THE EFFECTS OF VANADIUM ON THE PARENT PLATE AND WELDMENT PROPERTIES OF API 5LX-80 LINEPIPE STEELS

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SUMMARY

The effects of vanadium, nitrogen and titanium levels and cooling rate through the transformation on the parent plate and weldment properties of laboratory cast, controlled rolled and accelerated cooled API 5LX-80 linepipe steels have been investigated. Parent plate yield strengths >600MPa have been achieved by control of processing conditions and chemical composition.

Superior transverse Charpy toughness, (>90J @ -100°C), has also been obtained in steels which contained a small (0.01%) titanium addition.

The steels examined all had very fine ferrite grain sizes (2-3μm) and the mechanical properties, particularly the yield strength, have been shown to depend on grain size, substitutional strengthening and precipitation strengthening. In addition, the titanium-treated steels have been shown to contain a lower volume fraction of M-A phase and this reduction has had a significant beneficial influence on the impact properties.

In two-pass welds, in which the second pass was carried out at 3.9kJ/mm, acceptable levels of both weld metal and heat affected zone impact properties, weldment hardness levels and crossweld tensile performance have been achieved. The addition of titanium resulted in refinement of the HAZ microstructure. The levels of weld metal and HAZ toughness have been compared with published information and have been shown to be broadly typical of such steels.

INTRODUCTION

Linepipe producers are responding to the increasingly severe demands which are being placed on pipelines. The need for higher strength, from an economic standpoint, and high toughness, for safety, are challenges that the linepipe producers are addressing in the production of API 5LX-80 grade steel. Arctic grade X-70 linepipe properties have been achieved in steels with polygonal ferrite-pearlite structures, in which strength is derived from the combination of a fine ferrite grain size, precipitation of vanadium and to a lesser extent niobium carbonitrides in ferrite, plus some strengthening from dislocations. A relatively small contribution to the strength also comes from manganese, silicon and other elements in solid solution. In these steels, the fine ferrite grain size has been achieved by additions of niobium, typically in the range 0.02 - 0.05wt%. This forms precipitates of niobium carbide which prevents recrystallization of austenite and, during the heavy controlled rolling schedule just above Ar_3, results in a heavily deformed pancaked austenite which transforms to fine polygonal ferrite grains. In some cases the processing technique involves rolling into the ferrite and austenite field in which a substantial contribution to strengthening is produced from dislocations in ferrite. Rolling into the $\alpha + \gamma$ phase field has been examined in the production of API 5LX-80 grade steels [1-4] and application of this process has resulted in commercial production.

The use of controlled rolling, followed by accelerated cooling after rolling, combined with suitable additions of Nb-Ti, Nb-V or Nb-V-Ti, has also been employed in the production of X-80 grades. The potential benefits of on-line accelerated cooling in microalloy steel plate processing include [5-8]:-

control and refinement of the transformation products.

increased precipitation strengthening from microalloys.

lower impact transition temperatures.

leaner composition required to meet high strength specifications.

lower carbon equivalent, resulting in improved weldability.

X-80 properties have been achieved in controlled rolled and accelerated cooled steels containing Nb-Ti[9] and Nb-V-Ti[10]. It is evident on examining the potential X-80 compositions that the microalloy content may vary significantly. For example, in Nb-Ti steels, although Nb levels of 0.03-0.05 wt% are fairly typical, they may be as high as 0.10% as recommended by Hulka et al [11]. In those steels containing Nb-V-Ti, the niobium is usually approximately 0.05% and vanadium may vary between 0.04 and 0.08%[12]. In most cases titanium is present primarily to provide stable TiN precipitation to limit grain growth in the heat-affected zone (HAZ) and minimize loss of toughness in this region.

Given the variation in potential process routes and composition for this relatively new grade of high strength steel, a study was initiated to examine the effect of cooling rate, through the transformation, on steel which had been controlled rolled with a heavy deformation above the Ar_3 followed by accelerated cooling to 550°C. The study also examined the effects of vanadium, nitrogen and titanium on the parent plate and weldment properties of the steel, which had a nominal composition of 0.08%C/1.8%Mn/0.25%Si/0.05%Nb. This paper gives some details of the study.

EXPERIMENTAL PROCEDURE

Composition

The chemical compositions of the experimental steels under investigation are presented in Table 1.

Steel 1 had a base chemical composition containing microalloying additions of Nb and Ti. The remaining steels had two levels of V and were manufactured with and without a Ti addition. Furthermore, the higher V steels were examined at two levels of nitrogen.

The steels were made in the CANMET experimental casting laboratory, using an induction furnace of heat size 230 kg (500 lb). Each heat was cast into four 50 kg (110 lb) ingots in cast iron moulds, to simulate a typical solidification rate in a continuously-cast steel slab. The measured cooling rate at mid-thickness of the ingots was 22°C/min from 1470°C to 1100°C, which is equivalent to the post-solidification cooling rate at 1/3 to 1/16 depth of a 230mm thick continuously cast slab. Each 125 x 150 x 300mm cast ingot was cut to a 125 x 125 x 150mm (h x w x l) slab for hot rolling.

Processing

The ingots were reheated to 1150°C for 3h and rolled to 16mm plate on CANMET's pilot-reversing mill, using the processing conditions given in Table 2. The temperature at mid-thickness of the plate was monitored using an embedded thermocouple. The 15-pass schedule introduced a total reduction of 87%, of which 67% was applied between 850°C and the finish rolling temperature of 800°C. The plates were moved immediately from the finish rolling pass to an in-line, water spray, accelerated cooling system where they were cooled to 550°C, at measured cooling rates in the range of 5-36°C/s. Details of the processing conditions

Table I. Chemical Compositions of the Steels Investigated

Steel	C	Mn	Si	Al	Nb	V	Ti	Cu	Ni	N	O	S	P
1	0.089	1.80	0.35	0.033	0.057	-	0.012	0.21	0.11	0.0062	0.0028	0.0018	0.0062
2	0.079	1.70	0.25	0.013	0.053	0.045	-	0.19	0.10	0.0069	0.0063	0.0019	0.0077
3	0.079	1.69	0.27	0.021	0.047	0.05	0.009	0.19	0.11	0.0057	0.007	0.0017	0.0091
4	0.076	1.90	0.25	0.017	0.061	0.089	-	0.20	0.10	0.0058	0.008	0.0016	0.0062
5	0.083	1.80	0.34	0.034	0.058	0.079	0.008	0.20	0.10	0.006	0.005	0.0023	0.0062
6	0.08	1.90	0.32	0.031	0.05	0.07	0.01	0.20	0.09	0.0065	N/A	0.0026	<0.006
7	0.08	1.90	0.32	0.032	0.053	0.07	0.01	0.20	0.09	0.0062	N/A	0.0026	<0.006
8	0.093	1.90	0.24	0.014	0.045	0.085	0.008	0.21	0.11	0.0066	N/A	0.0016	<0.006
9	0.087	1.70	0.31	0.035	0.052	0.091	0.011	-	-	0.0068	N/A	0.0017	<0.006
10	0.081	1.70	0.19	0.036	0.053	0.085	-	0.20	0.10	0.01	0.005	0.0019	0.0088
11	0.081	1.70	0.19	0.036	0.053	0.085	-	0.20	0.10	0.01	0.005	0.0019	0.0088
12	0.078	1.80	0.26	0.022	0.049	0.09	-0.01	0.20	0.12	0.0105	0.0058	0.0017	0.0070

Table II. Aim Processing Conditions

Ingot reheating temperature	1150ºC for 3 h
Rolling schedule:	67% deformation below 850ºC FRT 800ºC
Cooling schedule:	10s delay to start cool 5-36ºC/s cooling rate 800-500ºC 550ºC finish cooling temperature
Final plate dimensions	15mm x 125mm x 2500mm

experienced by each plate are given in Table 3.

Mechanical Properties and Metallography

Specimens in the thickness x longitudinal direction, cut from near the thermocouple location of the plate, were prepared for microstructural examination using standard metallographic techniques. Etching was firstly carried out using a 4% picral solution and was followed by a 2% Nital solution. Micrographs were taken from the 1/4 thickness region of the plates at a magnification of 500x. Further examination was carried out at 1000x and linear intercept and point counting techniques employed to determine the ferrite grain size and volume fraction of polygonal ferrite, respectively. Samples were also etched using LePera's reagent and quantitative measurements of volume fraction of M-A constituent were made using a Quantimet.

Tensile tests were conducted for each plate using two transverse specimens of 6.35 mm (0.2 in) diameter gauge length, taken from near the thermocouple location. Charpy V-notch impact tests were performed on both longitudinal and transverse specimens at temperatures ranging from ambient to -140ºC.

Welding

Plates were prepared for welding as shown in Fig 1. Two-pass welds were produced using SD3Mo wire and

Table III. Processing Conditions Obtained

Steel	FRT °C	Cs-t secs	Cs-T °C	Ts °C	800-500°C °C/sec	End cool °C
1	805	10	790	690	25	540
2	805	8	795	690	22	545
3	810	7	800	650	27	530
4	810	8	800	670	31	515
5	805	10	795	670	36	505
6	805	7	790	710	5	580
7	805	8	795	700	21	550
8	805	10	795	650	33	510
9	790	9	775	680	28	500
10	800	10	780	680	15	580
11	810	9	795	680	26	525
12	810	8	800	670	20	555

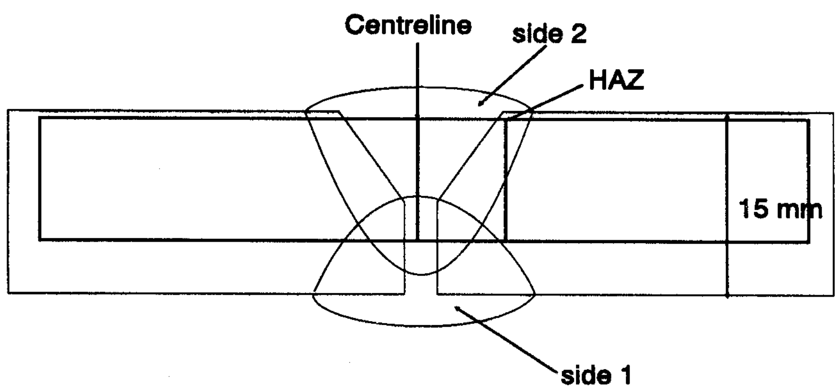


Figure 1. Schematic Diagram of Weld Preparation

OP122 basic flux, at heat inputs typical of those used in the seam welding of high-strength linepipe steels. The actual welding conditions are given in Table 4.

Table IV. Welding Conditions

	Side 1	Side 2
Lead	DC 600A x 32V x 30ipm	DC 750A x 32V x 29ipm
Trail	AC 400A x 40V x 30ipm	AC 600A x 40V x 29ipm
	Total heat input 2.75 kJ/mm	Total heat input 3.90 kJ/mm

A full chemical analysis was carried out for each weld metal. This consisted of spectrographic analysis for most elements, Leco analysis for carbon and sulphur and wet analysis for nitrogen and oxygen. Extended Charpy specimens were machined from the sub-surface on side 2 and etched in 5% Nital. Specimens from each weld were then scribed at both weld metal centre line and fusion line locations, notched and cut to length. Charpy V-notch testing was performed to generate full transition curves for both weldment locations. It should be noted that, in the case of the weld centre line impacts, the weld bead on the second side was large enough for the specimens to be extracted from the single bead on that side. In addition, full thickness flat cross weld tensile specimens were machined, tested and examined to determine the location of failure.

Each weld was sectioned, polished using standard techniques and etched in 2% Nital. Optical micrographs were taken on typical weld metal and heat-affected zone microstructures.

RESULTS AND DISCUSSION

Parent Plate

The processing conditions and tensile properties for each of the steels examined are given in Tables 3 and 5 and the results of quantitative metallography are presented in Table 6. In general, the microstructures of all of the steels examined revealed the presence of a fine mixture of polygonal ferrite intermixed with ferrite containing second phase. The ferrite grain size was typically between 2 and 3 μm and was achieved primarily as a result of the controlled rolling and cooling schedule. It can also be seen from Table 6 that the volume fraction of polygonal ferrite varied from 12% to 46%. Within the remaining major constituent, i.e, ferrite with second phase, the second phase was mainly in the form of carbide. However, varying levels of M-A constituent were also observed, typically in the range of 0.6 - 4.4%. In the case of Steel 6, it was as high as 12.6%, with a small volume fraction of pearlite (<5%), being present. Generally, the grain size became finer and the amount of polygonal ferrite decreased and ferrite with second phase increased, with increase in cooling rate. Typical microstructures depicting these changes are shown in Figure 2.

Table V. Tensile Properties

Steel	YS N/mm^2	UTS N/mm^2	R of A %	El %	Luders strain, %
1	555	561	76	25	1.8
2	526	639	73	27	1.7
3	541	644	74	24	2
4	624	727	61	19	0.6
5	611	699	45	16	2
6	481	705	68	28	0
7	561	703	74	25	1.1
8	611	710	68	26	1.2
9	557	632	67	27	2.1
10	542	665	54	26	1.5
11	633	715	52	20	0.8
12	593	676	75	21	1.6

Cooling rate was one of the main factors affecting the parent plate properties of the steels examined in the present work. Figure 3 shows the effect for steels 5 - 8, which were of the same nominal composition. In this figure, the small differences in substitutional hardening, arising from varying levels of Si, Mn, Cu and Ni have been normalised to 0.25% Si, 1.8% Mn, 0.2% Cu and 0.1% Ni. The vectors used are detailed later in this section. Increasing the cooling rate was then observed to increase yield strength by 4 N/mm^2 per °C/s increase in cooling rate. Further, to achieve a yield strength of 550 N/mm^2, a cooling rate of at least 20°C/s was indicated as being required.

Figure 4 shows the effects of vanadium and nitrogen on the yield strength, normalised to a constant cooling rate of 20°C/s by the above vector and at constant Si, Mn, Cu and Ni levels. It should also be noted that, in the case of the titanium treated steels, it has been assumed that all the titanium would combine stoichimetrically with nitrogen to form TiN. This nitrogen has been deducted from the total nitrogen to give that available for reaction with vanadium, niobium, aluminium etc. The yield strength was observed to increase by around 10 N/mm^2 per 0.0001 increase in V x N product. At the chosen cooling rate of 20°C/s, a V x N product of around 4×10^{-4}, corresponding to a vanadium level of 0.08% and a nitrogen level of 0.005%, was required to achieve a yield strength of 550 N/mm^2. Clearly increasing the cooling rate and/or increasing the nitrogen level would enable a reduction in vanadium level to be achieved.

Table VI. Microstructure of Steels Investigated

Steel	Polygonal ferrite %	Av. Ferrite grain size, μm	$d^{-1/2}$ $mm^{-1/2}$	MA, %
1	32	2.5	20	1.1
2	40	2.9	18.6	2.4
3	17	2.7	19.2	1
4	13	2.1	21.8	4.5
5	19	2.2	21.3	0.7
6	46	3.2	17.7	12.6
7	36	2.5	20	n/a
8	26	2.2	21.3	n/a
9	43	3.2	17.7	n/a
10	31	2.5	20	3.8
11	12	2.2	21.3	n/a
12	34	2.6	19.6	0.6

To determine the absolute contribution to yield strength from precipitation, the following extended Hall-Petch relationship [15] was used.

$$\sigma y = \sigma_o + \sigma_s + \sigma_p + \sigma_d + \sigma_t + kyd^{-1/2}$$

where σ_o is a term due to lattice friction and includes a contribution from C and N remaining in solution. It has been suggested [15] that its value should be of the order of 70 N/mm². σ_s results from solid solution hardening and was taken to be 32.%Mn[13] + 84.%Si[13] + 38.%Cu[13] + 43.%Ni[14]. σ_d is hardening due to the presence of dislocations and is, generally, considered to be of importance, where a significant proportion of the rolling schedule has been carried out below the Ar_3 or the transformation temperature is low, both of which could result in an increase in dislocation density. In the present case where the finish rolling temperature was above the Ar_3 and the bainite start temperature was ≈600°C, the contribution from dislocations is likely to be small[18]. σ_t is due to texture hardening and because the finish rolling temperature was above the Ar_3, can be considered to be at, or near to, zero. d is the ferrite grain size and ky is a constant which, for controlled rolled steels, has been suggested to equal 18.1 $N/mm^{3/2}$ [15].

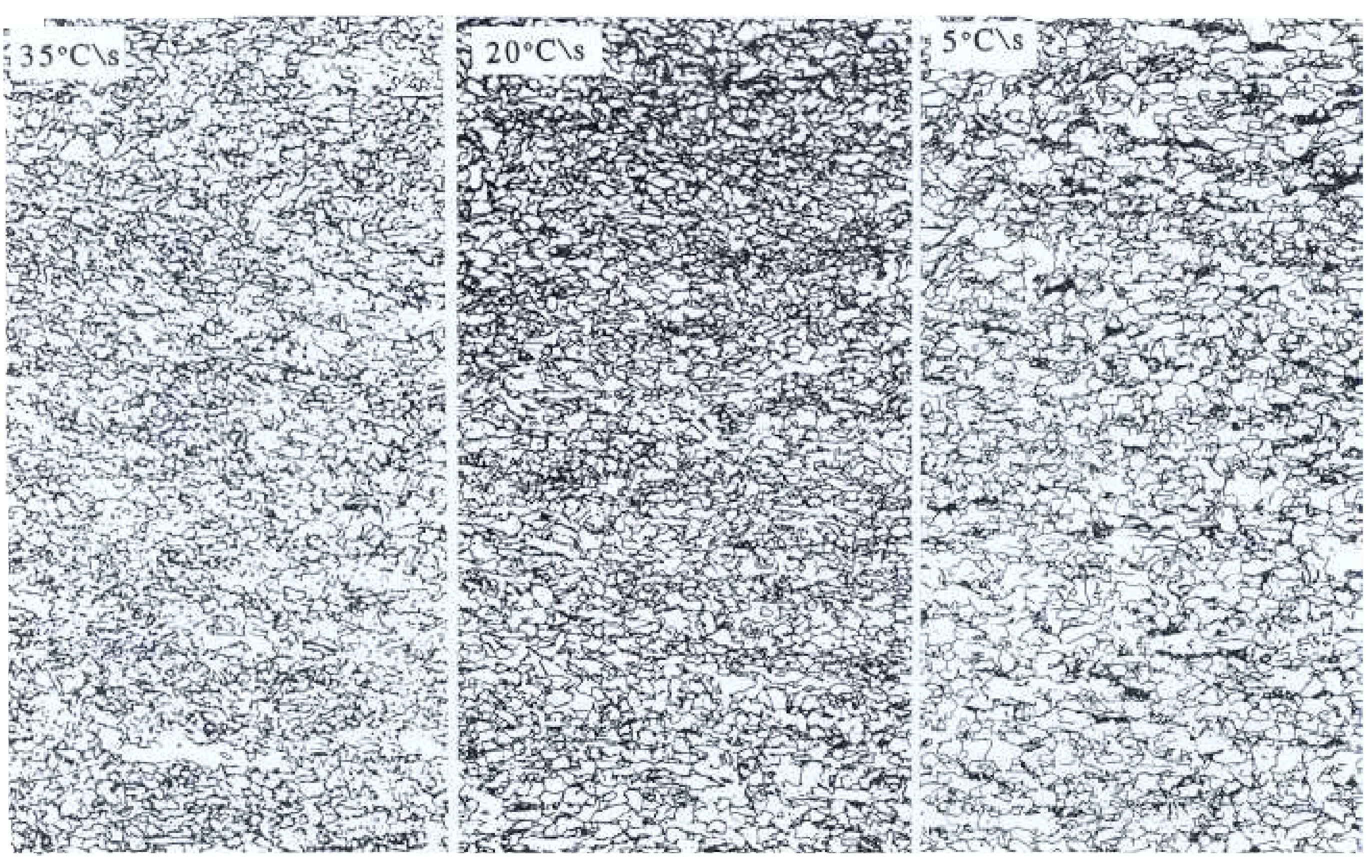


Figure 2. The Effect of Cooling Rate through the Transformation (800 - 500°C) on the Microstructure of Accelerated Cooled API 5LX80 Steels (Steel 2) x500

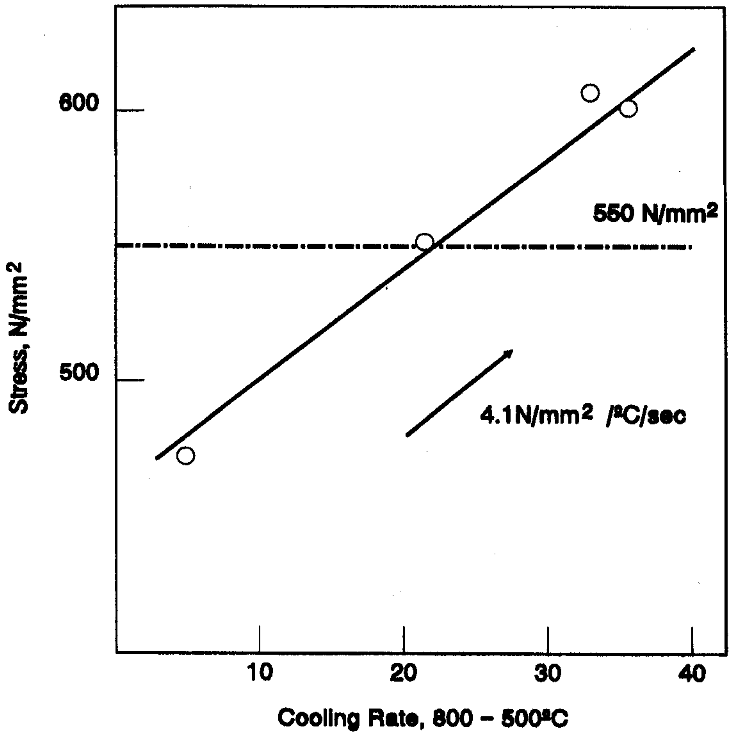


Figure 3. The Effect of Cooling Rate on the Yield Strength of API 5LX80 Accelerated Cooled Steels containing 0.08%V/0.05%Nb/0.01%Ti/ 0.006%N at Constant Si, Mn, Cu & Ni

The contributions to yield strength from σ_o, σ_s and d were subtracted from the observed yield strength and the residue plotted as a function of V x N product (Figure 5), once again making allowance for the effect of Ti on the available nitrogen. This results in a reasonably good straight line which shows that, given the above assumptions on σ_d and σ_t, up to 100 N/mm^2 was obtained from precipitation strengthening. The line also intersects the ordinate at approximately 25 N/mm^2. This suggest that, within any 100 N/mm^2 from precipitation strengthening, 75 N/mm^2 would arise from VN precipitation while the remaining 25 N/mm^2 would be due to NbCN precipitation. Examination of the literature indicates that in vanadium-containing microalloyed steels it is reasonable to assume that the precipitate is VN rather than VC [16,17] and the contribution from NbCN is not unreasonable for a steel of the composition and rolling schedule used [19].

Table 7 and Figure 6 present the results from Charpy V-notch testing of the steels. As would be expected from a steel of very fine grain size, transition temperatures were all very low, with some being below -140°C. Although all the steels contained less than 0.0026%S, a difference in toughness was observed between the longitudinal and transverse directions. This arose because the steels were not treated with calcium during manufacture. It is also clear from Figure 6, that the steels which contained titanium had superior low temperature toughness compared with those which did not. Quantitative analysis revealed a variation in the volume fraction of M-A constituent (Table 6). Steels containing titanium had a level of M-A generally less than 1.1%, while those without titanium contained between 2.4 and 4.4% M-A. It has been observed [22,23] that such small differences in M-A can result in significant changes in toughness, the M-A acting as microcrack initiation sites at low temperatures.

The exact role played by titanium in reducing the amount of M-A, is not precisely understood. However, it is possible that the presence of titanium has reduced the level of available nitrogen in the steel and hence reduced the effect of nitrogen as an austenite stabiliser. Alternatively, it may be that titanium acts as a carbide promoter, thus increasing the degree of

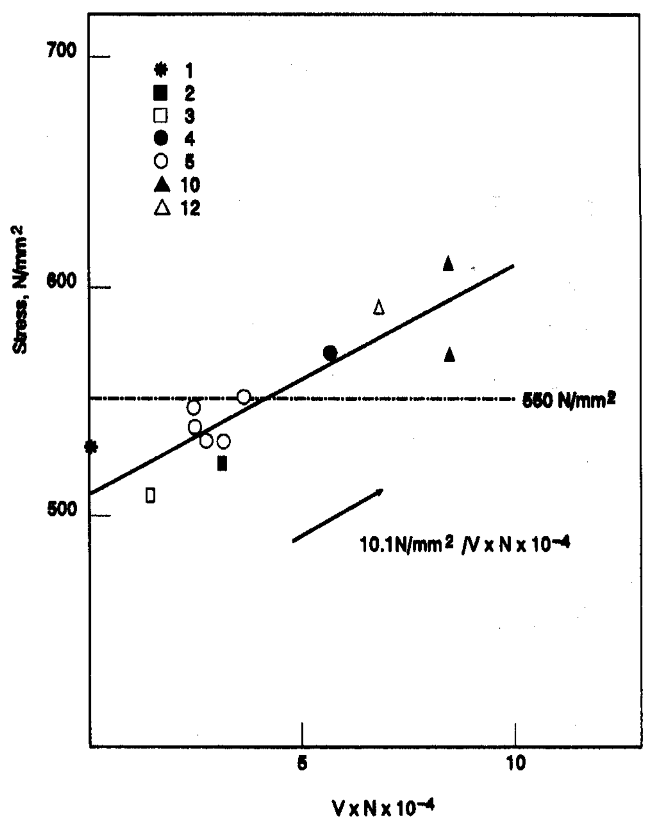


Figure 4. The Effect of Vanadium and Nitrogen on the Yield Strength of API 5LX80 Accelerated Cooled Steels at 0.25%Si/ 1.8%Mn/ 0.2%Cu/ 0.1%Ni and 20°C/s

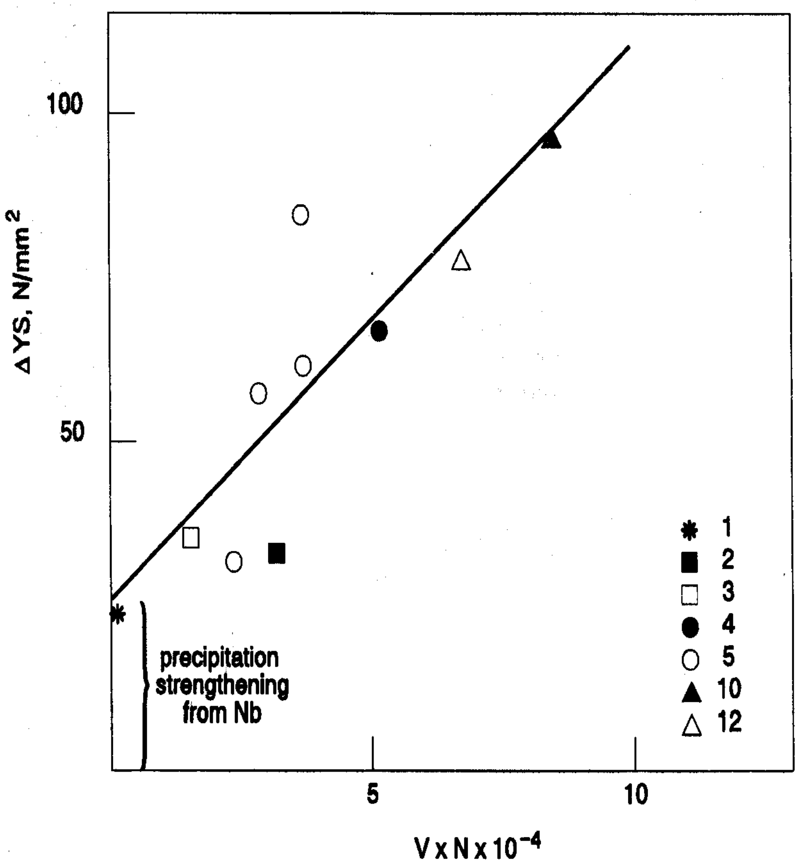


Figure 5. The Effect of Vanadium and Nitrogen on the Increase in Yield Strength due to precipitation in Accelerated Cooled Steels.

transformation to carbide of the high carbon austenite remaining towards the end of transformation.

In summary, it has been shown that the parent plate mechanical properties of API 5LX-80 steel, containing 0.08%C/1.8%Mn/0.25%Si/0.055%Nb, which has been controlled rolled using a rolling schedule incorporating 67% reduction below 850°C (FRT 800°C) followed by accelerated cooling between 800°C and 550°C at a rate of 20°C/s, can be achieved using vanadium and nitrogen additions. In such a steel a V x N product of 4×10^{-4} is required to ensure achievement of yield strength. This is equivalent to a vanadium level of 0.08% and a nitrogen level of 0.005%. The steel had a very fine grain size (2 - 3μm), resulting from the rolling and cooling schedules and, consequently, had excellent toughness. Treatment with 0.01%Ti imparted superior low temperature toughness with a 50% FATT <-140°C. This resulted from a reduction in the amount of M-A phase obtained in the microstructure.

Table VII. Charpy V-notch Impact Properties

Steel	Energy absorbed, J				50% FATT, °C
	-50°C		-100°C		
	L	T	L	T	
1	233	166	180	112	<-140
2	184	106	60	40	<-80
3	296	107	235	100	<-140
4	172	87	90	30	<-80
5	245	122	200	90	<-140
10	216	91	85	28	<-80
12	254	119	195	95	<-140

Weld Metal and Heat Affected Zone

The chemical compositions of the weld metal are given in Table 8. The differences in composition in the weld are inherited from the parent plate. For those steels with the lower vanadium content (0.045 - 0.050%), dilution of 60% has resulted in a weld metal vanadium content of 0.024 - 0.030%. At the higher vanadium level, the same level of dilution produced a weld metal vanadium level between 0.043 and 0.056%. A difference in nitrogen level also existed, depending on the original base metal content. This was between 72 and 82ppm for the low nitrogen steels and 104 - 107pmm for the high nitrogen steels.

Figure 7 gives the results of CVN testing of the weld metal. No significant variation in microstructure occurred for the steels examined in this study, all of which contained a high volume fraction of acicular ferrite (>90%), as shown in Fig 8. A comparison of notch toughness for steels 4, 5, 10 and 12, which had similar weld metal chemistries, shows that some scatter existed in the results. For example, steel 4 had an absorbed energy value of 55J at -20°C, whereas steel 5 had 104J at -20°C. This variation can be attributed to the fracture path relative to the microstructure. It has been shown that, for weld centre line tests of microstructures containing grain boundary ferrite, depending on the position of the crack front with respect to micro-

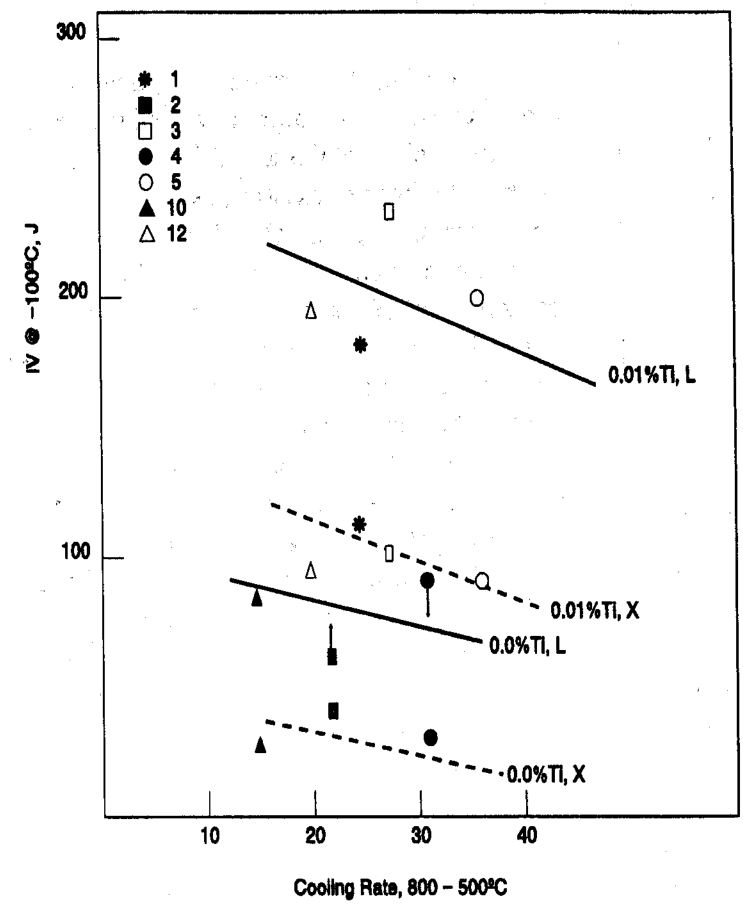


Figure 6. The Effect of Cooling Rate on the Charpy V-notch Energy Absorbed at -100°C, in non-Calcium Treated, Accelerated Cooled API 5LX80 Steels

structure, it is possible for the crack to follow veins of grain boundary ferrite, giving a reduced energy value.[25] Given this variation, it is still valid to state that, on increasing the vanadium content in the weld from 0 to 0.056 wt%, no deterioration in weld metal toughness occurs.

The results of CVN testing at the fusion line are given in Fig 9. Once again no significant effect of vanadium on the fusion properties was observed. However, as expected, in general, those steels which contained Ti tended to have higher absorbed energy values compared with those which did not. This was particularly noticeable in the case of steels 10 and 12 as shown in Fig 9.

Fig 10 presents optical micrographs of the fusion line for these two steels. This clearly shows a significant difference in the grain size adjacent to the fusion line. Steel 12, which contains 0.01% Ti, has the smaller grain size. Similarly, a comparison of steels 2 and 3 again revealed the expected improved fusion line toughness with the presence of Ti. Steels 4 and 5, however, did not follow this trend. It can only be concluded that this was due to experimental scatter.

The results of cross weld tensile tests, which are used to indicate potential problems with heat affected zone softening, were all satisfactory, with fracture occurring in the parent plate. Further evidence of adequate levels of weldment hardness is given in Table 9. In this table it can be seen that the hardness of the HAZ was either comparable to, or slightly higher than, that of the parent metal. In addition, the weld metal hardness matched that of the HAZ quite closely. It should be noted that the overall level of hardness obtained in both the weld metal and HAZ was also acceptable and no potential problems in use can

Table VIII. Chemical Composition of Weld Metal

Steel	C	Mn	Si	Al	Nb	V	Ti	Cu	Ni	N	O	S	P
1	0.077	1.60	0.24	0.009	0.029	-	<0.005	0.18	0.08	0.0076	0.0336	0.0038	0.015
2	0.071	1.60	0.22	0.006	0.028	0.03	<0.005	0.19	0.09	0.0075	0.0337	0.0042	0.014
3	0.068	1.60	0.20	0.01	0.023	0.024	<0.005	0.12	0.07	0.0079	0.0352	0.0045	0.015
4	0.066	1.70	0.20	0.011	0.023	0.043	<0.005	0.12	0.06	0.0082	0.0394	0.0044	0.013
5	0.071	1.70	0.23	0.009	0.027	0.053	<0.005	0.19	0.09	0.0072	0.0406	0.0043	0.016
10	0.072	1.60	0.20	0.014	0.026	0.055	<0.005	0.18	0.08	0.0107	0.0416	0.0038	0.016
12	0.072	1.60	0.23	0.009	0.028	0.056	<0.005	0.18	0.1	0.0104	0.0311	0.0039	0.016

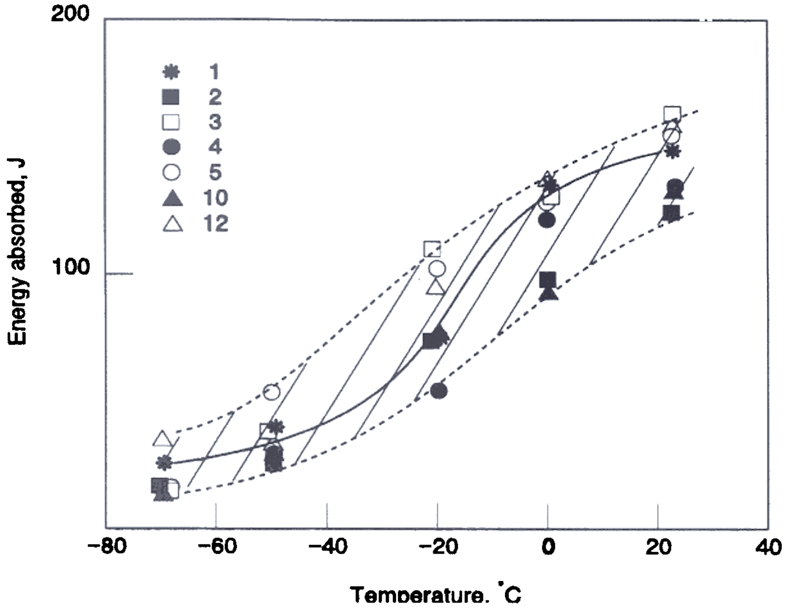


Figure 7. Charpy V-notch transition curves of weld Metal in Accelerated Cooled API 5LX80 Steels

be envisaged.

In considering the potential applications of these steels, for linepipe, it is useful to compare the weld metal and HAZ results of this study with those of existing experimental and commercial X-80 steels. Table 10, which summarises this data, indicates that a comparable level of notch toughness has been obtained in both the weld metal and HAZ. The heat affected zone results are a little lower than that obtained by Hulka et al[11], and are very close to those given by Nakasugi et al[27]. Bearing in mind the potential for variability in weldment testing, as illustrated in the results of Matousu et al[26], it can be safely said that they are within the range expected from this type of steel. It could also be reasonably expected that development work to enhance weld metal toughness would lead to an overall increase in the values obtained.

In summary, therefore, it can be said that the weldment properties of seam welds in accelerated cooled API 5LX-80 steels, containing different levels of vanadium and nitrogen, are satisfactory and are not affected by vanadium or nitrogen level, within the ranges examined.

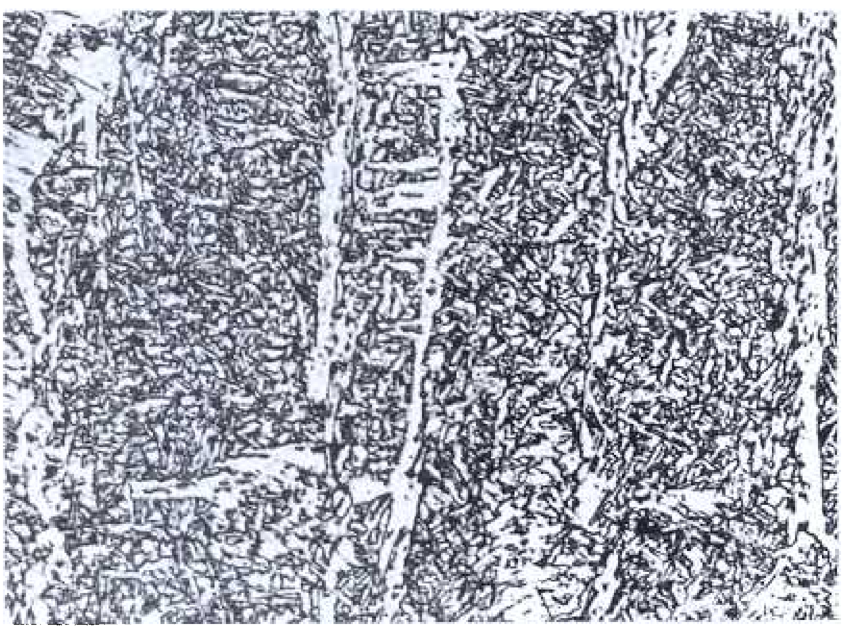


Figure 8. Typical Weld Microstructure

The presence of a small (0.01%) titanium addition has resulted in a beneficial effect on toughness in the grain coarsened HAZ. The level of toughness observed in both the weld metal and heat affected zone was within the range expected from this type of steel. No evidence of heat affected zone softening was observed and the levels of hardness obtained are unlikely to result in any problems during manufacture, or the envisaged use, of these steels.

CONCLUSIONS

The effects of vanadium, nitrogen and titanium levels and cooling rate through the transformation on the parent plate and weldment properties of a range of API 5LX-80 steels, of base composition 0.08%C/1.8%Mn/0.25%Si/0.055%Nb, which had been subjected to a controlled rolling schedule involving 67% deformation below 850°C (FRT 800°C), followed by accelerated cooling from 800°C to 550°C, have been examined. It can be concluded that:-

A cooling rate of 20°C/s through the transformation coupled with a vanadium and nitrogen level, such that %V x %N = 4 x 10^{-4}, were required to ensure a satisfactory level of tensile properties in the parent

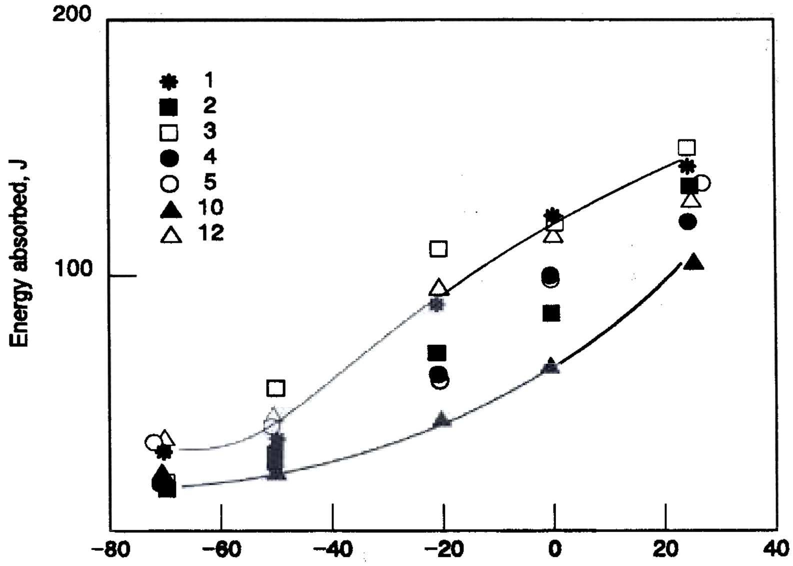
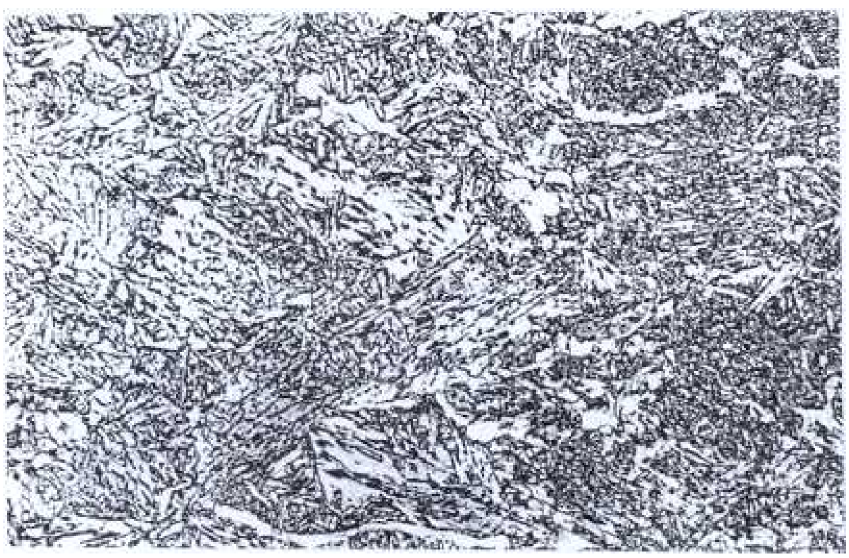


Figure 9. Charpy V-notch transition curves for the Heat Affected Zone in Accelerated Cooled API 5LX80 Steels

Steel12



Steel 10

Figure 10. Typical HAZ Microstructures depicting the Effect of Titanium

plate.

2. Impact properties were excellent and were enhanced by a small (0.01%) addition of titanium. In this latter case, impact transition temperatures below -140°C were obtained.

3. The steels were characterised by having extremely fine grain sizes (2 - 3 μm) and a microstructure which consisted mainly of polygonal ferrite and ferrite containing a second phase. Both the grain size and the proportion of polygonal ferrite decreased with increasing cooling rate. In addition, it was observed that titanium-treated steels had a lower proportion of M-A phase than non-treated steels.

4. The main factors contributing to yield strength have been shown to be grain size, substitutional strengthening and precipitation strengthening. The main contribution to precipitation strengthening came from vanadium nitride, although there was a smaller contribution from NbCN.

5. The impact transition temperature was affected by the presence of M-A phase. Titanium-treated steels with lower levels of M-A phase exhibited a superior level of toughness.

6. The levels of weld metal and heat affected zone properties obtained in the present steels have been shown to be acceptable and independent of vanadium and nitrogen levels within the ranges examined.

7. The addition of titanium refined the grain coarsened heat affected zone microstructure and, in general, resulted in an improvement in toughness.

8. Cross welded tensiles all fractured in the parent plate and the levels of weld metal and heat affected zone hardness obtained were satisfactory and would present little or no problem in use.

9. The levels of weldment properties obtained are similar to those already published in the literature, indicating that the present steels are within the same broad range. In addition, it is suggested that development of tougher weld metals would probably enhance these properties.

Table IX. Vickers Hardness of the Parent Plate, Weld Metal and Grain Coarsened Heat Affected Zone

Steel	Parent Plate	Weld Metal	GCHAZ
1	188	230	224
2	214	225	220
3	215	226	221
4	232	229	233
5	225	230	224
10	219	225	222
12	198	238	234

Table X. Comparison of Weld Metal and Heat Affected Zone Toughness

Reference	Absorbed energy at 0ºC, J	
	Weld metal	HAZ
This study	94 - 138	65 - 95 (no Ti) 95 - 120 (Ti)
9	50 - 150	
26	73 - 150	42 -236
11	25 - 125	245
27		28 - 40 (at -40ºC cf 26 - 50 in this study)

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