rolled grain size still exists.

c) Effects of Vanadium during Rolling

During rolling it is recognised that vanadium does not significantly inhibit austenite recrystallisation or precipitation until relatively low temperatures are reached^(9, 10, 11). Consequently, rolling loads are similar to those obtained with C/Mn steels⁽¹²⁾. In addition, it has been demonstrated⁽¹³⁾ that, when recrystallisation occurs in C/Mn/V steels, the recrystallised grain size remains reasonably constant over a fairly wide range of temperature. Furthermore, it has also recently been shown⁽¹⁴⁾ that, when warm working vanadium-containing steels, a significant degree of strengthening is obtained from recovered ferrite. Some implications and applications arising from these observations will now be considered.

Firstly, in the rolling of wide (>1500mm), or thin (<3mm), high strength (>350N/mm²) strip steel, considerable attention requires to be paid to achieving gauge and shape control and to ensuring that rolling loads remain within the capability of the mill. The facts that vanadium steels recrystallise during rolling and that precipitation does not take place until low temperatures are reached significantly assist in this. The flow

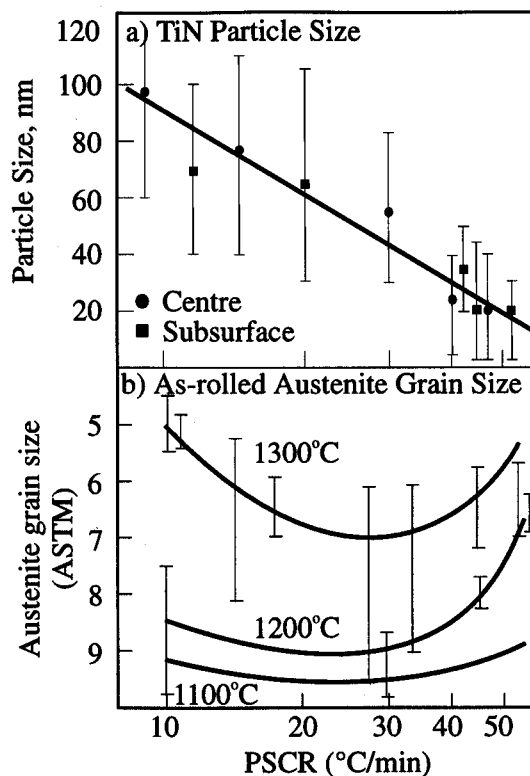


Fig. 2. Effect of Post Solidification Cooling on a) the Particle Size of TiN Particles b) the As-rolled Austenite Grain Size in a 0.12%C/1.6%Mn/0.45%Si/0.05%V/0.01%Ti/0.008%N Steel (Ref. 8)

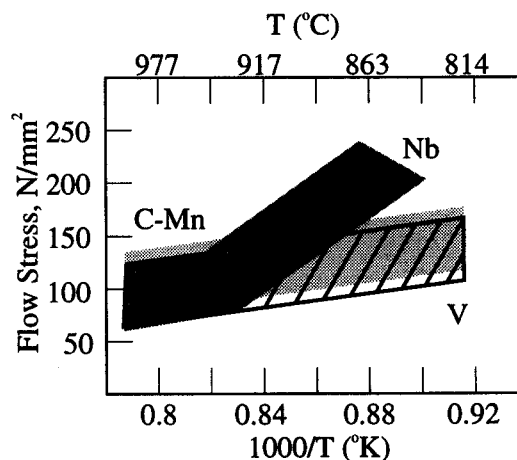


Fig. 3. Mean Flow Stress as a Function of Hot Rolling Temperature at the Finishing Stand (Ref. 15)

stress during rolling for vanadium microalloyed steels is compared with that of some other steels in Figure 3⁽¹⁵⁾. Clearly the flow stress curve for the vanadium steel continues to increase steadily with falling temperature, in a manner similar to that of C/Mn steel, without exhibiting the significant increase shown by the niobium steel at temperatures below about 930°C. Additionally, because the vanadium steel recrystallised until relatively low temperatures were reached there was little opportunity for the hard and soft spots, associated with partial recrystallisation, to form. Consequently, gauge control along the length of the coil improved, (Figure 4), resulting in greater control of run out length.

Secondly, when rolling material in modern high productivity mills, it is desirable to use rolling schedules which incorporate few, if any, delays. In so doing, finish rolling temperatures tend to increase. Also, in older mills, where the degree of process control which can be exercised is limited, rolling can result in significant variability in finish rolling temperature. In the manufacture of high strength low alloy steels such changes and variability in finish rolling temperature may lead to a deterioration of and increase in the scatter of mechanical properties, particularly toughness. However, it has been demonstrated⁽¹⁶⁾ that, while the toughness of vanadium-containing steels with 0.05-0.23%V and 0.009-0.014%N does deteriorate with increasing finish rolling temperature, this reduction in toughness can be accommodated, especially at lower nitrogen contents. This is confirmed in Figure 5⁽¹⁷⁾, which shows that as the finish rolling temperature of 25mm thick, 0.14%C/1.4%Mn/V, 350 N/mm² yield strength, structural steels increased from 820 to 1020°C, the 40J Charpy vee-notch impact transition temperature increased from approximately -80 to -60°C. This increase was accompanied by only a small reduction in yield strength and a slight coarsening of the grain size.

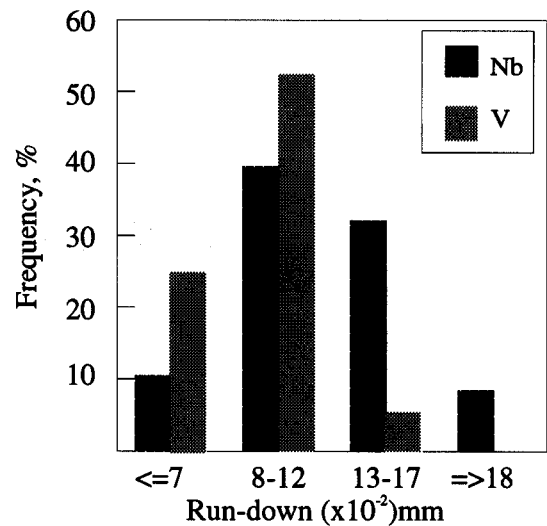


Fig. 4. Distribution of Thickness Run-down for V and Nb Steel (Ref. 15)

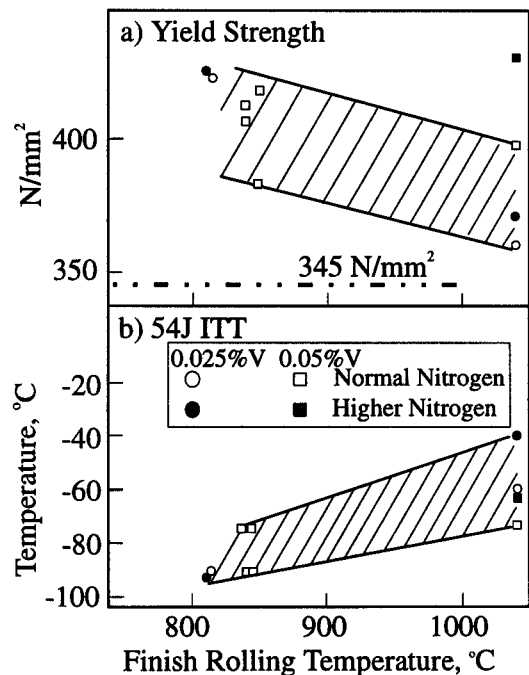


Fig. 5. The Effect of Finish Rolling Temperature on the Mechanical Properties of 25mm Thick Plate of CEV=0.38 (Ref. 17)

Thirdly, in thermo-mechanically treated steels, where the achievement of very fine grain size, accompanied perhaps by some precipitation strengthening is required, then niobium has normally been the chosen microalloying element. Niobium steels rolled at low temperatures exhibit sluggish, or no, recrystallisation and result in the development of highly deformed austenite grains of large surface area/volume ratio which transform to fine ferrite and pearlite associated, perhaps, with some precipitation of NbCN. However, in the absence of accelerated cooling, achievement of high strength levels can require rolling in the two phase austenite/ferrite region to generate a high dislocation density in the ferrite. Recent work⁽¹⁴⁾ has examined and compared the effects of rolling in the two phase region of the microstructure and properties of 0.15% vanadium, 0.06% niobium and 0.05% vanadium-0.03% niobium steels, given up to 50% total reduction at around 700°C. Increasing the degree of deformation in the two phase region increased the grain elongation ratio of the transformed ferrite grains and resulted in an increase in yield strength due to precipitation (σ_p) and dislocation substructure (σ_d). As can be seen from Figure 6, the increase in $\sigma_p + \sigma_d$ was greater in the case of vanadium steels than in the cases of the niobium or niobium + vanadium steels. At the same time, the impact transition temperatures of some of these vanadium steels, were significantly better than would have been predicted from the normal understanding of the effect of precipitation on toughness (Figure 7).

PROPERTIES OF VANADIUM-CONTAINING STEELS

- a) **Structural Steels**
Vanadium has been widely

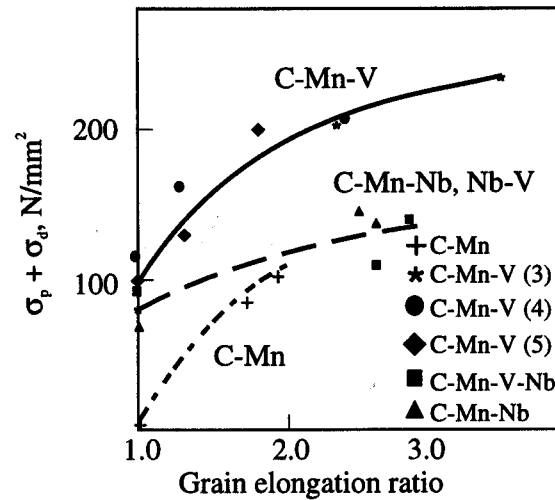


Fig. 6. Relationship between Grain Elongation Ratio and Strengthening from Dislocations in C-Mn Steel and Steels containing Vanadium and Niobium (Ref. 14)

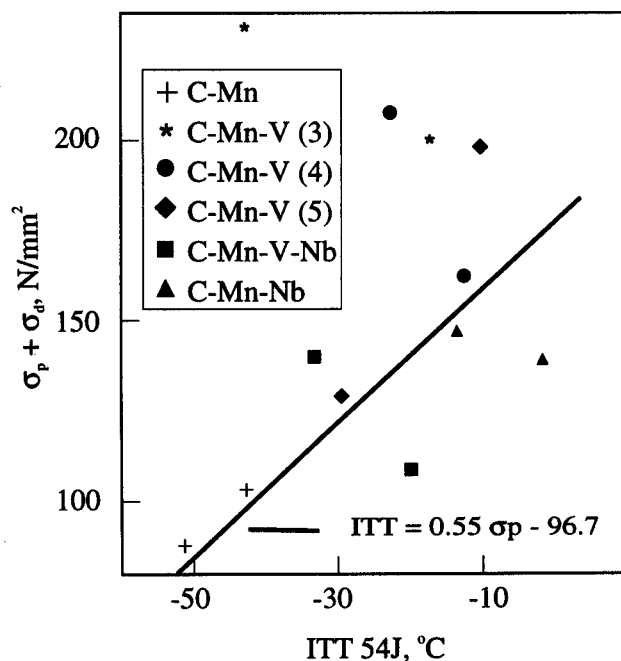


Fig. 7. The Influence of Precipitation and Dislocation Strengthening on the 54J ITT Corrected to a Grain Size of $10\text{mm}^{-1/2}$ (Ref. 14)

used in the manufacture of structural steels for many years, primarily, as noted above, because it acts as a good precipitation strengthener while at the same time providing reasonable toughness, even at high finish rolling temperatures. Recently, attention has focused on the provision of steels which have been lightly controlled rolled with finish rolling temperatures within the range of temperatures frequently associated with normalising treatments i.e. 850-950°C. These steels depend on a combination of moderately fine grain size and precipitation strengthening for achievement of properties and are known as temperature controlled rolled or normalised rolled steels.

Figure 8⁽¹⁷⁾ shows the effect of vanadium on the tensile and impact properties of 25mm thick plate of 0.1-0.14%C/1.4%Mn temperature controlled rolled steel with the tensile properties being shown at constant CEV = 0.38. Many steels of this type are manufactured to a specification which includes 345 N/mm² yield strength, 490 N/mm² UTS, with 40J Charpy vee-notch energy absorption at -40°C. From Figure 8 it is clear that ability to achieve the UTS is the controlling feature in achieving tensile properties of these steels, both the yield strength and impact transition temperature being achieved with relative ease. Even in the case of the UTS, an addition of vanadium of the order of 0.03-0.05%, would be sufficient to ensure achievement of the specification requirements up to at least 40 mm thick. Furthermore, if a lower level of UTS (say 460 N/mm²) was required, then this could be achieved by reducing the CEV to around 0.35, by reducing the carbon content, while maintaining the other factors constant. This would have the additional benefit of improving both toughness and weldability.

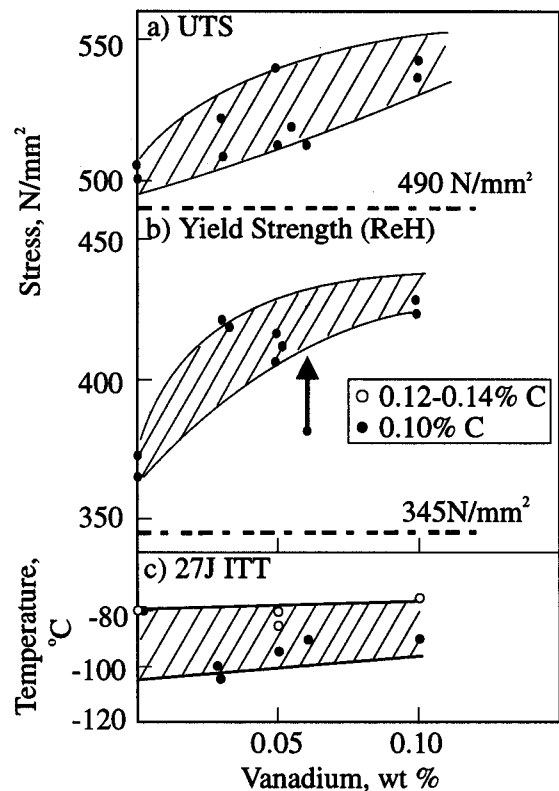


Fig. 8. The Effect of Vanadium on the Mechanical Properties of 25mm Thick Temperature Controlled Rolled Steels. Tensile Properties Corrected to Constant CEV=0.38, Impact Properties Uncorrected for Differences in CEV. (Ref. 17)

While it is possible to achieve relatively high strength and toughness levels by temperature controlled rolling, the most acceptable method of obtaining even higher strength properties in structural steels is by accelerated cooling. When optimised, this process ensures a fine grain size, accompanied by precipitation strengthening and a good distribution of carbon rich phases in the microstructure. It has been shown⁽¹⁷⁾ that a yield strength of 420 N/mm², UTS of 550 N/mm², accompanied by a 54J transition temperature of -70°C can be obtained in 25mm thick plate of 0.34 CEV, containing 0.05%V plus 0.5%(Cu+Ni),

finish rolled at 800°C and accelerated cooled at 10-15°C/sec between 800 and 550°C.

b) Linepipe

Vanadium is widely used, usually in combination with niobium, in the manufacture of higher strength and/or greater wall thickness linepipe steels. It is particularly useful in the manufacture of low carbon, low manganese API 5LX-65 steel, used for sour gas applications where an addition of 0.04-0.06%V is common.

Currently, much interest surrounds the development of linepipe steels of yield strength 80 ksi and greater. The preferred route for the manufacture of these steels is by accelerated cooling. Figure 9⁽¹⁸⁾ shows the effect of cooling rate between 800 and 500°C on the yield strength of a 0.09%C/1.8%Mn/0.25%Si/0.2%Cu/0.1%Ni steel containing 0.08%V, 0.05%Nb and 0.01%Ti. The yield strength increased by 4.1 N/mm² per °C increase in cooling rate, while at a constant cooling rate of 20°C/sec it increased by 10.1 N/mm² for each 10⁻⁴ increase in vanadium-nitrogen product (Figure 10). This latter figure illustrates the relatively strong interaction between vanadium and nitrogen and shows that in this type of steel, at this cooling rate, a vanadium level of 0.08% with 0.005%N accompanied by 0.05%Nb is required to achieve 550 N/mm² yield strength in the plate. It should be noted that the high yield strength in this steel was due to a combination of a very fine grain size (2-3mm) and substitutional and precipitation strengthening each contributing close to 100 N/mm². In this particular instance, the high strength level was accompanied by an excellent level of toughness of around 100J @ -100°C in the transverse direction. Work is continuing with this steel to assess

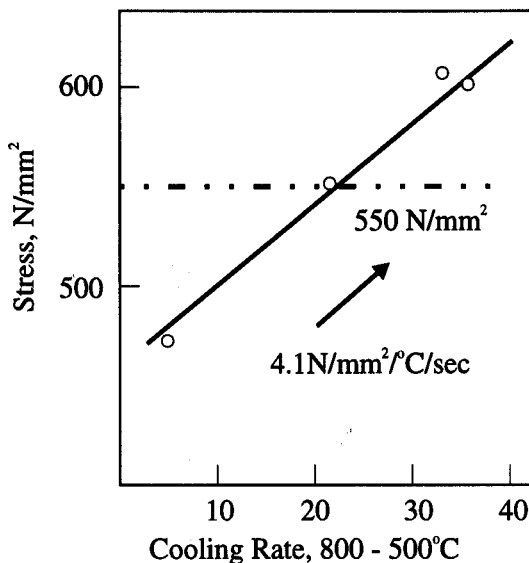


Fig. 9. The Effect of Cooling Rate on the Yield Strength of API 5LX80 Accelerated Cooled Steels Containing 0.08%V/0.05%Nb /0.01%Ti /0.006%N at Constant Si, Mn, Cu and Ni. (Ref. 18)

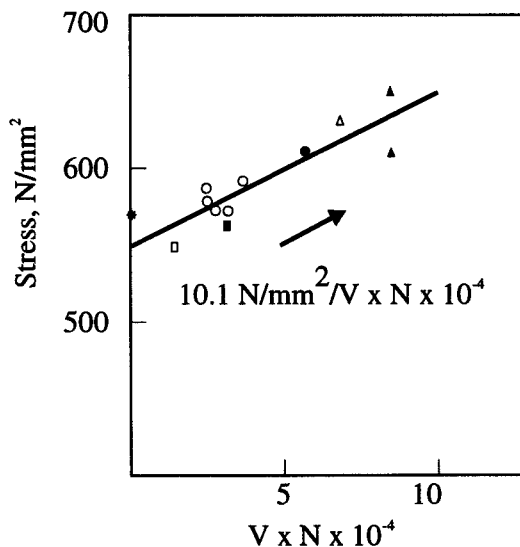


Fig. 10. The Effect of Vanadium and Nitrogen on the Yield Strength of API 5LX80 Accelerated Cooled Steels at 0.25% Si/1.8% Mn/0.2% Cu/0.1% Ni at Constant Cooling Rate from 800-500°C of 20°C/sec. (Ref. 20)

its capabilities for achieving a strength level of 100ksi or greater. Initial results⁽¹⁹⁾ indicate that an addition of up to 0.25%Mo may be required.

WELDABILITY

There have been many papers over the years which have dealt with the weldability of microalloyed steels. However, since the work of Hannerz⁽²⁰⁾ comparatively few have given details of systematic investigation of the effects of vanadium on the heat affected zone microstructure, hardness and toughness and this will now be discussed, especially with respect to the grain coarsened heat affected zone (GHAZ) close to the fusion boundary. In addition, the use of titanium as a means of controlling the GHAZ microstructure in vanadium treated steels will be considered.

a) The Effect of Vanadium on HAZ microstructure

Easterling⁽²¹⁾ and his co-workers have demonstrated that, at the temperatures and times involved in welding processes, most, if not all, of the vanadium compounds present in conventional HSLA steels will be taken into solution in the GHAZ. As most of the other microalloying elements, with the exception of titanium, will also be taken into solution grain coarsening can occur, the extent of which depends on the maximum temperature and the time at temperature. Thus, as Wang et al⁽²²⁾ have shown for structural steels containing 0.09%V, increasing the heat input from 3 kJ/mm to 6 kJ/mm resulted in an increase in the HAZ austenite grain size from 114 to 203mm.

The addition of titanium can significantly improve austenite grain size. Nishio et al⁽²³⁾, among others, have demonstrated that with an 0.01-0.03%Ti addition, which was slightly hypostoichiometric with respect to oxygen and nitrogen, an austenite grain size of 50 μ m was achieved in simulated welds with $T_{max} = 1350^{\circ}\text{C}$ and $\Delta t = 40$ secs. As shown in Figure 11, Zajac et al⁽²⁴⁾, using a 0.09%C/1.4%Mn/0.08%V/0.01%Ti steel, have determined that increasing the nitrogen content from 0.003% to 0.013% resulted in significant refinement of the austenite grain size at 1350 $^{\circ}\text{C}$. However, as the temperature increased to 1450 $^{\circ}\text{C}$, coarsening occurred even in the high nitrogen steel.

As previously noted, Sage et al⁽⁸⁾ have demonstrated that the post solidification cooling rate has a significant effect on the TiN particle size and composition and on the austenite grain size during reheating for rolling. There is a similar effect on the heat affected zone grain size during welding and, in this work, an austenite grain size of 35-40 μ m was obtained in a 0.05%V/0.01%Ti/0.01%N steels which had been cooled at 15 $^{\circ}\text{C}/\text{min}$ during casting and reheated to 1300 $^{\circ}\text{C}$ for 20 seconds. Thus, while titanium

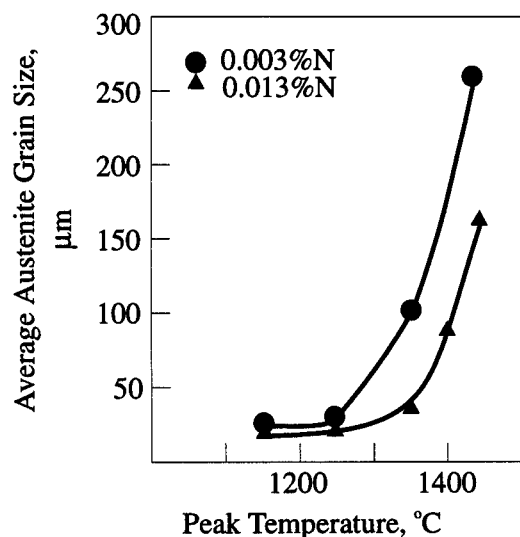


Fig. 11. Grain Growth of Austenite as a Function of Peak Temperature in the HAZ of 0.01%Ti/0.08%V/N Steels (Holding Time 1 sec) (Ref. 24)

treatment may not result in significant refinement close to the fusion boundary it should result in a reduction in the width of the coarsest region, especially in steels manufactured by continuous casting.

Austenite grain size does, of course, have a significant effect on transformation temperatures and in the above work of Zajac it was reported that reducing the austenite grain size from approximately 100 μ m to 50 μ m resulted in an increase in transformation start temperature of 70-100°C. As would be expected this increase in transformation temperature led to a decrease in the amount of lower temperature aligned phases in the microstructure and was accompanied by an increase in the amount of grain boundary and intragranular ferrite.

In Ti-free 0.14%C/1.4%Mn steels, at a reasonably constant austenite grain size of 110-130 μ m, reheated to 1350°C with cooling times between 800 and 500°C ranging from 6 to 250 seconds, increasing the vanadium content from 0.0 to 0.1% had no effect on the transformation start temperature⁽²⁵⁾. Once transformation commenced the rates were similar to those observed in conventional C/Mn steels and the end transformation temperatures were also similar. Despite these similarities, the microstructures were significantly different, especially at the longer cooling times associated with high heat input welding (Figure 12). Here it can be noted that at longer cooling times vanadium steels exhibited less allotriomorphic and sideplate ferrite and slightly more grain boundary polygonal ferrite than the C/Mn steel. Furthermore, as can be observed in Figure 13, the nature of the acicular ferrite transformation product in the vanadium steel appeared to be different from that of the other steel. As has previously been noted^(26, 27, 28) it tended more towards an interlocking, intragranular, type of microstructure.

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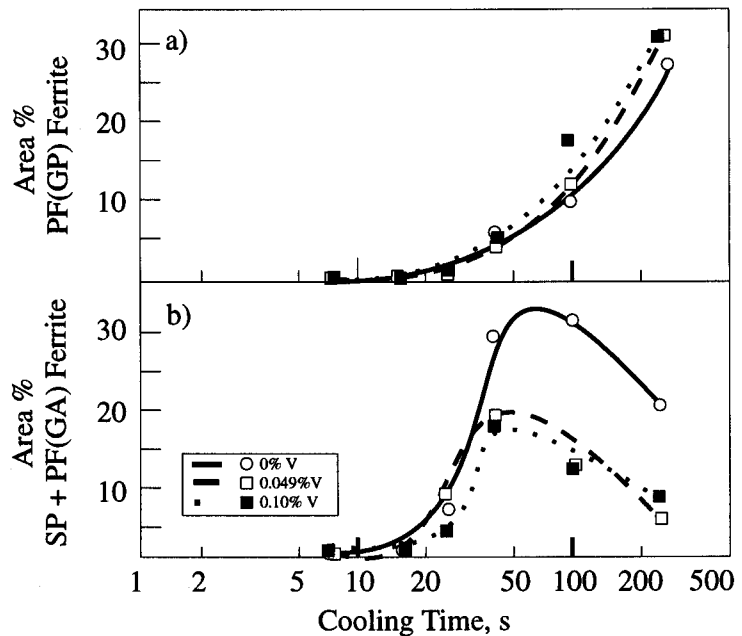
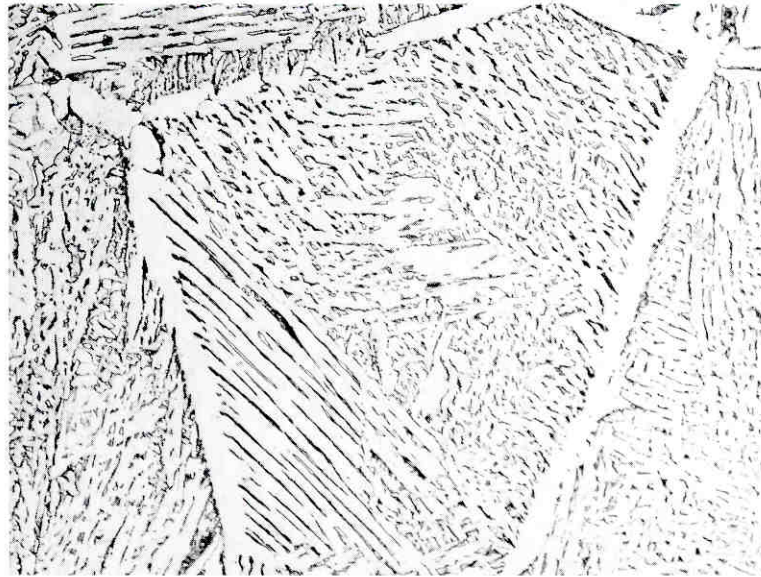


Fig. 12. The Effect of Cooling Time from 800-500°C on a) The Amount of Polygonal Ferrite b) The Amount of Allotriomorphic and Sideplate Ferrite in Simulated HAZ's of C-Mn and C-Mn-V Steels (Ref. 27).

It has also been suggested^(29,30) that vanadium can tend to promote the formation of M-A phase. However, as the steels investigated in these papers contained variable boron and/or aluminium levels and some additionally contained niobium this tendency must remain, as yet, unproven.



a) C-Mn



b) C-Mn-0.10%V

Fig. 13. Simulated HAZ Microstructures, Cooling Time
(800-500°C) 43 secs (Ref. 27) x500

In addition to modifying the microstructure, vanadium results in precipitation hardening in the heat affected zone. Figure 14⁽³¹⁾ indicates that the level of hardening increases with increase in vanadium level and that it is more pronounced in the post weld heat treated condition than it is in the as welded condition. This latter effect is undoubtedly due to tempering of the matrix accompanied, perhaps, by some further precipitation. The important thing to recognise is, however, that in the steels containing less than approximately 0.08%V, reported in this work, the hardness after post weld heat treatment was lower than that in the as welded condition. This work also indicated that increasing the nitrogen content from 0.007 to 0.017% resulted in 20-25Hv increase in hardness in both the as welded and post weld heat treated conditions. Furthermore, the work of Lau et al⁽²⁹⁾ indicates that as the heat input increased up to about 6 kJ/mm the as welded hardness decreased but from comparison of

the steels included in this work it would appear that it is the hardness of the matrix which is decreasing and not the degree of precipitation hardening per se.

b) The Effect of Vanadium on HAZ Toughness

As previously noted, Hannerz⁽²⁰⁾ carried out one of the first systematic studies of the effect of vanadium on HAZ toughness of simulated welds, over a range of cooling times $\Delta t_{8/5}$ 33 to 300 secs. In this work it was observed that at vanadium levels up to 0.1% there was little or no effect of vanadium on HAZ toughness. Hannerz also reported little or no effect of nitrogen on the HAZ toughness of the steels investigated.

More recently, the effect of vanadium on the HAZ toughness of multipass welds carried out on 25mm thick plate, welded at 2 kJ/mm ($\Delta t_{8/5} = 12$ secs), has been ascertained⁽³¹⁾. In this work, Figure 15, increasing the vanadium content up to 0.16% resulted in a continuous improvement in as-welded impact transition temperature of about 40°C while, in the same interval, the CTOD transition temperature increased by only 10°C. The improvement in toughness noted, particularly in the case of the Charpy vee-notch transition temperature, was thought to result from the formation of a microstructure containing intragranular ferrite which

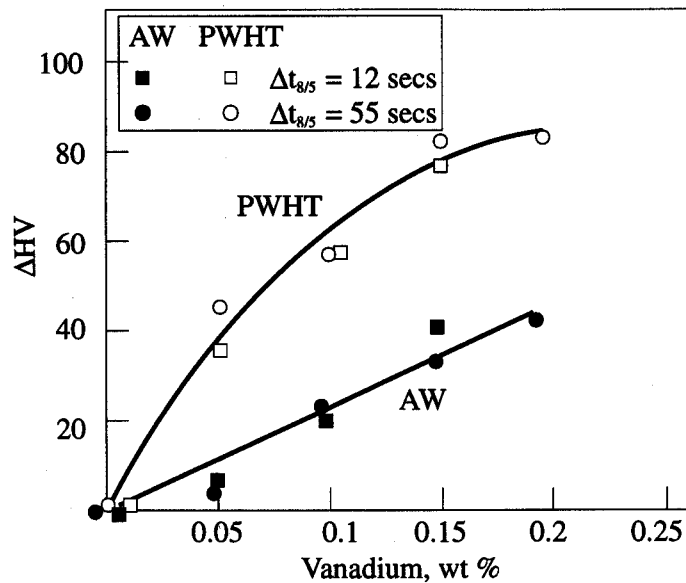


Fig. 14. The Effect of Vanadium on the Change in Maximum HAZ Hardness in the As Welded and Post Weld Heat Treated Conditions (Ref. 31).

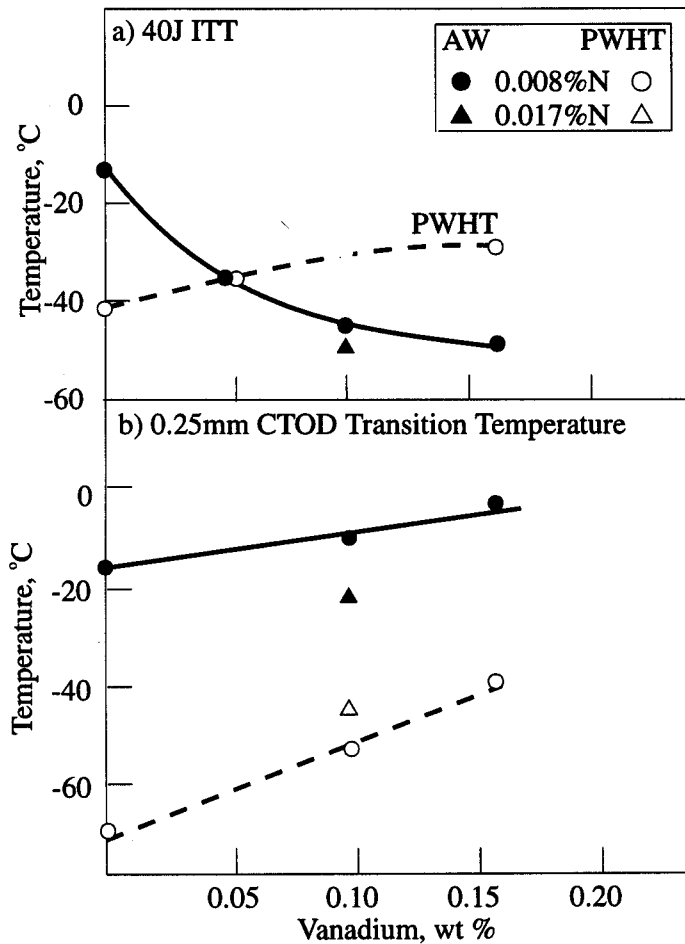


Fig. 15. The Effect of Vanadium on the Toughness of 0.12%C, 1.6%Mn Steels Multipass Welded at 2kJ/mm. ($\Delta t_{8/5} = 12$ secs) (Ref. 33)

more than offset any increase which may have accrued from precipitation hardening.

Figure 15 also depicts the effect of post weld heat treatment for one hour at 600°C. As the matrix tempered and precipitation strengthening became more pronounced the effect of vanadium on the Charpy vee-notch transition temperature was almost a mirror image of that in the as welded condition. Consequently, at vanadium levels of around 0.06% or less the toughness was the same or better than that in the as welded condition whereas, at vanadium levels greater than this a small increase in transition temperature of up to 20°C was observed. On the other hand, post weld heat treatment resulted in a 30-50°C improvement in CTOD transition temperature. The fact that the effect of vanadium on the CTOD transition temperature is similar in both the as welded and post weld heat treated conditions suggests that vanadium has had little effect on M-A phase formation in these steels. It should also be noted from Figure 15 that increasing the nitrogen content from 0.007 to 0.017% in steels containing 0.1%V had little or no effect on either Charpy vee-notch or CTOD transition temperatures in either as welded or post weld heat treated conditions.

The effect of heat input on the toughness of as welded vanadium-containing steels is shown in Figure 16. This figure has been compiled from a number of papers investigating steels of similar chemical composition and thickness(25, 27, 31). It should be noted that they include results from single pass bead in groove, single pass bead on plate and multipass welds notched in the grain coarsened HAZ. From this figure it would appear that as the heat input increased up to 6 kJ/mm the Charpy vee-notch transition temperature exhibited a small deterioration, while the CTOD transition temperature exhibited an improvement. It will be recalled from Figure 11 that as the cooling time increased vanadium containing steels tended to exhibit more grain boundary polygonal ferrite and less sideplate and allotriomorphic ferrite and that these changes were possibly accompanied by a relatively constant level of precipitation strengthening⁽²⁷⁾. It is these changes in

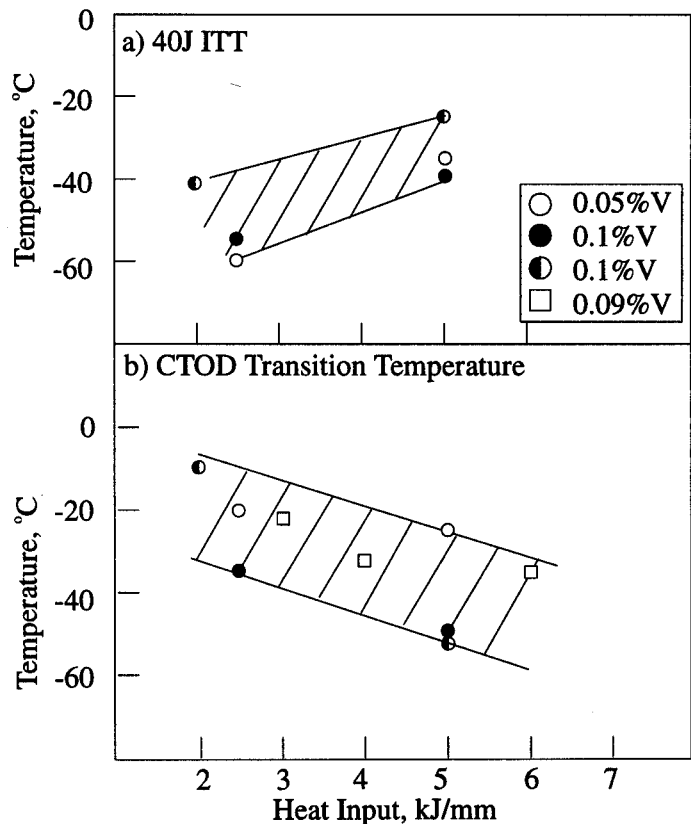


Fig. 16. The Effect of Heat Input on the As Welded HAZ Toughness of Vanadium Steels (Ref. 25, 27, 31)

microstructure which are thought to have contributed to the observed changes in toughness.

As with austenite grain size, a titanium addition which is slightly hypostoichiometric with respect to oxygen and nitrogen can be beneficial to HAZ Charpy vee-notch toughness⁽²³⁾. Zajac⁽²⁴⁾ has investigated the effect of heat input on the HAZ toughness of titanium-containing, 0.08%V, steels at nitrogen levels of 0.003 and 0.013% (Figure 17). Comparing his results with the titanium-free steels, given in Figure 16, the addition of titanium to the lower nitrogen steel resulted in an improvement in impact transition temperatures which was retained with increasing heat input up to approximately 8kJ/mm.

Although increasing the nitrogen content of the titanium treated steels from 0.003 to 0.013% was observed to refine the austenite grain size, this was not reflected in an increase in toughness and the level of impact transition temperature of the higher nitrogen steel was similar to that of the titanium-free steels up to a heat input of approximately 5kJ/mm.

The effect of heat input on the CTOD transition temperature of a vanadium-titanium steel is shown in Figure 17⁽²²⁾ where it is also compared with the results of the titanium-free steels of Figure 16. In this case there was a significant deterioration in CTOD transition temperature as the heat input was increased from 3 to 6 kJ/mm. From consideration of all these results it would appear that it is only at low heat input i.e. 3.5 kJ/mm that titanium-vanadium steels have similar or better HAZ properties to those of titanium-free vanadium-containing steels.

CONCLUSIONS

Some of the factors affecting the production, properties and weldability of vanadium containing steels have been discussed. It has been demonstrated that:

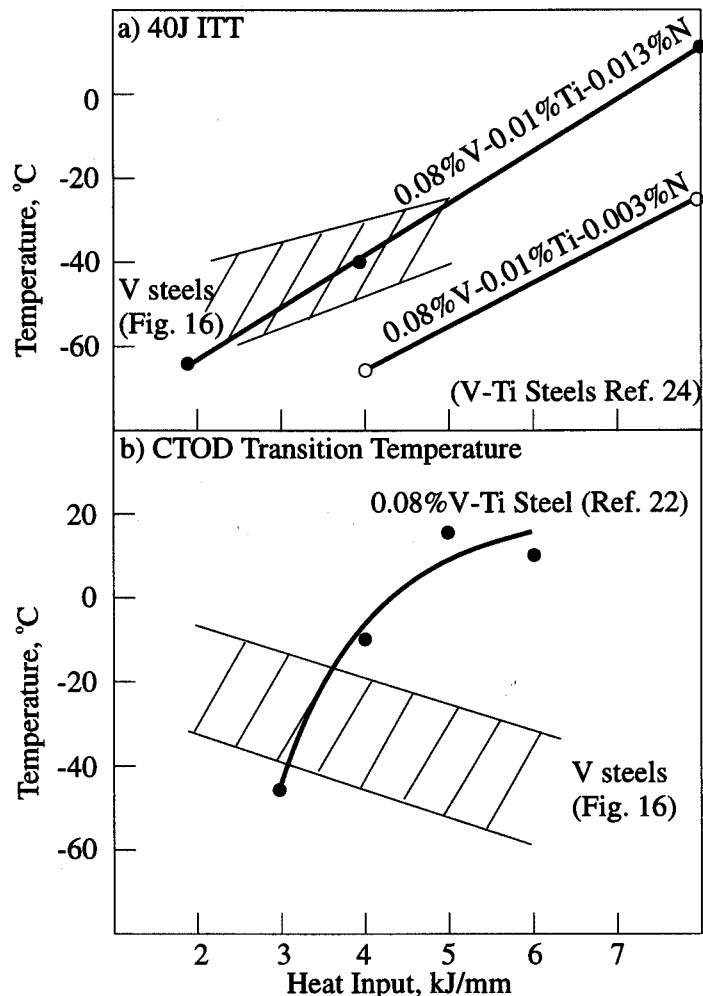


Fig. 17. The Effect of Heat Input and Titanium Addition on the HAZ Fracture Toughness of C-Mn-V Steels.

- 1) Providing the vanadium-nitrogen product is maintained at 1.2×10^{-8} , or less, the hot ductility of vanadium containing HSLA steels is relatively high leading to the possibility of a reduced level of transverse cracking during continuous casting.
- 2) In vanadium containing steels to which an addition of approximately 0.01%Ti has been made, the typical cooling rate at quarter depth of a continuously cast slab results in a TiN particle size and composition which ensure a significant increase in the grain coarsening temperature during reheating for rolling or during welding.
- 3) Vanadium containing HSLA steels tend to recrystallise during rolling leading to rolling loads similar to those obtained with C/Mn steels. Consequently, they are suitable for use when rolling wide, thin, hot rolled coil where improvements in gauge control have also been noted.
- 4) When vanadium containing steels recrystallise the recrystallised austenite grain size tends to be reasonably constant over a relatively wide range of temperature. As a result vanadium containing HSLA steels exhibit relatively small changes in properties with change in finish rolling temperature in the range 800-1000°C.
- 5) In vanadium containing steels which have been rolled with heavy deformation in the two phase region there is some evidence to suggest that recovery can occur, leading to a significant increase in strength, accompanied by a high level of toughness.
- 6) A structural steel of yield strength 345 N/mm², UTS 490 N/mm² and 40J Charpy vee-notch energy absorption at -40°C can be manufactured using controlled rolled vanadium containing steel, of approximate composition 0.12%C/1.5%Mn/0.03-0.05%V, in thicknesses up to at least 40 mm. Even higher strength and toughness levels can be obtained by accelerated cooling.
- 7) In the manufacture of API 5LX80 linepipe steels containing 0.09%C/1.8%Mn/0.25%Si/0.05%Nb/0.01%Ti cooled at 20oC/sec from 800-500°C a vanadium addition of around 0.08%, at 0.005%N, can be used to ensure the required strength level in the skelp.
- 8) Vanadium appears to promote the formation of relatively tough microstructures in weld heat affected zones. Consequently, in multipass welds carried out at 2 kJ/mm, increasing the vanadium content from 0.0 to 0.16% led to an improvement in the Charpy vee-notch toughness of the grain coarsened heat affected zone and was accompanied by only a small increase in CTOD transition temperature.
- 9) Depending on the vanadium level, post weld heat treatment can result in an improvement or a small reduction in Charpy vee-notch toughness accompanied by a significant improvement in CTOD transition temperature.
- 10) In vanadium containing steels increasing heat input during welding appears to promote the formation of polygonal ferrite in the prior austenite grain boundaries. This is accompanied by a small reduction in Charpy vee-notch

toughness, accompanied by an improvement in the CTOD transition temperature.

- 11) While a titanium addition which is slightly hypostoichiometric with respect to oxygen and nitrogen can result in a significant improvement in heat affected zone austenite grain size, such additions, on balance, appear to result in little or no overall benefit in vanadium containing steels welded at heat inputs greater than about 3.5 kJ/mm.

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