THE WELDABILITY OF MODERN NIOBIUM MICROALLOYED STRUCTURAL STEELS

P. Kirkwood

The Old School House, Cresswell, Northumberland, NE61 5JT, England

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Abstract

This paper focuses on the concept of weldability in the context of modern structural steels and contrasts where we are today with the situation in earlier decades. The increasing trend towards lower carbon levels has dramatically opened the door to more effective utilisation of the unique benefits that can be derived through the use of niobium microalloying and my paper addresses the question; “What are the implications for weldability, if any?” The structural segment of the global steel market is of increasing strategic importance and a thorough understanding of all the issues relating to the development of optimum and progressive specifications is essential if the full benefits of niobium microalloying are to be realised for the benefit of the world’s steelmakers and fabricators. This paper explores these issues.

Introduction

Whilst niobium has become an almost irreplaceable microalloying element in HSLA steels for linepipe applications [1] and has, more recently, established a particularly significant niche for X80 pipe production through the effective utilisation of the low carbon (0.04%), higher niobium (0.1%) High Temperature Processing (HTP) route [2], it remains the case that about 60% of total world steel production is accounted for by the structural steels market. The strategic importance of this segment of the global market has been dramatically increased by the rapid worldwide expansion of the developing economies and this is already having a significant impact on world steel consumption. At present, only some 10% of this structural market is made up of steels microalloyed with niobium [3]. However, the potential benefits of higher niobium levels are now being more widely understood, particularly in conjunction with the significantly lower carbon levels of modern steels, and the possible overall environmental savings of using higher strength structural steels has introduced a factor, which has previously been inadequately appreciated [4].

It is unfortunate that one of the factors which has inhibited the more effective utilisation of niobium in a range of structural steel products and applications has been a technically unjustified concern about how the element affects certain aspects of weldability and although such myths have more recently been effectively dispelled [5] the legacy of decades of prejudice lingers on in countless worldwide specifications. This, in turn, is preventing steelmakers and fabricators enjoying the full benefits of niobium microalloying. It is, therefore, an opportune and appropriate time to re-examine the role of niobium in the welding of the new generation of structural steels.

Weldability

It is pertinent to address the question - what do we actually mean by the term weldability? The literature abounds with a variety of definitions but I found myself dissatisfied with most of them. For the purposes of this paper I have, therefore, prepared one of my own viz;

The ability of a material to be joined using a wide range of appropriate welding processes to produce joints effectively free from significant defects and strong enough and tough enough, in all areas of those joints, to be fit for the purpose intended.

The role of niobium in the welding of microalloyed steels was comprehensively reported in 1981, in the era of higher carbon steels [6], and more recently reconsidered in detail in 2001[7]. Unfortunately some of the key messages of those papers have, apparently, been all too readily forgotten and it is essential that we re-examine the subject in the light of the improved understanding of the fundamental metallurgy of how niobium behaves in HSLA steels and their weldments. As already alluded to, the trend towards much lower carbon contents has dramatically changed our perception of the benefits to be gained from an increased use of niobium but it is also important to understand the additional benefits which have accrued from steelmaking developments and the advent of ‘clean’ steelmaking practices.

In the following paragraphs I will contrast how the focus of what is understood by the term “weldability” has changed in the last forty years and how the emphasis has shifted from types of welding defects which were prevalent with the older high carbon equivalent steels to those which are more often encountered in the new generation of lower carbon equivalent structural steels.

Steels of Earlier Decades

In the early 1970s and even considerably later, structural steels, in many product forms, with yield strengths around 340 N/mm^2 would have, typically, had a chemistry as shown in Table I.

Table I. Typical Chemistry of Early 1970s’ Vintage Structural Steel wt.%

C %	**Mn %**	**Si %**	**S %**	**P %**	**Nb %**	**Al %**
0.23	1.56	0.50	0.032	0.025	0.035	0.04

With such combinations of carbon, sulphur and phosphorous and with steel processing routes of that era it was not uncommon to find significantly banded microstructures with copious elongated ‘Type II’ manganese sulphide inclusions [8]. These microstructural features in high carbon material resulted in three particularly unwelcome defects in welded joints.

Liquation Cracking

Even moderate heat input (3 kJ/mm) butt welds often exhibited a defect known as liquation cracking. As its name suggests this occurred simply because the heat of welding partially melted some inclusions and delamination at the steel/inclusion interfaces resulted. See Figure 1.

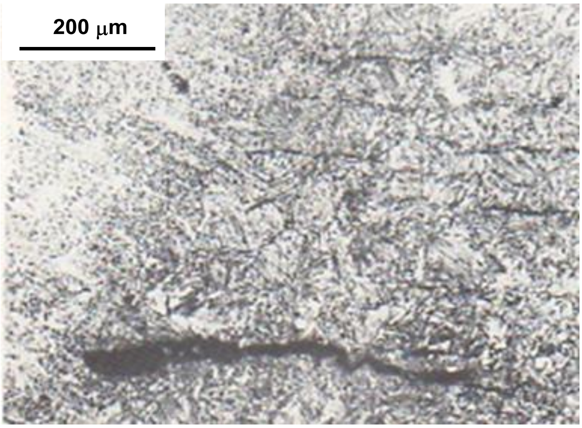

Figure 1. Liquation cracking in an HSLA steel butt weld, circa 1971.

Fortunately this is not a defect that today's generation of welding engineers are likely to encounter but it is, nevertheless, instructive to remind ourselves of what earlier generations had to contend with. *Clearly niobium plays no role in this particular cracking mechanism*.

Lamellar Tearing

The same elongated inclusions, which were responsible for liquation cracking, also caused a quite common, and potentially damaging, type of defect known as lamellar tearing. This, as shown in Figure 2, was quite regularly encountered in T-butt welds and resulted because of the build-up of through thickness stresses on the base plate to which the attachment was being added.

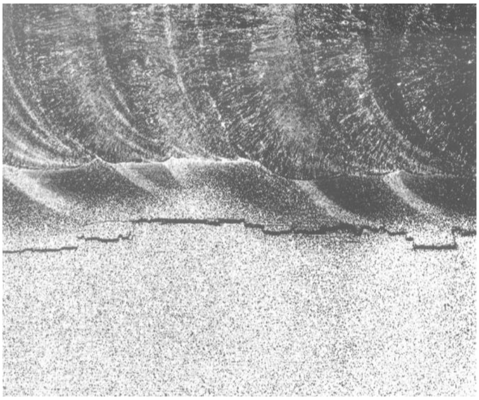

Figure 2. Lamellar tearing in a T-joint weld, source TWI [9].

Defects of this nature were responsible for the evolution of steels with much lower sulphur and phosphorous levels and for the advent of inclusion shape control through 'ladle' calcium injections. This was particularly important in node joint steels for offshore structures and a through thickness tensile test ductility criterion was introduced into specifications to ensure that steelmakers were producing plate products which would be resistant to this problem [10].

Once again, this is a problem that is not influenced in any way by microalloying.

Solidification Cracking

A further common defect in these higher carbon steel weldments was solidification cracking, that is cracking within the weld bead itself due to lack of high temperature ductility during solidification. Figure 3 illustrates the common (but not unique) location of such defects.

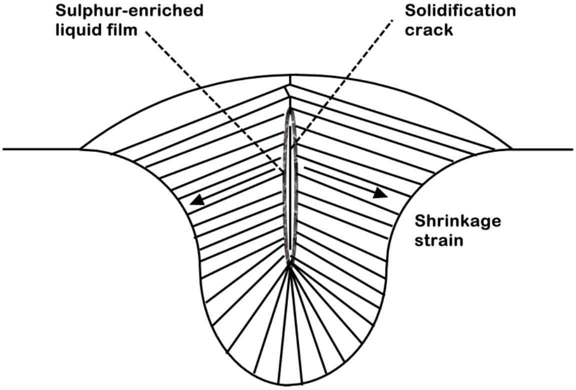

Figure 3. Schematic of solidification crack within a weld bead - image adapted from: http://offshore-tech.blogspot.co.uk/2011/01/crack.html.

Whilst centreline solidification cracking can in some cases be triggered by weld bead shape and profile [11], it is chemical composition which primarily controls the incidence of the phenomenon. The mechanism of this form of cracking was very thoroughly researched many years ago [12] and, for the present purposes, it suffices to note that it is caused by inter-dendritic segregation effects in the solidifying weld metal and internal stresses resulting from weld bead cooling. Studies on a wide range of C-Mn and microalloyed steel welds using transverse restraint testing [12] resulted in a cracking susceptibility formula which appeared, at the time, to indicate the relative importance of the key chemical elements. UCS stands for units of crack susceptibility.

$$\text{UCS} = 230\%\text{C}^* + 190\%\text{S} + 75\%\text{P} + 45\%\text{Nb} - 12.3\%\text{Si} - 5.4\%\text{Mn} - 1 \qquad (1)$$

If C <0.08, C* was taken as 0.08%

Although it depended to a degree on the type of welded joint, values of UCS >30 were considered to represent poor resistance to solidification cracking whilst values <10 suggested excellent resistance [11].

In this context, it is noteworthy that the early 1970s' steel of Table I produces a calculated value of UCS of 46.9 which clearly suggests significant susceptibility. Table II is useful and highlights the dominating effects of carbon and sulphur and also explains why this type of cracking is much less likely to be encountered in steels of the current era.

Table II. UCS Values for Selected Steels

Steel Type	**Units of Crack Susceptibility (UCS)**
1970s' vintage structural steel from Table I	46.9
Typical 1980s' low sulphur node steel (C = 0.14%, S = 0.004%)	22.6
Example of an HTP low carbon, higher niobium linepipe steel (C = 0.04%, S = 0.002%)	13.3*

*This figure was calculated using C = 0.08%, as per the UCS formula, but the actual steel only contained 0.04%C.

As any steelmaker will tell you, many elements like C, Mn, S and Nb can segregate during steel casting and the effect in solidifying weld metal is analogous. However, the effects of carbon and sulphur are so overwhelming that it can, categorically, be concluded that niobium is of no concern in the context of weld metal solidification cracking. Indeed it has been shown for HTP low carbon, higher niobium steels that segregation, even during steel casting, is minimal and readily controllable [13].

At this juncture, we can, therefore, conclude that niobium has played no significant role in the formation of any of the welding defects described to date and that, with the advent of lower carbon contents and cleaner steels, liquation cracking, lamellar tearing and solidification cracking are problems unlikely to be encountered by today's welding engineers.

We can, therefore, proceed to consider one final important weld cracking issue which was prevalent in the past and which, together with two other specification issues, is still relevant to the much improved structural steels which fabricators are presented with today. I refer to hydrogen induced cold cracking and will deal with this in the following section.

Modern Structural Steels

By the early 1980s, the weldability benefits of lower carbon and cleaner steel practices were becoming much more widely appreciated [10] and this was not only in the context of the defect types discussed to date. Perhaps the most important welding problem of all time, in structural steels, has been hydrogen induced cold cracking. This is a defect which usually manifests itself in the heat affected zone (HAZ) adjacent to completed lower heat input welds, often manual metal arc fillet welds, in thick section materials where inadequate weld pre-heat has been applied. It is now widely appreciated that such cracking is initiated by a combination of three factors:

Hydrogen Induced Cold Cracking

- The presence of hydrogen from extraneous sources (eg moisture in welding consumables);
- Stress from highly constrained or complex joints (usually in thicker sections);
- A susceptible HAZ microstructure (usually containing martensite).

Figure 4 shows the typical appearance of this type of cracking.

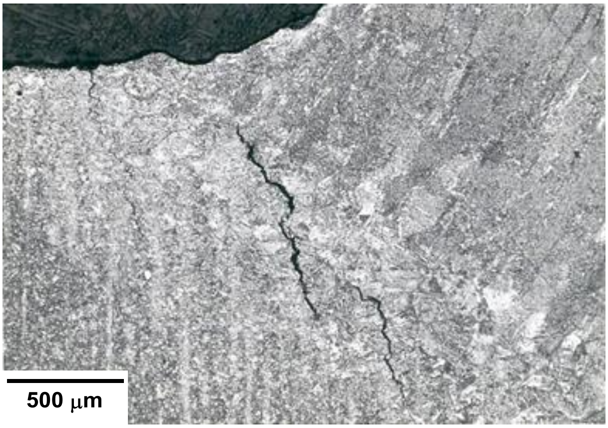

Figure 4. Typical hydrogen induced HAZ cold cracking.

In this paper we are primarily concerned with the role of chemistry in the generation of a susceptible microstructure. It is widely considered that susceptible microstructures are hard martensitic microstructures associated with fast thermal cycles which are encountered, for example, during manual metal arc welding. Maximum hardness is believed to be dictated by carbon content but hardenability is conferred by a combination of elements. The contribution of all the other elements is normally made through the use of a “carbon equivalent” formula which, in effect, is a numerical expression which attempts to assess the overall contribution of chemistry to hardenability. The most commonly used, and long-standing, equation for C-Mn steels, is that employed since the 1960s by Lloyd's Register for the assessment of shipbuilding steels and shown as Equation 2 [14].

$$CE = C + \frac{Mn}{6} + \frac{(Cr + Mo + V)}{5} + \frac{(Ni + Cu)}{15} \qquad (2)$$

It was considered that if a steel had a CE >0.45, based on a ladle analysis, then low hydrogen electrodes and preheating would be required to avoid cold cracking. On the other hand if CE was <0.41 then the precautions described above could probably be avoided except in conditions of extremely high restraint. This formula was generally expected to be applicable for any carbon – manganese type steel containing "minor quantities of grain refining elements, for example niobium, vanadium or aluminium." Experience in the late 1960s confirmed that, while CE was not a precise measure of weldability, it provided an excellent control guideline [15].

Historically, in the older generation of structural or ship building steels, CE values well above the 0.45 figure were frequently encountered and, for example, the steel composition in Table I provides a calculated CE figure of 0.49 from carbon and manganese alone; there would, inevitably be some additional contribution from the other alloying elements not included in Table I so this is a very high CE steel. By the early 1970s correlations between 'Lloyd's' CE, hydrogen potential and cracking susceptibility, as measured using Controlled Thermal Severity (CTS) tests, were available and the whole approach became more scientifically founded [16]. Over the years the methodology has been further refined and the nomograms, now available in BS EN 1011-2: 2001 [17], employ CE to provide fabricators with reliable guidance for the avoidance of hydrogen cracking. The advent of lower carbon steels, particularly those with carbon below 0.1%, such as HSLA steels for linepipe or thermomechanically controlled processed steels (TMCP), has led some workers to prefer an alternative carbon equivalent formula which can also be used to predict weldability through correlations with implant tests. This new formula is often referred to as the Ito Bessyo equation, Equation 3 [18].

$$P_{CM} = C + \frac{Si}{30} + \frac{(Mn + Cu + Cr)}{20} + \frac{Ni}{60} + \frac{Mo}{15} + \frac{V}{10} + 5B \qquad (3)$$

Using the Ito Bessyo formula, a typical medium carbon offshore structural steel would have a P_{CM} of about 0.26. There are many other CE formula variants and while some have incorporated a small factor for niobium [19] the commonly used ones do not.

But doesn't niobium also increase hardenability?

Well yes, it does, but the following diagram, Figure 5, derived from data in various published sources [5,20,21] provides an explanation for this apparent anomaly.

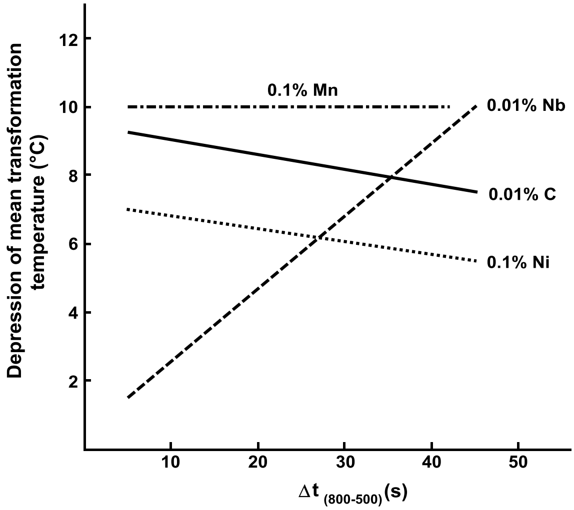

Figure 5. The effect of various important elements on HAZ transformation temperature.

It is important to be clear about what this diagram is portraying. At each weld cooling rate, as measured by $\Delta t_{(800\text{-}500)}$, the individual lines show the depression of mean transformation temperature that would be expected from the small percentage of alloying element indicated. So, for example, to establish the effect of 0.06%C, at a particular cooling rate, the figure read from the vertical axis of Figure 5 should be multiplied by six.

This is an extremely useful diagram derived as it is from real data. The dominant role of carbon is emphasised and the importance of manganese and nickel are also apparent very much in line with what we might expect from the Lloyd's CE prediction. However, it is the line illustrating the niobium effect which is most remarkable. Confirming practical observations and theoretical analysis, the extremely powerful effect of niobium on HAZ hardenability is confirmed. In contrast with the conventional alloying elements and carbon however, Figure 5 reveals that the effect of niobium is progressively more potent as cooling rate **decreases**, that is at higher welding heat inputs. In fact at the fastest cooling rates, that is with the lowest heat input welds, niobium has little impact on transformation temperature. Thus in low heat input manual metal arc welding, where hydrogen induced cracking is most frequently a potential problem, niobium has no significant effect on martensitic hardenability.

This confirms a conclusion I reached many decades ago without fully understanding why [6] but it is now widely appreciated that niobium exerts its hardenability effect when it is in solution in austenite and not when its fine precipitates remain undissolved at lower heat inputs. In fact the grain refining effect of fine dispersions of Nb(CN) may actually result in finer colonies of martensite in the HAZs of low heat input welds in niobium microalloyed steels which may therefore actually be more resistant to cold cracking than even their CE value predicts.

We can conclude therefore that niobium has no adverse role to play in HAZ hydrogen induced cold cracking and may even be beneficial.

Hydrogen can also, in certain circumstances, result in transverse **weld metal** cracking sometimes referred to as chevron or staircase cracking because of its unique and characteristic appearance on longitudinal sectioning [22]. Cracks are usually angled at approximately 45 degrees from the plane of the plates in a butt joint. The precise mechanism is often hotly debated but the phenomenon is usually associated with significant stress in under-matched weld metal where hydrogen removal precautions have not been strictly adhered to. *Weld metal chemistry is not considered to be a factor of any specific importance.*

Weld Metal Toughness

By the late 1960s and early 1970s the advantages of niobium treated steels were widely recognised for shipbuilding and general structural steel applications but, **in spite of their trouble-free fabrication history**, the effect of niobium on weld metal toughness suddenly became an issue. The submerged arc welding process, with its associated higher heat inputs, was being increasingly employed and in some instances certain researchers attributed difficulty in obtaining adequate weld metal Charpy V-notch toughness to the presence of niobium [23-25]. In reality there was little definitive evidence to substantiate these claims, but until much more systematic work was carried out a few years later [26], a degree of credence was associated with the claims being made. In fact, before long, the matter was debated by the International Standards Organisation and, on the basis of fairly flimsy evidence, it was proposed that niobium levels should be limited to 0.06% in a range of structural steel specifications [27].

In spite of the improved understanding which was available by 1974 [26] individual workers continued to attribute difficulties to the presence of niobium. In fact, quite interesting work at the Polish Welding Institute in 1976 [28] actually resulted in a new formula which was intended to predict the “weldability” of C-Mn-Nb steels. Good weldability in the context of this work actually referred to the ease of achieving adequate sub-zero Charpy properties in high heat input submerged arc welds. The weldability index C_W was proposed as follows:

$$C_W = C\% + \frac{Mn\%}{10} + 3Nb\% \qquad (4)$$

As an example of the use of this formula it was suggested [28] that, if $0.35 < C_W < 0.45$, the heat input of submerged arc welding should be restricted to 3.5 kJ/mm max.

In my experience it is foolish to dismiss systematic work of any origin if its experimental validity stands up to scrutiny and this work had a defendable experimental design and was executed professionally. It is therefore important to examine just what it is telling us and what it might still mean for steels of the present generation. I therefore prepared the following Table III.

Table III. The Effects of Carbon and Niobium on C_W at 1.4% Mn

C %	Mn %	Nb %	C_W
0.18	1.40	0.033	0.42
0.10	1.40	0.06	0.42
0.04	1.40	0.08	0.42

It is clear that the same 0.42 value of C_W can be achieved by many different combinations of carbon and niobium but, with hindsight, I suspect what the Polish work is telling us is that high carbon levels do make the achievement of adequate weld metal toughness more difficult in niobium treated steels. There are parallels here between these observations and work on the effects of niobium on HAZ toughness which we will discuss in the next section of this paper. The critical factor is transformation temperature and it is probable that in the Polish work the particular combinations of carbon and niobium studied led, at the slower cooling rates associated with submerged arc welding, to side plate ferrite or classical upper bainitic microstructures.

However, more recent experience tells us that the range of validity of the C_W formula must be limited to the older steels examined at the time because we know that modern steels of the low carbon higher niobium type (like Example 3 in Table III) can be submerged arc welded with ease to produce very high levels of Charpy V-notch weld metal toughness. This is not really surprising because we now know that much lower carbon levels permit the retention of much more niobium in austenite which will lead to a larger effect on weld metal hardenability than the Poles were experiencing in their 1976 work on higher carbon steels.

There have been major advances in our understanding of what produces good weld metal toughness in microalloyed steel weld metals. The objective is to produce a predominantly acicular ferrite microstructure which forms at an intermediate transformation temperature, Figure 6.

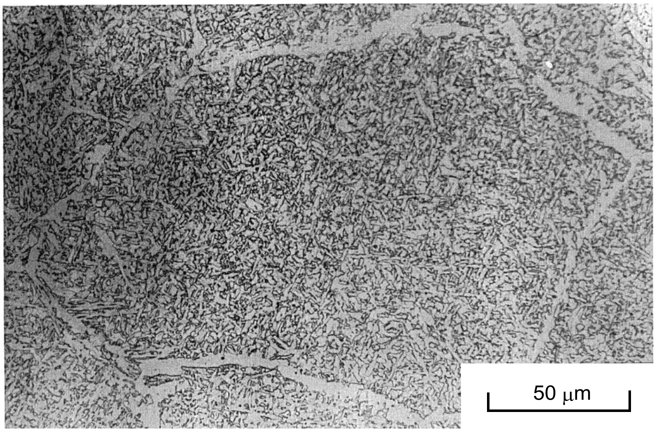

Figure 6. Intragranularly nucleated weld metal acicular ferrite.

The key to the production of acicular ferrite, which it is now known nucleates intragranularly [29], is an adequate dispersion of fine oxide or complex inclusions of a size suitable to act as nucleation sites for this type of transformation product. In addition, it is necessary to have the correct balance of other elements which control transformation temperature. Inadequate alloying will result in an excess of polygonal ferrite and too much alloying may lead to solid solution hardened acicular ferrite with retained MA constituent between the bainitic ferrite laths. This **may,** in some circumstances, reduce toughness.

In the latter context, niobium has a key and unique role to play. Because of its powerful hardenability effect, as shown for the HAZ in Figure 5, niobium facilitates the achievement of adequate weld metal hardenability (by dilution from the parent material) to ensure that acicular ferrite is readily obtained. The effect is similar to that demonstrated many years ago for Mo-B-Ti alloyed submerged arc weld metals [26].

Thus in the context of the achievement of good weld metal microstructure and toughness, niobium, in the new generation of lower carbon steels, is a positive benefit providing an excellent opportunity to manipulate transformation temperature to maximum advantage. It may even be appropriate, in some instances, to consider the development of welding wires which deliberately set out to increase weld metal niobium levels above those available by dilution from the parent steel.

Heat Affected Zone Toughness

Finally, we can turn our attention to the area which has, perhaps, proved most contentious over the years.

What is the effect of niobium on heat affected zone (HAZ) toughness?

The myth that niobium is specifically detrimental to the toughness of the heat affected zone (HAZ) in HSLA steels has recently been systematically investigated and dispelled [5]. However, for completeness, the key features of the latter work are repeated herein. Many individual areas of the HAZ can be defined, but the two of most immediate interest are the Coarse Grained Heat Affected Zone (CGHAZ) and the Intercritically Reheated Grain Coarsened Heat Affected Zone (ICGHAZ). These are considered separately below.

Coarse Grained Heat Affected Zone Toughness. It is widely accepted that the toughness of the CGHAZ is controlled by microstructure which, in turn, is critically dependent on composition and cooling rate. The latter parameters control transformation temperature and, therefore, the nature of the resulting microstructure. Previous work has suggested that optimum microstructure is achieved at different transformation temperatures for different carbon contents [30] as portrayed in Figure 7. It must be stressed that the model described in this diagram is only indicative of trends and every individual steel composition will behave slightly differently. The role of niobium is best considered in the context of such diagrams and clearly the secret of success is to utilise the hardenability effect of niobium by tailoring the remainder of the steel composition to optimise both parent material properties and transformed CGHAZ microstructure.

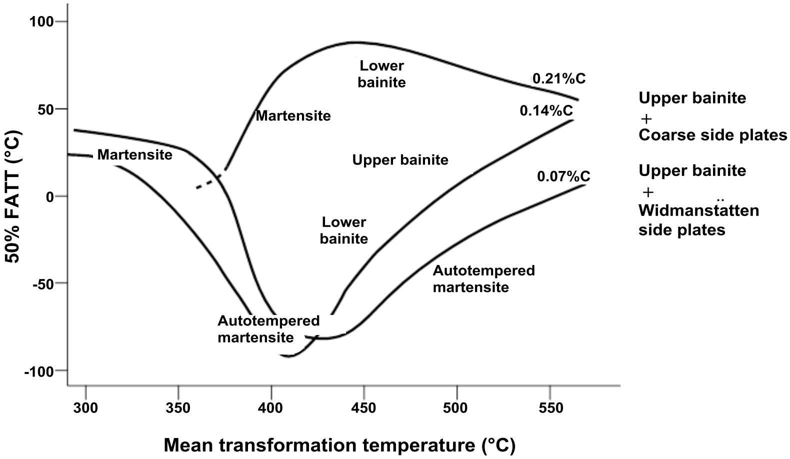

Figure 7. The effect of carbon and transformation temperature on CGHAZ toughness [30].

The dominant effect of carbon is emphasised by this theoretically constructed diagram and is dramatically confirmed by a systematic study on real welds in C-Mn-Nb steels [31]. The referenced paper explores the effects of carbon and niobium over a range of welding heat inputs and Figure 8, reproduced from that source, demonstrates very clearly that in niobium-bearing HSLA steels, **carbon is bad news** if optimum CGHAZ toughness is required. Similar patterns of behaviour were noted at other heat inputs.

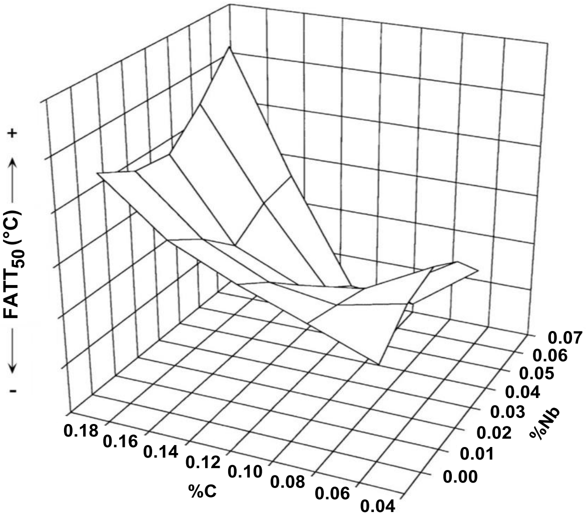

Figure 8. Carbon/niobium interactions and CGHAZ toughness at 3 kJ/mm, $\Delta t_{(800\text{-}500)}$ = 18 secs. After Hattingh & Pienaar [31].

The microstructures encountered in the CGHAZs of HSLA steels are predominantly bainitic and in the optical microscope often appear as per the example in Figure 9.

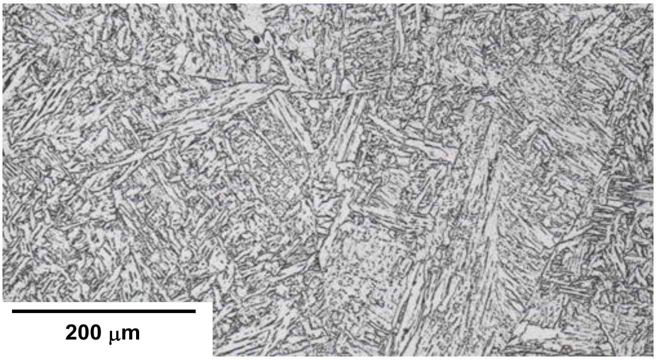

Figure 9. Typical bainitic microstructure of a CGHAZ in a C-Mn-Nb HSLA steel, $\Delta t_{(800\text{-}500)}$ = 30 seconds.

As demonstrated by many workers niobium ‘pushes’ the nose of the CCT diagram in HSLA steels well to the right as illustrated in Figure 10 and in consequence it is much easier to avoid the presence of polygonal ferrite in CGHAZ microstructures.

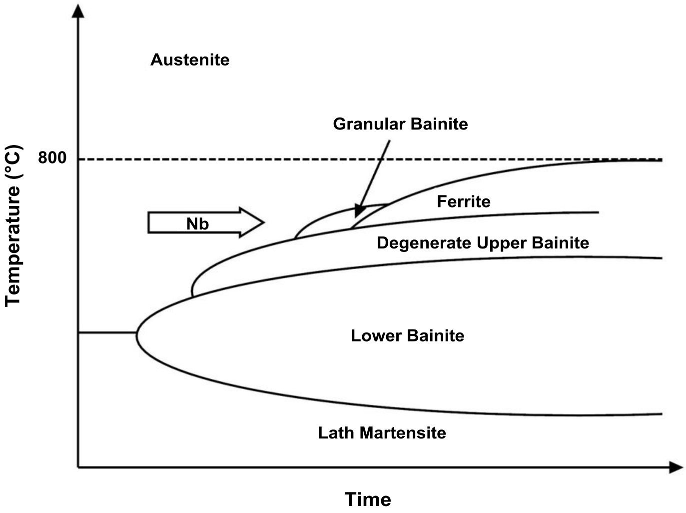

Figure 10. CCT diagram for low carbon HSLA steels. Adapted from Koo et al [32].

The diagram in Figure 10 portrays the situation for low carbon HSLA steels. It is important to note this point, because, with higher levels of carbon, the nature of the bainite which will be encountered in the CGHAZ can be very different at the fine detail level. In higher carbon steels, even when the polygonal ferrite nose is avoided during transformation, classical upper bainite with cementite films at lath boundaries can readily be formed at intermediate heat inputs such as those associated with submerged arc welding. Higher carbon, classical upper bainite, is generally of poorer toughness and faster cooling rates are required to produce tougher lower bainite or auto-tempered martensite. Even then, as indicated in Figure 7, carbon level will dictate the optimum toughness which can be achieved and the advantages of lower carbon become more and more obvious.

In fact, in Figure 10, the phase referred to as granular bainite would be difficult to distinguish in higher carbon CGHAZs and that referred to as degenerate upper bainite would not be observed at all. These phases would actually simply be replaced by classical upper bainite, as described earlier. It is the low level of carbon in modern steels which, in combination with higher niobium levels, allows optimum CGHAZ transformation behaviour to be observed over a wide range of cooling rates. This results in what has been referred to as degenerate upper bainite or lower bainite [32].

Figure 11 describes the key features of some of the bainitic phases described in the preceding paragraphs.

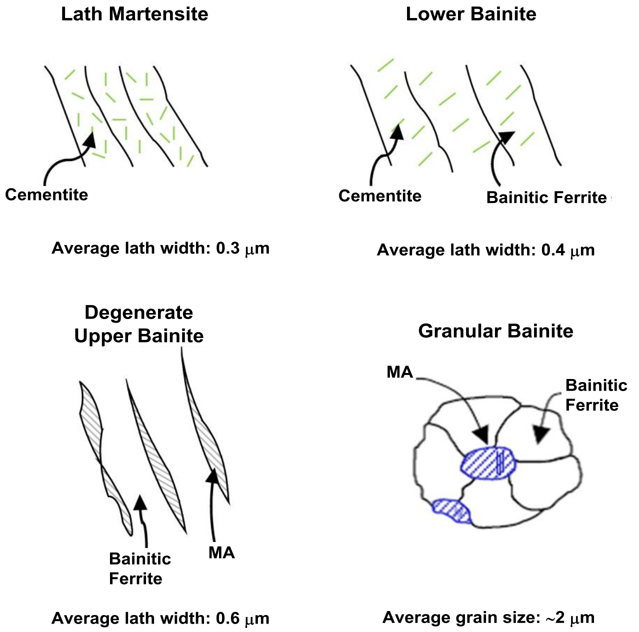

Figure 11. Austenite decomposition products in low carbon HSLA steels.
Adapted from Koo et al [32].

It has been suggested [32] that the fine MA films in low carbon, higher niobium, degenerate upper bainite may actually be beneficial to CGHAZ toughness and other on-going work at some of the world’s top universities will shortly be published confirming this contention. This theory has been more than adequately confirmed in practice by the results of many different welding trials on samples of HTP X80 linepipe steels [33] which have resulted in excellent CGHAZ toughness and there is no reason to suppose that structural steels, of comparable analysis, will react any differently.

Intercritically Reheated Grain Coarsened Heat Affected Zone Toughness (ICGHAZ). This particular region of a welded joint has been adequately defined elsewhere [7] and much has been written about the role of martensite austenite constituent (MA) in controlling toughness. Bainitic transformations occur by a displacive mechanism and it is inevitable that, as the bainitic ferrite laths form, the remaining small pools of austenite become progressively enriched in carbon and certain other partitioning elements. At slower cooling rates these carbon enriched areas degenerate into ferrite carbide aggregates but at faster cooling rates they eventually transform to martensite or combinations of martensite and retained austenite.

These pools of MA phase can be distributed in many different ways throughout the transformed microstructure and their nature and morphology are important in both the CGHAZ and, following subsequent weld passes, in the ICGHAZ. Figure 12 shows the appearance of MA in a weldment sample etched in Picral.

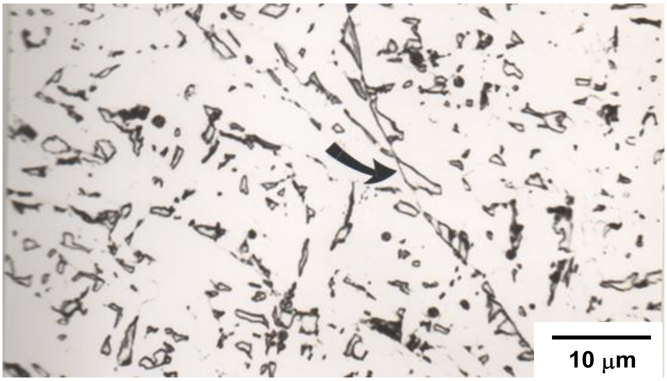

Figure 12. MA constituent in various morphologies, Picral etch.

Claims have been made to the effect that niobium promotes the formation of MA and that this is subsequently detrimental in the ICGHAZ. These claims have been carefully studied [5] and are rebutted by other evidence. It is clear, consistent with the description of the formation of MA provided earlier, that cooling rate and carbon level are primary factors in determining the volume fraction and nature of the MA observed in the CGHAZ. See Figures 13 and 14. Silicon is similarly implicated.

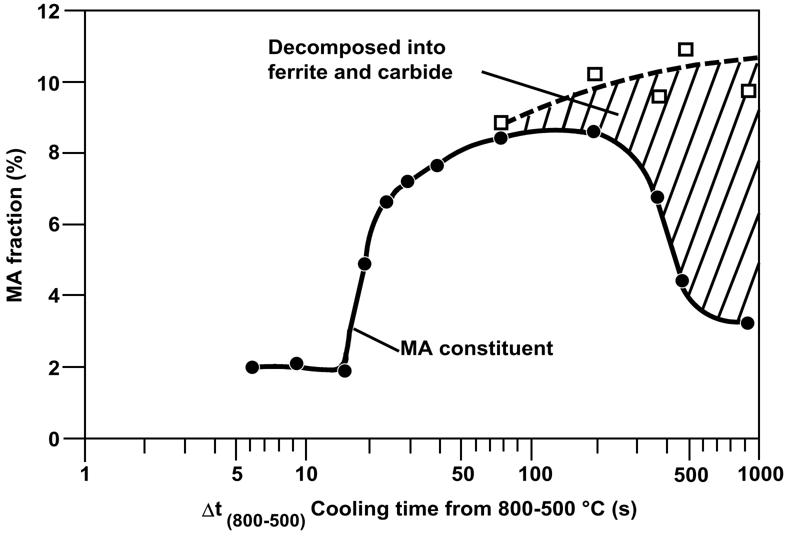

Figure 13. The effect of cooling time on MA formation in the CGHAZ. After Ikawa et al [3].

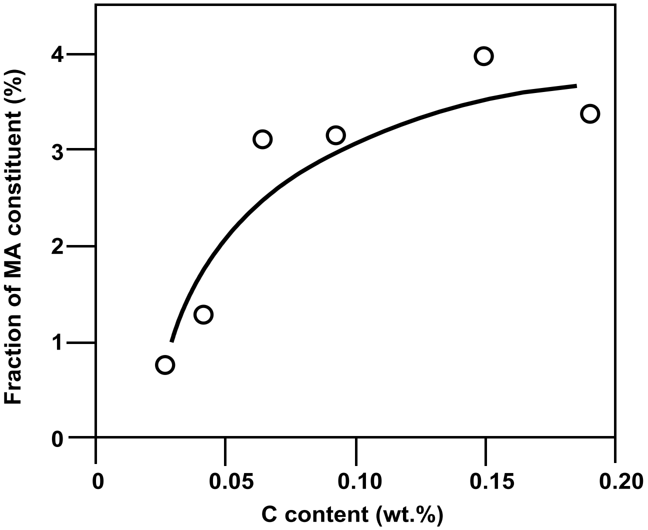

Figure 14. The effect of carbon on MA fraction in the CGHAZ. After Komizo and Fukado [35].

The contention that niobium increases MA in the ICGHAZ is not definitively supported in the literature [5] but what is clear is that the percentage of MA in the ICGHAZ is primarily dictated by how much was present to start with in the CGHAZ. This is not surprising because, being carbon rich, small regions of MA are first to re-austenitise when subjected to a second thermal cycle. The effect of a second thermal cycle on the size and proportion of MA is perfectly illustrated in the following Table IV.

Table IV. The Effect of a Second Thermal Cycle on MA Size and Volume Fraction $\Delta t_{(800\text{-}500)}$ = 30 seconds. After Bonnevie et al [36]

		Total MA population	**MA with L max >2 µm**	**MA with L max >3 µm**
	Mean size	**1.4 µm**	**2.8 µm**	**3.9 µm**
CGHAZ (1350 °C)	Proportion of total population	100%	20%	5%
	Mean size	**2.0 µm**	**3.2 µm**	**4.3 µm**
ICGHAZ (1350 °C + 850 °C)	Proportion of Total population	100%	36%	15%

It has previously been concluded that niobium does not specifically increase MA and that if it is considered an area of concern then it is best addressed by reducing carbon and silicon levels to the lowest practical levels commensurate with acceptable alloy design [5].

There is, therefore, no need to be concerned about any aspect of the welding of niobium-bearing HSLA steels in respect of the development of good HAZ microstructure and the achievement of adequate toughness. This is increasingly the case the lower the carbon content of the steel being welded.

In the sections of this paper concerned with weld metal and heat affected zone toughness I have deliberately confined my attention to the as-welded situation since modern low carbon steels, even in thicknesses up to 100 mm and beyond, rarely require to be subjected to post weld heat treatment. However, for the few situations where stress relief may be a consideration the following remarks, largely reiterated from an earlier publication [5], are pertinent.

"When niobium-bearing steels and weldments are stress relieved there are numerous effects to consider. In many cases the tempering and effective removal of MA constituent from HAZ or weld metal microstructures offsets any potentially detrimental effects of precipitation hardening but this will not always be the case [26]. Fortunately, post weld heat treatment is usually carried out with the intention of reducing residual stresses rather than effecting any specific metallurgical changes and in this context it should always be remembered that the successful elimination of residual stresses significantly increases the tolerance of the welded joints to the presence of defects even if the fracture toughness is moderately reduced by the procedure."

Specifications

In spite of the successful exploitation of low carbon, higher niobium (HTP) X80 linepipe steels in China and elsewhere, many national and international specifications, covering a wide spectrum of steel products for pipe and structural applications, still contain arbitrary restrictions on niobium; these frequently control the level of the element to 0.05% or below. There is no doubt that this is, primarily, a legacy of decades of largely unjustified adverse comment in the literature relating to the purported influence of niobium on weldability. This current paper taken together with the comprehensive re-evaluation presented in 2011 [5] must surely signal the time for a global reconsideration of this issue.

Niobium is an essential element in almost all HSLA structural and linepipe steels and yet its full benefit, in conjunction with lower carbon levels, is being unreasonably restricted thus preventing steelmakers and fabricators taking full advantage of the special benefits which can accrue from the element's unique properties. It is time for relevant specifications to be brought into the twenty-first century. The importance of this issue is highlighted by the following two diagrams.

Figure 15 shows how decreasing carbon and increasing niobium levels have led to progress in the linepipe sector allowing strengths up to X80 and beyond to be realised through the HTP route but, significantly, this has only been achieved where confident, forward thinking, specifiers have been prepared to develop and employ their own private company specifications and not be bound by the myths and prejudices of the past.

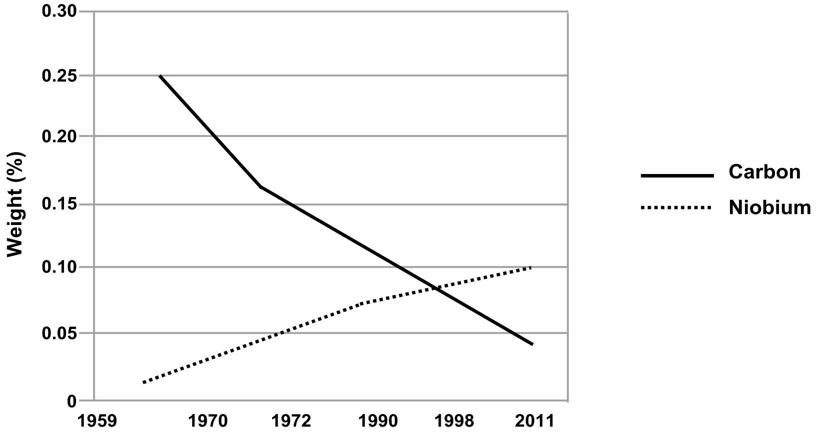

Figure 15. Changes in carbon and niobium levels in linepipe steels. After J. M. Gray [37].

In contrast, the arbitrary restrictions on niobium levels, for example, in European structural steel specifications have prevented similar developments and Figure 16 shows how an initial trend towards lower carbon and higher niobium levels in the early 1970s was halted in its tracks by the adverse technical press referred to earlier in this paper. Many structural steel specifications now restrict niobium to 0.05% max and this is clearly technically unjustified and is preventing the full benefit of lower carbon and higher niobium being exploited across the whole spectrum of relevant products.

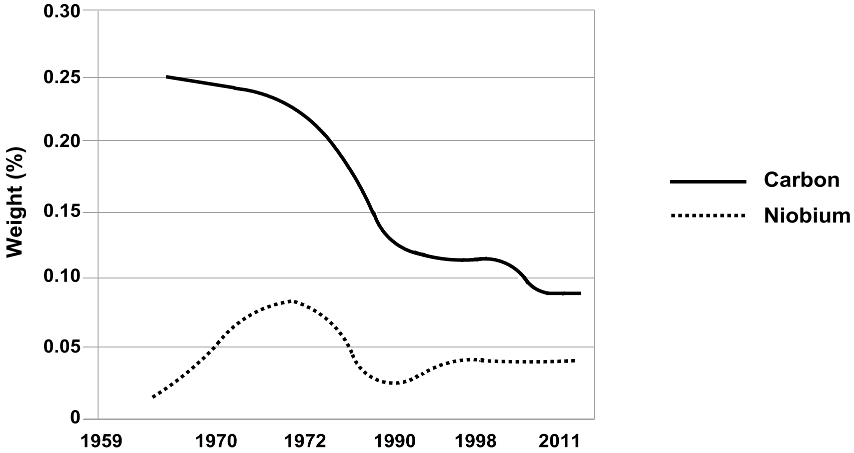

Figure 16. Trends in carbon and niobium levels in European structural steels.

Fortunately, there are now growing numbers of progressively thinking individuals who understand the metallurgy and who recognise what has to be done. Hopefully the next few years will see long overdue action being taken on the steel standards front.

Conclusions

In almost every aspect of weldability, niobium microalloyed structural steels can be considered to be comparable to any other HSLA steel of an equivalent strength. Niobium has no influence on the incidence of the types of defects which were prevalent in steels of earlier decades such as liquation cracking, lamellar tearing and solidification cracking.

Equally, as regards susceptibility to hydrogen induced cold cracking, it has been demonstrated that the effect of niobium is either neutral or, potentially, even beneficial. It is further suggested that, in lower carbon weld metals, niobium may be specifically beneficial to toughness by providing adequate hardenability without excessive solid solution embrittlement, thus encouraging the formation of acicular ferrite in the presence of an adequate distribution of nucleating second phase particles.

Finally, addressing the area of greatest contention, there is no longer any reason to be concerned about adverse effects of niobium on heat affected zone toughness in modern structural steels. Any justifiable concern of earlier decades undoubtedly related to the high heat input welding of higher carbon steels and both theory and recent practical evidence suggest that the combination of higher niobium and lower carbon is, potentially, one of the most rewarding areas for future structural steels development.

Appropriate changes to national and international specifications must now be facilitated to enable specifiers and fabricators to take full advantage of the special benefits and enhanced properties which will accrue from higher levels of niobium in the new generation of lower carbon HSLA steels.

References

1. J.M. Gray and F. Siciliano, “High Strength Microalloyed Linepipe: Half a Century of Evolution,” (Paper presented at the Pipeline Technology Conference, Oostende, October 2009).

2. J.M. Gray, “Evolution of Microalloyed Linepipe Steels with Particular Emphasis on the ‘Near Stoichiometry’ Low Carbon, 0.1 Percent Niobium “HTP” Concept”, *Proceedings of the 6th International Conference on High Strength Low Alloy Steels (HSLA Steels 2011),* Beijing, China, (May/June 2011).

3. S.G. Jansto, “Niobium-Bearing Structural Steels for the 21st Century,” *Niobium-Bearing Structural Steels,* (TMS, 2010), 1-27.

4. L.M. Silvestre et al., “High Strength Steel as a Solution for the Lean Design of Industrial Buildings” (Paper presented at the Value Added Niobium Microalloyed Construction Steels International Symposium, Singapore, TMS/CBMM, 5-7 November 2012).

5. P.R. Kirkwood, “Niobium and Heat Affected Zone Mythology” (Paper presented at the Welding of High Strength Pipeline Steels International Seminar, Araxa, Brazil, TMS/CBMM, 27-30 November 2011).

6. P.R. Kirkwood, “Welding of Niobium Containing Microalloyed Steels,” Niobium Science and Technology, *Proceedings of an International Symposium, Niobium,* (1981), TMS, San Francisco, USA.

7. A.D. Batte, P.J. Boothby and A.B. Rothwell, “Understanding the Weldability of Niobium-Bearing HSLA Steels,” *Proceedings of the International Symposium Niobium 2001*, Orlando, Florida, (December 2001), 931-958.

8. L. Luyckx et al., “Sulphide Shape Control in HSLA Steel,” *Metallurgical Transactions*, 1 (12) (1970), 3341-3350.

9. J.C.M. Farrar and R.E. Dolby, *Lamellar Tearing in Welded Steel Fabrication*, (The Welding Institute (TWI) Booklet, 1972), 4.

10. P.R. Kirkwood, "Offshore Structural Steels – A Decade of Progress, Institution of Gas Engineers Communication 1334", *Proceedings of the 124th AGM, Harrogate International Centre*, 4-7 May, 1987.

11. K. Hakansson, "Submerged Arc Welding - A Test for Centerline Cracking," *Hot Cracking in Welds II*, (New York, NY: Berlin Heidelberg, April 2005), 116.

12. N. Bailey and S.B. Jones, "The Solidification of Ferritic Steel During Submerged Arc Welding," *Welding Journal*, August 1978, 217s-231s.

13. K. Hulka and J.M. Gray, "High Temperature Processing of Line-Pipe Steels," *Proceedings of the International Symposium Niobium* 2001, Orlando, Florida, USA.

14. J.E. Clark, "High Yield Ship Steels," North East Coast Institute of Engineers and Shipbuilders, April 1971.

15. R.M. Carter et al., "Materials Problems in Offshore Platforms," (Paper presented at the First Annual Offshore Technology Conference, OTC 1043, 18-21 May 1969, Houston, Texas, USA).

16. N. Bailey, *Welding Steels Without Hydrogen Cracking,* (The Welding Institute (TWI), Cambridge, Published 1993), 11-12 and 35-36.

17. British Standard BS EN 1011-2: "Welding Recommendations for Welding of Metallic Materials - Part 2, Arc Welding of Ferritic Steels, March 2001 (Amended A1, Dec 2003).

18. Y. Ito and K. Bessyo, "Weldability Formula of High Strength Steels Related to Heat Affected Zone Cracking," IIW Doc IX-576-68, 1968.

19. N. Yurioka, "Carbon Equivalents for Hardenability and Cold Cracking Susceptibility of Steels," (Paper presented at the Select Conference on Hardenability of Steels, Paper 3, 17 May 1990, Derby, England).

20. P. Bufalini, F. Bonomo and C. Parrini, "Effects of Composition and Microstructure on HAZ Toughness and Cold Cracking," *Proceedings of the ASM-AIM International Conference on the Welding of HSLA Microalloyed Structural Steels*, Rome, (November 1976).

21. A.B. Rothwell, "Heat Affected Zone Toughness of Welded Joints in Micro-Alloy Steels" Part 1, Results of Instrumented Impact Tests, IIW 1X-1147-80 and Subsequent CCT and Microstructural Studies (Private Communication).

22. ASM International 1997, "Weld Integrity and Performance", Materials Park, Chapter 1, Weld Solidification, 20.

23. P. Colvin and A.F. Bush, "Multipass Submerged Arc Welding of High Yield Strength Notch Ductile Ferritic Steels, *Iron and Steel,* 42 (1969), 361.

24. P. Colvin, “Notch Toughness Properties of Single Pass Submerged Arc Weld Metal on EH 2 Shipbuilding Quality Steels,” *Welding and Metal Fabrication*, February 1971.

25. N.E. Hannerz, G. Valland and K. Easterling, “Influence of Niobium on the Microstructure and Mechanical Properties of Submerged Arc Weld Metals for C-Mn Steels,” IIW Documents II-612-72, XII-A-46-72 and IX-798-72.

26. J.G. Garland and P.R. Kirkwood, “Towards Improved Submerged Arc Weld Metal Toughness,” *Metal Construction*, 7 (5 and 6) (May and June, 1975).

27. International Standards Organisation Document - Draft Recommendations by The ISO/TC 17/SO3 Committee.

28. J. Wegrzyn, “Weldability of Microalloyed Steels” (Polish Institute of Welding Report, 1976).

29. S.S. Babu, “The Mechanism of Acicular Ferrite in Weld Metal Deposits,” *Current Opinion in Solid State and Materials Science*, 8 (2004), 267-278.

30. A.D. Batte and P.R. Kirkwood, “Developments in the Weldability and Toughness of Steels for Offshore Structures,” *Proceedings of Microalloying’ 88 held in conjunction with the 1988 World Materials Congress,* Chicago, Illinois, USA, (September 1988).

31. R.J. Hattingh and G. Pienaar, “Weld HAZ Embrittlement of Niobium Containing C-Mn Steels,” *International Journal of Pressure Vessels and Piping 75* (1998), 661-677.

32. J.Y. Koo et al., “Metallurgical Design of Ultra-High Strength Steels for Gas Pipelines,” *Proceedings of the Thirteenth International Offshore and Polar Engineering Conference*, 25-30 May, 2003, Honolulu, Hawaii, USA.

33. K. Hulka, “The Role of Niobium in Low Carbon Bainitic HSLA Steel,” *Proceedings of the First International Conference on Super High Strength Steels*, November 2005, Rome, Italy.

34. H. Ikawa, H. Oshige and T. Tanoue, “Effect of Martensite - Austenite Constituent on HAZ Toughness of a High Strength Steel,” *Transactions of The Japan Welding Society*, 11 (2) (October 1980), 87-96.

35. Y.I. Komizo and Y. Fukado, “CTOD Properties and MA Constituent in HAZ of C-Mn Microalloyed Steel,” *Proceedings of the 5th International Symposium of the Japan Welding Society*, Tokyo, (April 1990).

36. E. Bonnevie et al, “Morphological Aspects of Martensite - Austenite Constituents in Intercritical and Coarse Grained Heat Affected Zones of Structural Steels,” *Materials Science and Engineering A*, 385 (2004), 352-358.

37. J.M. Gray, “Weldability of Niobium Microalloyed Linepipe Steels: 50 Years of History” (Paper presented at the Welding of High Strength Pipeline Steels International Seminar, TMS/CBMM, 27-30 November 2011, Araxa, Brazil), 1-21.