WELDING OF NIOBIUM CONTAINING MICROALLOYED STEELS

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Synopsis

This paper reviews a selection of the more important contributions of the past decade on the influence of niobium on various aspects of the weldability of high strength low alloy steels. It is apparent that niobium has insignificant effects on many aspects of weldability, but that considerable controversy has existed in relation to its influence on weld metal and heat affected zone toughness. The apparently contradictory observations of various earlier authors have been carefully examined and rationalized in the light of all the relevant data now available in the world literature.

It is concluded that the complex role of niobium can best be understood when its overall effects are examined in relation to its influence on transformation temperature, and in this context, the importance of carbon content and cooling rate following welding is emphasized.

Finally, it is suggested that the most judicious use of niobium will only be made when the liaison between the steelmaker and the end user is improved. It is therefore proposed that all steel compositions should be tailored, not simply with a view to obtaining optimum base plate mechanical properties, but with equal consideration to the end product and in particular, to the welding process envisaged for use in practice.

Introduction

Since the first successful utilization of niobium in a high strength low alloy linepipe steel some twenty years ago (1) there have been continuing advances in microalloying technology, and it is now clear, as witnessed by the scope of the present conference, that niobium has found application in a very wide variety of product forms. Almost exclusively, these products, plates, shapes, bars, pipe etc. require to be welded either during manufacture or subsequent fabrication and a great wealth of information on their response to all the common welding processes has now been accumulated.

Certain aspects of the welding of individual products such as thin strip and bars are unique to the less common processes sometimes applied in these areas, and any specific effects of niobium in this context are dealt with by the authors of other papers in the appropriate sessions. The purpose of this paper is to provide a general review of the influence of niobium on steel weldability, and since a great deal of the published literature relates to studies on plate material, it is inevitable that there is a bias towards those welding processes most commonly applied to this product. Nevertheless, the trends established and the general metallurgical conclusions formulated, are equally applicable to a wide variety of situations and it is hoped that the presentation of the data will be of value to all end users of niobium bearing steels.

General Aspects of Weldability

A study of the available literature, makes it quite clear that niobium has no significant effects on general aspects of weldability such as heat affected zone hydrogen cracking, liquation cracking or weld metal solidification cracking and in fact the various carbon equivalent formulae used to predict weldability (2), do not incorporate niobium as an element requiring consideration. Thus, e. g., in the U.K., the British Standard used to derive safe pre-heat levels for structural steels (3) considers C-Mn-Nb steels as equivalent to C-Mn steels and much practical experience confirms the validity of this well established practice. Niobium is not considered to have any influence on martensitic hardenability and it is therefore not surprising that it should have no effect on the susceptibility of steels to HAZ hydrogen induced cracking.

In fact, the present author has found no credible reference to any specific effect of niobium on general weldability and it is clear that there are only two areas of potential conflict in the literature:

a. The effect of niobium on weld metal toughness and,

b. The effect of niobium on heat affected zone toughness.

A cursory examination of the literature in these two areas immediately reveals that there has been considerable controversy over the past decade with some authors claiming beneficial effects of niobium and others indicating quite marked detrimental effects. The remainder of this paper is, therefore, concerned with an evaluation of a selection of the more important contributions in the world literature in these areas, and an attempt to rationalize the results of various authors to provide clearer guidance on the role of niobium over a wide range of practically relevant welding conditions.

Historical - Niobium Detrimental to Weld Metal Toughness

In the late Sixties and early Seventies, there were a number of publications which suggested that niobium was an undesirable element in situations where it was essential to obtain good weld metal toughness *(4,* 5, 6). The most serious problems appeared to be associated with welds deposited at higher heat inputs e.g. with the Submerged Arc process which was being increasingly used for the welding of shipbuilding and structural steels, and, in spite of the general acceptance of niobium treated steels and their trouble free fabrication history, the use of microalloying elements such as niobium began to be questioned.

The importance of niobium was subsequently debated by the International Standards Organization and it was soon proposed that the element should be limited to 0.06 percent in specifications (7). In 1971 and 72 following extensive work by Hannerz et al (6), the Swedish Delegation to the International Institute of Welding proposed niobium maxima first of 0.05 percent and subsequently 0.03 percent for steels intended to be fabricated using submerged arc welding where reasonable impact properties were required at sub-zero temperature.

Some of the results recorded by Hannerz et al *(6)* on 15mm plates are reproduced in Figure 1, and it is clear that with heat inputs in the range 5.7 to 7.2 KJ/mm (Δ T $\frac{800}{500}$ = 150 → 220 secs) a significant embrittling effect of niobium was experienced. A full analysis of the microstructural changes accompanying the additions of niobium is not given in the paper, but from those comments which are made and the chemical analysis figures included, it is possible to conclude that Hannerz's higher heat input welds had microstructures dominated by ferrite side plates and polygonal ferrite and it is unlikely that much acicular ferrite was present. Hannerz et al *(6)* recorded significant increases in yield strength and hardness when niobium was added and postulated that the cause of the low impact strength was the presence of a fine dispersion of coherent niobium carbonitrides and an increase in weld metal dislocation density. Considering all their results, Hannerz et al *(6)* suggested that, for submerged arc weld metal, the Charpy v-notch transition temperature would increase by about 2 C per 0.001 percent Nb added, i.e. about 20 C per 0.01 percent Nb.

Further investigations in later years which also employed high heat inputs with plate thickness ⩽ 25mm provided similar conclusions and following very extensive systematic work at the Welding Institute in Poland, Wegrzyn (8) actually attempted to rate the "weldability" of C-Mn-Nb steels by means of a type of carbon equivalent formula. Good "weldability" in this context of course referred to the ease with which adequate sub-zero weld metal Charpy results could be achieved. Wegrzyn (8) defined the index C_w by the formula.

$$C_w = C\% + \frac{Mn\%}{10} + 3\ Nb\% \qquad (1)$$

The suggested range of validity of the formula was for carbon contents from 0.1 percent to 0.18 percent, manganese contents from 1.0 to 1.6 percent, and niobium contents from 0.025 percent to 0.095 percent and the C_w results were applied as follows:

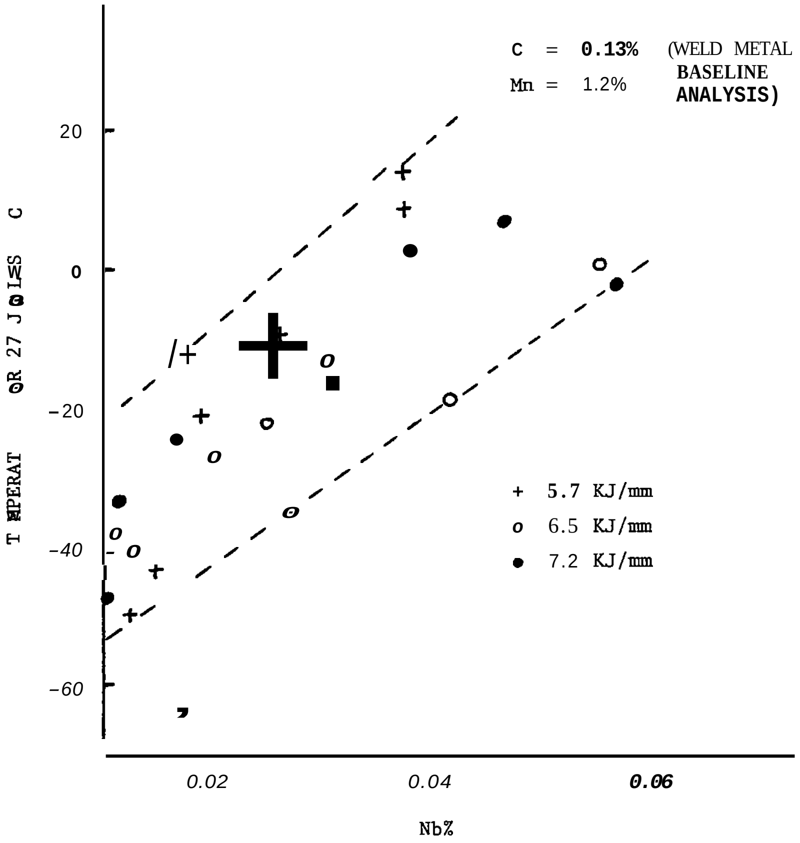


Figure 1. **The influence of niobium on high heat input weld metal toughness** ΔT_{500}^{800} **= 150 to 220 secs (6).**

$C_w \leqslant 0.35$ - No restriction on heat input or welding process

$0.35 < C_w \leqslant 0.45$ - Heat input should be restricted to 3.5 KJ/mm if submerged arc welding is employed.

$C_w > 0.45$ - Only manual metal arc welding with basic coated electrodes should be considered.

This was certainly a novel analysis and the relative weightings given to carbon and niobium indicate that the importance of the former element was very much appreciated. In the light of present day steelmaking trends, it is interesting to reflect that Wegrzyn's formula results in identical C_w values for the two steels detailed below:

	C%	Mn%	Nb%	C_w
Steel 1	0.18	1.40	0.033	0.42
Steel 2	0.10	1.40	0.06	0.42

Niobium as a Promoter of Tough Acicular Ferrite Microstructures

By 1973, the adverse publicity being received by niobium was sufficient to cause concern amongst European and American Steelmakers who had developed specific product ranges to take advantage of the economic strengthening which niobium provided, and in 1974 the British Steel Corporation published the results of a very comprehensive investigation into the effects of niobium on Submerged Arc weld metal toughness (9). This work was carried out on both 12.5mm and 25mm plates and incorporated welds with heat inputs in the range of 1.9 to 7.6 KJ/mm, welds with amphoteric and basic fluxes, changes in welding wire compositions variations in interpass temperature and finally the effecs of stress relieving.

It was suggested that the effect of niobium depended critically on a number of factors including plate thickness, heat input, flux type and welding wire analysis but the most significant observations were that, there were many instances where niobium had no overall detrimental effect on as welded toughness and some cases where niobium bearing welds were clearly superior after stress relieving.

These latter points were highlighted by Garland & Kirkwood (10) in a subsequent publication which summarized the findings of the complex B.S.C. studies and Figure 2 reproduces some of their results which illustrate the points made above. A detailed quantitative microstructural analysis of these welds was carried out, with particular attention being given to the nature and morphology of "pools" of retained martensitic microphases in the acicular ferrite matrix, since these had previously been identified as of potential importance by Biss & Cryderman (11). The results of this microstructural analysis for the welds considered in Figure 2 are given below:

Heat Input 3.3 KJ/mm (25mm Plate),

$\Delta T_{500}^{800} = 8.6$ Secs.

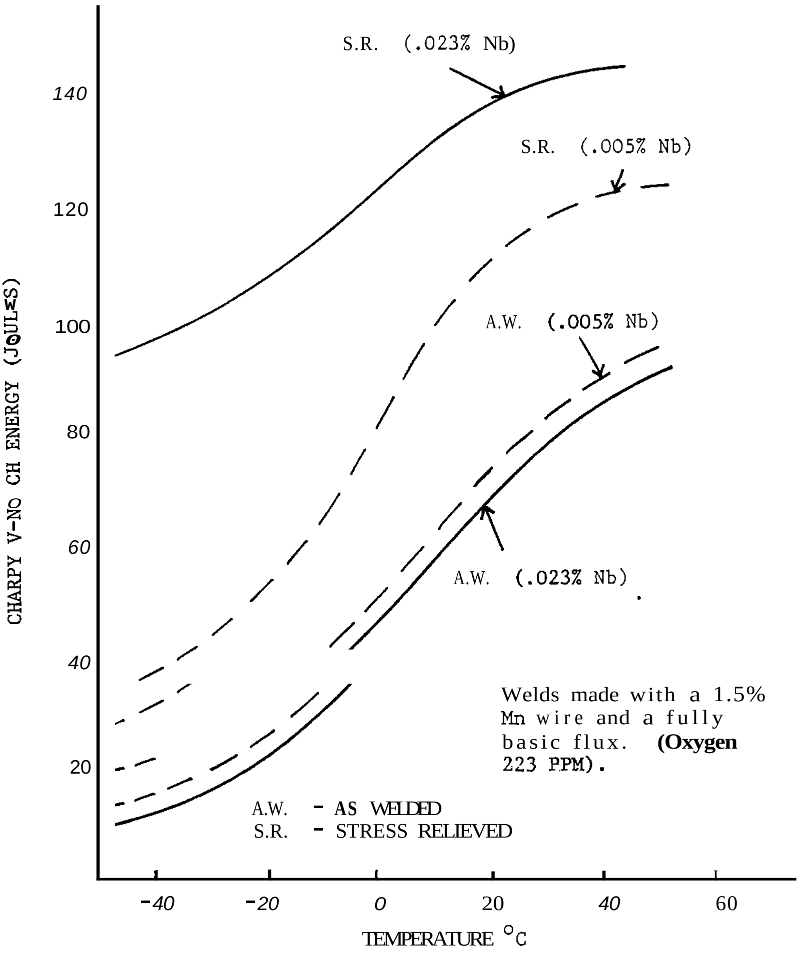


Figure 2. The effect of weld metal niobium level on charpy V-notch toughness (10).

	Acicular Ferrite %	Ferrite Side Plates[x] %	Polygonal Ferrite %	Acicular Ferrite Martensitic micro phases % Lath	Twinned
0.005% Nb	47	5	48	2	7
0.023% Nb	78	3	19	15	1

x Given the misnomer "Upper Bainite" in the original publication

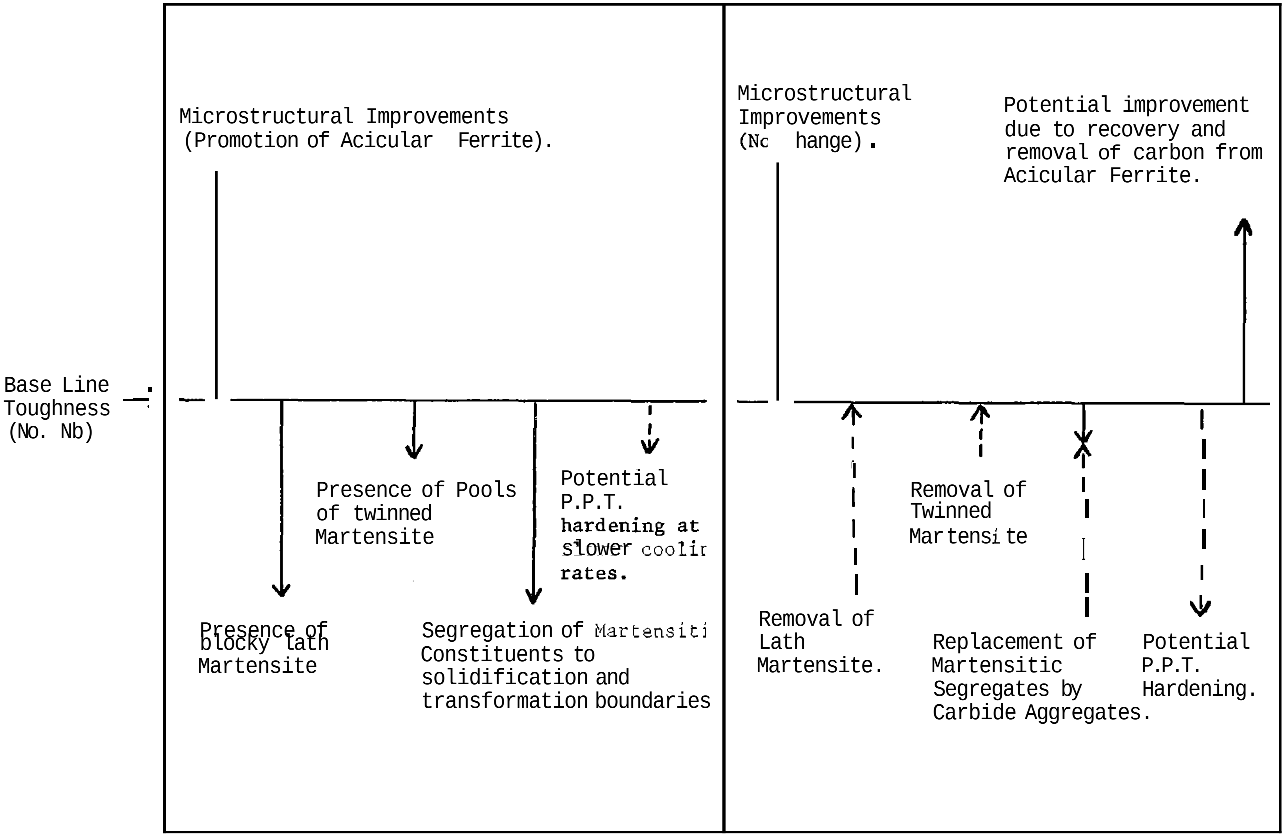
It is clear that the presence of niobium has had a marked beneficial effect on gross microstructure though it has significantly increased the amount of lath martensite present in the acicular ferrite matrix. Garland & Kirkwood noted, as had previous authors that the presence of niobium resulted in significant increases in weld metal strength, particularly after stress relief and the results appropriate to Figure 2 where as follows:

	0.2% PROOF STRESS (N/mm^2) As Welded	Stress Relieved
0.005% Nb	426	431
0.023% Nb	493	571

These strengthening effects were considered to be due to, in the as welded condition, a combination of transformation hardening and perhaps limited precipitation hardening, and in the stress relieved condition, very definitely precipitation hardening.

Combining these observations with all their other results (9, 10), Garland & Kirkwood proposed a vector model which could be used qualitatively to explain how opposing effects of niobium could interact to produce either overall beneficial or detrimental effects on toughness depending on the precise compositions and welding conditions under consideration. A schematic presentation of this model is given in Figure 3, and if we bear in mind that the "vectors" in the diagram indicate only the direction of effects and not their relative magnitude then in the context of Figure 2, the as welded results are explained by postulating that the gross microstructural improvement resulting from the addition of Nb is counterbalanced by the sum of the negative vectors shown in Figure 3a. In the stress relieved condition Figure 3b applies, and in this instance the removal of the martensitic constituents and other beneficial effects, such as recovery and removal of carbon from the acicular ferrite, offset the increased precipitation hardening, and allow the potential of the improved gross microstructure to be realized as shown in Figure 2.

Since 1974 there have been numerous papers published in the world literature and an excellent review of these has recently been carried out by Dolby (12). Many of the results reported can be rationalized on the basis of the Garland & Kirkwood model (10) but there are certain exceptions which



(a) AS WELDED (b) STRESS RELIEVED

Figure 3. Schematic representation of factors affecting weld metal

require further consideration. It is not the intention of this review to deal exhaustively with all the contributions which fall into this latter category and two, have, therefore, been specifically selected since these best demonstrate the additional effects which niobium has been reported to produce.

A notable feature of Garland & Kirkwood's work *(10)* was that niobium always promoted an increase in acicular ferrite content at the expense of grain boundary polygonal ferrite (and indeed many other investigators have made a similar observation) but there are clearly circumstances in which this does not occur.

Niobium Promoting Ferrite Side Plate Structures

In 1975, Jesseman *(13)* using a calcium silicate flux (higher oxygen deposits– 380 ppm) produced a series of submerged arc welds on 12.7mm plates which clearly indicated that, with welds lean in manganese (~ 1%), niobium, while it did decrease the percentage of polygonal ferrite present, did not promote acicular ferrite formation. Instead, even in the presence of limited additions of Ni and Mo, niobium increased the amount of lamellar side plate ferrite in the microstructure. Very modest increases in hardness were recorded with the addition of niobium, and on the whole (excepting one result at 0.053 percent Nb in the C-Mn deposit series) niobium had marginal beneficial effects on toughness, see Figure *4*. On stress relief, however, the hardness data clearly indicate an important role of Nb in precipitation hardening, the base welds without Nb actually decreasing in hardness and the Nb bearing welds showing modest increases. As shown in Figure *4*, Nb is then seen to be very detrimental to weld metal toughness.

It is noteworthy, that in this work, the addition of nickel was specifically detrimental to toughness, particularly in the absence of Nb (see Fig. *4)* probably because its addition also promoted the formation of side plate ferrite in this situation. The significance of such observations is often overlooked when we are concentrating on the effects of niobium itself, and it is worthwhile reflecting that there are few references in the literature which suggest that nickel is detrimental to weld metal toughness.

The weld metal microstructures observed by Jesseman (13) are considered by the present author to be of fairly high transformation temperature, perhaps not unlike those noted in some of Hannerz's lower heat input welds considered in his 1972 paper (6).

Niobium Promoting Lath Type (Bainitic) Microstructures

A further important contribution to the literature was made by Levine & Hill when following a useful review *(14)* in which they made a significant contribution to our present understanding of the way in which weld metal microstructures develop, they published the results of a series of submerged arc welds deposited with a basic flux (low oxygen– 270 ppm) on 13m plates (15). These welds were prepared on plates of much higher manganese content (1.70%) than in e.g. Jesseman's work (~ 1%) and the resultant weld metals were therefore as expected of greater hardenability. Levine and Hill studied Nb levels from 0 to 0.048 percent, in welds containing C – 0.09 percent and Mn– 1.65 percent with and without 0.2 percent Mo and these results are reproduced in Figure 5.

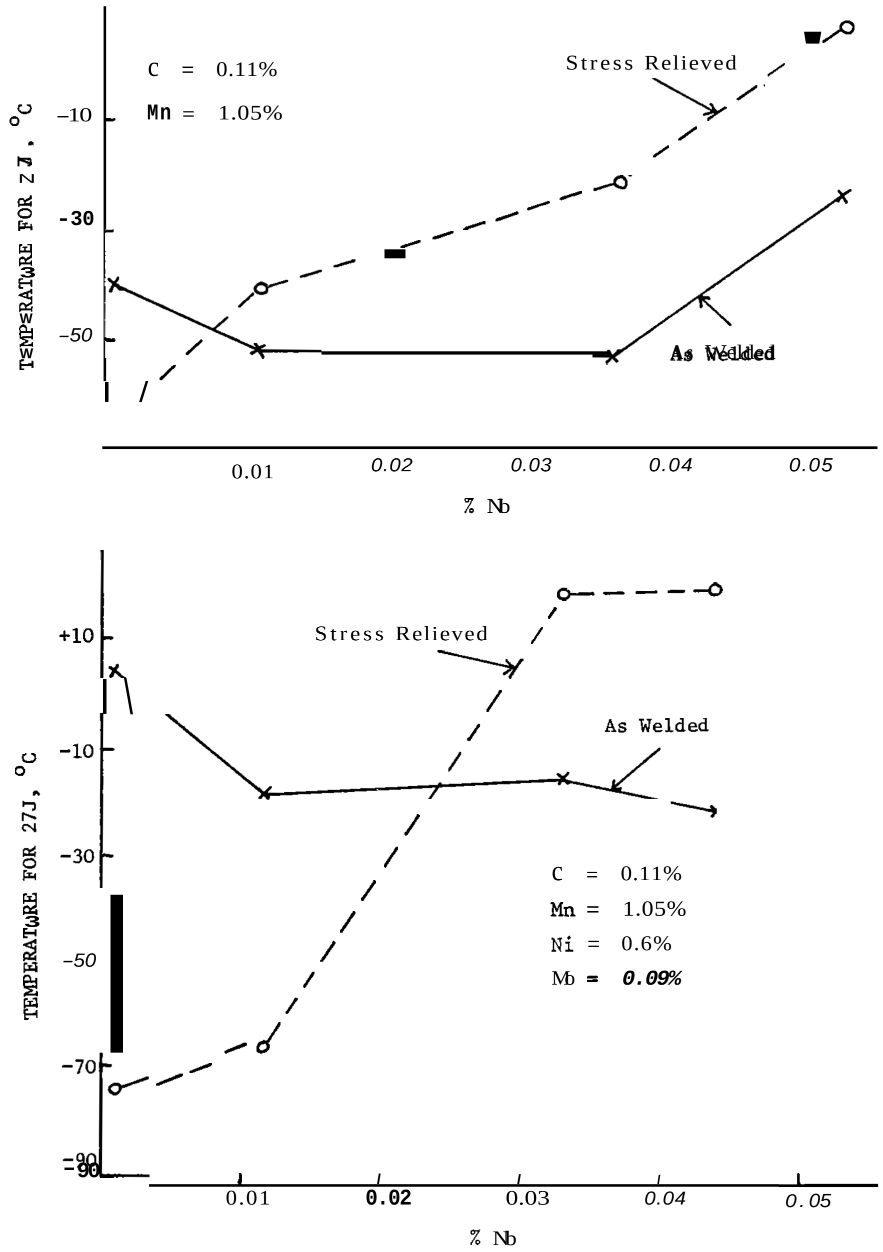


Figure 4. The effect of Nb on the toughness of low manganese, higher oxygen (0.038%) sub-arc welds deposited on 12.7 mm plates at 2.2KJ/mm, ΔT_{500}^{800} = 23 secs (13).

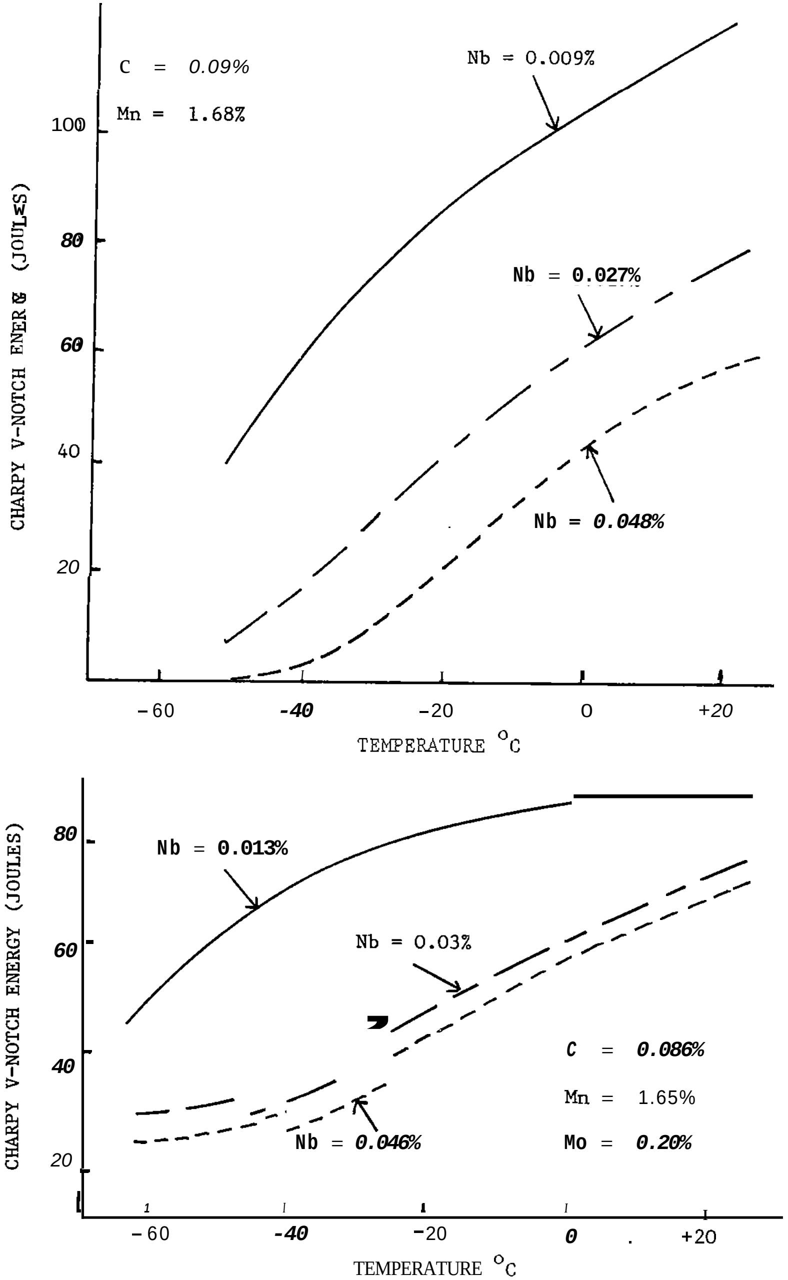


Figure 5. **The effect of Nb on the toughness of low oxygen (sub-arc) welds deposited on 13 mm plates at 2.4KJ/mm,** ΔT^{800}_{500} = **30 secs** (15).

In contrast with Garland and Kirkwood's (10) and Jesseman's (13) work, Levine & Hill found marked detrimental effects of Nb on toughness of similar magnitude to that noted by Hannerz et al (6) in much higher heat input welds. However, in contrast to Hannerz et al, Levine & Hill's study showed that Nb had no effect on as welded yield strength, the values being consistently high ~ 545 N/mm^2 with or without Nb (~ 565 in the presence of 0.2% Mo). Fortunately, Levine & Hill carried out a thorough metallographic examination of their microstructures and they concluded that, as in all other studies, niobium reduced the amount of grain boundary ferrite present. Once again, however, niobium (except at very low levels) did not promote the presence of acicular ferrite; instead it progressively replaced acicular ferrite with a lath type structure with low angle grain boundaries (revealed by transmission electron microscopy). It seems likely that this lath structure is similar to that previously described by Ito and Nakanishi (16) in low oxygen niobium free weld metals, and is akin to a form of true "bainite". This constituent, according to the latter authors, forms at temperatures below those normally associated with acicular ferrite formation. The morphology of the lath structure is such that cleavage initiation and propagation are easy and this explains the detrimental effects of niobium recorded by Levine & Hill (15).

The Effect of Niobium on Weld Metal Transformation

The common experience of the various investigators is that Nb reduces the amount of polygonal ferrite in the weld metal and it is clear, therefore, that one of its primary effects, in common with other alloying elements such as Mn, Ni and Mo, is to increase hardenability. However, overall hardenability depends on a number of other factors and in the case of submerged arc welding, even the type of flux and the associated oxygen level have been shown to be important variables (17, 18).

It follows, therefore, that the effect of niobium will vary depending on the precise circumstances and in particular on what the microstructure is like in the absence of niobium.

Cochrane & Kirkwood (17) have published the results of studies on the transformation behavior of weld metal samples reheated and cooled using high speed dilatometry and while it is unfortunate that there are no two identical welds to compare, it is possible to gain an appreciation of the magnitude of the Nb effect. The following results are from welds made with manganese silicate fluxes having oxygen contents about 600 ppm.

			TRANSFORMATION TEMPERATURE	
C%	Mn%	Nb%	Start	Finish
.14	1.25	<.005	725°C	600°C
.09	1.25	0.031	675°C	535°C

Nb Dilatometer samples cooled such that $\Delta T_{500}^{800} = 55$ secs.

The difference in carbon content of these two welds is, of course, unfortunate and prevents a true assessment of the effect of Nb. However, Abson (19) has shown that an increase in carbon would be expected to depress the weld metal transformation temperature and thus it seems reasonable to conclude that, at the 1.25 percent Mn level, the introduction of – 0.03 percent Nb has resulted in a drop in mean transformation temperature of at least 50 C. This is a very significant change and in fact Cochrane & Kirkwood have shown (17) that an increase in manganese from 1.0 percent to about 1.5 percent would be required in acidic weld metal to effect a change of similar magnitude.

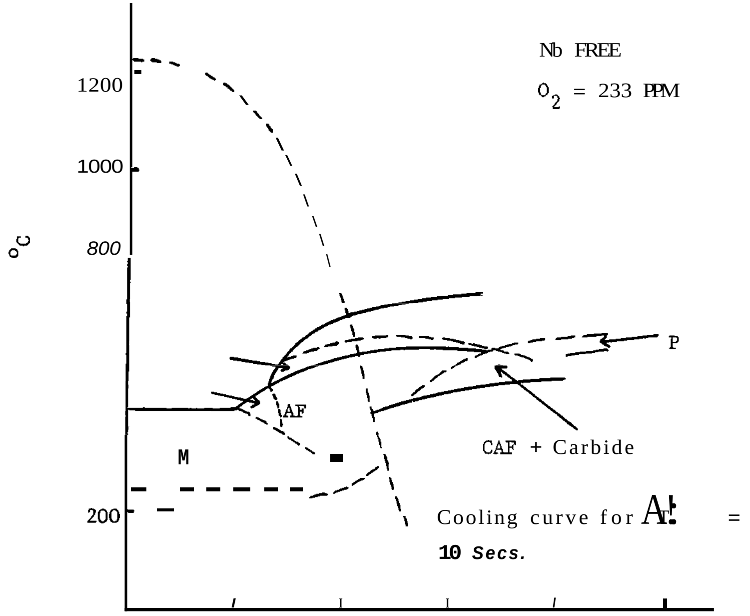
Recently, a more complete study of the effect of niobium on weld metal transformation has been reported by Harrison et al (20). In their study, high speed dilatometry and thermal analysis were combined to permit the construction of Continuous Cooling Transformation diagrams (C.C.T.) for a series of weld metals of varying C, Mn, Nb and O_2 contents. Two of Harrisons's diagrams are reproduced in Figure 6 for compositions not far removed from those studied by Garland & Kirkwood (see Figure 2), and these clearly indicate that, at a cooling rate resulting in a ΔT_{500}^{800} = 10 secs (c.f. 8.6 secs in Garland & Kirkwood's work), the primary effect of niobium should be to greatly reduce the amount of polygonal ferrite (P.F.) formed.

It is clear that, in this type of composition over a fairly wide range of cooling rates, the addition of niobium would increase the amount of acicular ferrite (A.F.) present without greatly changing the ferrite side plate (F.S.P.) content. The depression of ferrite start temperature recorded by Harrison et al is of a similar order to that noted by Cochrane & Kirkwood (17).

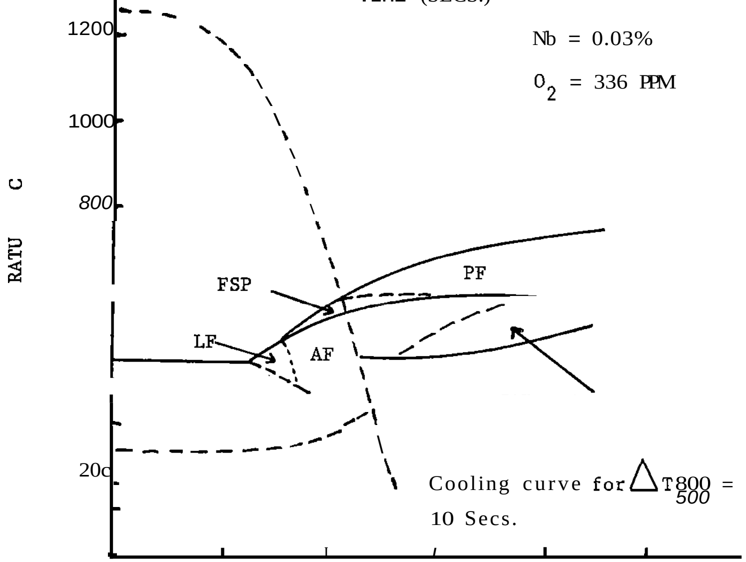
Harrison et al (20) also produced CCT curves for low manganese weld metals with higher oxygen contents, and in fact the compositions studied were not unlike those in Jesseman's work described earlier (13). These diagrams clearly show that at the cooling rate appropriate to Jesseman's studies i.e. 13 C/sec (ΔT_{500}^{800} = 23 secs) microstructures dominated by polygonal ferrie and ferrite side plates can be expected. Any acicular ferrite is likely to be of the coarse variety (CAF).

The CCT diagrams for these low manganese weld metals exhibit a much more extensive ferrite side plate regime than those shown in Figure 6, primarily due to the lower manganese itself, and perhaps with some effect of the higher oxyen in the deposits investigated (17). While Harrison's work is not definitive on the role of Nb in these lower manganese weld metals, it does appear to show that, as observed by Jesseman, an addition of 0.02 percent Nb increases the amount of ferrite side plates at the expense of polygonal ferrite.

Finally, all Harrison's diagrams indicate that at very fast cooling rates a lath ferrite (L.F.) intragranular constituent resembling bainite can form to replace some of the ferrite side plates or acicular ferrite. In Harrison's weld metals, the cooling rates required to introduce this constituent are much faster than those normally associated with typical welding procedures and it is unfortunate that to date no CCT curve has been prepared for a niobium bearing higher manganese (> 1.65%) weld metal deposited under a low oxygen potential flux. It seems probable in the light of Levine & Hills work (15) that the combination of lower oxygen and higher manganese, particularly in the presence of an additional hardenability element such as Mo or Nb promotes the formation of this constituent at very much slower cooling rates.


Nb FREE
O_2 = 233 PPM
1200
1000
800
°C
P
AF
M
CAF + Carbide
200
Cooling curve for ΔT =
10 Secs.

TIME (SECS.)


Nb = 0.03%
O_2 = 336 PPM
1200
1000
800
RATU C
PF
FSP
LF
AF
20c
Cooling curve for $\Delta T_{800-500}$ =
10 Secs.

Modified Vector Diagram To Illustrate the Various Effects of Niobium

Dolby (12) in a recent review prepared for the IIW commision IX-J proposed a modified vector diagram based on the original by Garland & Kirkwood, and with one minor addition this is reproduced in this paper in Figure 7. Comparing this diagram to the original (see Figure 3), it will be appreciated that it now takes account of all the various possible roles of niobium described in the previous paragraphs, and also introduces a yield strength vector which in the Garland & Kirkwood model was not considered separately from the precipitation hardening vector.

This new diagram while it is, of course, still schematic, and in no way a predictive model, does show how the addition of niobium could be beneficial or detrimental depending on the precise circumstances. Taken together, with the increased knowledge of the general effects of niobium on transformation behavior at different manganese levels, it should enable a fairly accurate assessment to be made of what to expect with a given wire flux combination *so* long as an anticipated heat input is available. Conversely, from a knowledge of the plate material, its thickness, and the proposed heat input, it should now be possible to select wire flux combinations to provide optimum toughness over a wide range of C, Mn and Nb levels.

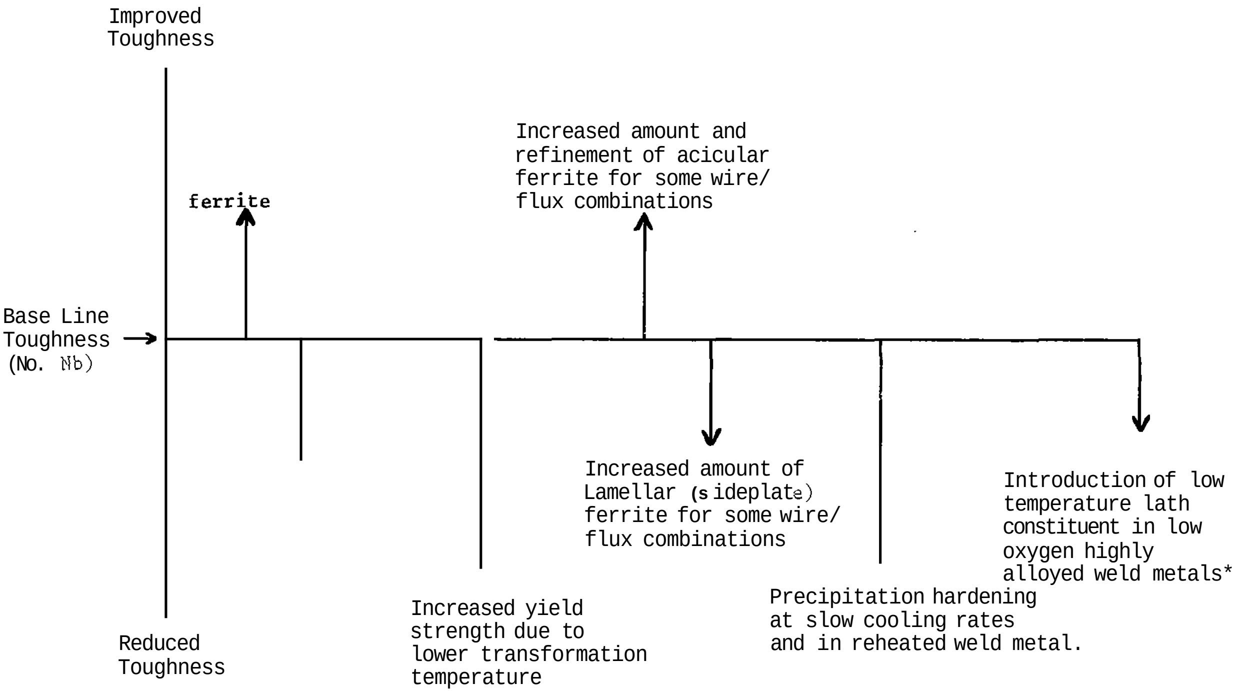
Precipitation Hardening During Stress Relief

The importance of stress relief in niobium bearing welds has already been emphasized, and results presented to indicate that its overall effects can be beneficial or detrimental depending on many other factors (see Figures 2 & *4)*. The detrimenal component is that associated with precipitation hardening, and as shown by Bosansky et al (21) the nature of the precipitation can vary depending on the type of overall microstructure under consideration. Additionally, the kinetics of precipitation will certainly depend on the thermal history prior to stress relief and on whether or not any precipitation or pre-precipitation clustering is present in the as-welded condition.

Rosansky et al (21) suggest that the fine strengthening precipitates which form during the initial cooling of the weld metal are only in the polygonal ferrite via the interphase mechanism, i.e. that the bulk acicular (or other) phases are relatively free from fine scale precipitation in the as welded condition. On stress relief, however, at a rate dependent on temperature, niobium carbo-nitride precipitates throughout the matrix microstructure with a preferential association for regions of high dislocation density. The precipitates are initially coherent with the matrix and result in significant increases in hardness and yield strength. Bosansky (21) suggests that the precipitates lose their coherency reluctantly, and that even after quite prolonged stress relief at *640* C their growth is *so* limited that they still act as good dislocation traps.

Some alloying elements have the ability to influence the kinetics of niobium carbo-nitride precipitation and Figure 8 shows how Mo pushes the peak hardness to longer ageing times.

The overall results of effects of this type of course vary depending on a number of factors, and since the duration of stress relief is often dependent on plate thickness, cases will arise where the Mo bearing weld appears to have been more prone to strengthening on stress relief. Reference to



* Further vector added by present author.

Figure 7. Modified schematic vector diagram showing general effects of Nb on as-welded toughness (12).

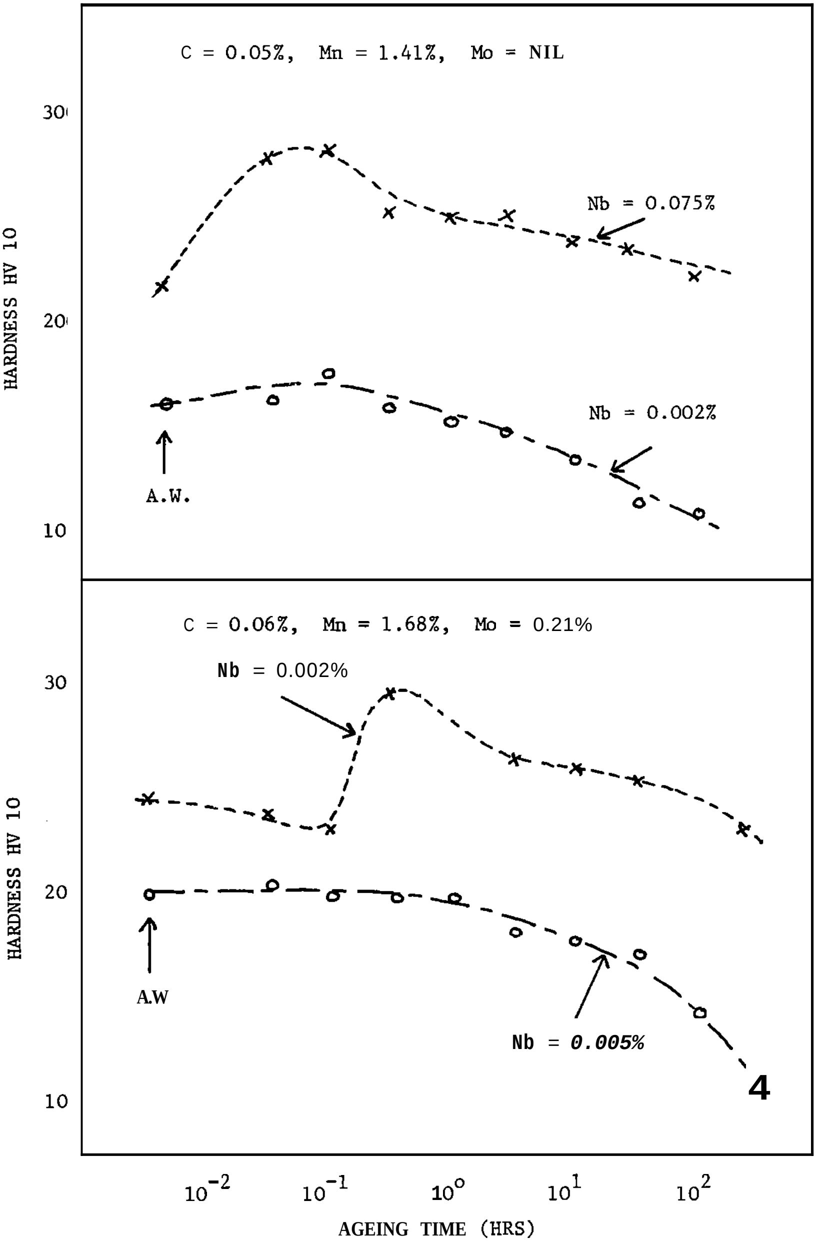


Figure 8. The effect of ageing time at *640* C on the hardness of basic submerged arc weld metals deposited on 15 mm plates at 8.2 KJ/mm, ΔT_{500}^{800} = *340* secs (21).

Figure 8 shows that if welds were stress relieved for between *30* mins and 1 hr. then it might be concluded that the Mo free weld was exhibiting less precipitation than the Mo bearing one. This is clearly not the case, since in 1 hr. the Mo free weld has in fact overaged at *640* C, and this simple example illustrates the care which must be taken in the interpretation of the hardness or yield strength changes recorded following stress relief.

Shiga et al *(22)* have actually shown that alloying the weld metal with boron can also prevent effective Nb (CN) precipitation on stress relief or in reheated regions and the advantages of this are indicated in Figure **9.** The level of boron required to get the maximum benefit from this approach has to be carefully controlled and the optimum level appears to be about 0.01 percent. This is well above the level of boron normally introduced into sub-arc weld metals when it is being added with Ti specifically for improvements to "as welded" toughness, 0.001 percent B being more typical (10), but this approach has apparently been successfully commercialized in Japan for stress relieved weld metals containing up to *0.04* percent Nb *(23).*

Niobium in Manual Metal Arc Weld Metal

In his recent review on the effect of niobium on sub-arc weld metal toughness, Dolby *(12)* concluded that the toughness of re-heated weld metal in multipass welds is likely to show an embrittling effect of niobium as a consequence of secondary hardening from Nb (CN) precipitation, and indeed there is considerable evidence to indicate that this can be a factor of major importance *(24,* 25).

Since the publication of Dolby's review, Abson & Evans *(26)* have reported what would seem to be the first systematic, and certainly the most comprehensive study of the efffect of niobium on manual metal arc weld toughness. These investigators used low hydrogen iron powder type basic electrodes and varied manganese and niobium levels in the. ferroalloy content of the coatings to achieve a wide range of final weld metal analysis. Niobium levels up to 0.17 percent were studied, but to illustrate the general trends recorded, some of their results at lower, more practically relevant niobium levels, are illustrated in Figures 10 & 11. A manganese level of 1.5 percent has been selected because Evans (27) has previously reported that this is the optimum in C-Mn manual metal arc deposits.

Figure 10 shows that Dolby's prediction was accurate and that quite marked detrimental effects on weld meal Charpy properties are observed at higher niobium levels. Figure 11 shows that "as welded" yield strength increased dramatically as Nb was added and this is attributed to a combination of transformation hardening and secondary hardening in the reheated regions of the low heat input multi-pass welds. It can also be seen from Figures 10 & 11 that stress relief *(2* hrs at 580 C) produced further hardening and embrittling effects.

Abson & Evan's microstructure of the as deposited regions in the absence of niobium consisted primarily of acicular ferrite and grain boundary ferrite, the remainder being grain boundary nucleated ferrite with aligned martensitic and carbide constituents. Low additions of niobium up to *0.02* percent, did not radically alter this structure, but above this level grain boundary ferrite decreased and acicular ferrite was progressively replaced by grain boundary ferrite with aligned martensitic and carbide constituents, and some intragranularly nucleated phase of a similar general appearance to the latter grain boundary phase. Thus, as Abson & Evans note (26), niobium does

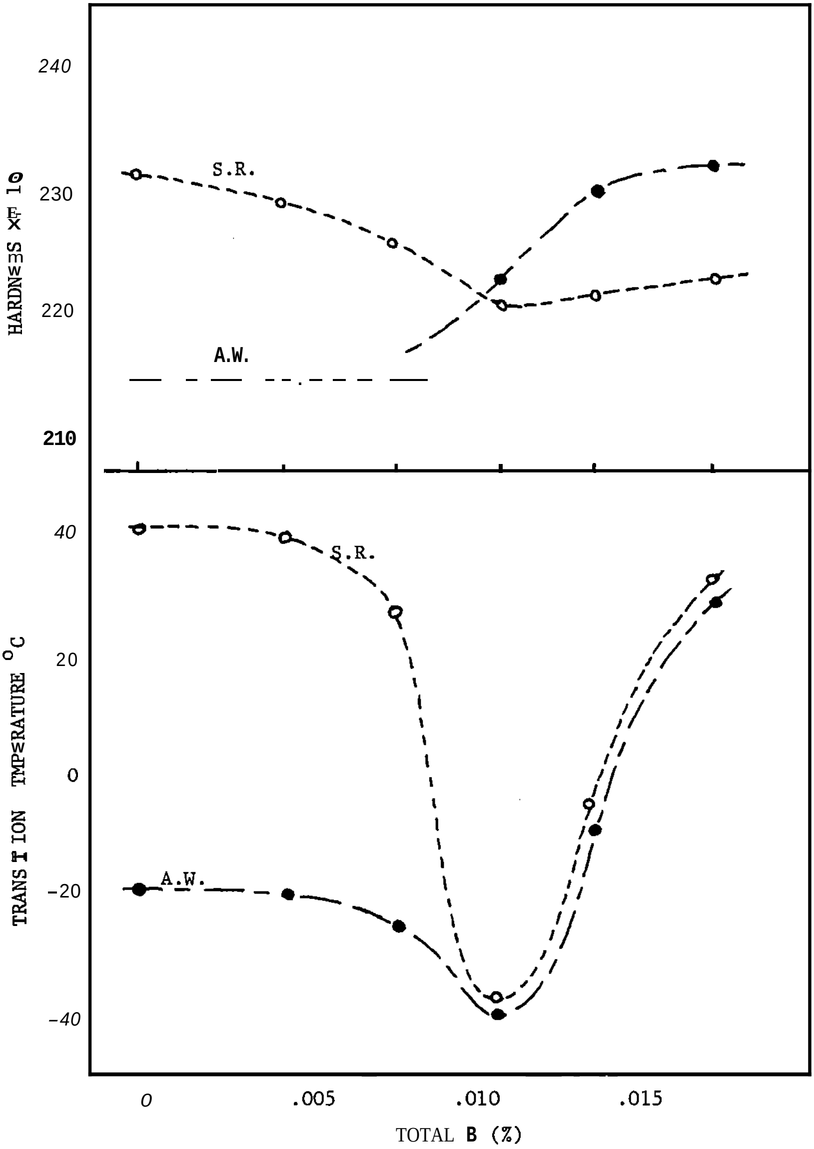


Figure 9. The beneficial effects of Boron on stress relieved niobium bearing submerged-arc weld metal (22).

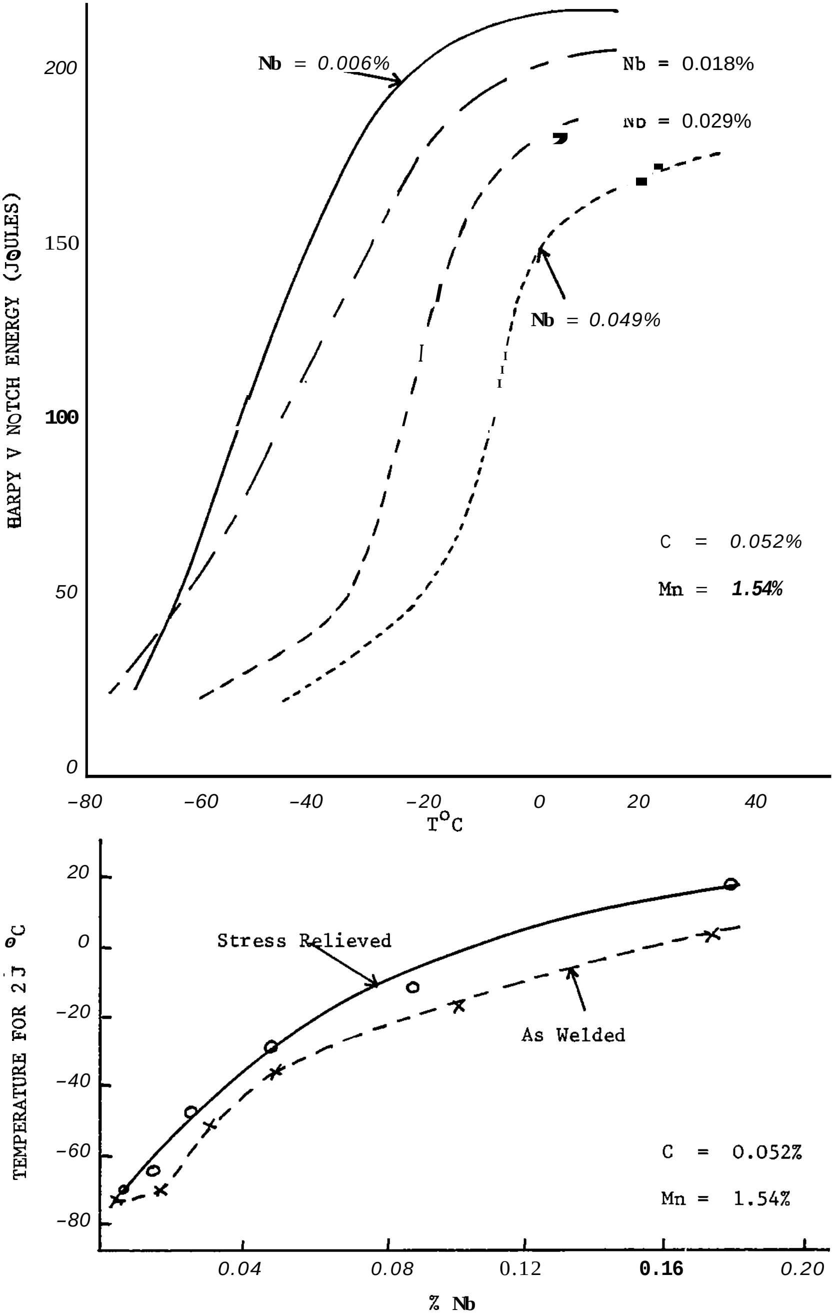


Figure 10. The effect of Nb on manual metal arc weld metal toughness; welds on 20 mm plates at 1KJ/mm heat input $\Delta T^{800}_{500} < 6$ Secs (26).

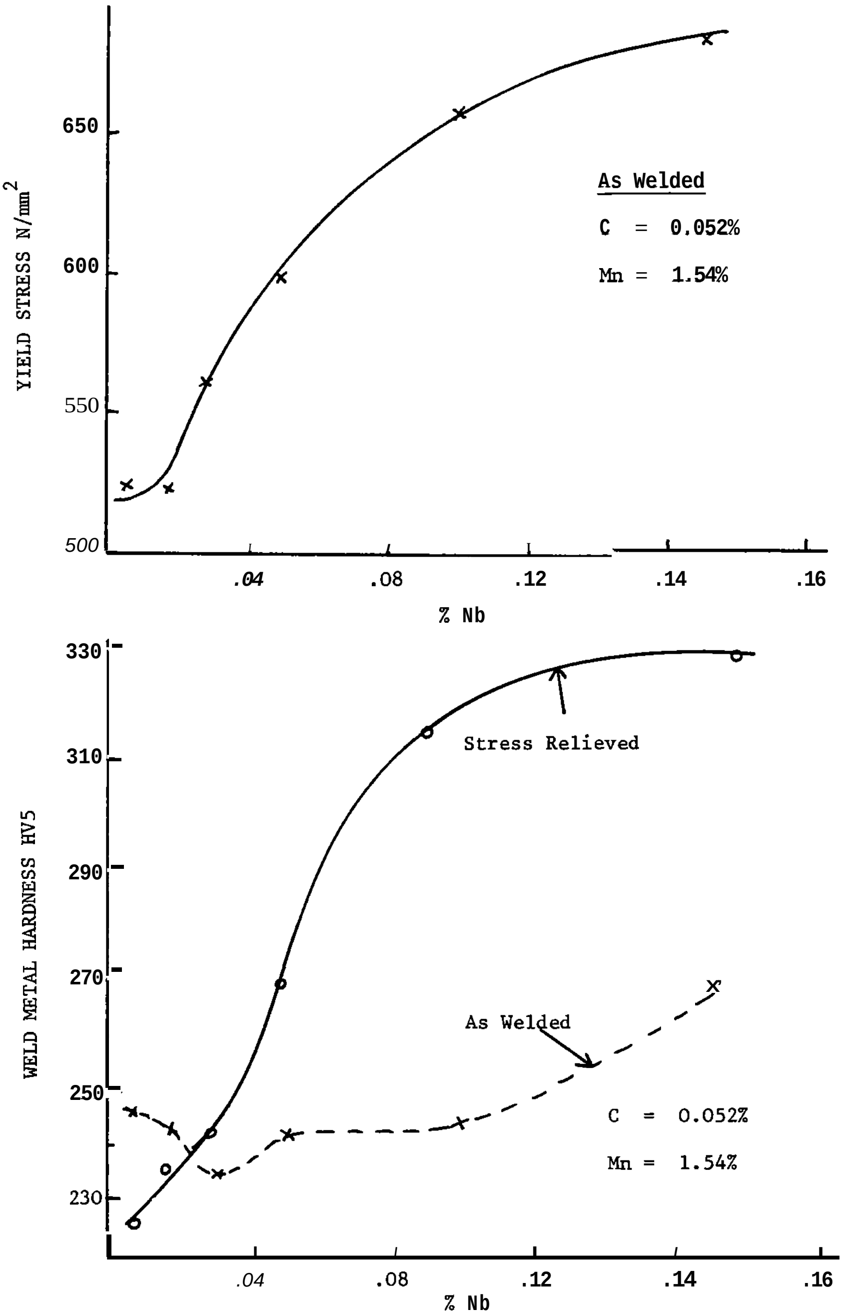


Figure 11. The effects of Nb on manual metal Arc yield strength and Hardness (26).

not, in this instance, further promote the presence of the predominant micro-structural constituent and no immediate explanation for this is apparent.

Again a CCT curve for this type of composition might provide the answer and it should be remembered that though high in manganese, these welds have an oxygen content > 450 ppm more akin to that found in submerged arc welds deposited using acidic fluxes. Higher levels of deposited oxygen are not uncommon with basic electrodes when the coatings contain substantial iron powder additions (28).

It is important that these results should be viewed in the proper perspective, and it must be noted that with steels of conventional niobium levels, i.e. < 0.06 percent, low heat input manual metal arc welding is unlikely to result in weld metal Nb levels much above 0.02 percent, and thus the practical effects on toughness in either the as welded or the stress relieved conditions would be of no consequence. However, some of the newer low carbon high niobium steels (Nb up to 0.15%) could introduce significantly higher niobium levels in the deposit and in this regime Abson & Evan's results could be of great practical significance.

The Effect of Niobium on Heat Affected Zone Toughness

The Nature of the Heat Affected Zone

A typical heat affected zone in a C-Mn microalloyed steel consists of various distinct zones each indicative of the thermal cycle experienced locally during welding. Immediately adjacent to the weld metal there is a coarse grained zone, followed with increasing distance from the fusion boundary by, a grain refined zone, a partially transformed zone, a spherodized zone and finally a region of heat affected base metal which, though it reveals no immediate evidence of the heat cycle may well have been affected, e.g. by dynamic strain ageing.

With the exception of steels which are susceptible to strain ageing embrittlement, the poorest zone in microalloyed steels is the coarse grained region, and it is this latter which is almost exclusively referred to in publications. This paper therefore is concerned solely with such grain coarsened zones and with the influence of niobium on the microstructure and toughness in this area.

Coarse Grained Heat Affected Zone Microstructure

The microstructure observed in the coarse grained H.A.Z. is dependent on transformation temperature, and this in turn is influenced by:

a. Overall chemistry

b. Austenite grain size

c. The inclusion level in the coarse grained austenite

d. The weld cooling rate ΔT^{800}_{500} °C

Item c requires some clarification, and it should be noted that there is some direct evidence and considerable circumstantial evidence to suggest that the HAZ's of cleaner steels transform at lower temperatures. The explanation forwarded for this observation is that inclusions present energetically favorable sites for early ferrite nucleation in the γ/α reaction (29).

Weld metal cooling rate of course depends on welding parameters and overall heat input, and the precise weld thermal cycle affects the degree of austenite grain growth which will occur in practice. Austenite grain size in the "coarse" grained HAZ can range from about 30 μm (in specially developed steels welded at low heat inputs) to ~ 400 μm in C-Mn steels subjected to electroslag welding.

The final microstructure observed depends on all the factors detailed above but it is important to realize just how dominant the effect of grain size can be.

Niobium in Heat Affected Zone

Niobium is generally present in steel to provide strengthening by a combination of precipitation hardening and grain refinement, and is therefore present as fine carbide, nitride or carbo-nitride precipitates (or perhaps partially in solid solution). These fine Nb (CN) precipitates are quite stable, and Hoogendoorn and Spanraft (30) have presented a solubility curve for a rapidly heated Nb bearing steel (samples held 1 hr prior to oil quenching) which indicates as shown in Figure 12 that a temperature of 1200 C was necessary before solution was virtually complete.

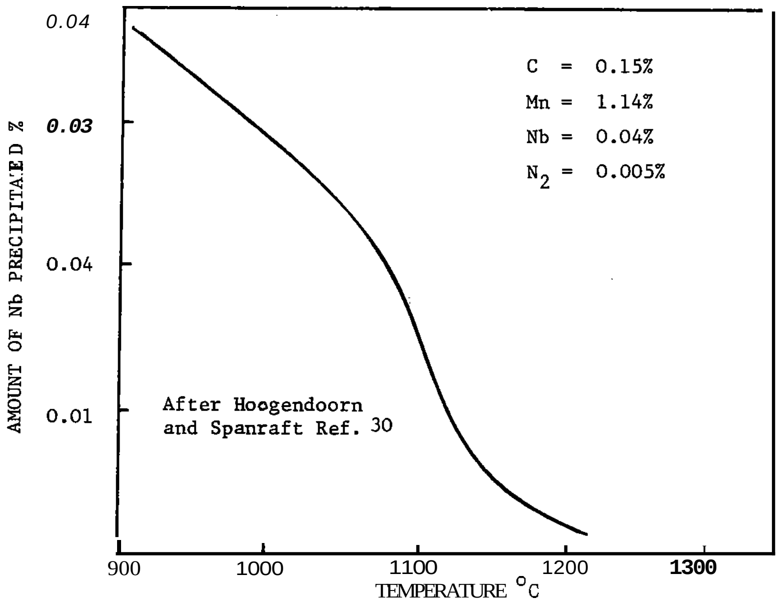


Figure 12. Solubility curve for Niobium Precipitates.

In the fast thermal cycles associated with welding, even with peak temperatures up to 1350 C, it is, therefore, not surprising that some Nb (CN) remains undissolved. These undissolved precipitates may in some circumstances act to retard austenite grain growth in the coarse grained HAZ, and Hannerz (31) has indicated the degree of this effect for 0.18 percent C, 1.3 percent Mn steels. As Figure 13 shows, the effect is markedly influenced by H.A.Z. thermal cycle and would therefore only be of practical significance in manual metal arc welds or lower heat input submerged arc welds.

Clearly, therefore, at the peak temperatures of the weld thermal cycle a proportion of the niobium will be in solution in the coarse prained austenite and on coolinp can act like many other alloying elements to depress transformation temperature. At slow cooling rates as shown earlier for high heat input weld metals, it has been suggested that Nb (CN) can reprecipitate during and after the γ/α transformation and such precipitation has been claimed to be a major factor contributinp to HAZ embrittlement in certain circumstances (31, 32).

Until fairly recently, convincing direct evidence of this precipitation was not available, and previous workers had inferred its role from other experimental observations. Weatherly (33) has, however, now made some direct observations of Nb (CN) precipitation using careful carbon replica techniques. His investipations were carried out on carbon (.09%), manganese

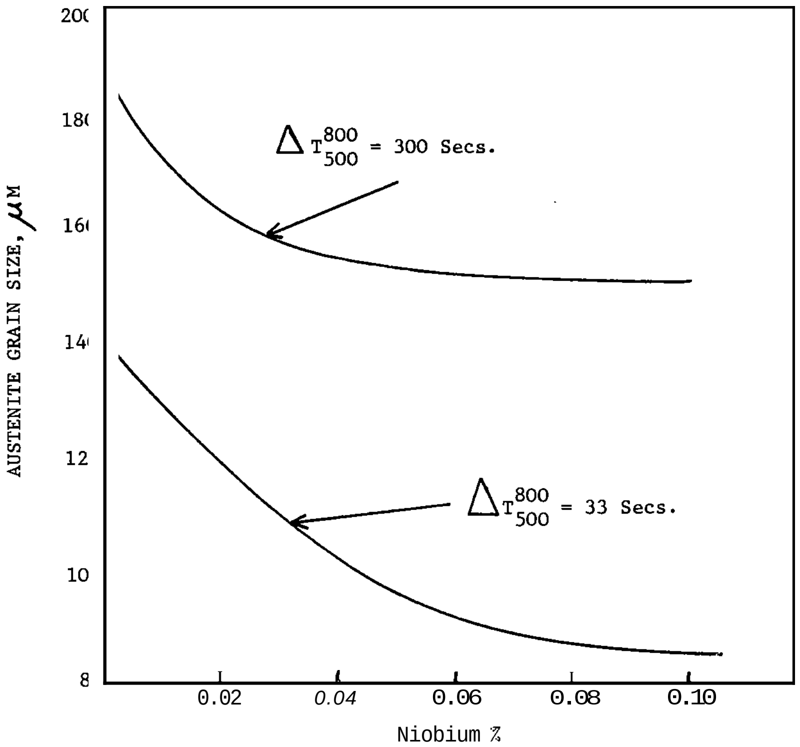


Figure 13. The influence of Nb on HAZ austenite grain size (31).

(1.30%) steel samples with niobium contents in the range .034 to .14 percent. The results as expected depend critically on cooling rate, and at $\Delta T \frac{800}{500} >$ 80 secs, Weatherly observed fine Nb (CN) precipitates in the 40 to 50A size range. At a faster cooling rate, however, $\Delta T \frac{800}{500} = 39$ secs, even in the steel containing 0.14 percent Nb only relatively coarse precipitates were observed, and these were concluded to be evidence of incomplete dissolution during the application of the weld thermal cycle.

The potential effects of niobium in the H.A.Z. can therefore be summarized as follows:

a. Niobium may have a significant effect ony grain size in the H.A.Z. of lower heat input welds (this will act to decrease hardenability).

b. Niobium in solution will usually increase hardenability.

c. Nb (CN) may precipitate at slow cooling rates.

Niobium Detrimental to H.A.Z. Toughness in High Heat Input Welds

The complex interacting effects described above, make the precise prediction of the role of niobium in any given situation difficult, and it is easy to understand why the available literature on the subject is, at first sight, so apparently confusing.

At higher heat inputs, where much of the niobium is in solution and at slow cooling rates, where precipitation effects are likely to be maximized, it might be expected that an overall detrimental effect of niobium would be recorded. This should be particularly the case where niobium has only a marginal influence on transformation behavior as might be expected in higher carbon steels (carbon dominating) or where niobium actually depresses the transformation into a temperature regime in which a poor microstructure is to be expected.

A good example of the former category comes from the work of Hannerz which, in 1975, sparked off much of the current interest in the effects of niobium in H.A.Z.'s (31). Hannerz studied a series of steels containing .17/.20 percent C and 1.29/1.40 percent Mn, and varied his niobium levels from zero to 0.11 percent. Some of his results from H.A.Z. thermal simulation samples, are reproduced in Figure 14, and while there were minor changes in gross microstructure, Hannerz invoked the theory of embrittling precipitation to explain the effects observed. In the absence of direct evidence for precipitation, Hannerz inferred the role of the latter from the very marked increases in hardness recorded in the presence of niobium.

Dolby (32) on the other hand, was able to explain his H.A.Z. results from high heat input electroslag welds in 0.16 percent C, 1.2 percent Mn steels by noting that the presence of 0.056 percent Nb inhibited the formation of pro-eutectoid ferrite at austenite grain boundaries, and promoted the presence of a microstructure dominated by ferrite side plates. This change in microstructure naturally resulted in poor toughness, and while Dolby did not overlook the possibility of precipitation hardening, its presence was not an essential feature of an adequate model to explain the detrimental effects of niobium in this instance.

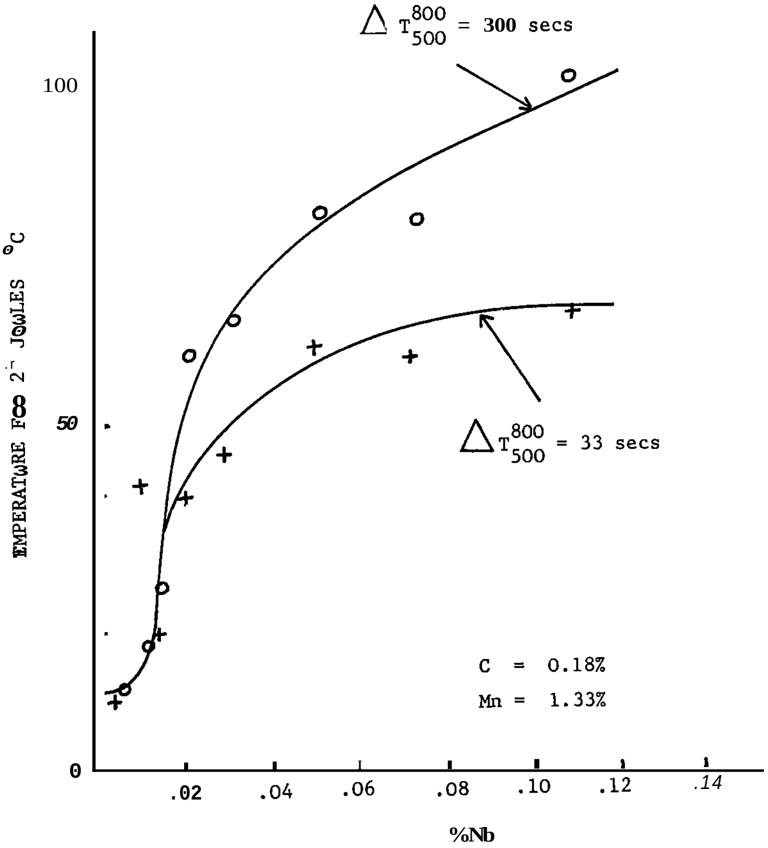


Figure 14. The influence of niobium on thermally simulated heat affected zones (31).

The Importance of Niobium Level and H.A.Z. Cooling Rate

Fortunately, other workers have shown that the overall effect of niobium need not always be detrimental and in doing so have illustrated the importance of other factors such as total chemical analysis, niobium level, and in particular, H.A.Z. cooling rate. Figure 15 shows some results from work at The French Welding Institute (34) on a steel of 0.16 percent C, 1.46 percent Mn and 0.043 percent Nb (with a Nb free steel for comparison). Like Hannerz (31), this French work relied on the use of thermally simulated samples to study the role of H.A.Z. cooling rate and, as Figure 15 shows, ΔT_{500}^{800} values of < 15 secs resulted in excellent toughness in the niobium bearing H.A.Z.'s. This was shown to be associated with a microstructure of predominantly autotempered martensite, and only when the cooling rate was progressively decreased did lower bainite and subsequently Widmanstatten ferrite appear with the resultant deterioration in toughness. In this particular study (34) it is not really possible to comment confidently on what would have happened to the C-Mn steel at fast cooling rates since the regime $\Delta T_{500}^{800} < 20$ secs has not been adequately covered, but it seems likely that the results would have been inferior to those with the niobium bearing steel.

That niobium can indeed be beneficial at fast cooling rates (and even intermediate cooling rates), has been comprehensively demonstrated by Rothwell (35). Figure 16 reproduces some of his results from real welds produced by the submerged arc process on 12 mm C(0.09%), Mn (1.29%) plates of varying niobium content. Rothwell used the instrumented Charpy test, and defined the transition temperature for each H.A.Z. as the point at which the fracture load/temperature curve intersects the extrapolated yield load/temperature line.

Figure 16 clearly shows the importance of cooling rate, and it is noteworthy that, at this lower carbon level, niobium was only detrimental at the slowest cooling rate studied. It would appear that at ΔT_{500}^{800} values < 40 secs niobium could have overall beneficial effects and this lends weight to the tentative conclusion drawn earlier from the French Welding Institute work.

Rothwell (36) went on to study the influence of niobium on the H.A.Z. transformation characteristics of his experimental steels using a high speed dilatomer and an extract from his very detailed results is included in Figure 17. Note that at very fast cooling rates Nb has no effect on transformation temperature but that a significant role is played, particularly in the early stages of transformation, as the cooling rate is progressively decreased.

It is concluded (36) that at the fastest cooling rates when mean (50%) transformation temperatures are of the order of 480 C that good toughness can be associated with the presence of predominantly martensitic microstructures. As the cooling rate is decreased, martensite is progressively replaced by parallel lath bainitic structures and ultimately when start temperatures rise above 680 C grain boundary ferrite appears. The role of niobium therefore depends, as in the case of weld metals (see Section 3.5), on what the H.A.Z. microstructure was in the absence of this element, and that, of course, depends on overall chemistry and cooling rate.

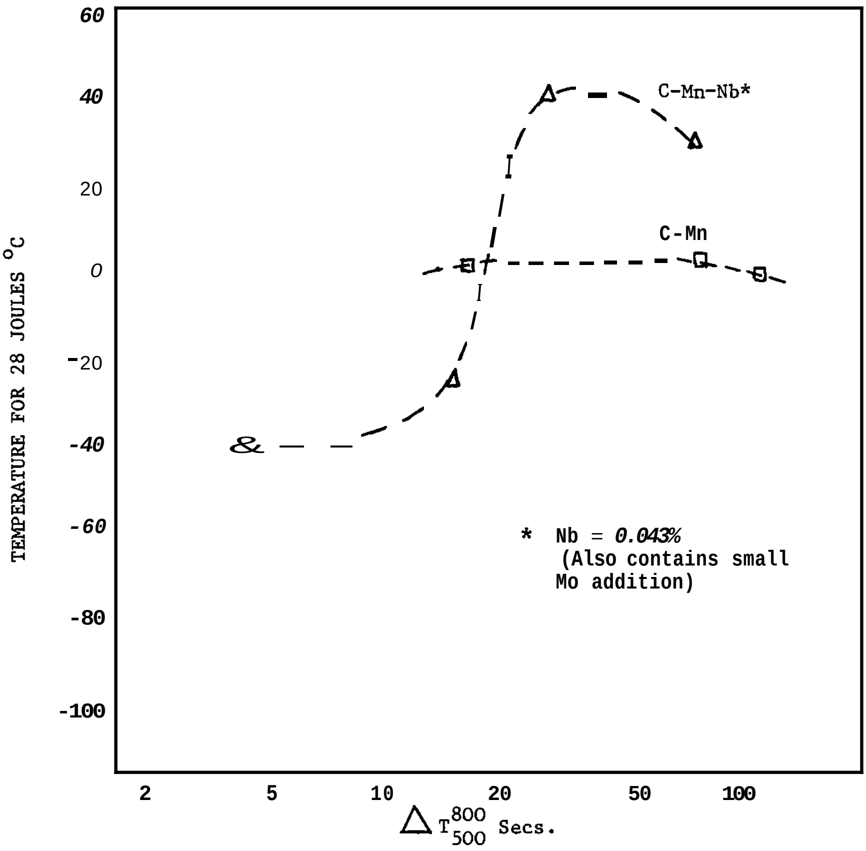


Figure 15. The effect of niobium on the Charpy V-notch impact behavior of thermally simulated HAZ's (34).

WELDS ON 12mm THICK PLATES

C = 0.09% Mn = 1.29%

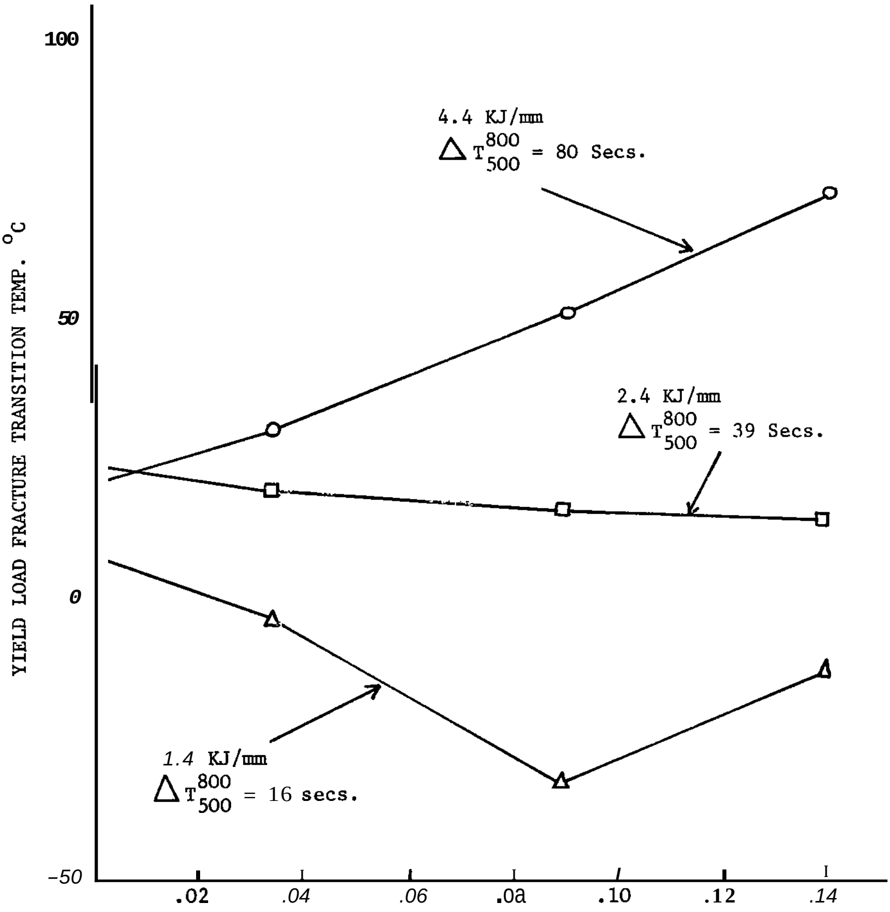


Figure 16. The importance of weld cooling rate on the effect of niobium on H.A.Z. **toughness;** (35).

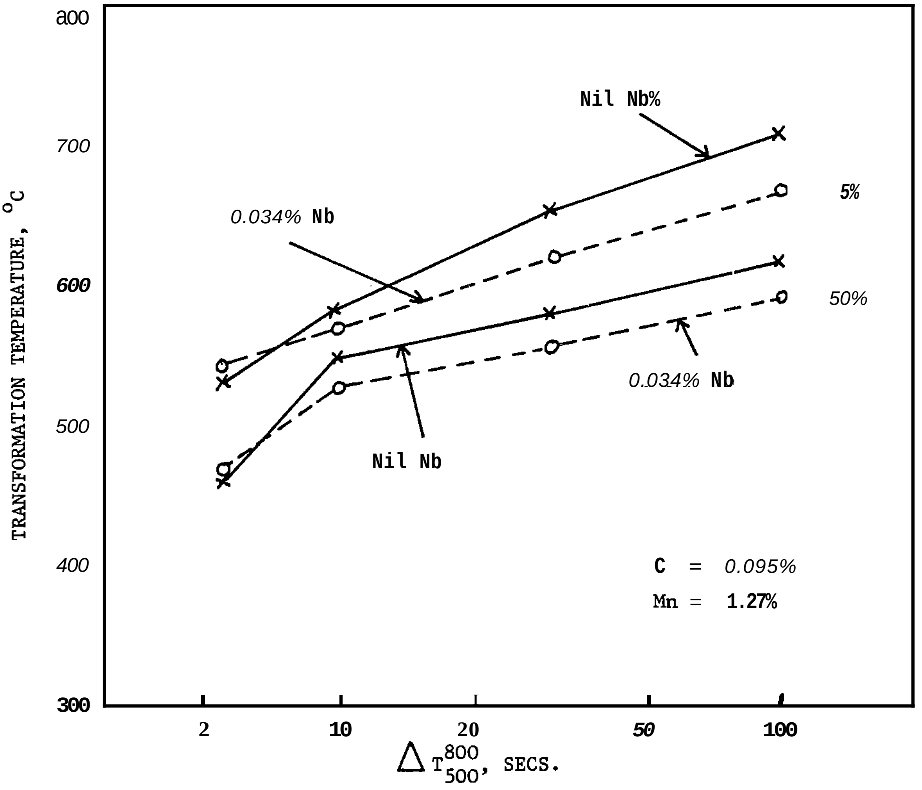


Figure 17. The influence of Nb on HAZ transformation temperature at an intermediate carbon level (36).

Carbon Content and Transformation Behavior of Niobium Bearing Heat Affected Zones

The dominant effect of carbon in relation to the importance of niobium is well illustrated by the very extensive investigation carried out by Rothwell & Bonomo (37), and Figures 18 and 19 present some of their thermal simulation results at carbon levels below and above that discussed in 4.5.

Figure 18 shows that as in the earlier work on 0.095 percent C steels reported by Rothwell (36) niobium is again beneficial at the 0.052 percent C level except at $\Delta T^{800}_{500} > 30$ secs. However, at a carbon level of 0.13 percent niobium is detrimental over the full range of cooling rates studied. Rothwell & Bonomo show that at even higher carbon levels in plain C-Mn steels this trend continues and their results therefore agree with those of Hannerz (31) who recorded the detrimental effects shown in Figure 14 at the 0.18 percent C level.

As is shown in Figure 19, carbon has a powerful influence on the effect of niobium on H.A.Z. transformation temperature. At the 0.052 percent C level the presence of 0.057 percent Nb depresses the mean transformation temperature by 30 to 50 C over the complete cooling range, and the effect of ΔT^{800}_{500} is clearly secondary. However, at the 0.13 percent C level niobium has virtually no effect and transformation temperature appears to be greatly influenced by HAZ cooling rate. Cross reference with Figure 17 from Rothwell's (36) 0.095 percent C, H.A.Z.'s indicates a clear and consistent trend, and it is obvious that an understanding of the importance of both C and Nb (at any given Mn level) is essential if all the available results in the literature are to be rationalized.

At the lowest carbon level, 0.052 percent, the transformation temperature is high (~ 550°C) even at the fastest cooling rate ($\Delta T^{800}_{500} = 5$ secs) and the suggestion in Rothwell & Bonomo's paper (37) that the H.A.Z. microstructure is upper bainitic is clearly correct. The addition of niobium drops the transformation temperature at the same cooling rate to just over 500 C (see Figure 19), and the toughness improves probably as a result of the replacement of the bainite with low carbon autotempered matensite.

However, at a carbon level of 0.13 percent, the situation is much more carbon controlled and since, as shown in Figure 19, there are only minor effects of niobium on transformation temperature, the toughness trends shown in Figure 18 must be indicative of either more subtle effects of niobium on microstructure and probably, at slower cooling rates, the influence of precipitation hardening.

In C-Mn steels at the 0.13 percent C level, the transition from upper to lower bainite takes place in the 440 to 480 C temperature range (38), but the precise temperature may vary for different austenite grain sizes and niobium levels, thus in the 400 to 500 C range, various mixtures of upper and lower bainite are possible.

With or without niobium, once the transformation temperature is below about 430 C mixtures of lower bainite and autotempered martensite would be expected at the 0.13 percent carbon level and good toughness should be, and in fact is, observed (see Figure 18).

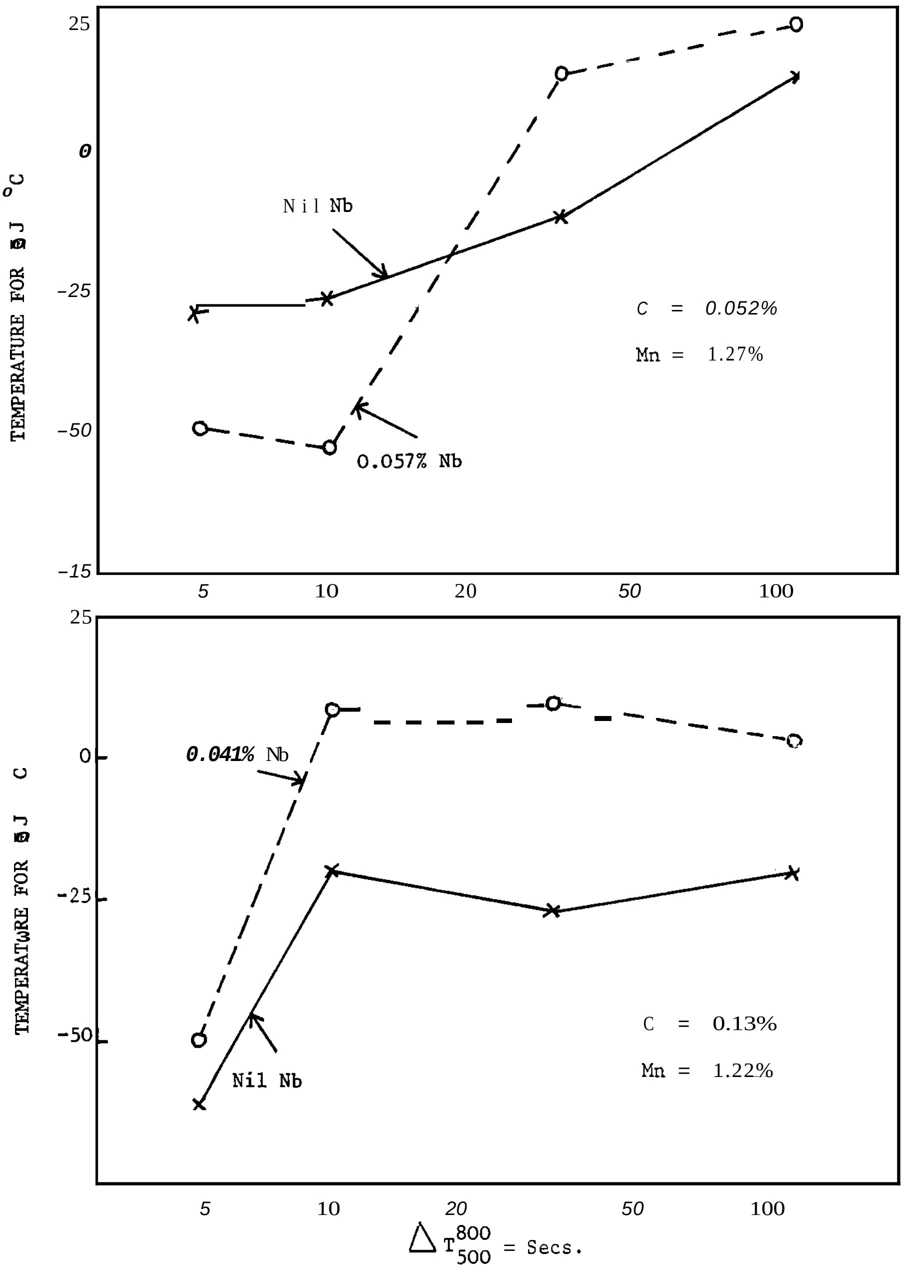


Figure 18. The influence of carbon on the effect of Nb on HAZ toughness (37).

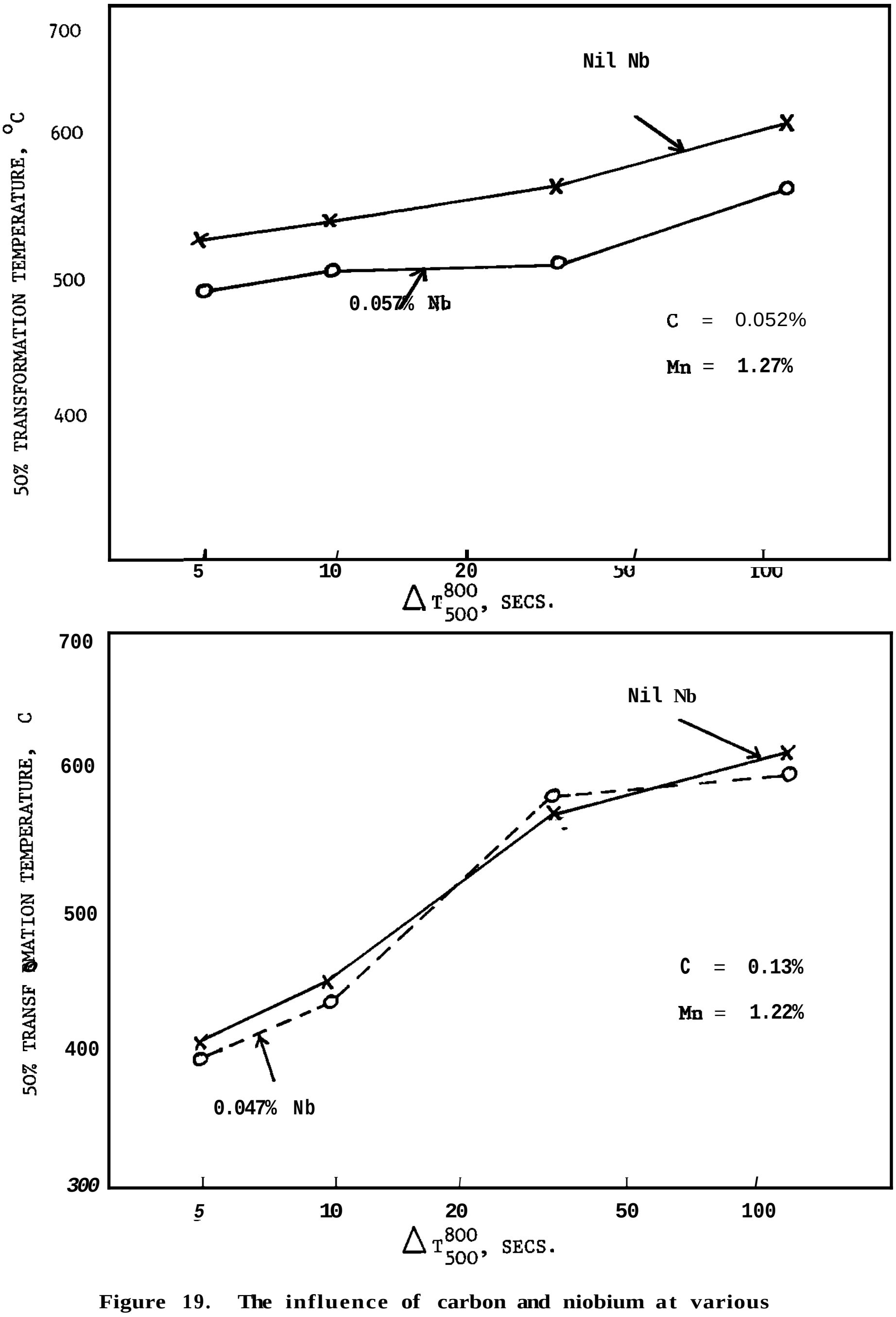


Figure 19. The influence of carbon and niobium at various cooling rates following welding in HAZ transformation temperature (37).

Further Alloying To Control the Toughness of Niobium Bearing H.A.Z.'s

It is clear from the above considerations that transformation temperature is of primary importance, and that Nb may appear to be beneficial or detrimental depending on a variety of other compositional and thermal history variables. However, by understanding the potential effects of Nb it is frequently possible by the judicious use of other minor alloying elements to tailor the steel composition to suit a particular set of welding circumstances and there are various examples of this in the literature.

Jesseman and Schmid (39) have provided a useful illustration of this approach for welds in 12.7mm ship plate. They observed that, in 0.12 percent C, 1.5 percent Mn, 0.25 percent Ni, 0.2 percent Cu aluminum killed steel, an addition of 0.05 percent niobium was specifically beneficial to coarse grained HAZ toughness over the heat input range 1-3 KJ/mm. A few of their results are reproduced below to illustrate this point.

	HEAT INPUT KJ/mm	TEMPERATURE FOR 27 JOULES
Without	1.0	< - 60°C
Niobium	3.0	- 40°C
With	1.0	- 75°C
Niobium	3.0	- 60°C

Fortunately Jesseman and Schmid (39) took the trouble to study the associated microstructural changes which occured when Nb was added and they made the following observations.

In the 2 to 3 KJ/mm range the improvement due to the addition of Nb was primarily attributed to the replacement of coarse upper bainite or ferrite side plate microsructures with either partially acicular or lower bainitic microstructures with a finer colony size. At the lower heat input, 1 KJ/mm, the cooling rate was sufficiently fast to produce martensite/lower bainite even in the Nb free HAZ. and the addition of the latter, therefore, produced a similar microstructure, but with a reduced colony size (perhaps as a result of austenite grain growth inhibition due to the presence of Nb (CN) precipitates).

It is important to note that if Cu and Ni had not been present then at this carbon level (0.12%), as noted in earlier, a quite different rôle of niobium would probably have been recorded.

Shiga et al (40) have illustrated similar beneficial effects of niobium in Cu and Ni treated linepipe steels (0.057% C, 1.70% Mn) and in fact have shown that a tolerance to very high levels of niobium can be realized, see Figure 20. This observation could have important implications for the development of a new generation of pearlite reduced hot rolled H.S.L.A. linepipe steels which could provide API-5L-X65 or X70 properties without controlled rolling. Considerable welding work on such steels has already been collated and reported by Gray and Stuart (41), and the progress of this development will be interesting to monitor.

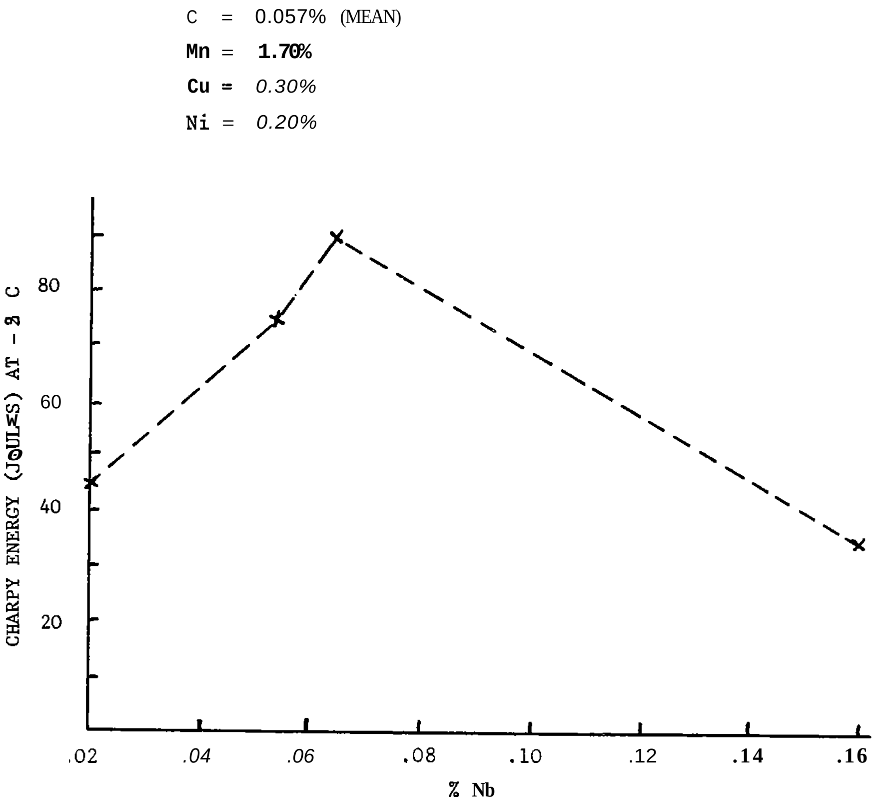


Figure *20.* **The effect of Nb on the HAZ notch toughness of Cu and Ni alloyed line pipe steels;** *12* **mm plates welded at** *2.34* **KJ/mm, ΔT_{500}^{800} = 36 Secs** *(40).*

Precipitation Hardening Embrittlement During Stress Relief

As with weld metals, precipitation hardening in niobium bearing H.A.Z.'s is different to avoid on stress relief, and as shown by various authors including Rothwell (35), it can produce marked detrimental effects on toughness, see Figure 21. Rothwell has demonstrated how this embrittlement is associated with significant increases in H.A.Z. hardness (see Figure 22) and has concluded, reasonably, bearing in mind the observations of Hannerz (31) and Weatherly (33), that precipitation hardening is responsible. In fact, Figure 22 shows how the addition of niobium has a much greater effect on stress relieved H.A.Z. hardness than it has in the as welded condition, and these results can only be explained by a significant precipitation hardening contribution.

In the case of weld metals, it was shown how alloying with boron could offset some of the worst effects of precipitation during stress relief (see 3.7), but in the case of the H.A.Z. it is much more difficult to conceive a successful approach along these lines. There are, however, of course many beneficial efffects of stress relief and the precise role of niobium will depend on what the microstructure of the H.A.Z. is in any given situation. Niobium bearing steels are widely used in many stress relieved applications, and it is suggested that care need only be taken when the specification toughness requirements are particularly onerous or where new unfamiliar steel types, with higher niobium levels, are being considered for the first time.

Summary and Concluding Remarks

This review suggests that niobium has no significant effects on general aspects of weldability such as heat affected zone hydrogen cracking, liquation cracking or weld metal solidification cracking, and that the controversy in earlier literature relates exclusively to its influence on weld metal and heat affected zone toughness.

It has been demonstrated for both weld metal and the heat affected zone, that toughness depends primarily on microstructure, which in turn, depends on hardenability and related transformation temperature. The role of a micro-alloying element such as niobium is therefore complex, and will be markedly influenced by other compositional and welding procedural variables.

For weld metals, the importance of overall composition in terms of carbon and manganese level has been emphasized and special factors related to individual welding processes have been highlighted, e.g. the role of flux selection in submerged arc welding which controls the oxygen level in the deposit, and for manual metal arc welding, the significance of the multi-run nature of the deposit (resulting in increased volumes of reheated weld metal).

It has been shown that niobium is just as likely to improve weld metal toughness as to impair it, and it is concluded that the literature of the late Sixties and early Seventies was unduly pessimistic, simply because the situations studied were those where, in the light of later experience, a detrimental effect could have been anticipated, i.e. moderately high heat input welds in high carbon lower manganese steels.

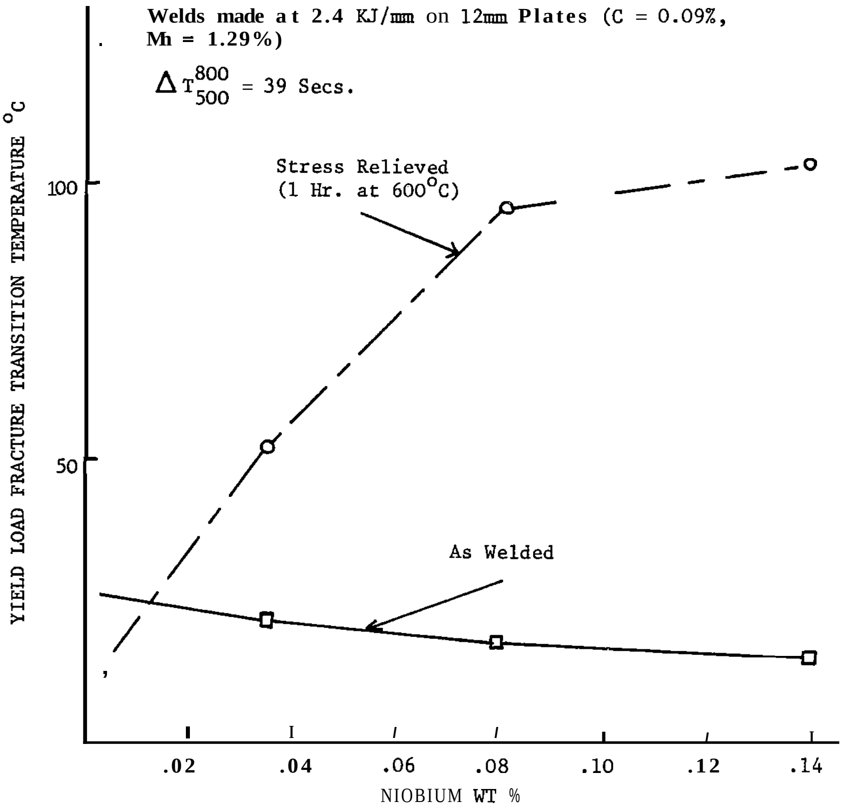
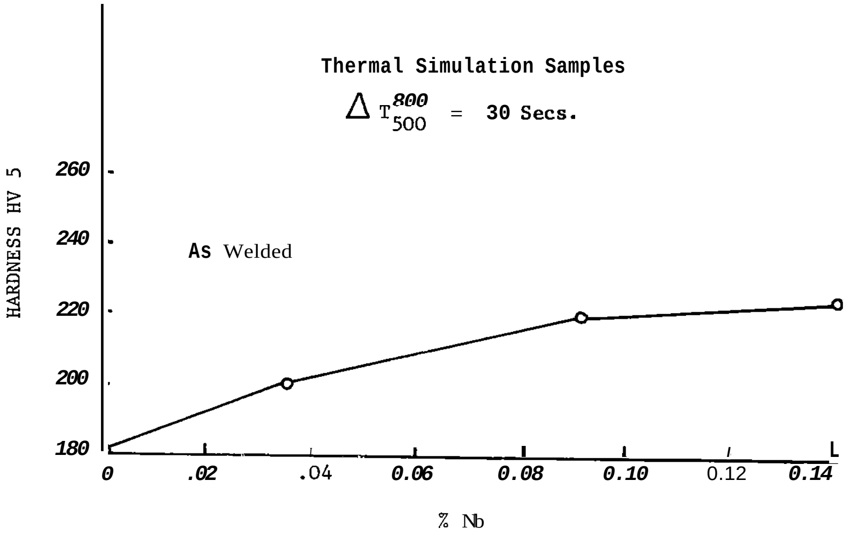


Figure 21. The effect of stress relief on the HAZ impact properties of Nb alloyed low carbon base plates; (35).



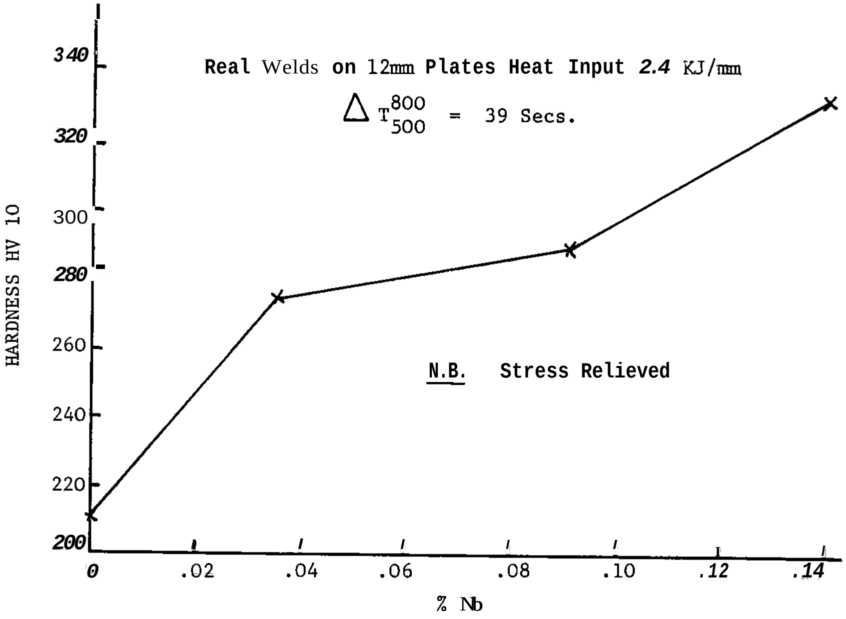


Figure 22. The effect of Nb on HAZ hardness (35).

Similarly, the early observations on the effect of niobium on H.A.Z. toughness which painted a black picture can now be understood and appreciated in their proper context. As for the weld metal, many of these early observations were from compositions and at heat inputs where we would now expect to record a detrimental effect of niobium, and it has been shown that there are many situations in which niobium is specifically beneficial.

The role of overall analysis, and in particular the importance of carbon content in controlling transformation temperature and microstructure of the H.A.Z., has again been emphasized and it is shown how the effects of niobium can be expected to vary over a wide range of plate compositions and welding conditions.

The role of precipitation hardening embrittlement has been considered for both the weld metal and the heat affected zone, and it has been suggested that in the "as welded" condition this is only likely to play a significant role in slow cooling situations, e.g. in high heat input submerged arc welding or in electroslag welding. In the stress relieved condition, however, care must be taken, and it is probable that Nb (CN) precipitation will occur to some extent. Whether or not this will produce overall detrimental effects depends on many other factors and these are adequately discussed in the text. It should, however, be noted that, at least for weld metals, there are many situations in which the overall effect of stress relief is beneficial and the reasons for this have been discussed.

Finally, it is suggested that niobium is no more, and no less important in the steel analysis than carbon, manganese, nickel and molybdenum, and that if steelmaking economics dictate, or favor the selection of a niobium treated steel for a particular application, it should be possible, with a knowledge of the welding thermal cycle to be met with in practice, for the welding metallurgist and steelmaker, between them, to tailor the overall composition to provide a satisfactory product.

Progress towards the most efficient use of niobium will, therefore, only be made when the liaison between end users and steelmakers is improved and it should be recorded that there are many signs that this has begun to happen over the past few years. We should therefore be optimistic in the present decade, and it is hoped that the increased knowledge now available will be reflected in future publications and specifications, and that a more balanced view of the role of elements such as niobium, will be seen to evolve.

Acknowledgement

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References

1. Altenburger, C. L. - Niobium Treated, Mild Carbon Semi-Killed Steel, AISI. Regional Technical Meeting No. 59, Buffalo, NY 1960.

2. Hrivnak, I. - The Mutual relationship and interdependance of Developments in Steel Metallurgy and Welding Technology. Welding In The World Volume 16, No 7/8 1978.

3. B. S. 5135: 1974 - Specification for Metal Arc Welding of Carbon and Carbon Manganese Steels.

4. Colvin, P. & Bush, A. F. - Multipass Submerged Arc Welding of High Yield Strength Notch Ductile Ferritic Steels. Iron & Steel 42, 1969, p. 361.

5. Colvin, P. - Notch Toughness Properties of Single Pass Submerged Arc Weld Metal on E.H.2. Shipbuilding Quality Steels. Welding & Metal Fabrication, February 1971.

6. Hannerz, N. E., Valland, G. and Easterling, K - Influence of Niobium on the Microstructure and Mechanical Properties of Submerged Arc Weld Metals for C-Mn Steels. I.I.W. Documents 11-612-72, Xll-A-46-72 and IX-798-72.

7. International Standard Organization Document - Draft Recommendations by the ISO/TC 17/SO3 Committee.

8. Wegrzyn, J. - Weldability of Microalloyed Steels. Polish Institute of Welding Report 1976.

9. Garland, J. G. & Kirkwood, P. R. - The Noth Toughness of Submerged Arc Weld Metal in Micro-Alloyed Structural Steels. I.I.W. Document IX-892-74.

10. Garland, J. G. & Kirkwood, P. R. - Towards Improved Submerged Arc Weld Metal Toughness. Metal Construction, Volume 7 Nos. 5 & 6, May & June 1975.

11. Biss, V. & Cryderman, R. L. Met. Trans 2.2267, 1971.

12. Dolby, R. E. - Review of Work on the Influence of Niobium on the Microstructure and Toughness of Ferritic Weld Metal. I.I.W. Document IX-1175-80 (IX-J-27-80).

13. Jesseman, R. J. - Niobium Pick-Up in High Solution, Sumberged Arc Weld Deposits. Microalloying 1975 - Proceedings of International Symposium, October 1-3 1975, Washington, USA. Union Carbide, p. 578.

14. Levine, E. & Hill, D. C. - A Review of the Structure and Properties of Welds in Niobium and Vanadium Containing High Strength, Low Alloy Steels. Welding Research Council Bulletin No. 213, p. 20.

15. Levine, E. & Hill, D. C. – Toughness in H.S.L.A. Steel Weldments. Metal Construction, Volume 9, No. 8, August, 1977, p. 346.

16. Ito, Y. & Nakanishi, M - Study on Charpy Impact Properties of Weld Metal with Submerged Arc Welding - The Sumitomo Search No. 15, May 1976, p. 42.

17. Cochrane, R. C. & Kirkwood, P. R. - The Effect of Oxygen on Weld Metal Microstructure. Welding Institute International Conference "Trends in Steels and Consumables for Welding", London 13-16 November 1978, Paper 35.

18. Abson, D. J. et al - The Role of Non-Metallic Inclusions in Ferrite Nucleation in Carbon Steel Weld Metals. Conference as above, Paper 25.

19. Abson, D. J. - A Study of the Influence of Carbon and Silicon on the Microstructure and Toughness of Submerged-Arc Welds in 25 mm C-Mn Steel Plate. Welding Research International Vol. 9, No. 5 1975, p. 1.

20. Harrison, P. L. et al - How Niobium Influences S. A. Mild Steel Weld Metals - Part 2 - Continuous Cooling Transformation (CCT) Diagrams. Welding and Metal Fabricaton, April 1981 p. 161.

21. Bosansky, J. et al - The Effect of Stress Annealing Treatments on the Structure of High Heat-Input Welds Containing Nb and Mo. Scandinavian Journal of Metallurgy No. 6 1977, p. 125.

22. Shiga, A. et al - Effects of Niobium and Vanadium on Toughness of Submerged Arc Weld Metal. I.I.W. Document IX-1049-77 (XII-A149-77).

23. Shiga, A. - Private Communication June, 1981.

24. Abson, D. J. & Hart, P. H. M. - An Investigation into the Influence Niobium and Vanadium on the Microstructure and Mechanical Properties of Submerged Arc Welds in C-Mn STeels, Part 1. Welding Institute (UK) Report No. 122/1980.

25. Yoshino, Y. & Stout, R. D. - Effect of Microalloys on the Notch Toughness of Linepipe Seam Welds. Welding Journal, Research Supplement, 58 (3), 1979, p. 59s.

26. Abson, D. J. & Evans, G. M. - The Effect of Niobium on the Microstructure and Properties of C-Mn All Weld Metal Deposits. I.I.W. Document 11-A-530-81.

27. Evans, G. M. - Effect of Manganese on the Microstructure and Properties of All Weld Metal Deposits. Welding Journal, Research Supplement 59 (3), 1980, p. 67s.

28. Boniszewski, T. - The Effects of Iron Powder in Basic Low Hydrogen All Positional Electrodes. Welding Institute International Conference "Trends in Steels and Consumables for Welding", London 13-16 November 1978, Paper 15.

29. Hart, P. M. - Low Sulphur Levels in Carbon Manganese Steels and Their Effects on H.A.Z. Hardenability and Hydrogen Cracking. Welding Institute International Conference "Trends in Steels and Consumables for Welding", London 13-16 November 1978, Paper 20.

30. Hoogendoorn, T. M. & Spanraft, M. J. - Quantifying the Effect of Microalloying Elements on Structure During Processing. Proceedings of Microalloying 75, October 1-3 1975, Washington, U.S.A., p. 75.

31. Hannerz, N. E. - Effect of Niobium on H.A.Z. Ductility in Constructional H.T. Steels. Welding Journal Volume 54 No. 5, 1975, Research Supplement, p. 162s.

32. Dolby, R. E. - The Effects of Niobium on the H.A.Z. Toughness of High Heat Input Welds in Carbon Manganese Steels. The Welding Institute Research Bulletin, September 1977, p. 250.

33. Weatherly, G. C. - Heat Affected Zone Toughness of Welded Joints in Micro-Alloy Steels - Part III Section B - Precipitation Effects. Canment Report 79-6, January 1979 (I.I.W. Document IX-114-80).

34. E.C.S.C. Final Report - Convention No. 6210-93-3-304 Properties in Use, Weldability of Steels, Weldability of Carbon Manganese and Microalloyed Steels. Report by The Institute of Recherches De La Siderurgie Francaise, St. Germain-En-Laye.

35. Rothwell, A. B. - Heat Affected Zone Toughness of Welded Joints in Micro-alloy Steels, Part 1, Results of Instrumented Impact Tests. Canmet Report 79-6 January 1979 (I.I.W. Document IX-1147-80).

36. Rothwell, A. B. - Above Report, Part III - Continuous Cooling Transformation Characeristics.

37. Rothwell, A. B. & Bonomo, F. - Weldability of H.S.L.A. Steels in Relation to Pipeline Field Welding. Paper presented at the AWS 57th Annual Meeting in St. Louis, Missouri, May 1976.

38. Pickering, F. B. & Clark, B. R. - I.S.I. Special Report 93, 1965, p. 143.

39. Jesseman, R. J. & Schmid, G. C. - Influence of Some Steelmaking Variables on Ship Plate Weldment Toughness. Paper presented at the AWS 60th Annual Meeting held in Detroit, Michigan, April 2-6 1979.

40. Shiga, C. et al in Effects of Nb and V Contents on Mechanical Properties of Controlled-Rolled Plates and Their Welded Portions for X65-X80 Line-pipe. Kawasaki Technical Reports Volume 10, No. 1 1978.

41. Gray, J. M. & Stuart, H. - Developments of Superior Notch Toughness in High Dilution Weldments of Microalloyed Steels. A Niobium Products Co. Ltd. Publication, Application Report 010-1980.