THE EFFECT OF VANADIUM ON THE MICROSTRUCTURE AND TOUGHNESS OF WELD HEAT AFFECTED ZONES

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Abstract: Work on factors affecting heat affected zone toughness is summarised and includes the effects of vanadium, nitrogen and heat input on the toughness of the coarse grained HAZ and thr intercritically reheated, grain coarsened HAZ as well as the effect of vanadium on the toughness of laser weldments.

Key words: vanadium, microstructure, toughness, heat affected zone.

1. Introduction

Hannerz and Jonsson-Holmquist(1), in 1974, carried out one of the first systematic investigations into the effects of vanadium on weld heat affected zone microstructure and toughness. They used simulated welds in a steel containing 0.15%C – 1.4%Mn, the welds having cooling times between 800^{o}C and 500^{o}C of 33 secs, 100 secs and 300secs. At cooling times of 33 secs and 100 secs they observed little or no effect of vanadium on HAZ toughness up to 0.1%V, or perhaps more. Indeed, at the lower cooling time, thought to be similar to that experienced by a 20mm thick plate welded at 3.6kJ/mm(2), the HAZ toughness appeared to exhibit a small improvement at a vanadium level of 0.06%. Even at the longest cooling time, equivalent to welding 20mm thick plate at 10.8kJ/mm the deterioration in HAZ toughness, in steel containing 0.1%V, was small. It was only at higher vanadium levels that significant reductions in toughness occurred at any cooling time.

Since 1974 there have been many papers which have examined different aspects of the effect of micro-alloying on HAZ toughness. It is the purpose of the present paper to summarise work on factors affecting HAZ toughness which, has been carried out on behalf of VANITEC supplemented, as required, by other published material. This includes an examination of the effects of vanadium, nitrogen and heat input on the toughness of the coarse grained HAZ, the effect of vanadium on the toughness of the intercritically reheated HAZ and the effect of vanadium on the toughness of laser weldments.

2. Factors Affecting the Heat Affected Zone Toughness of Vanadium Microalloyed Steels

Among the more important factors affecting the heat affected zone toughness of vanadium micro-alloyed steels are likely to be : -

- The solubility of VN and VC and the size of any particles present.
- The austenite grain size in the coarse grained heat affected zone.
- The effects, if any of vanadium on transformation temperatures, rates of transformation and the resulting microstructures.
- The size and amount of any M-A phase which may form in the intercritically reheated heat affected zone.
- Precipitation hardening within the HAZ.

2.1 Solubility

Equilibrium solubility data for VC and VN in austenite and ferrite, as shown in the equations in Table 1, indicate significant solubility of vanadium compounds in both phases, that of the carbide being higher than the nitride. In addition their solubility in austenite is greater than that in ferrite.

Table 1. Solubility Products of VC and VN in Austenite and Ferrite

	Phase	A	B	References
VN	Austenite	-8330	3.46	3
	Ferrite	-7830	2.45	4
VC	Austenite	-9500	6.72	5
	Ferrite	-12265	8.05	6

$Log_{10}[\%M]^a[\%X]^\beta=A/T+B$

In welds equilibrium conditions are rarely observed. However, Easterling and his co-workers(7) have shown that, even under welding conditions, the greatest majority, if not all, of the VC and VN present should be in solution in the coarse grained region next to the fusion boundary. The further away from the fusion boundary the more likely will VC and VN remain out of solution and in the intercritically reheated region there will, almost certainly, be some VC and VN present. The possibility of obtaining some dissolution and/or precipitation in the sub-critically reheated heat affected zone cannot be overlooked.

2.2 Grain Coarsening

In steels micro-alloyed with vanadium (or with vanadium + nitrogen) there will be few, if any, particles to impede austenite grain growth close to the fusion boundary. Consequently, at temperatures above the natural grain coarsening temperature, grain growth can occur. For vanadium containing steels the natural grain coarsening temperature is likely to be of the order of 1000°C - 1100°C or, perhaps, higher bearing in mind the non-equilibrium nature of the weld. The maximum austenite grain size which is observed close to the fusion boundary has been shown by Lau et al[8] to depend on the weld heat input and it increased from 100µm approx to 200µm approx as the heat input increased from 3kJ/mm to 6kJ/mm.

It should also be borne in mind that, in multi-pass welding, grain refinement of previously coarsened grains can be obtained as the heat affected zone from one pass is reheated through the austenite-ferrite transformation by subsequent passes. It is also possible to restrict the amount of austenite grain growth which occurs by adding titanium to form TiN and TiO_2[9]. However, this will not be considered here.

2.3 Transformation from Coarse Grained Austenite

The effect of vanadium on the transformation of austenite to ferrite in simulated, coarse grained, weld heat affected zones is shown in Figure 1(a). This shows that for a steel containing 0.13%C – 1.45%Mn containing 0.1%V, with austenite grain size typically 110 - 130µm and with cooling times between 800°C and 500°C ranging from 5 - 250 seconds, i.e. a range typical of most welding conditions, there was little or no effect of vanadium on the transformation start temperature when compared with that of a C-Mn steel. In addition, as can be seen from Figure 1(b) both the rate of transformation and the range of temperature over which transformation occurs were largely unaffected by vanadium additions up to 0.1%V, at the weld cooling times shown.

Although there was little difference in transformation temperature between vanadium-containing and vanadium-free steels, there can be significant differences in microstructure. Figure 2 depicts the effect of simulated weld cooling time on the microstructure of the above C-Mn and C-Mn-V steels.

While there were little or no differences in the amounts of grain boundary polygonal ferrite and of ferrite with aligned second phase between the three steels shown there were significant differences in the amount of grain boundary allotriomorphic (PF(GA)) + sideplate (SP) ferrite, especially at longer cooling times. The amount of SP + GA ferrite present in the C-Mn steel was, at peak, almost double that observed in the C-Mn-V steels. In addition, it was noted that the ferrite in the vanadium steels was more intra-granular in nature and the effect of vanadium on intra-granular ferrite formation in a steel containing 0.07%C – 1.6%Mn, two pass welded at 5kJ/mm, is shown in Figure 3.

Vanadium has a similar effect on the microstructure of CO_2 laser welds. Figure 4 shows that, as the vanadium level increased up to 0.1%, the amount of acicular ferrite in the microstructure increased, the increase being greater at 14 kW than at 18kW.

As can be seen in Figure 5 a significant proportion of the acicular ferrite appears to have resulted from intra-granular nucleation.

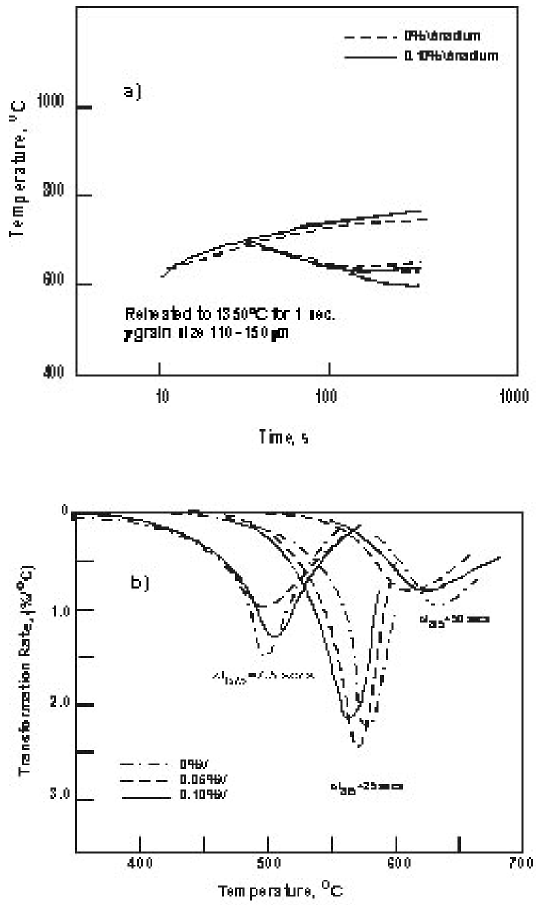


Figure 1 The effect of vanadium on a) The temperature for start of transformation from austenite to ferrite and b) the rate of transformation and the temperature range for transformation.

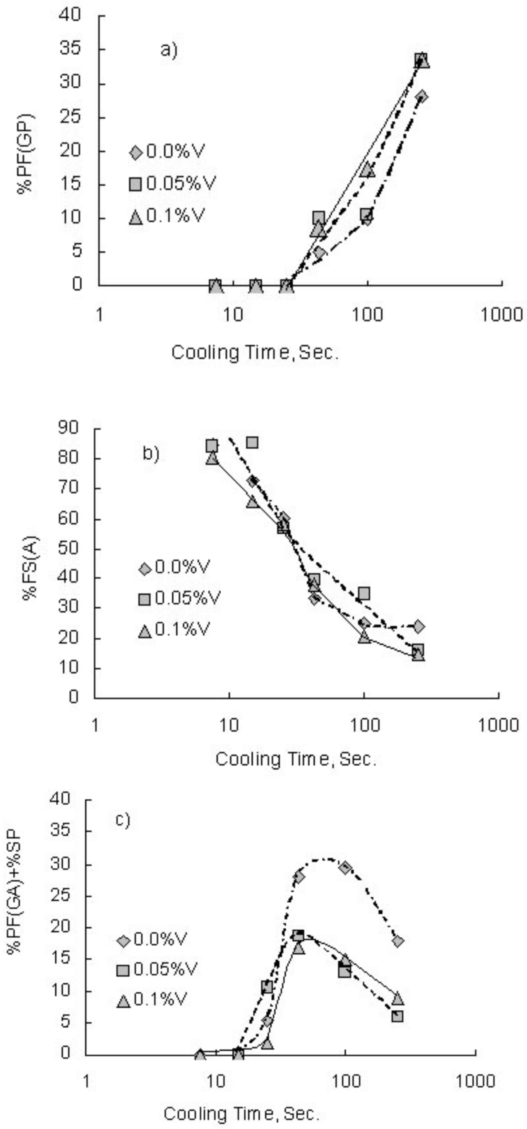
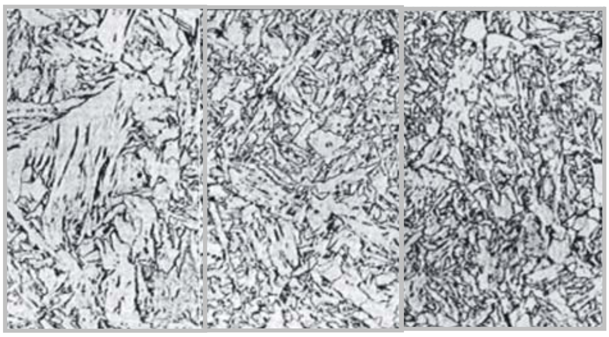


Figure 2 The effect of weld cooling time on the amount of a) grain boundary polygonal ferrite. b) ferrite with aligned second phase c) grain boundary allotriomorphic + sideplate ferrite, in simulated welds.



0.0%V 0.1%V 0.19%V

Figure 3 Effect of vanadium on intra-granular ferrite formation.

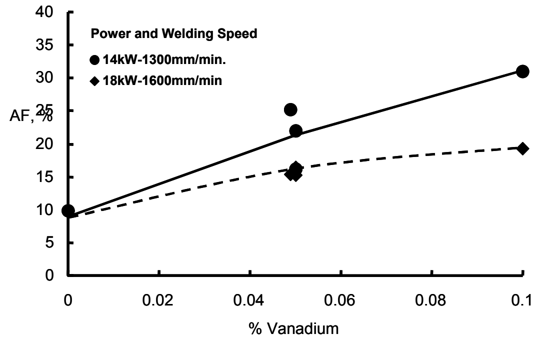


Figure 4 Effect of vanadium on acicular ferrite formation in laser welds.

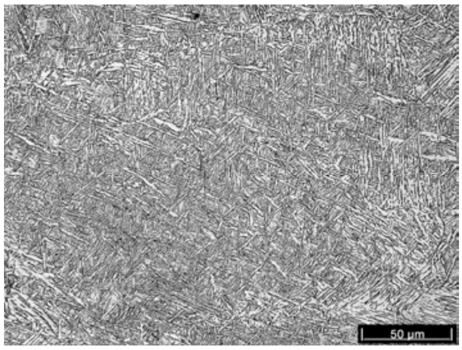


Figure 5 Intra-granular acicular ferrite in a laser weld.

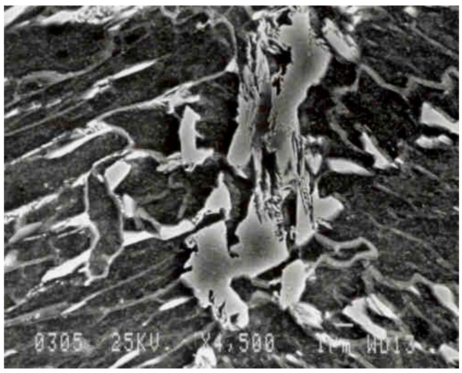


Figure 6 M-A phase from an inter-critically reheated HAZ.

2.4 Transformation from Intercritically Reheated Austenite

In the intercritically reheated heat affected zone the critical feature of the heat affected zone is the appearance of brittle M-A phase which results from partial transformation of the as-welded microstructure to austenite, followed by rapid cooling. The stability of the M-A phase is to a large extent dependent on its carbon content and this microstructure, an example of which is shown in

Figure 6, tends to form after reheating to relatively low temperatures in the inter-critical range. Any steel constituents such as B, Cr, Mn, Mo, P etc which tend to stabilise austenite will accentuate the amount of M-A phase. Constituents which tend to promote the formation of ferrite will tend to reduce the amount of M-A phase.

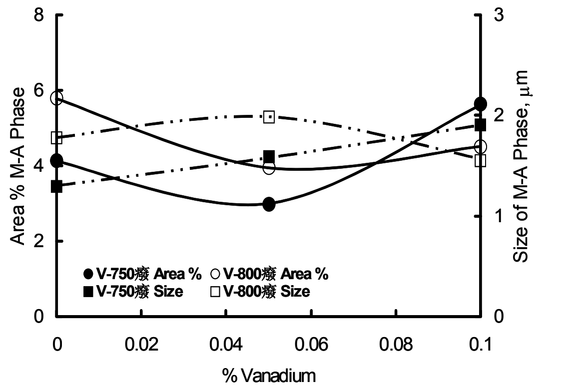


Figure 7 Effect of vanadium on the amount and size of M-A phase in HAZ's inter-critically reheated to 750°C and 800°C.

Figure 7 shows the effect of reheating to 750^0C and 800^0C a steel containing 0.1%C / 0.2%Si / 1.4%Mn / 0.005%P / 0.003%S / 0.039%Al / 0.005%N with 0.0%V, 0.05%V and 0.1%V, care having been taken to ensure that there were few, if any, additional components. From this Figure it is clear that vanadium up to 0.1% has had little or no effect on the amount, or size, of the M-A phase which had formed.

2.5 Precipitation Within The HAZ

In vanadium containing steels it is likely that any precipitation which occurs will be of VCN. In any give steel, such precipitation is likely to significantly depend on the vanadium and nitrogen levels and the cooling time of the weld. If it occurs it will, almost certainly, increase the hardness of the welded joint.

Figure 8(a) shows the effect of vanadium level on the change in as-welded hardness of steels containing typically 0.07%C – 1.6%Mn – 0.03%Al – 0.008%N, with $T_{(800)/500}$ = 55 secs (4.9kJ/mm). This shows that the hardness increased by approximately $25H_v$ / 0.1%V, reflecting the increased strength of the steel.

Figure 8(b) shows the effect of nitrogen and cooling time on the as-welded hardness for a steel containing 0.06%C – 1.4%Mn – 0.08%V – 0.03%Al. The effect of nitrogen on the HAZ hardness appears to be very similar at the cooling times investigated. This suggests that the increase in hardness due to increasing precipitation hardening, as a result of increasing the nitrogen content is of the order of $16H_v$ / 0.01%N, Once again, this reflects the higher strength expected in the steel.

The decrease in hardness observed with increasing heat input is probably a result of the changes in HAZ microstructure described above.

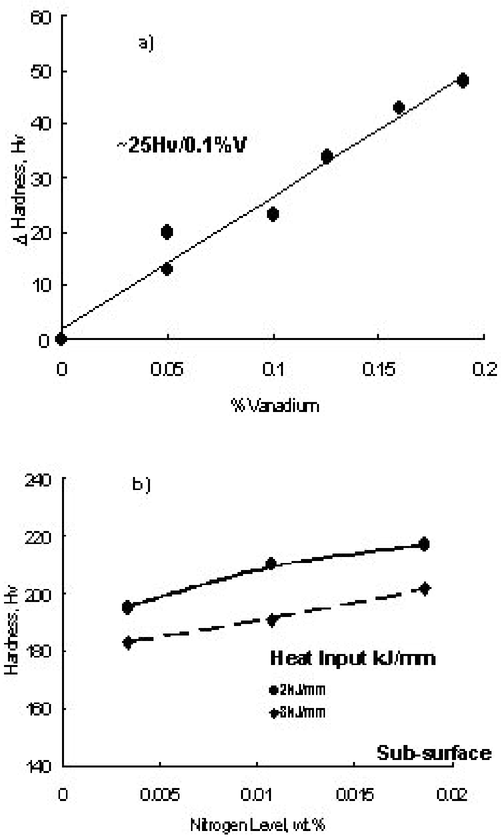


Figure 8 Effects of a) vanadium and b) nitrogen and heat input on the hardness of multi-pass welds.

3 Heat Affected Zone Toughness

3.1 Coarse Grained HAZ in Multi-Pass Welds

3.1.1 Effect of Vanadium

The effect of vanadium on the HAZ toughness of as welded, multi-pass welds, welded at 2kJ/mm ($\Delta T_{800-500}$ = 12 secs) in a steel containing 0.12%C – 1.6%Mn – 0.04%Al – 0.008%N is shown in Figure 9 a) and b).

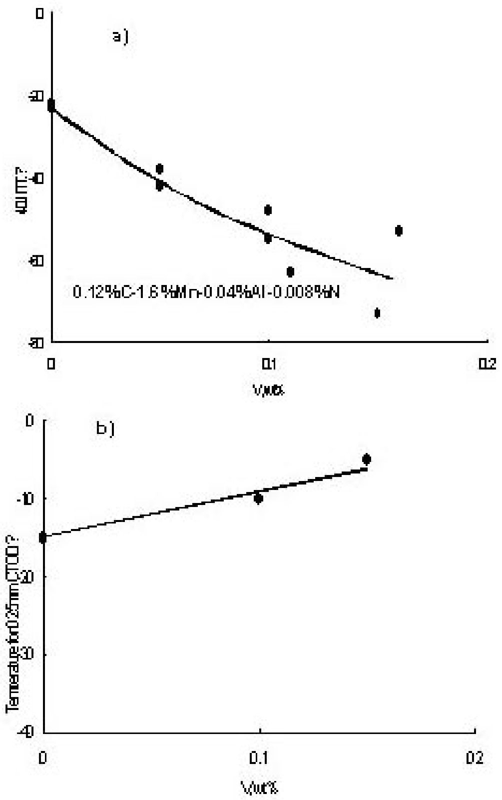


Figure 9 The effect of vanadium on a) the 40J ITT and b) the 0.25mm CTOD transition temperature, at constant CEV= 0.38, in multi-pass welds welds welded at 2kJ/mm ($t_{800/500}$ = 12 secs), as-welded. N.B. The CTOD transition temperature was a lower bound value.

Despite the increase in hardness noted in Figure 8(a) the addition of vanadium resulted in an improvement in the as-welded 40J impact transition temperature, from a level of –20°C at 0.0%V to a level of –60°C at 0.15%V. This improvement in Charpy vee-notch toughness was accompanied by little change in the level of the 0.25mmCTOD transition temperature, this latter parameter increasing from –15°C to –5°C for the same increase in vanadium content. This behaviour is considered to be due to the nucleation of intra-granular ferrite, the refinement of microstructure observed compensating for the increase in hardness.

3.1.2 Effects of Weld Heat Input (Cooling Time) and Nitrogen Level

Figure 10(a) shows the effects of weld heat input, in the range 2kJ/mm to 8kJ/mm ($t_{800/500}$ = 9 seconds to105 seconds), at two nitrogen levels (0.003%N and 0.018%N), on the 40kJ Charpy vee-notch impact transition temperature of 25 mm thick, butt welds, in a low carbon steel of base composition 0.06%C – 1.4%Mn – 0.08%V – 0.03%Al. At heat inputs up to 4kJ/mm (26 seconds) there was little or no effect of heat input on the weld impact transition temperature and it is only at heat inputs greater than this that any deterioration was observed. However, reflecting the higher strength of the base material and the higher hardness of the weld, the transition temperature of the 0.018%N steel was higher than that of the 0.003%N steel. At heat inputs less than 4kJ/mm (26 seconds) the difference was 25°C to 30°C, the impact transition temperature of the weld in the 0.003%N steel being –80°C, while that of the 0.018%N steel was –50°C to 55°C. With increase in heat input to 8kJ/mm (105 seconds) the weld ITT of the 0.003%N steel rose to –45°C while that of the 0.08%N steel rose to 25°C.

The effects of heat input and nitrogen level on the 0.1mm, lower bound, CTOD transition temperature of the 0.08%V steel, described above are shown in Figure 10(b). The main conclusion which can be drawn from this figure is that there is an increase in the weld CTOD transition temperature from a level of –70°C / -80°C to a level of –25°C / -45°C, as the heat input increased from 2kJ/mm to 4kJ/mm, at both levels of nitrogen. With further increase in heat input to 8kJ/mm the additional increase in CTOD transition temperature was 25°C in the case of the 0.003%N steel and 75°C in the case of the 0.018%N.

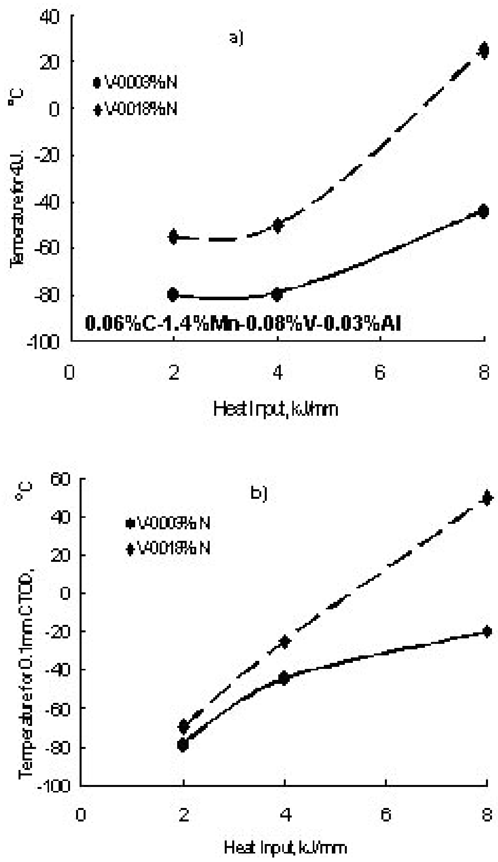


Figure 10 The effects of weld heat input and nitrogen level on a) the 40J ITT and b) the 0.1mm CTOD transition temperature of multi-pass welds in low carbon steel.

These results suggest that any effect of nitrogen on the toughness of the weld is dependant on the weld heat input and that at heat inputs up to at least 4kJ/mm (26 seconds) the effect of nitrogen is small. The results also suggest that at lower nitrogen levels, vanadium-containing steels may be suitable for welding at quite high heat input, in the present case up to at least 8kJ/mm.

3.2 Intercritically Reheated, Coarse Grained Heat Affected Zone (ICGCHAZ)

The effect of vanadium on the Charpy vee-notch impact and CTOD transition temperatures of a simulated, coarse grained (100 - 150μm), heat affected zone which had been intercritically reheated to 750^0C or 800^0C is shown in Fig. 11a) and b).

The base steel composition was 0.1%C – 1.4%Mn - -0.03%Al – 0.005%N, with vanadium levels of 0.0%, 0.5% and 0.1%V. In the figure, both transition temperatures have been plotted as a function of the hardness of the ICGC HAZ. As noted previously, care had been taken to minimise the effects of austenite stabilisers such as B, Cr, Mo, P etc. Also included in these results is data for a similar steel containing 0.03%Nb.

In the case of the impact transition temperature, inter-critically reheating to 750^0 – 800^0C resulted in an increase in I.T.T. of between 25^0C and 50^0C, irrespective of the level of vanadium. Similarly, the 0.1mm CTOD transition temperature increased by 30^0C to 100^0C.

These results indicate that the most important feature is the presence of the inter-critically reheated heat affected zone itself and that any effect of vanadium is of secondary importance. It will be recalled from Figs 6 and 7 that the main microstructural feature in the ICGC HAZ was the presence of brittle M-A phase and that the amount of M-A phase was relatively independent of the vanadium present in the steel, at least up to 0.1%V. Thus, on intercritically reheating the transition temperature increased irrespective of vanadium level, and the relatively small effect of vanadium itself, was almost certainly due to the increase in overall hardness (and strength) resulting from it's addition. Similar behaviour was noted in the case of niobium.

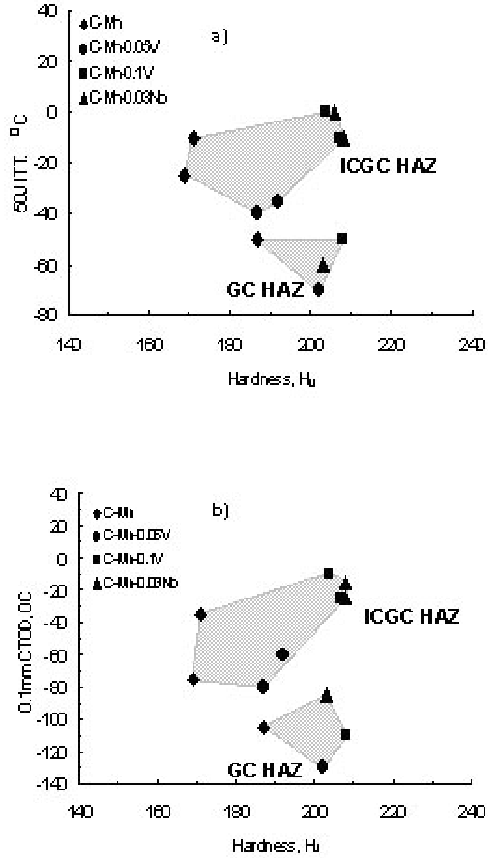


Figure 11 Effects of hardness and vanadium (and Nb) on a) the 50J ITT and b) the 0.1mm CTOD transition temperature of simulated welds, inter-critically reheated to 750°C and 800°C.

3.3 Laser Welded Steels

The effect of vanadium and nitrogen, expressed as a product, on the 27J I.T.T. of CO_2 laser welds, under two levels of power and welding speed is shown in Figure 12.

While all the transition temperatures were very good under the welding conditions examined, as the V.N. product increased from 0.0 up to 0.005, equivalent to a steel containing 0.1%V – 0.005%N, the I.T.T. improved from a level of -80 / -90^0C to a level of -100 / -110^0C. This broadly corresponds with the increase in the level of acicular ferrite shown in Figure 4.

Thus, just as tending to improve the heat affected zone microstructure and toughness in normal welds, vanadium can also improve the microstructure and toughness of CO_2 laser welds.

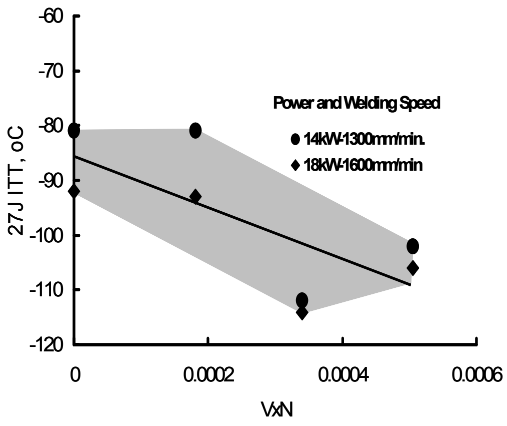


Figure 12 Effect of vanadium on the 27J ITT of CO_2 laser welds

4. Conclusions

The effects of micro-alloying with vanadium on weld HAZ microstructures and toughness have been considered and it has been demonstrated:

1. The austenite grain size in the coarse grained HAZ is between 100-200µm, depending on heat input.

2. Compared with C-Mn steel there is no significant effect of vanadium, up to 0.1%, on the temperature for start of transformation from → or on the range of temperature over which transformation occurs, in coarse grained HAZ's

3. Vanadium refines the HAZ microstructure in both multi-pass and laser welds by promoting the formation of intra-granular acicular ferrite

4. In the inter-critically reheated HAZ of steels where care has been taken to minimise the effects of austenite stabilisers such as B, P, Cr, Mo etc., vanadium, up to 0.1%, has little or nor effect on both the amount and size of M-A phase which forms.

5. Vanadium increases the hardness of the HAZ by 25Hv/0.1%V and nitrogen by 13.5Hv/0.1%N, both in line with the increases in strength arising from the additions.

6. The effects of the change in microstructure and hardness are such that in both multi-pass and laser welds increasing the vanadium level gives rise to an improvement in impact toughness accompanied, in the case of multi-pass welds, by only a small reduction in fracture toughness

7. Increasing both the nitrogen level and heat input of multi-pass welds increases the ITT and reduces the fracture toughness. However, at heat inputs up to at least 4kJ/mm, the ITT and fracture toughness were both satisfactory in a 0.08%V steel containing 0.003%N.

8. At 8kJ/mm, the ITT and fracture toughness were both satisfactory in a steel containing 0.08%V and 0.003%N.

9. In the inter-critically reheated HAZ, in the absence of austenite stabilisers, the main factor affecting both ITT and fracture toughness was the presence of the inter-critically reheated zone, itself. Vanadium, up to 0.1%, had little or no effect on toughness other than that arising from the increase in hardness, as a result of the use of a steel of higher strength.

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