EFFECT OF VANADIUM ON THE TOUGHNESS OF WELDS IN STRUCTURAL AND PIPELINE STEELS

ABSTRACT

There are few data in the literature on a systematic study of the influence of vanadium on HAZ toughness of structural and line pipe steels. An extensive programme of work on a series of steels containing up to 0.20% vanadium, with nitrogen in the range 0.005 - 0.018% and niobium additions of up to 0.05% was therefore carried out to evaluate the influence of vanadium additions on HAZ toughness and the results now presented.

Two pass welds, representative of the line pipe seam welds, made at a heat input of 4.9kJ/mm and multipass butt welds, representative of welds in structural steels, made at 2 and 5kJ/mm were made in plate produced from laboratory heats of steel. The HAZ toughness was assessed, in both the as welded and post weld heat treatment conditions, using both Charpy and CTOD transition curves for specimens notched at the fusion boundary. In addition relevant hardness data was produced and accompanying metallographic examination was carried out to observe microstructural changes.

It was found that, particularly in the as-welded condition, substantial additions of vanadium could be tolerated with no adverse effect on the HAZ toughness despite an increase in hardness, indeed, in some instances HAZ toughness was observed to significantly improve with vanadium additions up to 0.15%. It was seen that the vanadium additions promoted the development of a finer transformed microstructure in the HAZ. This beneficial change in the microstructure is believed to be the principal reason for the observed good HAZ toughness behaviour.

INTRODUCTION

Although both vanadium and niobium are used as microalloying elements in the production of linepipe steels, niobium is perhaps the more common microalloying element to be found in structural and pressure vessel steels. Partly this arises because some manufacturing[1,2] specifications apply quite low maximum limits to the vanadium content. It would seem that the origin of these low maximum contents stems from experience of the behaviour of older higher carbon, high nitrogen steels untypical of modern day concentrations of these elements. Since there is an extreme lack of relevant data on the influence of vanadium on HAZ toughness in modern lower carbon, lower nitrogen steels, a programme of work has been carried out at TWI on the HAZ properties of welds in current vanadium containing steels. This paper presents and summarises the considerable body of data which has been obtained from two pass welds representative of pipe production and from multipass welds representative of structural and pressure vessel welding.

EXPERIMENTAL PROCEDURE

Materials

All the steels studied were produced as small scale (50kg) laboratory heats and rolled to approximately 25mm (1 in) plate. Although most of the study was carried out on steels with 0.12% carbon, some two pass welds were made on higher (0.22%) carbon steels and some two and multipass welds were made on lower (~0.07%) carbon steels to help study the influence of this element. They were supplied in the hot rolled condition which was considered acceptable since the parent plate microstructure does not exert a significant influence on the high temperature HAZ microstructure. The compositions of these steels are given in Table 1 for two pass pipe welds and in Tables 2 and 3 for the multipass weld studies. In general it will be seen that close control of composition was maintained with the exception of intended variations in microalloy element content. The carbon equivalent was calculated from the IIW formula without the V term as it was intended to study the individual effect of this element (see below). The impurity and residual alloying element levels were in the range: 0.009 - 0.014%S, 0.006 - 0.010%P, 0.02 - 0.04%Ni, 0.01 - 0.02%Cr, <0.01%Mo, 0.01 - 0.03%Cu, <0.001%Ti and <0.0003%B.

Welding

Two pass welds

Plates were given a double-vee preparation, as shown in Fig.1a, and submerged arc welded with tandem wire welding using a 1.5Mn, 0.5Mo wire in conjunction with a basic flux. The first pass was made at an arc energy of 4.0kJ/mm (100kJ/in) and the second with an arc energy of 4.9kJ/mm (125kJ/in), the latter giving a cooling time of about 55 seconds between 800 and 500°C, details are given in Fig.1b. Charpy transition curves were produced using specimens taken from the position shown in Fig.1c, from the second side weld with the notch located so that it crossed the fusion boundary at its mid width. Thus both weld metal, HAZ and some parent material were sampled. For CTOD testing, surface notched specimens, with an a/w ratio of 0.3 and with the notch tip located in the grain coarsened HAZ of the second side weld were taken as shown in Fig.1c, and tested to produce transition curves.

In addition to producing data in the as-welded condition, tests were also carried out on welds given a postweld heat treatment (PWHT) of 1hr at 600°C. While such a heat treatment would not usually be given to pipe seam welds in practice, it was carried out as part of a general study of the influence of vanadium on HAZ toughness.

Multipass welds

A single bevel joint preparation with a straight side was used for these welds and the material to be tested was on the vertical face of the weld allowing Charpy and CTOD specimens to sample only HAZ close to the fusion boundary. The weld bead positioning was controlled to ensure a realistic proportion of grain coarsened HAZ remained in the completed weld. Welding was by submerged arc using a basic flux and a 1.5%Mn wire at a heat input of 2kJ/mm (50kJ/in), details are shown in Fig.2. This heat input produces an 800-500°C cooling time of approximately 12 secs which is comparable to that of ~16 secs produced at 3kJ/mm (75kJ/in) on 50mm (2 in) material. Charpy tests were conducted on specimens with a notch located as shown in Fig.2 and tested to generate a transition curve. Full plate thickness, square section CTOD specimens, notched through the thickness at the fusion boundary were also carried out so as to produce transition curve behaviour. For all steels, welds were tested in the as-welded condition, while selected welds were additionally studied after PWHT of 1 hour at 600°C.

Certain steels were also welded at a higher heat input of 5kJ/mm. This was done in conjunction with a thicker plate of 50mm section thickness, giving a $t_{8/5}$ of 27 seconds. Since the experimental steel was only able to be obtained in 25mm thickness, this was achieved by making a "sandwich" panel by bolting the 25mm experimental steel to a carbon manganese steel of 25mm thickness, after machining the two butting faces to produce a close fit, details are shown in Fig.3. Toughness testing was done in a similar manner to that of the 2kJ/mm welds, taking samples from the experimental steel part of the weld.

It should be noted that, for all the weld procedures examined, in both two pass and multipass welds, the interpretation of Charpy impact results was based on an average transition curve. In the case of the CTOD data the lower bound was taken. In both cases about 10 specimens were used to produce a transition curve.

Metallographic examination and hardness testing

In addition to the toughness testing, transverse metallographic sections were taken from welds to examine the HAZ microstructure and to measure the hardness using a Vickers method. The indentation load on the two pass and on the 5kJ/mm multipass welds was 10kg. For the 2kJ/mm multipass welds the load was 5kg, except where noted. For the two pass and multipass welds the HAZ indents were made within one indent from the fusion boundary and for the two pass welds the weld metal indents were made 5mm from the fusion boundary.

Metallographic examination was also carried out on the CTOD specimens, after testing, to examine the region sampled by the fatigue crack and, in those specimens showing brittle fracture, to determine the fracture initiation point.

RESULTS AND DISCUSSION

Two Pass Welds ($\Delta t_{8/5}$ ~ 55 sec)

The results of the Charpy testing are given in Table 1 in terms of 27 and 40J transition temperatures. A multiple linear regression analysis was carried out regressing the 40J transition temperature against the IIW carbon equivalent, vanadium, niobium and nitrogen. It should be noted that the vanadium term in the IIWCE formula was omitted to enable the individual effect of vanadium to be determined. Similar regression analyses were carried out for weld metal and HAZ hardness. The resulting vectors are given in Table 4.

Considering first of all the as-welded results it can be seen (Fig. 4a) that, as expected, increasing CEV, in this study principally carbon, significantly raised the 40J transition temperature. In order to plot the effect of the microalloying elements, as in Fig. 4b, the vector for CEV shown in Table 4 was used to correct the transition temperature of each steel to a common carbon equivalent of 0.34, enabling all the steels of differing carbon content to be compared. It can be seen from Fig. 4b that there was no significant effect of plate vanadium content of up to 0.20%, on the HAZ 40J transition temperature. This is in contrast to the trend of hardness in both the weld metal and HAZ (see Table 1 and Fig.5a) which showed a significant increase. Usually an increase in hardness, if this is not accompanied by an increase in microstructural refinement, will result in a deterioration of toughness. It was found that the addition of vanadium produced a beneficial change in the grain coarsened HAZ microstructure. This arose from the tendency to promote intragranular decomposition of the austenite, thereby reducing the effective grain size of the overall transformed product. This is shown in Fig.6. It is considered that this beneficial microstructural effect balanced any deleterious decrease in toughness which may have accompanied the observed increase in hardness.

Increasing the nitrogen content of, or adding 0.03% niobium to, the 0.05 and 0.10% vanadium steels had little or no effect on the HAZ 40J transition temperature within the range of results obtained in Fig.4b. Indeed, it can be seen from Table 1 that reducing the carbon from 0.07 to 0.04% ensured that the HAZ toughness was maintained when the combined V + Nb content was raised to 0.15%.

Turning to the effect of PWHT on the Charpy toughness, this is indicated in Table 4 and in Fig. 7. Again the expected adverse effect of increasing carbon and CEV is apparent, the vector being slightly larger than for the as-welded condition. The data also show that the vectors for vanadium, niobium and nitrogen are all significantly positive, being greatest for nitrogen. However, it can also be seen that for vanadium levels of less than about 0.010% the 40J transition temperature is improved by PWHT.

The changes in toughness on PWHT broadly followed the trends in the change in hardness after PHWT, where Fig.5b and Table 1 show that there was an increase in HAZ and weld metal hardness of the microalloyed steels following PWHT.

The results of the fracture toughness testing of the HAZ are presented in Table 1 in terms of the temperature for a CTOD of 0.25mm and the CTOD at -10°C. It can be seen from Fig. 8 that compared to the base C:Mn steel, an initial vanadium addition of 0.05% produces a substantial improvement in fracture toughness. Although this improvement is less at higher levels of vanadium, even at the highest level of vanadium studied (0.20%), the toughness was at least as good as that of the base C:Mn steel. The addition of 0.03% niobium at 0.07% carbon, and 0.05% niobium at 0.04% carbon to steels containing 0.10% vanadium had no significant effect on the as-welded fracture toughness behaviour. Following PWHT for the two steels in which this was studied, 0.05% and 0.20% vanadium, the trend was for the toughness to decrease (Table 1). The changes in fracture toughness are believed to arise from the net effect of the observed beneficial change in transformed microstructure and the tendency for hardness to rise with increasing microalloy content and following PWHT.

Multipass Welds

25mm thick, 2kJ/mm arc energy ($\Delta t_{8/5}$ ~ 12 secs)

The results of the Charpy testing are given in Table 2. Regression analyses were conducted in an analagous manner to that described for the two pass welds and the resultant vectors given in Table 4. As for the two pass welds in order to plot the effect of the microalloying elements, as in Fig. 9b, the vector for CEV established in Table 4 was used to correct the transition temperature of each steel to a common carbon equivalent, in this case 0.38. Again this allowed all the steels of differing carbon and manganese contents to be compared. Table 4 shows that reducing CEV improves HAZ toughness while Fig. 9a shows shows a beneficial effect of increasing vanadium content in the as-welded condition. Although not shown in Fig. 9 it can be seen from Table 2 that the beneficial effect of vanadium is greatest in the lower carbon steels studied and that improvement was still occurring at the highest level (0.15%) studied. This trend was not significantly altered by the addition of nitrogen to the 0.1% vanadium steels. The addition of 0.03% niobium to a 0.07% vanadium steel produced no significant effect on 40J transition temperature in the 0.12% carbon material. Further, the addition of 0.05% niobium to the 0.04% carbon, 0.10% vanadium steel still allowed excellent levels of HAZ toughness to be obtained.

In the PWHT condition (Fig. 9b), the effect of CEV seemed more pronounced, as in the two pass welds, and the trend with vanadium was the reverse of that in the as-welded condition, the level of toughness decreasing with increasing vanadium content. However, PWHT resulted in an improvement in joint toughness for steels containing up to about 0.1% vanadium at 0.38 CEV. Moreover, even at the highest vanadium level examined (0.15%) the 40J transition temperature was reasonable at about -30°C after PWHT. Little or no effect of nitrogen was observed, somewhat in contrast to the two pass results. In the case of the steels containing 0.07% vanadium + 0.03% niobium, the 40J transition temperature after PWHT was some 20-30°C higher than those of the equivalent steels.

Figures 10a and 10b show the trend of HAZ hardness with vanadium content; that in both as-welded and PWHT conditions, increasing vanadium content increased the HAZ hardness. However, even at 0.16%V the hardness after PWHT was only 285 HV5. For lower levels of vanadium, (<0.1%) particularly in the 0.12% carbon materials, PWHT resulted in a net reduction in HAZ hardness.

The influence of vanadium content on the fracture toughness of the as-welded grain coarsened HAZ is shown in Fig.11a where it can be seen that increasing vanadium up to 0.15% gave rise to a small reduction in the fracture toughness. The level of fracture toughness in the vanadium containing steels was similar to, but marginally better than, that in a comparative niobium microalloyed material. The addition of 0.03% niobium to a 0.07% vanadium steel produced a lower level of toughness than in a straight vanadium steel. It was found that decreasing the carbon content from 0.12 to 0.07% in the 0.10% vanadium steel produced a marked improvement in as-welded toughness. A similar excellent level of toughness was observed in the case of the 0.04% carbon, 0.1% vanadium, 0.05% niobium steel.

The application of PWHT to the 0.12% carbon vanadium-containing steels produced a substantial improvement in fracture toughness (Fig. 11b) and this was also true of the application of PWHT to niobium and niobium-vanadium materials. For the low carbon, 0.1% vanadium steel, PWHT resulted in little or no change in fracture toughness which was already at a very acceptable level. Increasing the nitrogen content to 0.017% had little effect on the toughness, in both the as-welded and PWHT conditions for the 0.12% carbon, 0.1% vanadium steel. However, for the lower carbon steel, the higher nitrogen content resulted in lower fracture toughness in both the as-welded and PWHT conditions, although PWHT did result in an improvement in the temperature for 0.25mm CTOD of ~20°C, giving a transition temperature of -40°C.

As was observed in the two pass welds, the addition of vanadium produced some modifications to the grain coarsened HAZ microstructure. Possibly the most important of these was the tendency to promote intragranular decomposition of the austenite, reducing the extent and particularly the size of the colonies of ferrite with aligned second phase. This microstructural change is considered to be beneficial for improving cleavage resistance. As was noted earlier above, increasing vanadium also gave rise to an increase in hardness and in terms of the measured fracture toughness, these two features are in opposition to each other.

50mm thick, 5kJ/mm arc energy ($\Delta t_{8/5}$ ~ 27 secs)

The HAZ toughness data and hardness measurements for these welds are presented in Table 3. The effect of increasing heat input on the Charpy behaviour of the HAZ is shown in Fig.12. For all the steels investigated, this shows the traditional effect in that increasing heat input gives rise to a fall in Charpy toughness.

Considering the change in toughness following PWHT and how this is influenced by heat input, it can be seen that for all the steels studied, a higher heat input gave rise to small improvements in the 40J transition temperature. This is in contrast to the general trend at low heat inputs of a tendency for postweld heat treatment to raise the 40J HAZ transition temperature at the level of vanadium under consideration (ie 0.07% and 0.10%). This trend may be associated with a greater likelihood for some precipitation hardening to occur during the longer cooling times of the higher heat input process, thereby reducing the potential for precipitation hardening during PWHT and the often associated deterioration in toughness. Figure 13 shows that the general level of hardness of the higher heat input welds is lower than that of the low heat input welds, as would be expected, although for the combined microalloyed material the hardness is the same at both heat inputs. This is likely to arise from the greater precipitation hardening produced by the combined microalloy content. However, it should be noted that at the higher heat input the slower cooling will also produce a softer and coarser transformed microstructure, and the final HAZ hardness is a balance of the degree of tempering and any precipitation hardening occurring.

Turning now to the fracture toughness data, Fig.14, the trends with heat input depend on the steel type. In the as-welded condition, the best fracture toughness was achieved by both the 0.1% vanadium steel, and the 0.03% niobium material. In contrast, the vanadium:nitrogen material produced a poorer fracture toughness at the higher heat input. The reason for these different trends is believed to be associated with the influence of the different compositions on the development of the microstructure in the intercritically reheated grain coarsened HAZ. Other work [7] has shown that when the decomposition of the carbon enriched austenite formed during intercritical reheating tends towards ferrite and carbide, rather than martensite:austenite (MA), the toughness improves, other factors remaining constant. In general, the slower cooling associated with a higher heat input will tend to favour the formation of ferrite:carbide rather than MA, depending upon the precise overall composition and hence hardenability of the carbon-enriched austenite. The microstructure examination indicated that in the niobium and vanadium steels, at higher heat input, there was less MA constituent and more ferrite:carbide. However, in the vanadium:nitrogen material, more MA was observed at the higher heat input, presumably as a result of nitrogen enhancing the steel hardenability.

Considering the fracture toughness behaviour following PWHT, Fig.14b shows that again the trend is very dependent on the material type. For both the 0.03% niobium and 0.10% vanadium steels, PWHT produced a small reduction in toughness, while for the combined microalloyed material, the postweld heat treatment exhibited a marked improvement in toughness. Previous work has shown that when the as-welded fracture initiation is predominantly from the ICGC HAZ, which contains MA, PWHT usually produces a significant improvement arising from the tempering of the MA regions. This is probably the principal contributing factor to the improvement in toughness of the niobium:vanadium material. In contrast, both the 0.03% niobium and 0.10% vanadium steels, which may have contained a smaller proportion of MA in the ICGC HAZ, were observed to show a deterioration in toughness, presumably because the potential for improvement from the tempering of MA was significantly less. The decrease in toughness was small in the niobium steel but larger in the vanadium steel. Nevertheless the temperature for 0.25mm CTIOD of -15°C for the vanadium steel is quite satisfactory for most applications.

Practical Implications

The results presented in the previous sections demonstrate that, particularly in the as-welded condition, the presence of a significant level of vanadium does not have an adverse effect on the HAZ toughness. Indeed, under many circumstances it is shown that significant improvement in HAZ toughness can arise. Similar findings have been reported recently in several different studies including both experimental and commercial steels[3-6].

Specifically, for two pass welds, it was found that there was no effect on the 40J transition temperature of adding up to 0.20% vanadium and that the best fracture toughness in the grain coarsened HAZ was at 0.05% vanadium, while at 0.20% vanadium the fracture toughness was still similar to that of the base carbon manganese steel, despite in all cases increases in HAZ and weld metal hardness. In multipass welds there was a continuous improvement in the as-welded 40J transition temperature for additions of up to 0.16% vanadium (the highest studied), with only a very small fall in fracture toughness up to this level of vanadium, despite significant rises in HAZ hardness. For both the two pass and multipass weld studies, it was found that the tolerance of the vanadium steels to niobium additions in terms of an adverse effect on the HAZ toughness was improved by lowering the carbon content, such that at 0.04% carbon there was no deterioration when 0.05% niobium was added to a 0.10% vanadium steel.

In the present work it has been found that alloying with vanadium can produce a beneficial change in the transformed microstructure of the grain coarsened HAZ. This change was a notable tendency to promote intragranular decomposition of the austenite which resulted in an effectively smaller grain or colony size of the ferrite with aligned second phase, or bainitic products, typically found in carbon:manganese steels. This tendency for a finer grained transformed microstructure in the grain coarsened HAZ region is believed to be beneficial for cleavage resistance and to be the principal factor contributing to the observed toughness behaviour.

The situation is a little different when PWHT is considered since this can sometimes produce a net increase in hardness with some associated fall in toughness. Nevertheless, at low carbon levels (0.07%) alloying with up to 0.1%V produces no loss in toughness in multipass welds at 2kJ/mm. When fracture toughness data are considered it was found that in 0.12% carbon steels PWHT produced significant improvement in toughness, even at the 0.15% vanadium level. In structural steel applications there is a significant move to try and reduce costs by avoiding PWHT whenever it is possible to demonstrate adequate as-welded toughness for this to be so. The beneficial effects of vanadium, demonstrated in this study, particularly in the as-welded condition are of obvious considerable significance in respect of this trend.

Overall the data reviewed shows a substantially different picture to that often portrayed in the past concerning the adverse effect of vanadium on HAZ toughness. They show that under many circumstances there are significant advantages to be gained in respect of HAZ toughness by alloying with vanadium, particularly in the as-welded condition. The data indicate considerable scope for development of vanadium-containing low carbon:manganese steels having good levels of HAZ toughness.

CONCLUSIONS

1. For welds typical of the seam weld in linepipe, no deleterious effect of up to 0.20% vanadium was found on the as-welded HAZ toughness. The addition of 0.03% niobium to vanadium-containing materials also had little or no effect on the HAZ toughness.

2. In multipass butt welds at an arc energy of 2kJ/mm, the addition of up to 0.15% vanadium significantly improved the as-welded HAZ Charpy notch toughness and produced only a small decrease in the HAZ fracture toughness, despite significant increases in HAZ hardness. The addition of 0.03% niobium resulted in a decrease in HAZ toughness which was more marked for fracture toughness than for the 40J Charpy transition temperature.

3. For the multipass butt welds made at 2kJ/mm and given a postweld heat treatment of 1 hour at 600°C, the HAZ fracture toughness of 0.12% carbon steels significantly improved, while the 40J Charpy transition temperature improved for vanadium contents of up to 0.1% at 0.38 CEV.

4. Reducing the carbon content of vanadium-containing steels resulted in an increase in the ability of the steel for both two pass and multipass welds to tolerate vanadium plus niobium additions of up to 0.15% in a 0.04% carbon steel.

5. In high heat input (5kJ/mm) multipass welds the HAZ toughness of vanadium (0.1%) and niobium (0.03%) steels were similar, while vanadium plus niobium and vanadium plus higher nitrogen steels tended to give a lower level of HAZ toughness. In the case of the vanadium plus niobium steel this lower level of fracture toughness was recovered by PWHT. For the niobium and the vanadium steel decreases in toughness (small in the former, larger in the latter) were seen after PWHT. Nevertheless the fracture toughness of the vanadium steel was still quite satisfactory (0.25mm at -15°C). The effect of PWHT on the vanadium nitrogen steel was not examined.

6. Microstructural examination of the grain coarsened HAZ showed that the addition of vanadium promoted intragranular decomposition from austenite, producing a consequent reduction in

bainitic colony size, thereby providing a beneficial microstructural change. This is believed to be the principal cause of the general trend found for vanadium to either have no adverse effect or on some occasions a beneficial effect on HAZ toughness.

7. With the trend in structural steel welding to eliminate the need for PWHT by demonstrating adequate as-welded toughness, the present results indicate that the ability of vanadium-containing steels to produce good levels of HAZ toughness, particularly in the as-welded condition, make them particularly suitable candidates for this approach.

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									Hardness Hv10, (max)			ITT, °C		CTOD		
C	Si	Mn	V	Nb	Al	N	O	CEV		HAZ	WM	27J	40J	T°C for 0.25mm	mm at -10°C	Fracture initiation site
0.07	0.31	1.67	-	-	0.02	0.0088	0.004	0.337	AW	179	206	-47	-30	-8	0.05	GCHAZ
						(0.019)			PWHT	164	187	-75	-65			
0.07	0.31	1.48	0.05	-	0.024	0.0087	0.006	0.332	AW	192	217	-24	-12	-55	0.71	GCHAZ
			(0.03)			(0.017)			PWHT	202	240	-32	-24	-30	0.7	GCHAZ
0.07	0.4	1.58	0.05	-	0.029	0.0113	0.006	0.339	AW	199	227	-36	-15			
			(0.03)			(0.0158)			PWHT	206	239	-66	-15			
0.06	0.41	1.49	0.05	-	0.038	0.0177	0.006	0.314	AW	190	215	-32	-13			
			(0.03)			(0.0158)			PWHT	202	238	-47	-25			
0.07	0.32	1.51	0.1	-	0.025	0.0083	0.007	0.33	AW	202	225	-47	-34	-15	0.53	GCHAZ
			(0.07)			(0.0168)			PWHT	212	276	-42	-29			
0.06	0.33	1.56	0.16	-	0.025	0.0083	0.007	0.326	AW	213	220	-31	-16	5	0.04	GCHAZ
			(0.10)			(0.0083)			PWHT	235	228	-70	-42			
0.07	0.4	1.59	0.16	-	0.031	0.0122	0.006	0.343	AW	222	232	-27	-8			
			(0.11)			(0.0161)			PWHT	232	252	-27	-8			
0.06	0.31	1.56	0.19	-	0.038	0.0079	0.006	0.326	AW	227	242	-53	-34	-15	0.4	GCHAZ
			(0.12)			(0.0085)			PWHT	233	256	-46	-22	15	0.01	GCHAZ
0.07	0.31	1.54	0.05	0.03	0.026	0.0083	0.007	0.335	AW	198	219	-39	-20	-45	0.4	GCHAZ
			(0.03)	(0.018)		(0.0144)			PWHT	212	240	-26	-8			
0.07	0.31	1.61	0.16	0.032	0.026	0.0099	0.006	0.346	AW	215	228	-35	-12	-20	0.3	GCHAZ
			(0.10)	(0.018)		(0.0135)			PWHT	254	258	-8	17			
0.04	0.31	1.51	0.1	0.048	0.019	0.0086	0.007	0.292	AW	210		-45	-30	-15	0.45	GCHAZ

Table I. Compositions and properties of two pass pipe welds (4.9kJ/mm)

									Hardness Hv10, (max)			ITT, °C		CTOD		
C	Si	Mn	V	Nb	Al	N	O	CEV		HAZ	WM	27J	40J	T°C for 0.25mm	mm at -10°C	Fracture initiation site
0.11	0.4	1.6	-	0.026	0.053	0.008	0.006	0.335	AW	216	228	-15	4	20	0.054	GCHAZ
									PWHT	219	224	-40	-22			
0.12	0.3	1.54	0.05	0.03	0.042	0.0087	0.005	0.384	AW	212	228	-37	-14			
			(0.03)	(0.017)		(0.0133)			PWHT	235	258	-23	0			
0.12	0.31	1.59	0.15	0.03	0.042	0.0091	0.006	0.394	AW	228	238	-27	5			
			(0.09)	(0.010)		(0.0089)			PWHT	266	264	6	56			
0.23	0.37	1.47	-	-	0.036	0.0081	0.005	0.484	AW	229	233	-21	11	-20	0.12/ 1.35	GCHAZ
						(0.0081)			PWHT	182	213	-42	-27			
0.23	0.37	1.6	0.05	-	0.041	0.0084	0.005	0.504	AW	227	243	6	37			
			(0.03)			(0.0083)			PWHT	237	247	-29	-1			
0.22	0.29	1.64	0.1	-	0.037	0.0087	0.008	0.502	AW	260	252	-34	4			
			(0.06)			(0.0095)			PWHT	251	252	-34	4			
0.22	0.28	1.57	0.15	-	0.036	0.0085	0.005	0.491	AW	264	247	-13	27			
			(0.10)			(0.0103)			PWHT	274	268	13	52			
0.22	.0.3	1.61	0.2	-	0.044	0.0082	0.005	0.498	AW	266	252	-10	23			
			(0.12)			(0.0081)			PWHT	268	254	4	47			

Table I (continued). Compositions and properties of two pass pipe welds (4.9kJ/mm)

										Hardness Hv5, (max)	ITT, °C		CTOD		
C	Si	Mn	V	Nb	Al	N	O	CEV			27J	40J	T°C for 0.25mm	mm at -10°C	Fracture initiation site
0.07	0.38	1.63	-	-	0.031	0.0046	0.004	0.351	AW	212	-45	-30			
									PWHT	192	-95	-80			
0.07	0.3	1.58	0.05	-	0.032	0.006	0.006	0.342	AW	206	-65	-50			
									PWHT	223	-80	-65			
0.07	0.26	1.53	0.1	-	0.037	0.0068	0.005	0.333	AW	227	-70	-70			
									PWHT	225	-60	-43			
0.08	0.25	1.54	0.11	-	0.026	0.0065	0.005	0.345	AW	234	-70	-70	-89	0.45	SCGC/ICGCHAZ
									PWHT	265	-60	-43	-70	0.55	SCGCHAZ
0.07	0.24	1.48	0.1	-	0.034	0.0171	0.005	0.322	AW	252	-80	-50			
									PWHT	238	-85	-46			
0.08	0.32	1.6	0.1	-	0.025	0.0166	0.007	0.357	AW	262	-80	-50	-21	0.45	GC/SCGCHAZ
									PWHT	267	-85	-46	-40	0.45	GC/SCGCHAZ
0.07	0.46	1.7	0.15	-	0.034	0.0051	0.005	0.363	AW	260	-85	-75			
									PWHT	262	-40	-30			
0.04	0.29	1.51	0.11	0.048	0.018	0.0078	0.008	0.293	AW	247*	-70	-67	-70	1.1	SCGCHAZ
0.13	0.35	1.68	-	-	0.026	0.0077	0.007	0.414	AW	254	-30	-13	-15	0.35	GCHAZ
									PWHT	223	-60	-41	-68	0.7	GCHAZ
0.12	0.28	1.57	0.05	-	0.035	0.0068	0.005	0.391	AW	257	-55	-35			
									PWHT	246	-60	-35			
0.12	0.38	1.66	0.1	-	0.038	0.0075	0.005	0.404	AW	265	-65	-43	-10	0.27	ICGCHAZ
									PWHT	265	-50	-30	-52	0.72	GC/GRHAZ
0.13	0.34	1.6	0.1	-	0.029	0.017	0.006	0.407	AW	286	-80	-49	-23	0.33	GC/ICGCHAZ
									PWHT	289	-50	-27	-45	0.57	ICGCHAZ
0.12	0.34	1.63	0.16	-	0.035	0.0072	0.005	0.401	AW	289	-70	-47	-2	0.044	GC/ICGCHAZ
									PWHT	296	-50	-29	-38	0.75	ICGCHAZ
0.13	0.34	1.54	-	0.028	0.019	0.0065	0.004	0.395	AW	257	-63	-43	-5	0.2	SCGC/ICGCHAZ
									PWHT	229	-49	-40	-43	0.4	SCGC/ICGCHAZ
0.12	0.34	1.56	0.07	0.029	0.021	0.0067	0.005	0.386	AW	252	-60	-33	8	0.17	GC/ICGCHAZ
									PWHT	239	-50	-3	-52	0.56	GC/ICGCHAZ

Table II. Compositions and properties of multipass welds (2kJ/mm)

*Hv10

C	Si	Mn	V	Nb	Al	N	O	CEV		Hardness Hv10 (max)	ITT, °C		CTOD		
											27J	40J	T°C for 0.25mm	mm at -10°C	Fracture initiation site
0.12	0.31	1.59	0.1	-	0.028	0.0065	0.004	0.41	AW	235	-39	-24	-54	0.46	SCGC/ICGCHAZ
0.12	0.34	1.47	0.1	-	0.025	0.0062	0.006	0.39	PWHT	225	-50	-39	-15	0.3	SCGC/ICGCHAZ
0.11	0.28	1.52	0.1	-	0.025	0.0155	0.005	0.368	AW	247	-16	4	8	0.12	SCGC/ICGCHAZ
0.12	0.33	1.54	-	0.028	0.028	0.0062	0.004	0.382	AW	222	-15	-14	-49	0.38	SCGC/ICGCHAZ
									PWHT	224	-45	-30	-41	0.34	SCGCHAZ
0.12	0.24	1.57	0.07	0.025	0.035	0.0088	0.005	0.386	AW	247	-35	0	-5	0.07	SCGC/ICGCHAZ
									PWHT	249	-30	-8	-60	0.31	SCGC/ICGCHAZ

Table III. Composition and properties of steels used for multipass welds (5kJ/mm)

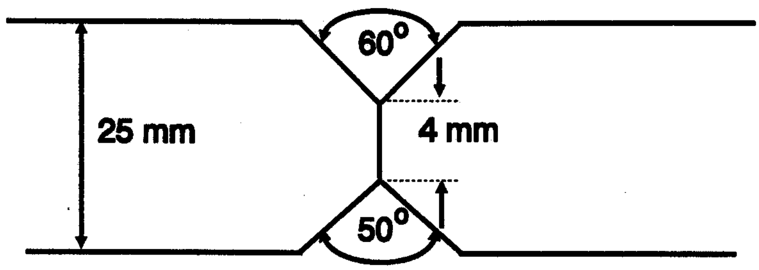
40J ITT					
Procedure	Condition	CEV	Vanadium	Niobium	Nitrogen
2kJ/mm Multipass	AW	225.1	-212.2	20.2	490
	PWHT	324.9	235.7	953.6	-44.3
5kJ/mm Two pass	AW	264.8	5	79.4	1,810.1
	PWHT	317.1	292.2	1,244.1	2,978.2

Maximum HAZ hardness					
Procedure	Condition	CEV	Vanadium	Niobium	Nitrogen
2kJ/mm Multipass	AW	535.1	259	613.2	2,252.6
	PWHT	467.1	442	88.7	1,628.3
5kJ/mm Two pass	AW	265.3	190.4	150.3	-156
	PWHT	205.9	341.9	833.8	1,405.9

Maximum weld metal hardness					
Procedure	Condition	CEV	Vanadium	Niobium	Nitrogen
5kJ/mm Two Pass	AW	225.4	166.9	6.9	171.6
	PWHT	259.5	385.2	738.1	3,944

Table 4. Vectors obtained from multiple regression analysis

a)



JOINT PREPARATION

Wire:	1.5%Mn, 0.5%Mo, 4 mm diameter
Flux:	highly basic (dried for 2 hrs at 450 °C)
Arc condition	
1st Pass:	
lead arc	650A, 30V, DC
trail arc	840A, 40V, AC
travel speed	800 mm/min
arc energy	4.0 kJ/mm
2nd Pass:	
lead arc	950A, 30V, DC
trail arc	840A, 40V, AC
travel speed	760 mm/min
arc energy	4.9 kJ/mm
Restraint:	Strong backs

b)

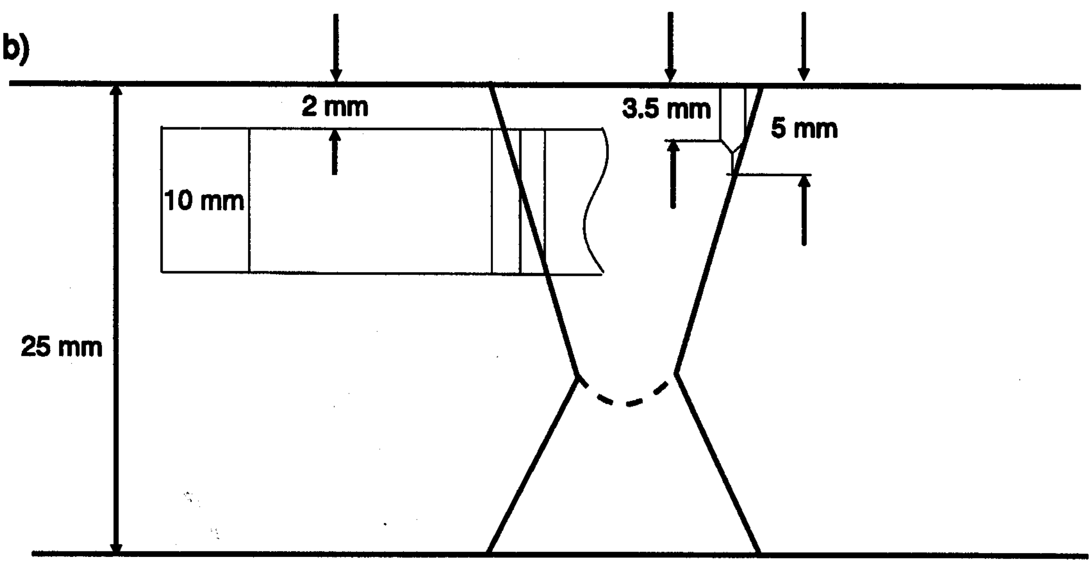
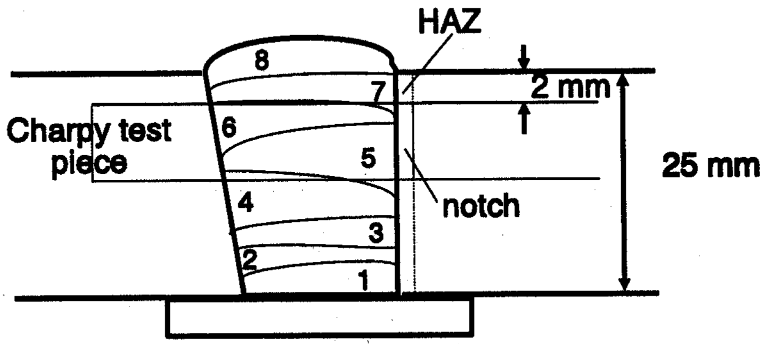


FIGURE 1 a) WELD PROCEDURE FOR TWO PASS SAW WELDS (5kJ/mm)
b) POSITION OF NOTCHES IN CHARPY AND CTOD SPECIMENS

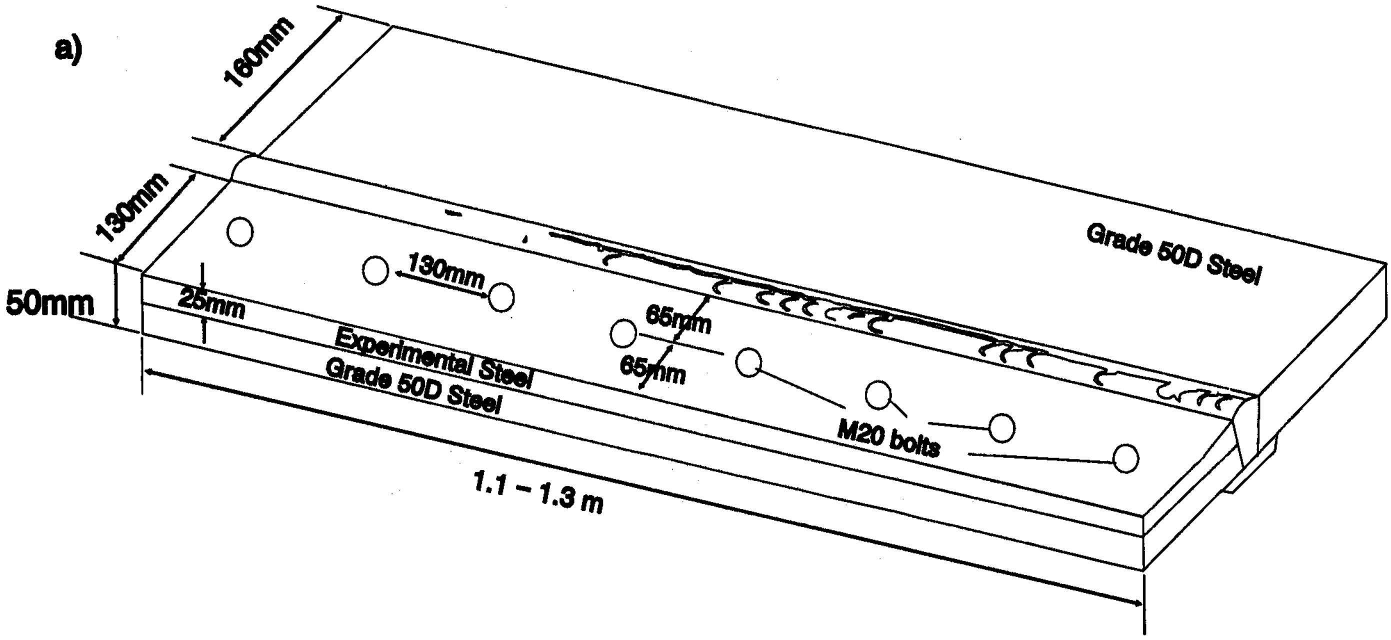


BEAD DEPOSITION

Welding preparation:	Single bevel with backing bar Included angle – 10° Root gap – 10 mm
Consumables:	Flux – semi basic Wire – 3.2 mm dia. 1.5% Mn
Welding conditions:	
Current	– 700A
Voltage	– 30V, DC +
travel speed	– 630 mm/min
Arc energy	– 2 kJ/mm ≅ 3 kJ/mm at 50 mm thickness
Restraint:	Strong backs

FIGURE 2 WELD PROCEDURE FOR MULTIPASS SAW WELDS (2kJ/mm)
SHOWING POSITION OF CHARPY NOTCH

a)



Schematic diagram showing the welding arrangement of the sandwich panel

b)

Welding process:	Submerged Arc (single wire)
Welding preparation:	Single bevel with backing bar Included angle 15 ^{o} Root gap 11 mm
Consumables:	Oerlikon SD3, 3.2 mm diameter Oerlikon OP121TT
Welding conditions:	
Current	680 – 700A
Voltage	30V, DC +
travel speed	250 mm/min
Arc energy	5 kJ/mm
Preheat Temperature	100 ^{o}C
Interpass Temperature	100 – 150 ^{o}C
Restraint:	Strong backs

FIGURE 3 WELD PROCEDURE FOR MULTIPASS SAW WELDS (5kJ/mm)

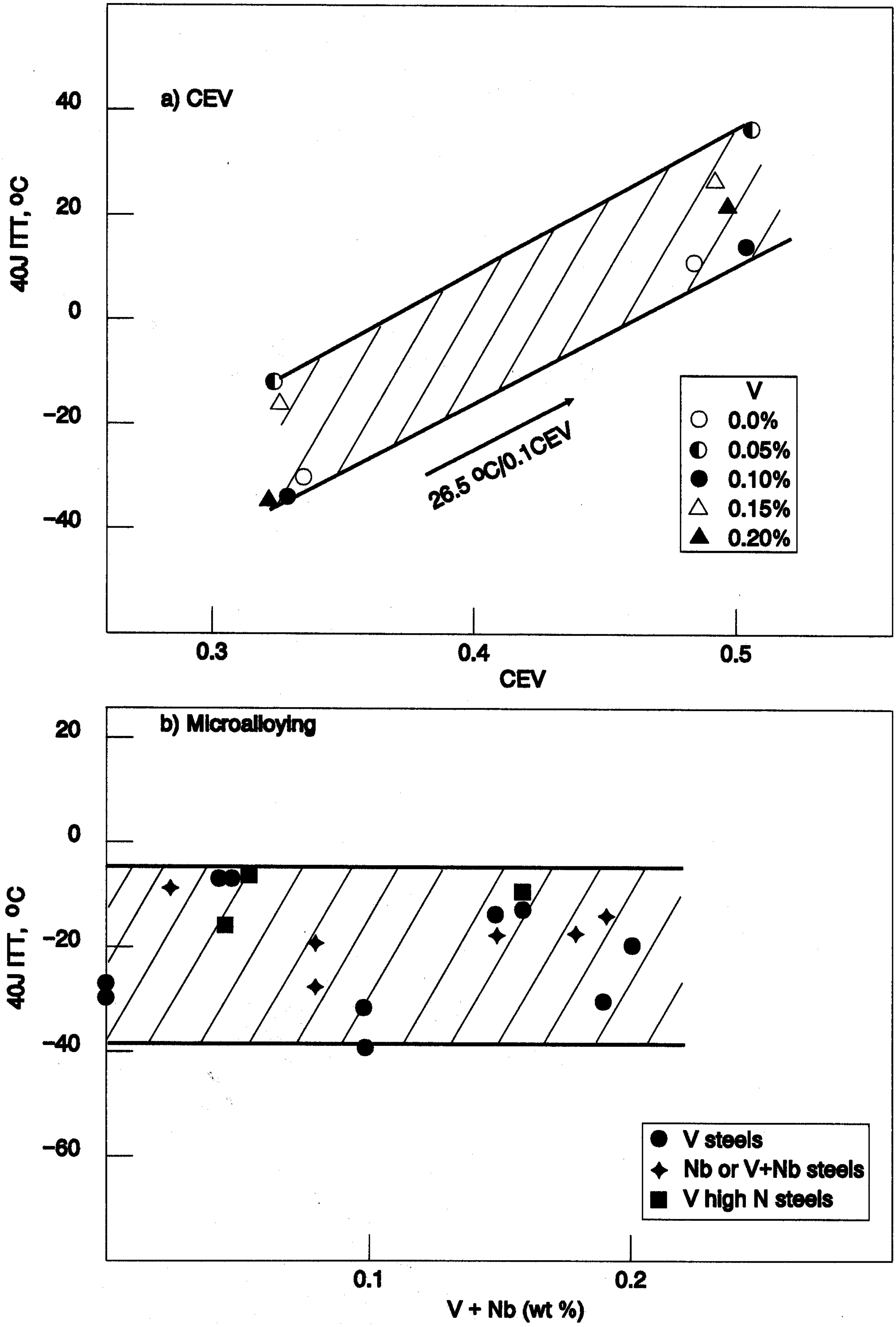


FIGURE 4 THE EFFECTS OF a) CEV b) MICROALLOYING AT CONSTANT CEV (0.34) ON THE 40J HAZ ITT, IN TWO PASS PIPE WELDS, AT 5kJ/mm, IN THE AS WELDED CONDITION.

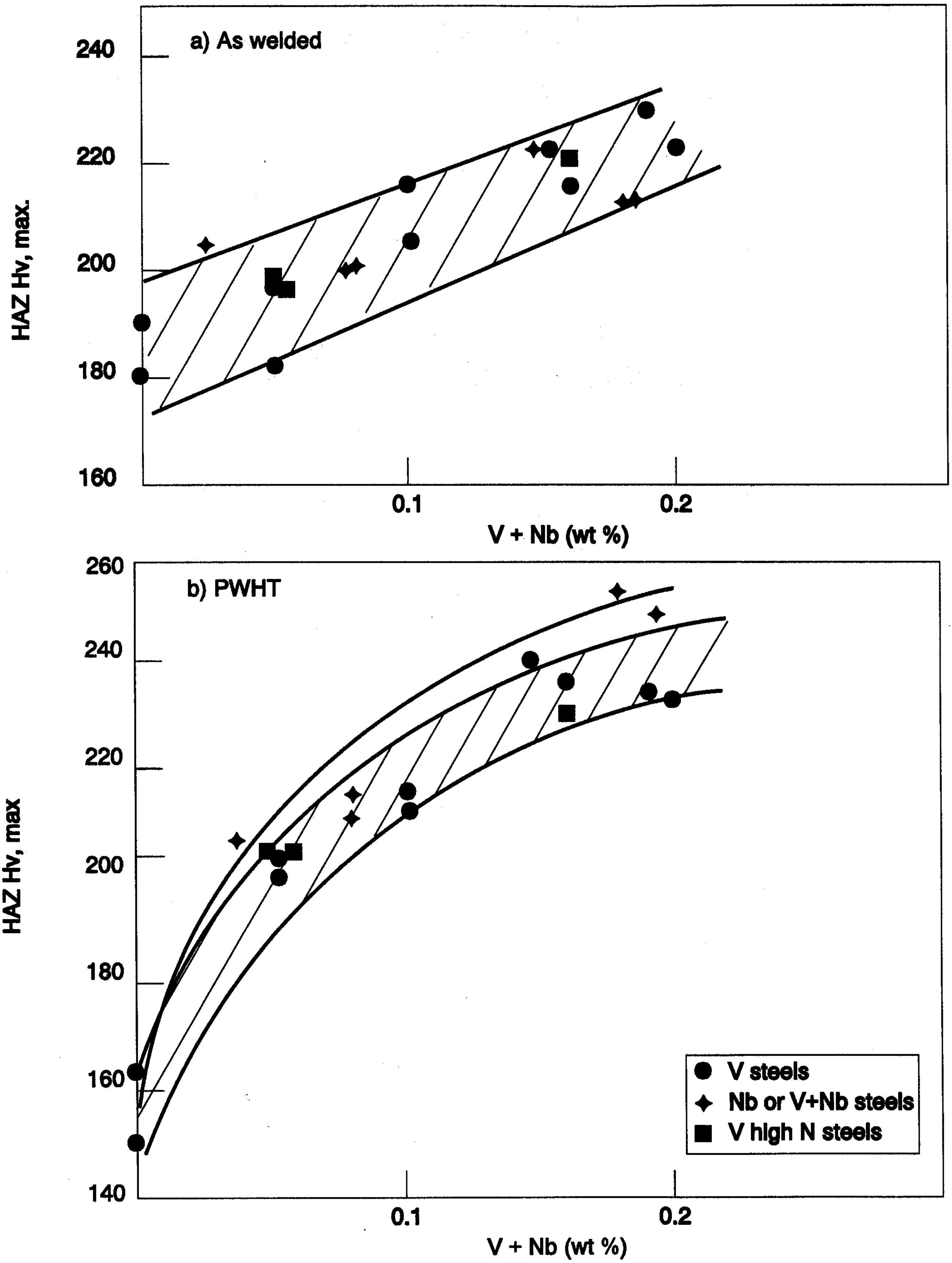
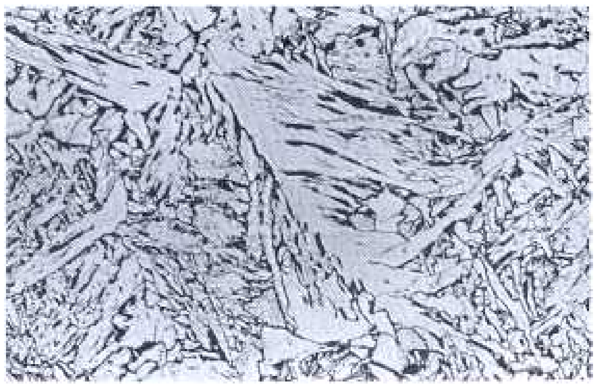
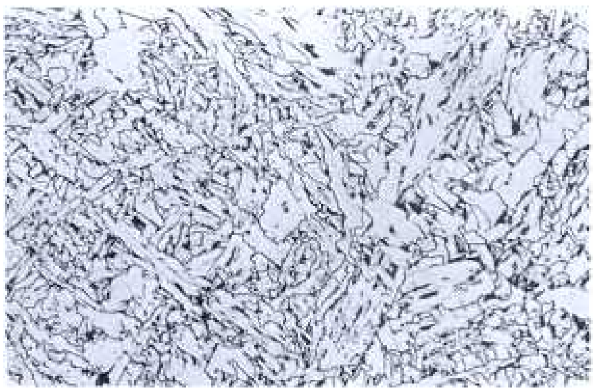


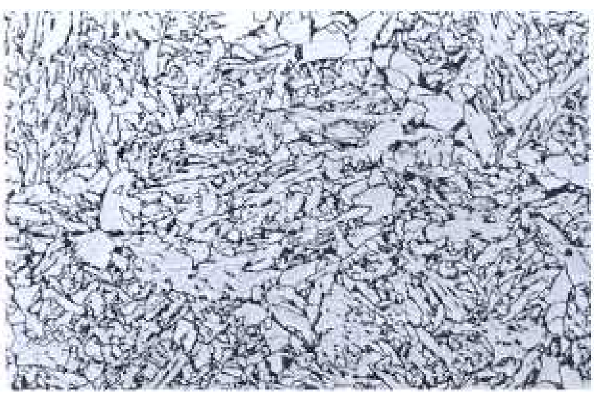
FIGURE 5 THE EFFECT OF MICROALLOYING, AT CONSTANT CEV (0.34), ON THE MAXIMUM HAZ HARDNESS IN TWO PASS WELDS, AT 5 kJ/mm a) AS WELDED b) AFTER 1 HOUR AT 600 ºC.



a) 0.0% Vanadium steel (x320)



b) 0.10% Vanadium steel (x320)



c) 0.19% Vanadium steel (x320)

FIGURE 6 THE EFFECT OF VANADIUM ON THE MICROSTRUCTURE IN THE HAZ OF TWO PASS WELDS (5kJ/MM) OF 0.07% CARBON STEELS

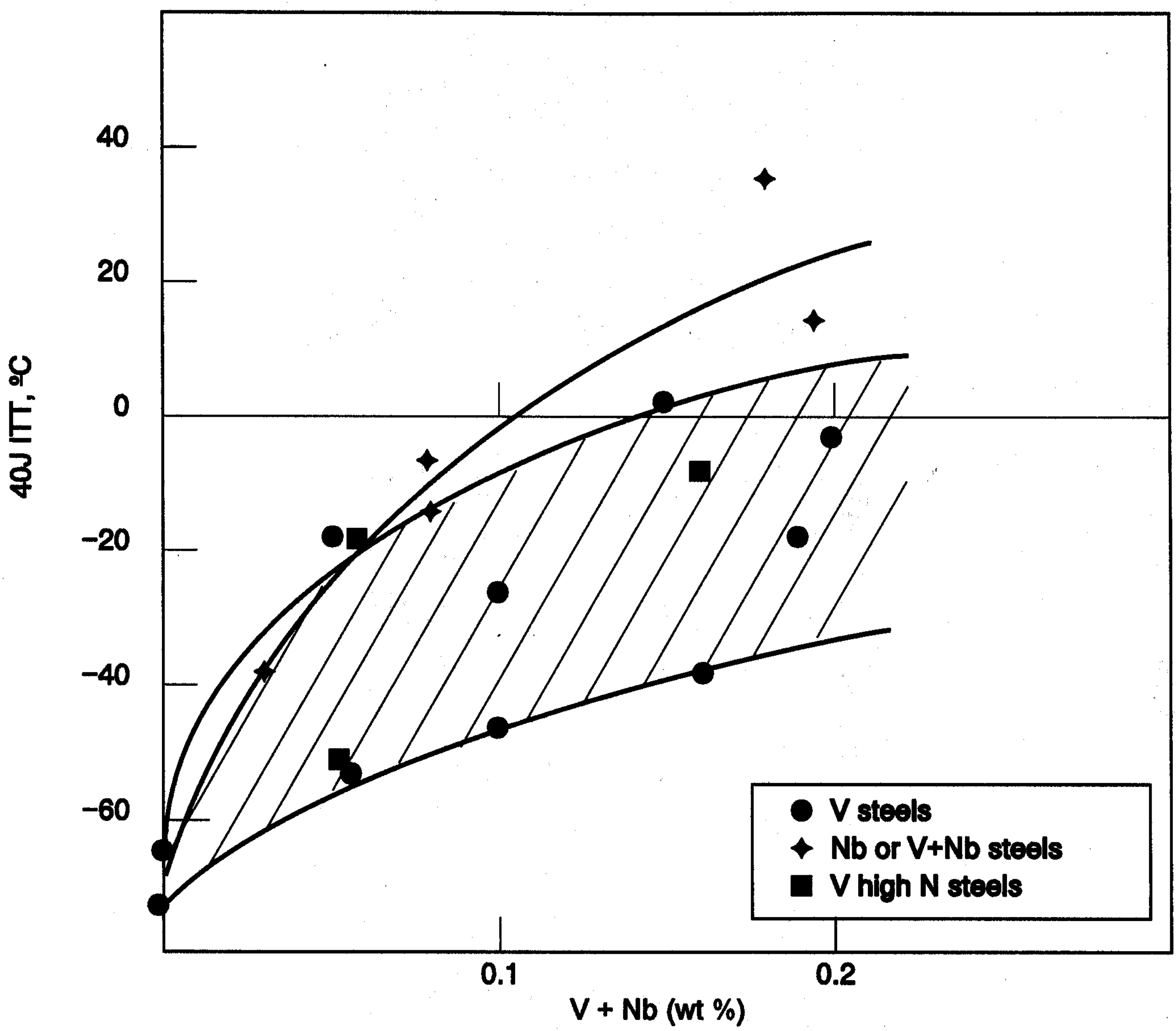


FIGURE 7 THE EFFECT OF MICROALLOYING ON THE 40J HAZ ITT, AT CONSTANT CEV (0.34), IN TWO PASS PIPE WELDS, AT 5kJ/mm, AFTER 1 HOUR AT 600 ºC.

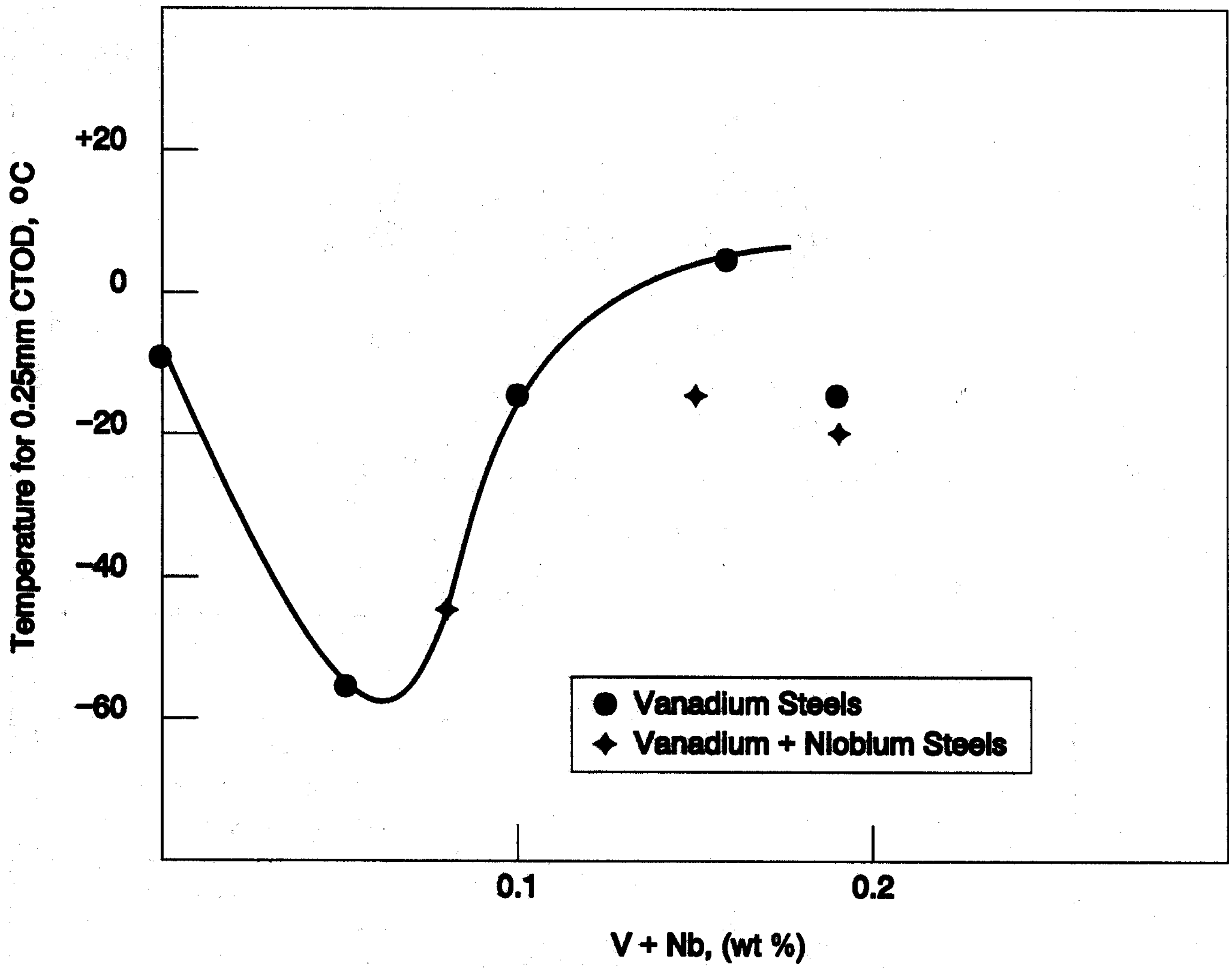


FIGURE 8 THE EFFECT OF MICROALLOYING ON THE TEMPERATURE FOR 0.25mm CTOD IN LOW CARBON (=<0.07%) LINEPIPE STEELS, WELDED AT 5kJ/mm, USING TWO PASSES, IN THE AS WELDED CONDITION.

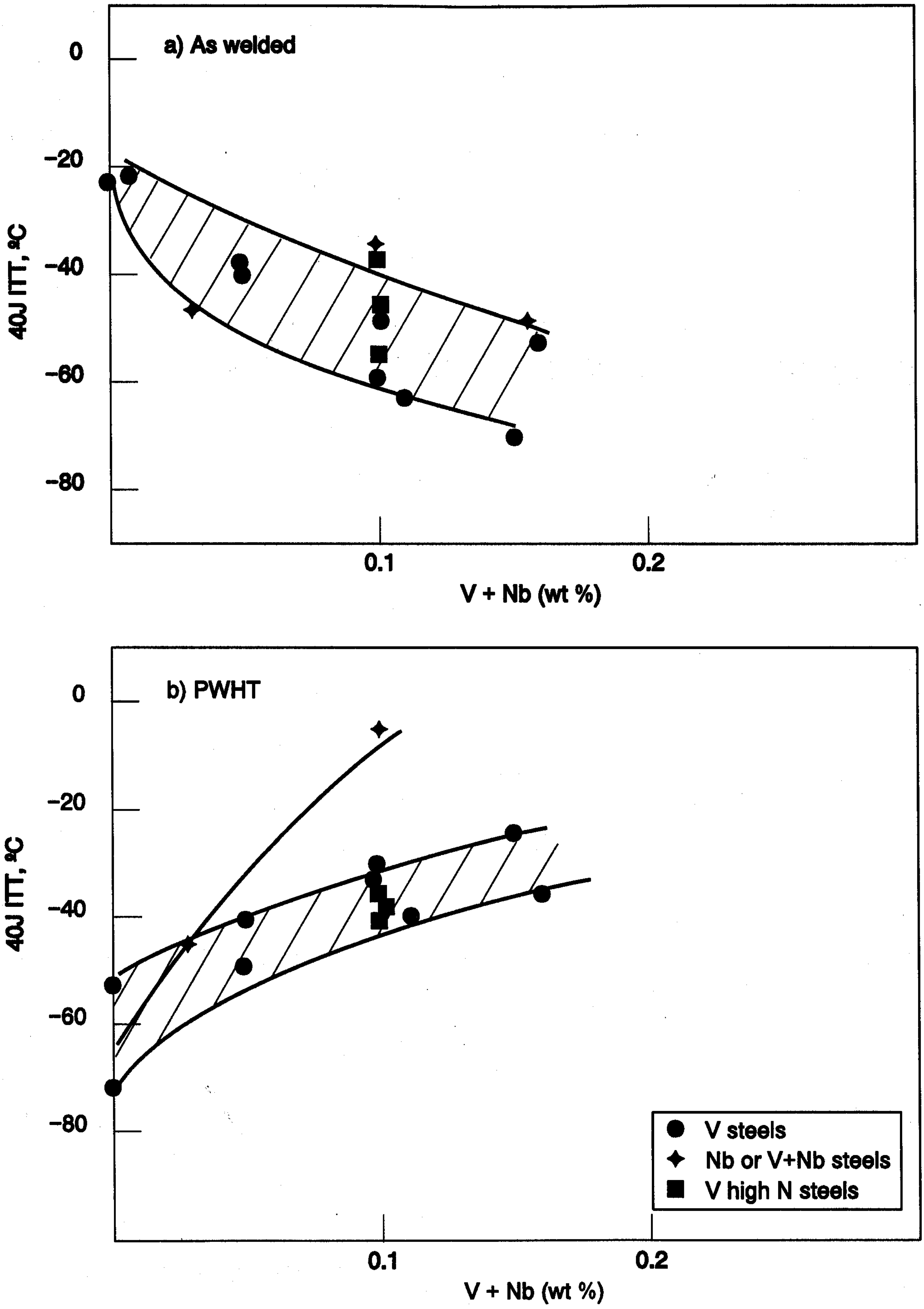


FIGURE 9 THE EFFECT OF MICROALLOYING, AT CONSTANT CEV (0.38), ON THE 40J HAZ ITT, IN MULTIPASS WELDS AT 2 kJ/mm a) AS WELDED b) AFTER 1 HOUR AT 600 ºC.

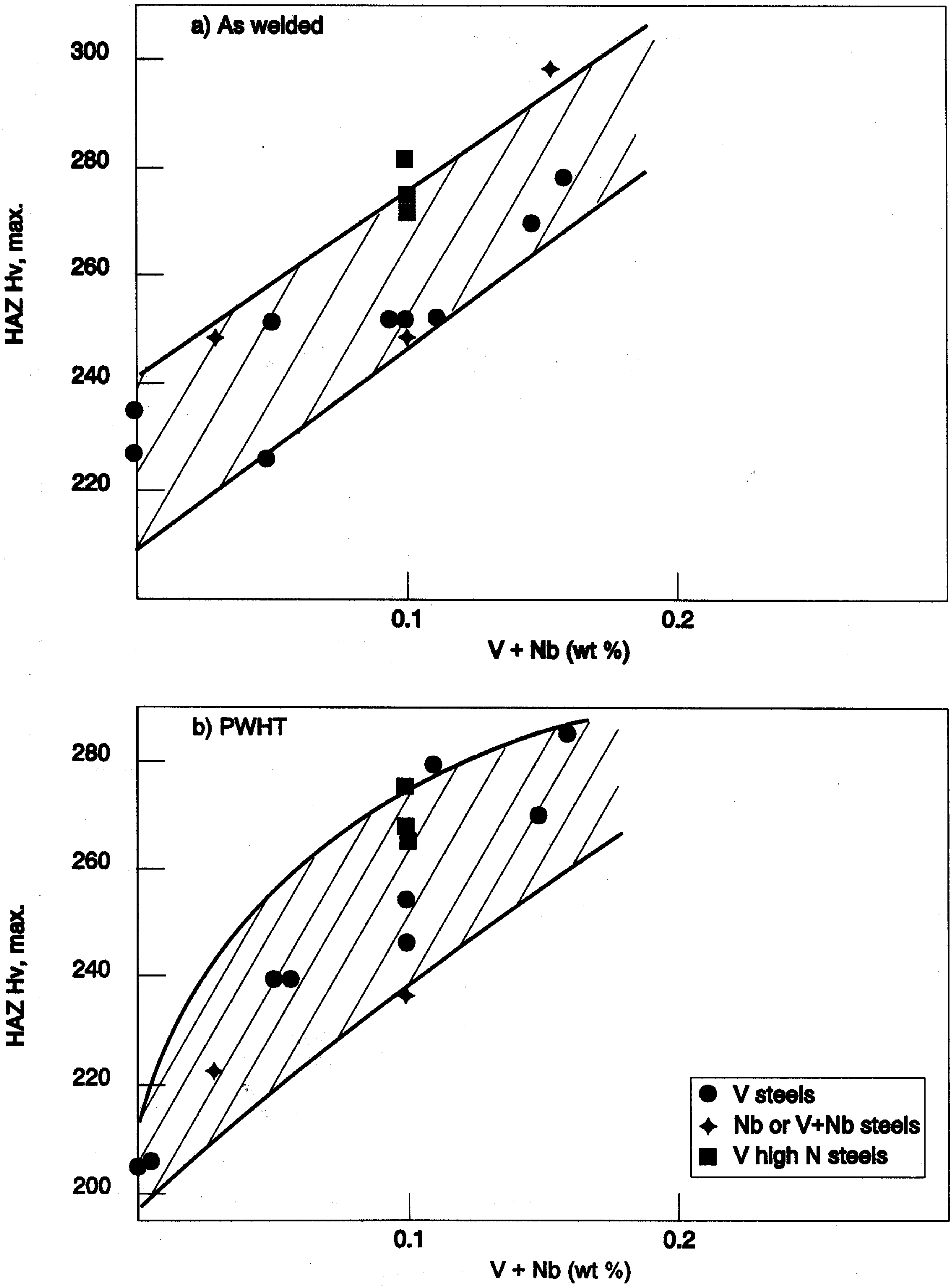


FIGURE 10 THE EFFECT OF MICROALLOYING, AT CONSTANT CEV (0.38), ON THE MAXIMUM HAZ HARDNESS, IN MULTIPASS WELDS AT 2 kJ/mm a) AS WELDED b) AFTER 1 HOUR AT 600 ºC.

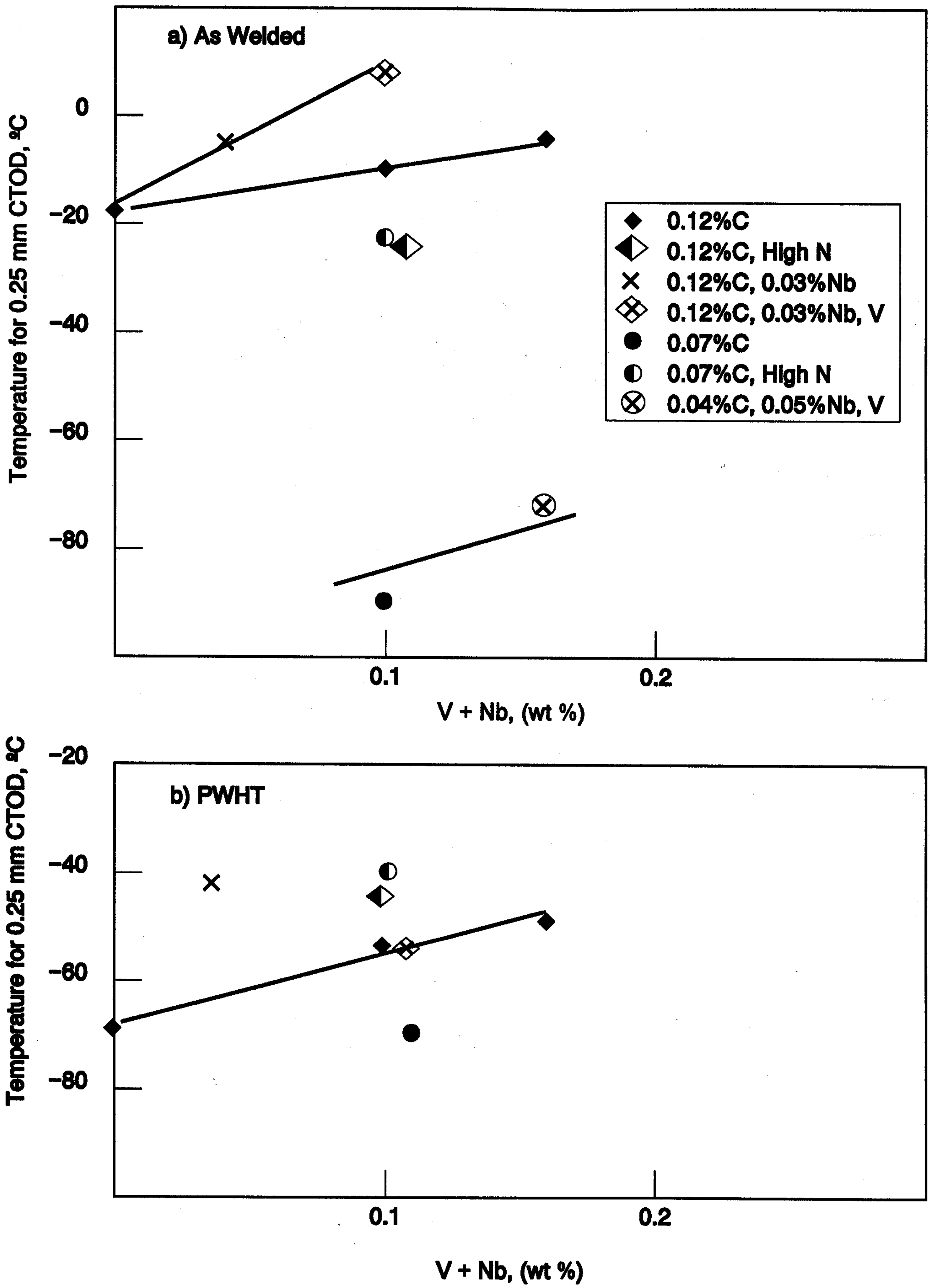


FIGURE 11 THE EFFECT OF MICROALLYING ON THE TEMPERATURE FOR 0.25mm CTOD, IN MULTIPASS WELDS AT 2 kJ/mm a) AS WELDED b) AFTER 1 HOUR AT 600 ºC.

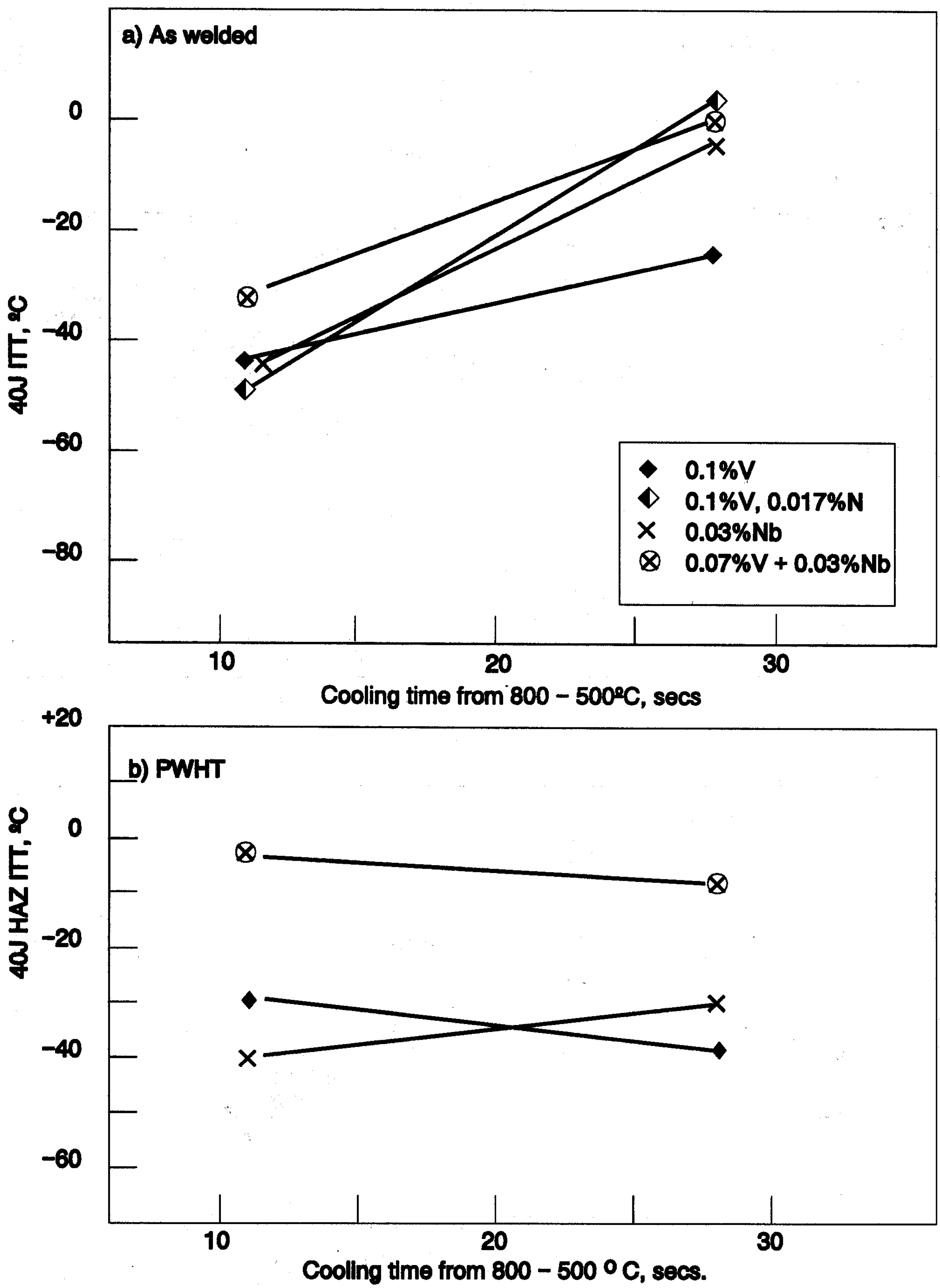


FIGURE 12 THE EFFECT OF HEAT INPUT ON THE 40J HAZ ITT IN MULTIPASS WELDS OF 0.12%C STEEL a) AS WELDED b) AFTER 1 HOUR AT 600 ºC.

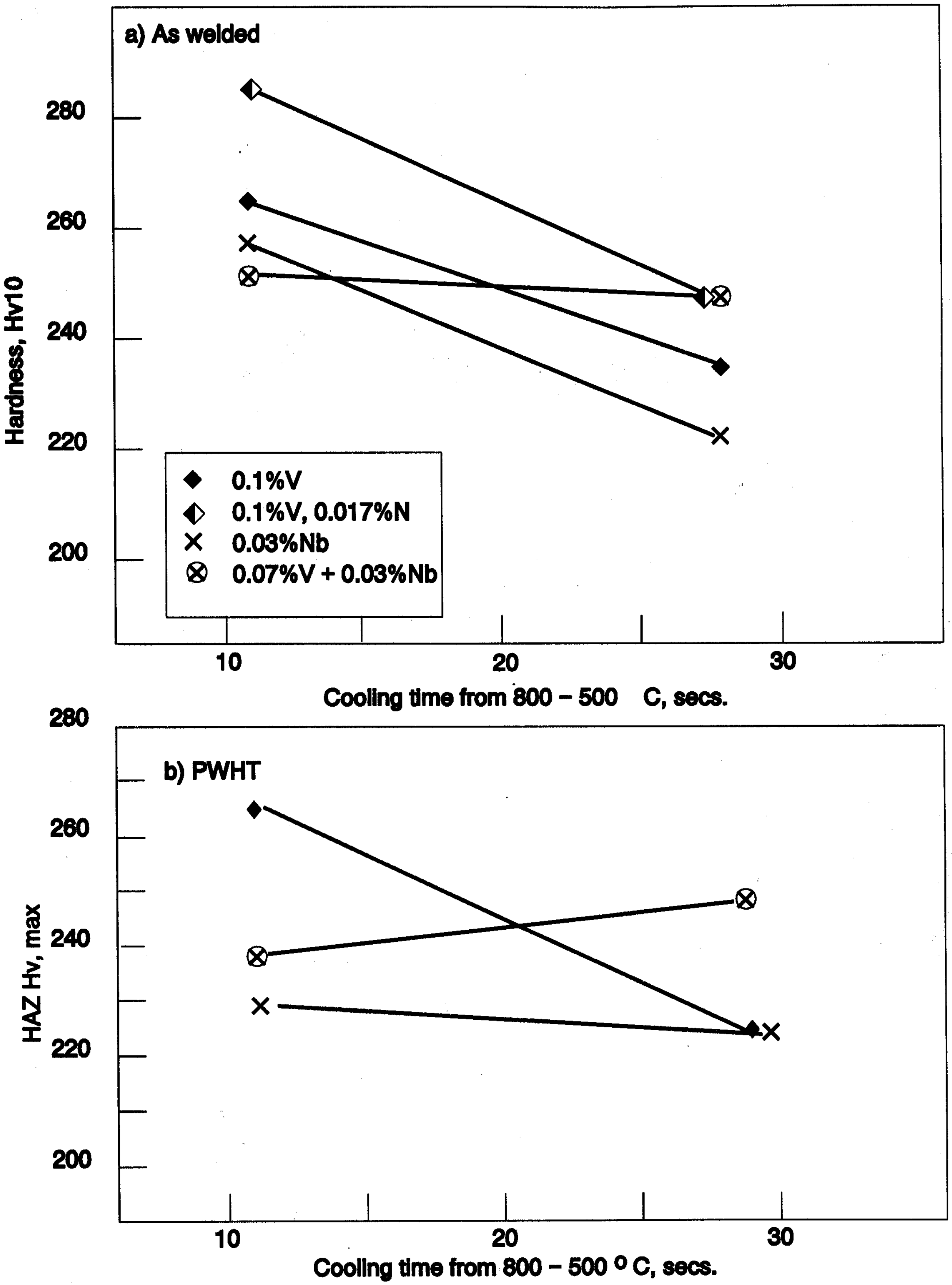


FIGURE 13 THE EFFECT OF HEAT INPUT ON THE MAXIMUM HAZ HARDNESS IN MULTIPASS WELDS OF 0.12%C STEEL a) AS WELDED b) AFTER 1 HOUR AT 600 ºC

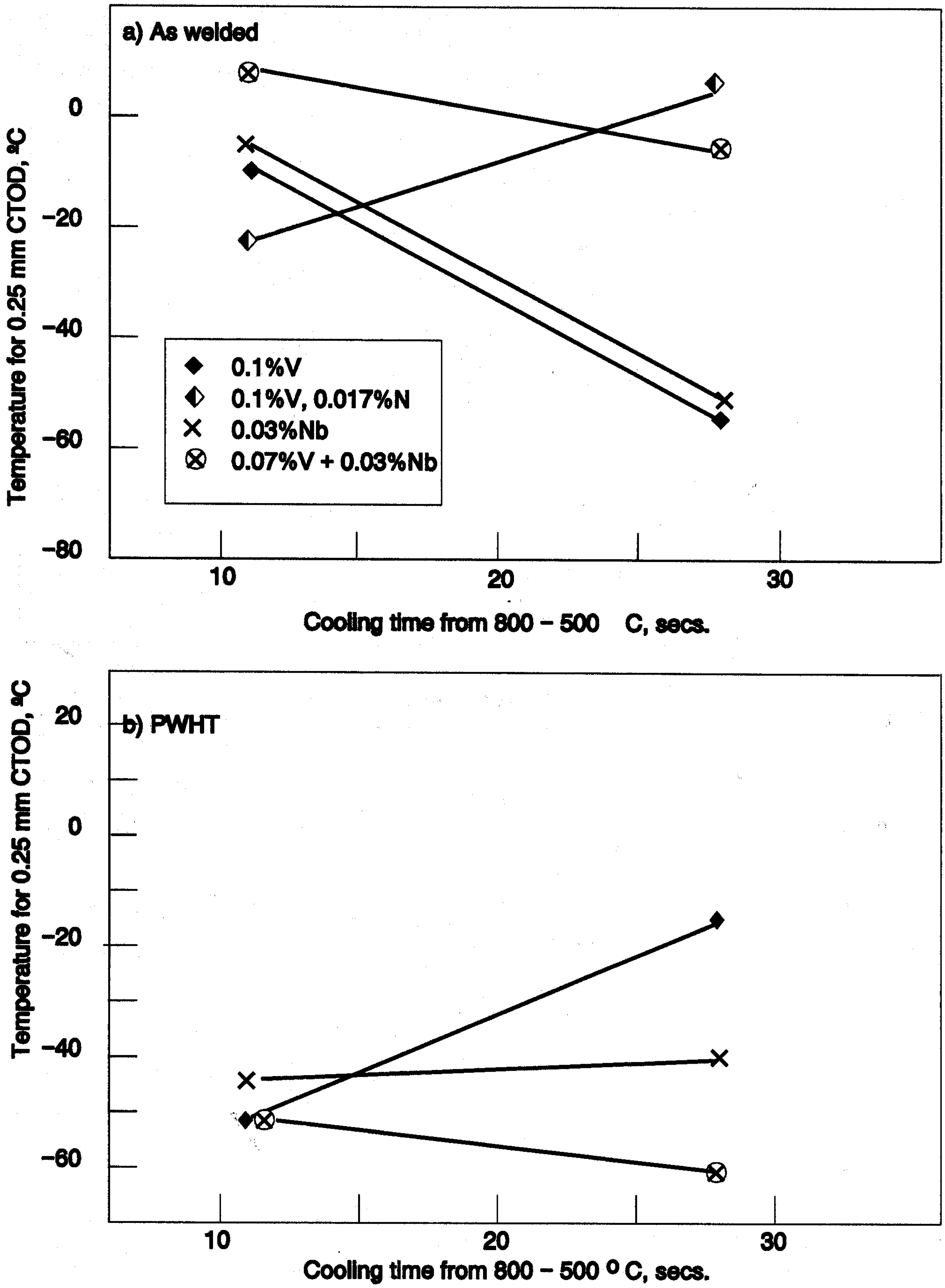


FIGURE 14 THE EFFECT OF HEAT INPUT ON THE TEMPERATURE FOR 0.25mm CTOD, IN THE HAZ OF MULTIPASS WELDS OF 0.12% C STEEL, a) AS WELDED b) AFTER 1 HOUR AT 600 ºC