Proceedings of the International Seminar on Welding of High Strength Pipeline Steels
CBMM and TMS, 2013

WELDING OF NIOBIUM MICROALLOYED LINEPIPE STEELS: 50 YEARS OF HISTORY AND EXPERIENCE

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Keywords: Welding, Microalloying, Niobium, Linepipe Steels, History

Abstract

This paper provides a brief overview of the history of the use of niobium in linepipe since its first application in 1959. The dramatic changes in pipe chemistry that have been enabled by the introduction of niobium are highlighted, in particular lower carbon and carbon equivalent levels at ever increasing pipe strengths, and the way in which these developments have progressively improved weldability are critically discussed.

The important roles of advances in steelmaking, facilitating consistently lower carbon levels, and plate processing, such as TMCP, in recent decades are emphasized and the paper then proceeds to explain the way in which lower carbon allows the maximum benefit to be derived from increased levels of niobium. To emphasize the latter and to point the way towards the future, the success of the Chinese in adopting the latest metallurgical thinking through the application of vast tonnages of High Temperature Processed (HTP) steels containing carbon below 0.05% and niobium levels up to 0.11% is described.

The HTP steel concept has provided the world with a dramatically improved, more economical route to the production of pipe strengths up to X80 and beyond and combined this advance with exceptional weldability over a wide range of welding processes and cooling rates.

Introduction

The first niobium microalloyed steel was introduced in 1959 for the Michigan Wisconsin Pipeline System and it is still in service today! It was produced by Great Lakes Steel [1]. It was a semi-killed steel with an average carbon content of 0.24 percent and an average niobium content of 0.008 percent, Table I, which was heralded by a dear colleague of the author Jerry Barkow [2] as an eminently weldable steel, especially considering the conventional much higher carbon (C-Mn) steels that it replaced.

Table I. Chemical Composition of Semi-Killed Niobium Great Lakes Steel – 1959

Chemical Composition	**Weight Percent**		
	C	**Mn**	**Nb (Cb)**
Min.	0.20	0.89	0.004
Max.	0.27	1.28	0.015
Mean	0.24	1.10	0.008

Today we might be hard pressed to find welders who could weld this marvellous steel! In fact if we look at the "Weldability Diagram" created by Brian Graville [3], Figure 1, the Great Lakes Steel falls on the zone boundary between "Very difficult to weld" and "Weldable with care".

Over the last half century yield strengths have increased from the Great Lakes X52 grade to X80 and beyond and simultaneously carbon contents have been steadily reduced. Steel compositions have become more complex, with mechanical properties achieved by a combination of rolling and post rolling water cooling i.e. Thermo-Mechanical Controlled Processing or TMCP, first introduced by NKK in 1982, but based on research by the British Iron and Steel Research Association (BISRA) in 1967 [4]. During this period, steelmaking, casting and rolling practices have become more advanced and very low carbon steels (< 0.05 percent) can now be produced economically in large tonnages. A comparison of steelmaking technology 50 years ago with the present day is presented in Table II. BOF steelmaking and vacuum decarburisation treatments now permit the development of linepipe steels with carbon contents as low as 0.02 percent or formable sheet steels with carbon contents of 20-30 ppm.

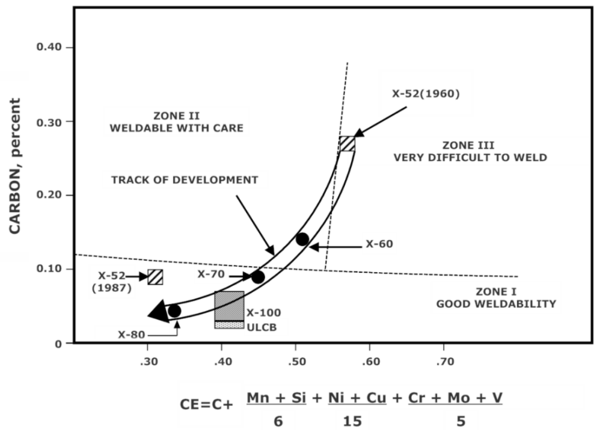


Figure 1. "Weldability Diagram" according to Graville [3].

Table II. 1959 – 2009 Contrasts

THEN	NOW
Semi-Killed Ingots	Concast Fully killed + calcium
Open Hearth	BOF + Ladle Metallurgy
Normal Rolling	TMCP
X52	X65 to X100
Integrated Mills	Non-Integrated
Onshore	Deep Offshore
China – Expanding existing plants	Building new plants

Weldability

The benefits of lower carbon contents on HAZ properties have been known for a long time [5, 6, 7] as shown in Figures 2, 3 and 4 and most high strength linepipe steels have carbon contents well below 0.10 percent, Figure 5 [8]. However, the welding process, which typically involves several thermal cycles, replaces the microstructure and mechanical properties that were produced during hot rolling with a vastly more complex range of microstructures formed in the overlapping thermal regions of multi pass welds, Figure 6 [9]. The response of each steel to a given thermal cycle depends on chemical composition and heat input and thus cooling rate. The metallurgical response is difficult to predict because every steel maker uses a different chemical composition and alloy design which affects the transformation and precipitation responses. Then there are complex interactions between chemical composition, precipitate dissolution and reprecipitation, heat input, and cooling rate that can produce a plethora of multi phase microstructures with varying hardnesses and toughness. Thus it is impossible to make generalized statements about weldment properties. Some of the metallurgical interactions are critically reviewed and discussed in Kirkwood's excellent paper at this seminar [10].

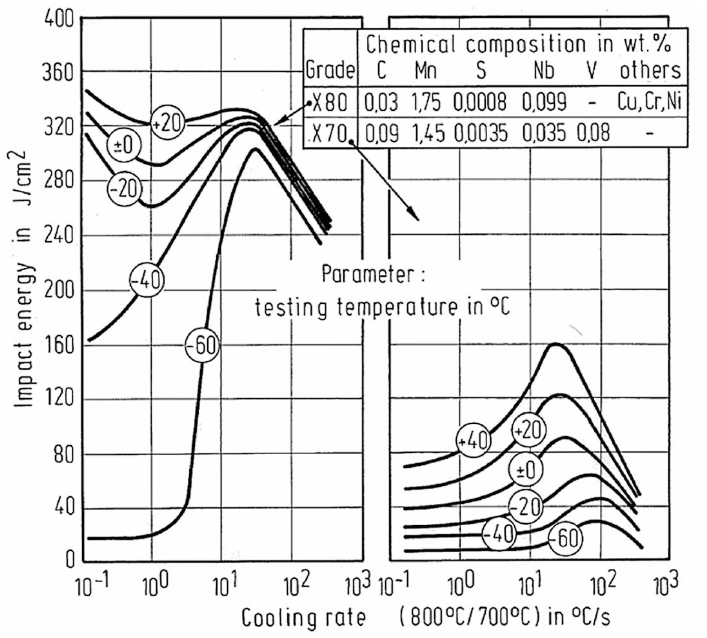


Figure 2. Comparison of HAZ toughness of conventional 0.09 percent carbon Nb-V steel and 0.03 percent carbon HTP (High Temperature Processed) steels in simulation tests [5].

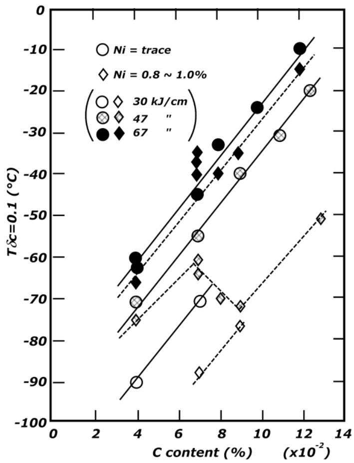
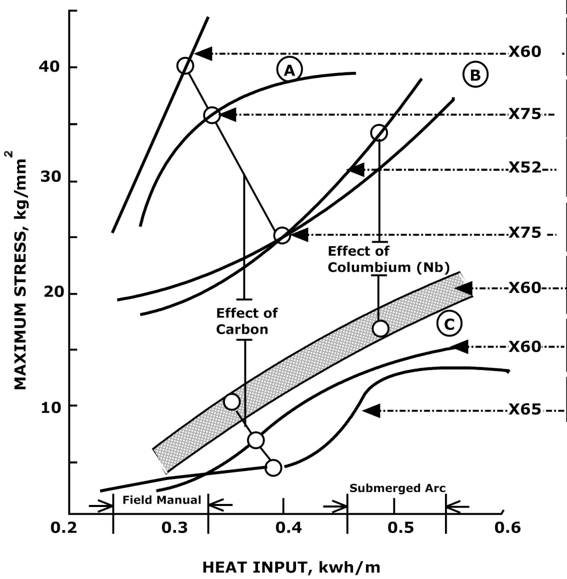


Figure 3. Effect of carbon content on HAZ Charpy V-notch transition temperature [6].



C	Mn	Mo	Cb (Nb)	V	C.E.
0.07%	1.40%	-	Total of 0.05 to 0.15%		0.31
0.06%	1.75%	0.30%	0.05%	0.05%	0.43
0.18%	1.20%	No Dispersoids			0.38
0.06%	1.95%	0.30%	0.07%	-	0.46
0.16%	1.30%	-	0.05%	•	0.38
0.17%	1.35%	-	0.033%	-	0.40
0.17%	1.47%	-	0.046%	-	0.42

$$C.E. = C + \frac{Mn}{6} + \frac{Mo+V}{5} + \frac{Ni}{15}$$

• V + N_2 alternative to Cb (Nb)

Figure 4. Implant test results for steels with various carbon contents [7].

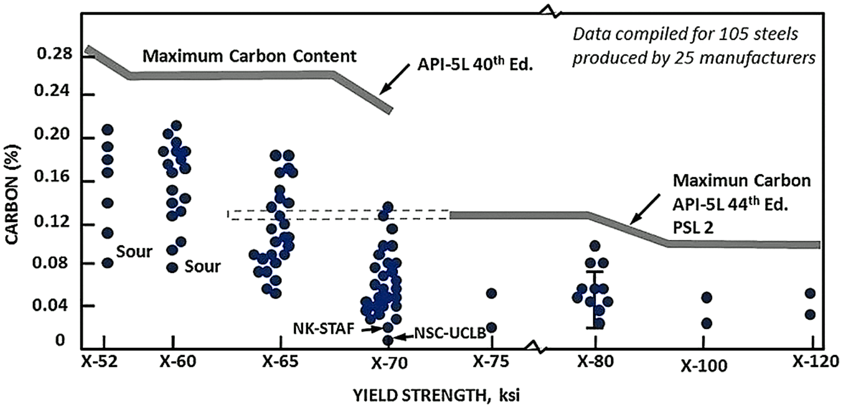


Figure 5. Trend of carbon content with pipe yield strength [8].

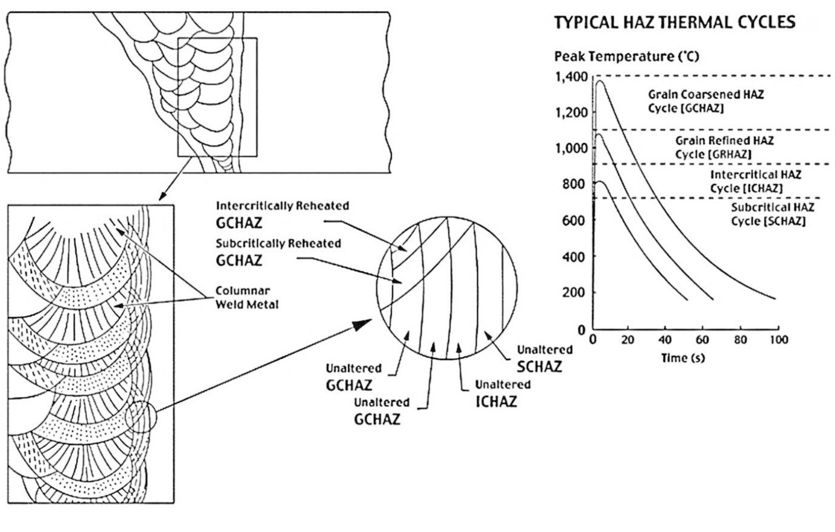


Figure 6. Range of overlapping thermally cycled regions in the weld HAZ [9].

The metallurgy of the HAZ cannot be summarized in simple terms, given the aforementioned interactions between grain growth, dissolution of microalloying solute, kinetics of austenite transformation and reprecipitation of carbo-nitrides, all of which vary with heat input and thus cooling rate. As a result a generic definition of weldability in terms of HAZ strength, hardness and toughness, either good or bad, is not realistic. The required microstructures and mechanical properties depend on the pipe application and service conditions. Opinions differ widely concerning metallurgical or engineering (fitness for service) solutions. Many welds with very poor HAZ toughness have been in service without event for countless years, presumably because of limited engineering loads, or overmatching weld metal as has been pointed out by Prof. Denys [11] for quite some time.

Nevertheless the objective of this seminar is to provide an opportunity to better understand all the competing factors so that the appropriate HAZ microstructures can be selected and optimized for each application. In this seminar there are many contributions which are using new tools such as EBSD, atom probe, as well as higher power electron microscopes. However, detailed fine scale investigations can be confusing if there is not a basic understanding of transformation behaviour, the bulk microstructures and morphologies of the components in these multi phase microstructures. Fortunately a "road map" was created by Kirkwood [12] and collaborators [13] almost 30 years ago, which is presented in Figure 7 [12].

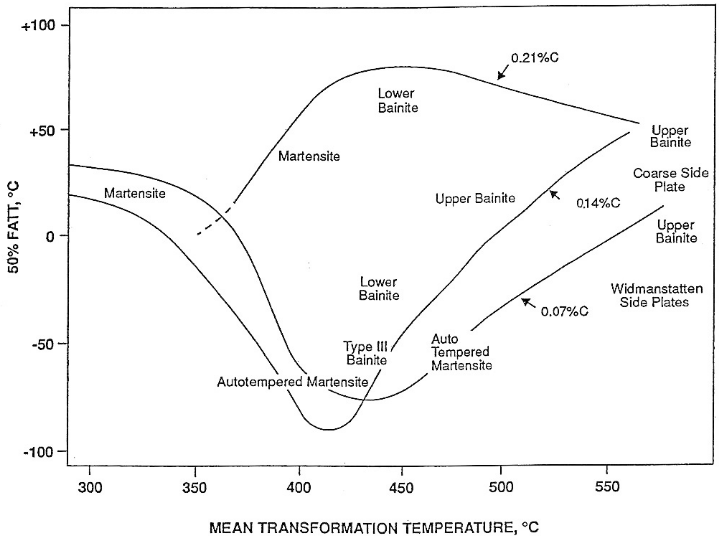


Figure 7. Effect of carbon content, and mean transformation temperature on HAZ 50 percent shear transition temperature [12, 13].

The diagram rationalizes the effects of carbon and austenite-to-ferrite decomposition temperature on HAZ microstructures. The ideas and concepts have been updated for this seminar and appear in Dr. Kirkwood's presentation [10]. The Kirkwood approach has further stimulated Prof. Subramanian to try to rationalize some of his new HAZ EBSD data which show the effect of heat input and the preservation of greater densities of high angle (> 45°) grain boundaries during the reverse transformation to austenite under certain conditions, Figure 8 in reference [14].

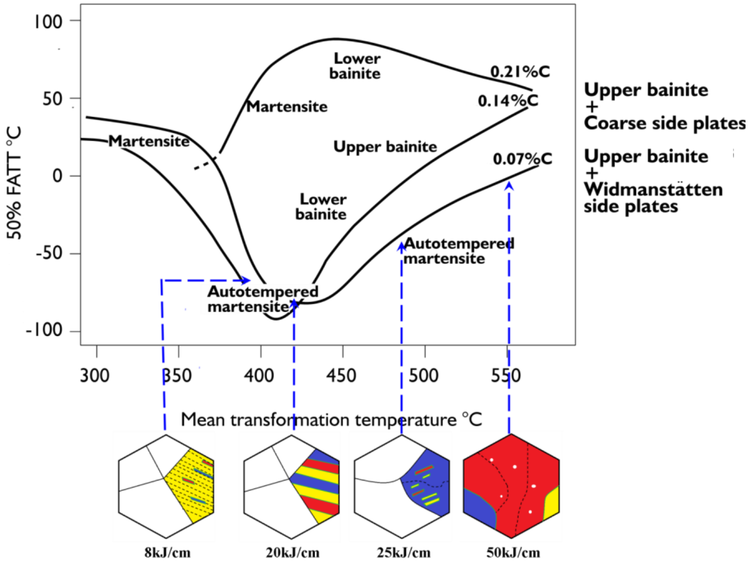


Figure 8. Range of microstructures associated with regions of high and low HAZ toughness[11].

Despite such progress in understanding transformation behaviour and the formation, morphology and significance of different phases in the HAZ, there is a growing semantics problem. In former days the terms diffusional and non-diffusional (shear) served us well. Now the term displacive has been put forward by Bhadeshia [15] to categorize non-diffusional transformation to Widmanstatten Ferrite, Bainite, and Martensite. Diffusional terminology includes Allotriomorphic or Proeutectoid Ferrite, Idiomorphic Ferrite, Massive or Irregular Ferrite and Pearlite, with proposals voiced at this seminar that acicular ferrite be reserved for weld metal microstructures nucleated intragranularly on non-metallic inclusions, Table III.

Table III. Terminology

DIFFUSIONAL	**NON DIFFUSIONAL (DISPLACIVE)**
ALLOTRIOMORPHIC FERRITE (PROEUTECTOID FERRITE) IDIOMORPHIC FERRITE MASSIVE FERRITE PEARLITE	WIDMANSTATTEN FERRITE BAINITE MARTENSITE
ACICULAR FERRITE NON-POLYGONAL FERRITE IRREGULAR FERRITE	

In addition to the primary phases mentioned above, a mixed martensite-austenite (MA) constituent can be found in weld metal and HAZ. The amount varies with chemical composition and cooling rate and the effects depend on size, distribution and hardness of the MA phase as well as the properties of the overall matrix (host phases). Haze and Aihara [16] have correlated CTOD toughness with volume fraction of MA constituent for high-heat submerged arc welds, Figure 9. Whilst the presence of MA constituent is generally considered undesirable, there is still considerable debate and confusion concerning the role of individual microalloying elements, particularly niobium, in affecting its formation. To put this in perspective there are also conflicting data and opinion concerning conventional elements such as manganese, nickel and carbon, again as discussed by Kirkwood at this seminar [10].

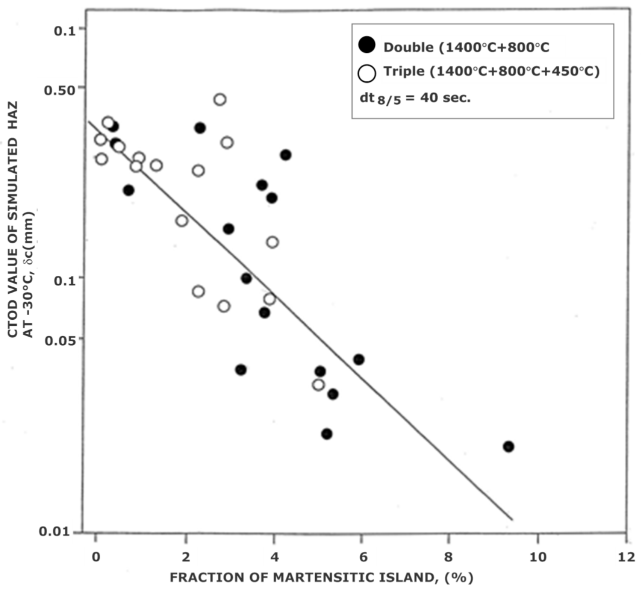


Figure 9. Effect of MA content on HAZ CTOD toughness in simulated welds [16].

There is no clear evidence correlating niobium content with the mechanism of MA formation whereas there is a direct correlation with carbon content and silicon content, Figure 10 [17], as well as with peak temperature reached in the intercritically reheated region. In the midst of this debate Kirkwood [10] presented mid-1970s' data developed by Hannerz [18] showing a beneficial effect of niobium in preventing HAZ grain coarsening, Figure 11 [18]. This grain refinement presumably improves the gross microstructure and affects the overall properties of the HAZ. Additional austenite grain coarsening data for samples held for much longer times than those associated with welding, show that the beneficial effect of niobium extends to 0.15 percent or more, Figure 12 [19].

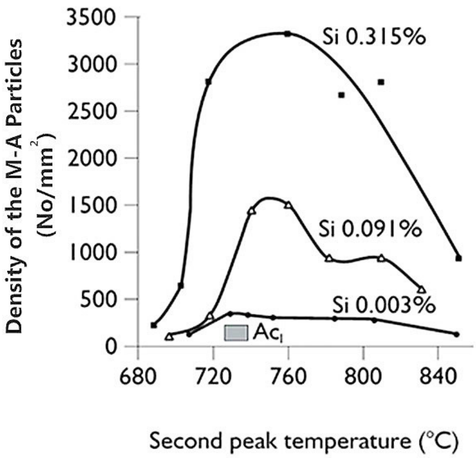


Figure 10. Effect of silicon on proportion of MA phase (Aspect Ratio>4; Lmax>2μm) in reheated HAZ regions [17].

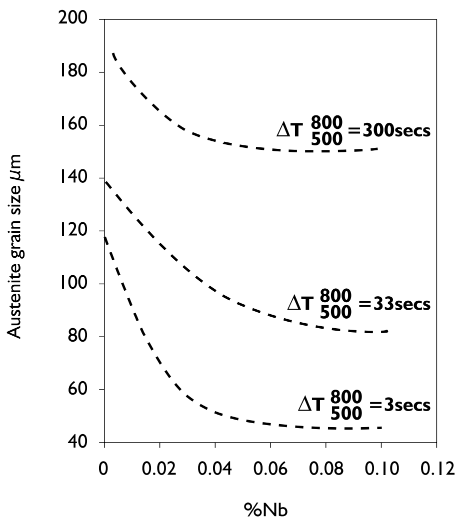


Figure 11. Effect of niobium content and cooling rate on austenite grain size in the HAZ [18].

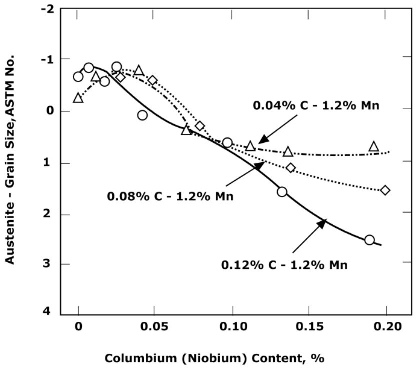


Figure 12. Austenite grain coarsening behaviour for steels reheated to 1200 °C for 1 hr and water quenched [19].

It is hoped that this seminar has led to a fertile interchange of ideas and opinions and to a better understanding of factors affecting the MA phase and its mechanism of formation. One particularly attractive contribution from Subramanian et al [14] is presented at this seminar, Figures 8 and 13. These authors have correlated the greater presence of high angle Bain variants with better Charpy toughness.

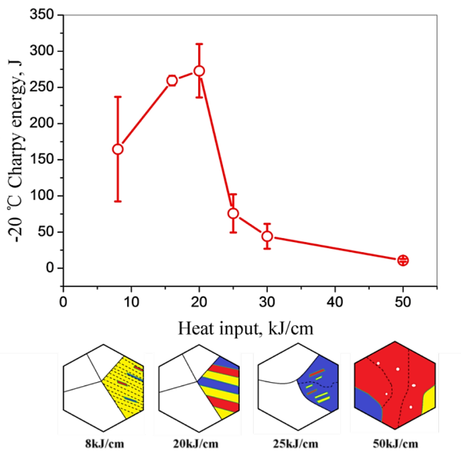


Figure 13. Density of high angle (> 45°) boundaries associated with high and low toughness microstructures [14].

The above discussion has centered on base-microstructural consideration, whereas in practice, parallel HAZ reactions involve dissolution and potential reprecipitation of the various carbide and nitride particles found in microalloyed steels. One can not generalize because, for example, titanium can be either beneficial when present as TiN, where it acts as a grain refiner, or detrimental at hyper stoichiometric levels (to nitrogen) where it forms titanium carbide. The effects of this, as well as the effects of vanadium and niobium, are additive to, or subtractive from, the influence of microstructural features on toughness.

The above discussion relates to the effect of the weld thermal cycles on HAZ properties. However, the properties of the unwelded pipe or plate should not be overlooked. This point is illustrated by analysis of a large volume of CTOD data developed in the onerous API RP2Z [20] HAZ and EEMUA 150 [21] prequalification testing procedures. It was found that when incoming (starting) plate properties were poor, such as when large TiN particles were formed during casting, or when carbon content was above 0.10 percent, and the resulting Charpy 50 percent shear, Fracture Appearance Transition Temperature (FATT) was above minus 55 °F, (minus 48 °C) Figure 14 [22] for a 4" (100 mm) thick plate, it was impossible to pass the API RP2Z Prequalification Test(s).

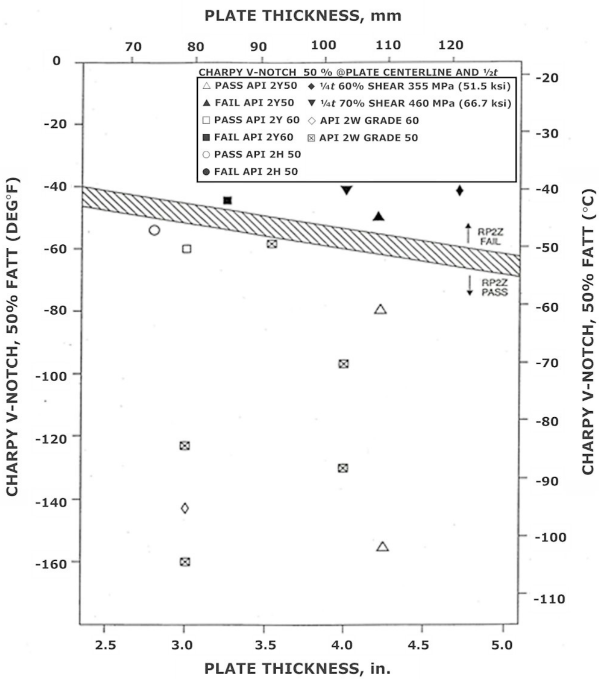


Figure 14. Effect of unwelded Charpy FATT on successful outcome of API RP2Z Prequalification Testing.

It is relevant to note Kirkwood's comments in his paper at this seminar, that when carbon contents are excessive, (> 0.14 percent), as in the case cited above, no amount of fiddling with welding procedures or alloying concept will produce high toughness HAZ microstructures. As far as the author is aware, there has been no systematic investigation into the effect of unwelded pipe properties on HAZ microstructures and toughness after welding. This would probably be a fruitful avenue for future research.

Since welding is not designed to improve HAZ microstructures, the best one can hope for is that there will only be moderate degradation. Since the RP2Z (heavy plate) prequalification concept is also being used to qualify the welding performance of heavy wall linepipe, especially for offshore applications, it is useful to recall its origin. HAZ cracking problems occurred in the early 1980s in steels being fabricated in Norway for an offshore structure. The cracking was related to high HAZ hardnesses, and complex, hard higher carbon phases such as MA in the intercritically reheated region of the weld, which gave rise to the term Local Brittle Zones (LBZs). The normalized, microalloyed steel which triggered the development of the API RP2Z and EEMUA 150 procedures had 0.14 percent carbon, exhibited centerline segregation and had a Charpy 50 percent FATT of -38 °F (-39 °C). Despite the alarm bells raised by the scattered hard microstructural phases, not everybody was concerned. In fact de Koning authored a paper entitled "Feeling Free with LBZs" [23] based on his assessment of the limited structural significance of such small brittle regions.

Application of Niobium Microalloyed Linepipe

Five decades of experience and discussion about the role of microalloying elements on weldability, specifically HAZ properties, have not prevented increasingly widespread adoption of microalloyed linepipe steels with niobium contents up to 0.11 percent or higher. A survey conducted in 1977 by the author [24], of prevailing niobium and carbon contents in X65 linepipe steel <u>installed</u> and <u>operating</u> in the UK, Canada and USA, revealed that relatively high niobium contents (up to 0.07 percent) combined with carbon contents of 0.18 to 0.20 percent, were in common use, Figure 15. Many of these steels undoubtedly had very poor HAZ notch toughness but the welds and pipelines have stood the test of time, and suggest that any present day preoccupation with small regions of MA phase may be overdone. Much of the linepipe being installed today utilizes the HTP (High Temperature Processing) alloy design which relies on up to 0.10 percent niobium, in combination with 0.03-0.06 percent carbon, see Figure 15, to develop an exceptional combination of strength and toughness. The trend in maximum niobium contents and carbon with time, since 1960 is presented in Figure 16.

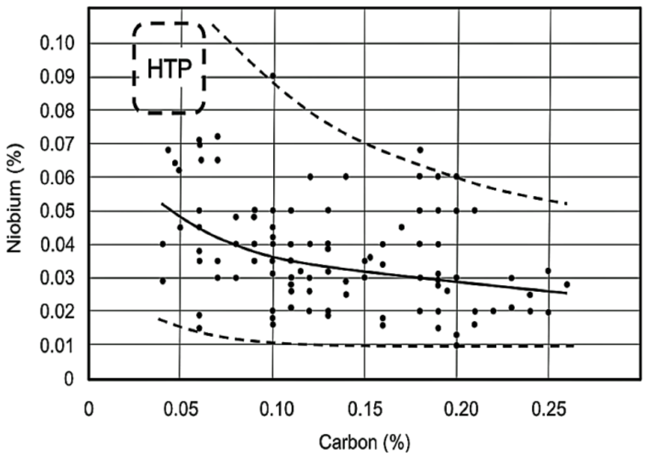


Figure 15. Combinations of niobium and carbon contents in operating pipelines in N. America, and UK circa 1977 [24, 27].

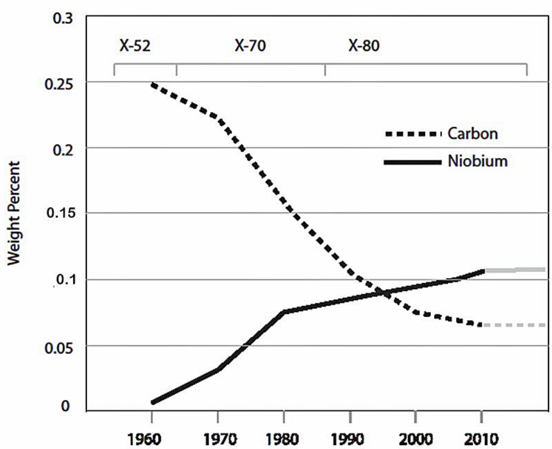


Figure 16. Changes in maximum niobium and carbon contents with time (and strength increase).

Table IV. Chemical Compositions of Chinese Hot Rolled Coils API Grade X80 [25]

Mill Name	**C**	**Mn**	**Cr**	**Mo**	**Nb**	**V**	**Ti**	**N**	**Thick-ness**
N. China Petroleum	0.06	1.88	-	0.33	0.056	-	0.023	0.005	14.6 mm
	0.07	1.89	-	0.24	0.055	0.05	0.011	0.004	14.6 mm
Jining	0.04	1.80	-	0.28	0.070	-	0.011	N.R.	18.4 mm
From TGRC Paper*	0.046	1.81	0.18	0.31	0.062	0.005	0.009	N.R.	18.4 mm
Angang	0.04	1.88	0.27	0.10	0.10	-	0.012	0.005	18.4 mm
Shougang	0.04	1.80	0.30	0.15	0.095	-	0.015	N.R.	18.4 mm
Nanjing	0.045	1.82	0.27	0.12	0.092	-	0.012	N.R.	18.4 mm

* Tubular Goods Research Center of the China National Petroleum Corporation

N.R. = Not Reported

Very large quantities of HTP X80 steel (> 2.0 million tons) have been installed in the CNPC Second West-East Pipeline [25]. Chemical compositions of the steel are presented in Table IV and a pipeline route map is presented in Figure 17. The pipeline is 48” OD x 18.4 mm, X80, with the main line having a length of 4772 km and utilizing 74 percent spiral seam and 26 percent longitudinal seam linepipe. Technical data relating to the metallurgy and welding of the plate and skelp used to manufacture the pipe are being presented at this seminar [26], selected data are reviewed below.

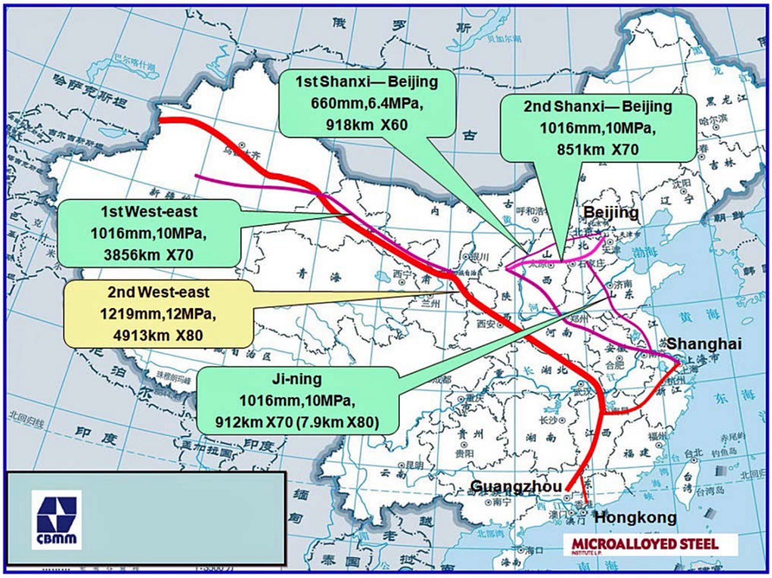


Figure 17. Map showing the Second West East Pipeline built by CNPC.

Studies of grain coarsening in single pass simulated welds were carried out on the steels shown in Table V [26].

Table V. Chemical Compositions of Experimental X80 Pipeline Steels (wt%)

No.	C	Si	Mn	Nb	Mo	Cr + Cu + Ni and others
1	0.08	0.17	1.64	0.05	0.21	$\geq$ 0.50
2	0.06	0.18	1.79	0.09	0.24	$\geq$ 0.50

The austenite grains coarsened more in the lower niobium steel, which is in line with data presented in Figures 11 and 12. Simultaneously the width of the grain coarsened region was reduced from 225 to 200 μm in the 0.09 percent niobium steel, Figure 18 [26].

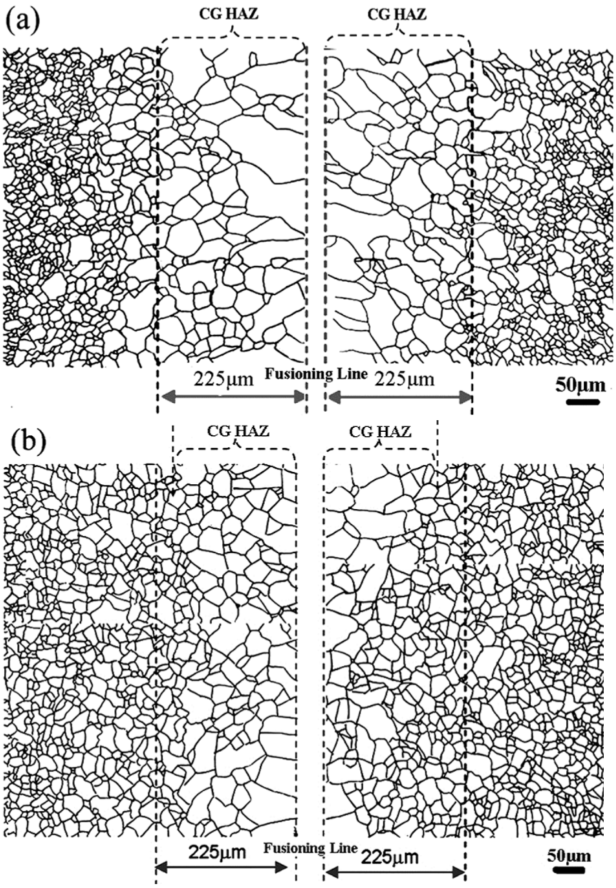


Figure 18. Distribution of prior austenite grains in actual HAZ (a) medium Nb steel (0.05 wt%); (b) higher Nb steel (0.09 wt%) [26].

Spiral seam HTP linepipe installed in the Second West-East Pipeline was manufactured by four separate mills, using representative skelp compositions shown in Table VI.

Table VI. Typical Composition of X80 Hot Strip Steel

	C	Mn	Si	P	S	Nb	V	Ti	Mo	Ni	Cr	Cu	Nb+V+Ti
Max.	0.06	1.88	0.24	0.01	0.003	0.10	0.02	0.020	0.24	0.26	0.27	0.25	0.13
Min.	0.03	1.76	0.16	0.01	0.001	0.06	0.00	0.008	0.02	0.15	0.01	0.02	0.10
Ave.	0.05	1.83	0.19	0.01	0.002	0.08	0.01	0.012	0.21	0.20	0.20	0.15	0.11

Charpy toughness data for the pipe body, weld metal and HAZ for four separate mills are presented in Tables VII, VIII and IX below:

Table VII. Charpy Impact Test Data of Pipe Body Made of Hot Strip Steels Supplied by Four Mills

Testing Temperature (-10 °C)	**Impact Energy (J)**			**Shear area (%)**		
	Min.	**Max.**	**Ave.**	**Min.**	**Max.**	**Ave.**
A (192 sets)	262	464	352	96	100	100
B (41 sets)	272	441	353	100	100	100
C (715sets)	251	497	352	86	100	100
D (349 sets)	215	477	343	83	100	100
Project Acceptance Criteria	Single $\geq$ 170J Average $\geq$ 220J			Single $\geq$ 80% Average $\geq$ 90%		

Table VIII. Charpy Impact Test Data of Weld Seam, with Steels Supplied by Four Mills (A,B,C,D)

Testing Temperature (-10 °C)	**Impact Energy (J)**			**Shear area (%)**		
	Min.	**Max.**	**Ave.**	**Min.**	**Max.**	**Ave.**
A (192 sets)	61	221	162	37	100	70
B (41 sets)	82	221	166	42	98	71
C (715sets)	66	241	167	38	100	72
D (349 sets)	72	265	158	41	100	68
Project Acceptance Criteria	Single $\geq$ 60J Average $\geq$ 80J			Single $\geq$ 30% Average $\geq$ 40%		

Table IX. Charpy Impact Test Data of HAZ, with Steels Supplied by Four Mills

Testing Temperature (-10 °C)	**Impact Energy (J)**			**Shear area (%)**		
	Min.	**Max.**	**Ave.**	**Min.**	**Max.**	**Ave.**
A (192 sets)	90	291	205	42	100	90
B (41 sets)	73	285	211	43	100	91
C (715sets)	62	295	206	38	100	90
D (349 sets)	62	296	199	41	100	96
Project Acceptance Criteria	Single $\geq$ 60J Average $\geq$ 80J			Single $\geq$ 30% Average $\geq$ 40%		

The results are considered excellent with no noticable effect of niobium content on weld metal and HAZ toughness. Additional excellent data for longitudinal seam linepipe is also presented by Shang et al [26] and readers are referred to that paper in these proceedings.

Summary and Conclusions

The utilization of niobium in linepipe steels began in 1959 and increased steadily throughout the past 53 years. Today niobium is considered an essential element when steels are produced by the TMCP route. Strengths have increased from X52 in 1959 to X80 and above today. Simultaneously niobium contents have increased twelve-fold from 0.008 to 0.10 percent and carbon contents have been reduced from 0.25 percent to 0.03 - 0.05 percent, the latter being made practical by introduction of oxygen steelmaking processes.

Hydrogen assisted cold cracking is no longer a concern even when using cellulosic (high hydrogen) electrodes and the term weldability is now associated with sophisticated issues related to HAZ microstructures and local properties. It has been demonstrated that niobium can be usefully employed in amounts at least up to 0.11% when carbon is reduced below 0.05% and when both carbon and silicon are simultaneously reduced to their lowest levels, commensurate with other requirements, optimum toughness is recorded in the coarse grained and intercritically reheated regions of the weld heat affected zone. Millions of kilometres of niobium containing linepipe have been successfully installed and operated since 1959 when niobium steels were first introduced. Perhaps the crowning glory is the successful adoption of the 0.10 percent, niobium, “near stoichiometry” steel [27] in the 9200 km China National Petroleum Second West-East Project [25]. Ongoing studies of the weld HAZ region will undoubtedly lead to better understanding of the microstructures and properties and optimization for particular circumstances and applications.

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