UNDERSTANDING THE WELDABILITY OF NIOBIUM-BEARING HSLA STEELS

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Abstract

The last four decades have seen remarkable progress in the development of High Strength Low Alloy (HSLA) steels for structural applications in pipelines, offshore structures, ships and buildings. These developments have been driven by the need to obtain improved combinations of weldability, toughness and strength in tonnage quantities at affordable prices. They have engaged all sectors of the industry; steelmakers, manufacturers, constructors, operators and safety regulators, and have spanned disciplines from theoretical modeling to manufacturing and site construction activities. At the heart of these developments has been the understanding of the physical and chemical metallurgy of the steel product, how the joining process influences it and how this in turn determines the performance, integrity and reliability of the final installation. Interest has focused particularly on the weld heat affected zone, its microstructure and properties, especially the effect of steel chemistry, weld process and heat input, preheat and post-weld heat treatment on weldability and toughness. While significant improvements in weldability have been achieved over the years through reductions in the carbon equivalent, a consequence of this has been the increased reliance on thermo-mechanical processing and a shift to microalloy-dominated strengthening mechanisms, away from the carbon-dominated mechanisms on which the original understanding was based. In turn, improvements in toughness in one region of the multi-pass weld heat affected zone have exposed other regions as the weakest remaining links. A resurgence of interest in yet higher strength, especially for pipelines, has prompted closer examination of many engineering-critical performance parameters associated with the integrity of the weld zone. The paper examines recent and current research that is directed towards obtaining a more coherent understanding of the effects of welding parameters and steel composition on both weldability and toughness, focusing particularly on the parent metal heat affected zone. It explores how the different influences are kept in balance, thereby ensuring the effective construction and reliable operation of next-generation pipelines and structures.

Introduction

The history of developments in HSLA structural steels during the last four decades has been driven by the need to obtain improved combinations of strength, toughness and weldability in tonnage quantities at affordable prices. Whether the applications have been for energy pipelines, offshore steel structures or ships, the objectives have been broadly similar, even though the balance of requirements may vary as determined by the specific design or operational needs.

The requirement to achieve this objective has engaged all sectors of the industry: steelmakers and product producers, constructors, operators and safety regulators. It is a tribute to those involved at the early stages of development that the need for a holistic approach was soon recognized and embraced to good effect; many of the significant steps were as a result of these interactions, building the broad industry-based confidence essential for making the often significant investments in capital plant and industry practice.

As a result of the specific thermodynamic and kinetic attributes of its carbide and nitride precipitates in steel, niobium can justifiably be considered the enabler of modern, controlled-processed HSLA steels. These steels themselves have certainly been the enablers of efficient and cost-effective design and construction technologies in a variety of applications. In the field of transportation pipelines, for example, the increases in the available strength level of line pipe that have taken place during the last forty years (from X52 to X80, with X100 on the horizon) have delivered cumulative benefits valued in the billion dollar range, largely because they have been achieved without compromising construction methods that were prevalent in the industry.

The similarities and differences in steelmaking, metal forming and welding processes applicable to pipeline and structural steels highlight the wide range of influences on the weld heat affected zone (HAZ). They also point to why so much attention has been focused on developing an understanding of the relationship between microstructure and mechanical properties within the different zones that make up the HAZ.

Similar considerations arise concerning the weld metal in HSLA steel weldments. The relationship between weld-metal composition, microstructure and mechanical properties is even more complex than for the HAZ because, in addition to all the issues identified above, the weld metal composition itself is a function of both the parent and consumable chemistries and the welding process variables.

Many papers have been written during the last 40 years exploring the influence of different parameters on properties and microstructure of both the HAZ and the weld metal. It is not the purpose of this paper to attempt yet another review; rather it is the aim to focus particularly on the parent metal HAZ and draw out some of the key trends and influencing factors that give an overall understanding; first how the metallurgical and thermal factors influence the development of the HAZ microstructure and second how the HAZ microstructure influences the important properties of weldability and toughness.

Trends in Steelmaking and Plate Forming

At the heart of the steelmaking and plate-forming developments has been the understanding of the physical and chemical metallurgy of the steel product; how it is influenced, firstly by the steelmaking and metal forming processes and secondly by the welding process. In combination, these processes determine the integrity and reliability of the final installation, matching the initial design requirements.

The significant advances in steelmaking technology have come about as steelmakers worldwide have invested in new plant and facilities to meet the requirements for low-cost high-quality steel. Major changes include the introduction of basic oxygen steelmaking, improvements in the accuracy of process control, and reductions in sulphur and phosphorus contents through techniques such as ladle steelmaking and ladle refining. In addition to reductions in residual elements, the overall control of composition and homogeneity has improved substantially. Product homogeneity has also been significantly improved by changes from ingot to continuous slab casting technology, and its subsequent extensive refinement, minimizing the consequences and extent of centerline segregation.

A second major trend has been the introduction and refinement of controlled rolling and thermo-mechanical processing. Controlled rolling of pipeline plate steels was first introduced nearly forty years ago. In the early years controlled rolling coupled with heat treatment was used to reduce the carbon equivalent of the normalized product. The later introduction of thermo-mechanical processing enabled further reductions of both carbon equivalent and microalloying content, while increasing parent metal strength without sacrificing weldability. The introduction and refinement of accelerated cooling during the last 20 years has added still further to the ability to achieve strength without extra carbon or microalloying.

Controlled rolling and thermo-mechanical processing were at first only applicable to the thinner products such as plates for pipelines, less than 25 mm (1 in) thick. In time the enhanced performance of processing plant has allowed the application of similar approaches to offshore structural steels, in excess of 50 mm (2 in) thick, resulting in similar benefits; however, because increased strength was not particularly beneficial for many offshore structural applications where plastic collapse and fatigue are more critical, the benefits were largely focused on improving weldability and weld zone toughness. Only for the thickest plates, used for offshore and shipbuilding applications, did normalizing remain as a major manufacturing route.

The original base material was, and remains, carbon-manganese-niobium steel. However, it is interesting to note that the changes in steelmaking and processing technology have led to a progressive fall in the amount of niobium added (Ref 1); between 1972 and 1981 for example the niobium level in offshore structural steels halved from around 0.05% to below 0.03% (Figure 1).

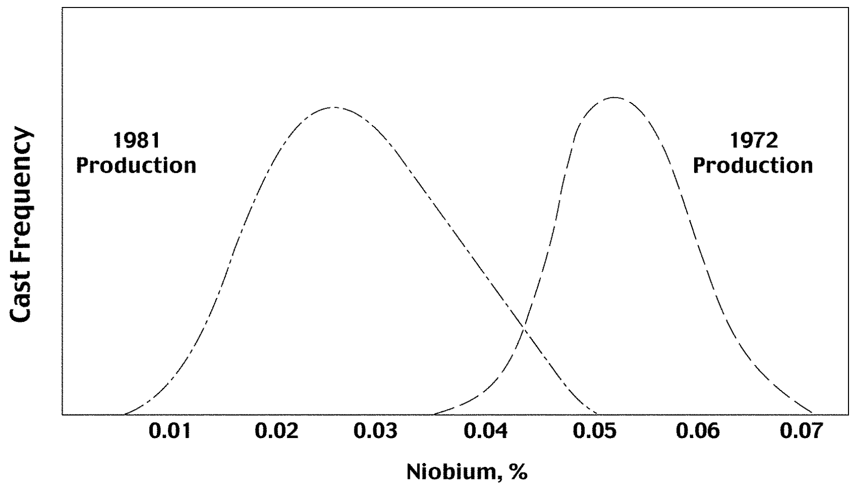

Figure 1: Change in niobium contents of offshore structural steel [from Kirkwood (1)].

This came about largely because of the more efficient use of niobium in conjunction with improved use and control of rolling processes, achieving the same effect of grain size control with less alloying addition. It should also be remembered that during the same period there was an almost proportional reduction in carbon content, and hence a much smaller effect on the balance of niobium to carbon.

Welding Processes

The other main influence on the integrity of the final installation is the welding process. For large-diameter pipelines, much of the early focus was on the achievement of adequate toughness in the submerged arc seam weld, to provide the same security against fracture initiation as in the parent metal (2,3). Such seam welds are typically made in two passes with heat inputs of 2 to 4 kJ/mm (50-100 kJ/in). Attention later turned to manual girth-weld toughness, once the significance of axial stresses arising from geomechanically induced and operational loads had been recognized. The introduction of semi-automatic and mechanized welding processes and the move towards assessment of weld defects on a fitness-for-service basis have focused attention on the toughness of both the weld metal and the heat affected zone (HAZ) of the girth weld (4-8). Typically such welds are made at heat inputs around 0.7-2 kJ/mm (15-50 kJ/in) for manual processes and 0.5-1 kJ/mm (10-20 kJ/in) for automated processes.

For offshore structural steels (9-12), manual and semi-automatic welding processes are widely used, but with the increased plate thickness there are more weld passes and the weld preparation can often be asymmetric. The drive to reduce construction times has led to a slightly higher range of heat inputs (typically 3 – 5 kJ/mm, 75 – 125 kJ/in) than for pipeline girth welds, but this is constrained on the one hand by the need to avoid hydrogen cracking during fabrication and on the other by the need for good toughness in the HAZ.

For shipbuilding applications (Ref 13), a primary requirement is the achievement of high weld deposition rates through high heat input welding. Toughness in the weld zone is less important. Heat inputs for submerged arc welds may be in the range 2 – 10 kJ/mm (50 – 250 kJ/in), while for electroslag welds they may typically be 25 – 50 kJ/mm (600 – 1300 kJ/in). Such high heat inputs can result in wide HAZs and correspondingly coarse microstructures, as will be discussed below.

The Development of Weld Zone Microstructure

The HAZ is made up of a wide variety of microstructures, each influenced by the heating rate, peak temperature and cooling rate during the deposition adjacent weld metal is solidification and cooling (Figure 2).

For the weld heat inputs applied to structural HSLA steels, the peak temperature (and the width of the peak) may be sufficient to cause significant coarsening of the austenite grains closest to the fusion line. The subsequent cooling times from 800°C to 500°C (1470 – 930°F) may vary from 15 seconds or less at around 1 kJ/mm (25 kJ/in) to 30 seconds at 5 kJ/mm (125 kJ/in) and around 200 seconds at 40 kJ/m (1000 kJ/in). The corresponding prior austenite grain sizes may vary from 50 µm to over 250 µm, while the intra-granular microstructure may vary from auto-tempered martensite to ferrite-pearlite.

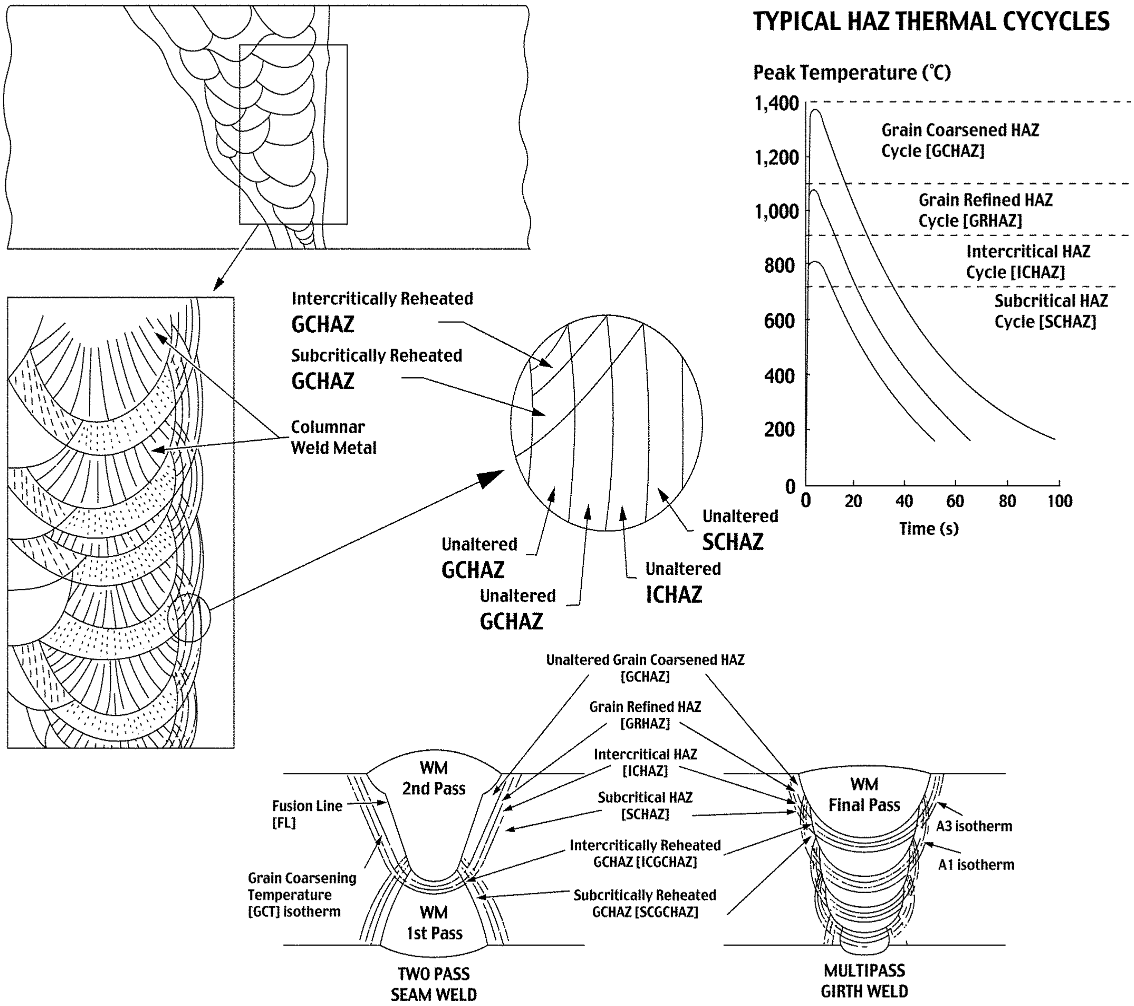

Figure 2: Heat affected zone microstructure in two-pass and multi-pass welds.

In a multi-pass weld, the HAZ from the second bead overlaps the HAZ from the first bead, giving rise to a variety of partially and fully reheated zones (Figure 2). The zones of most importance are the sub-critically or intercritically reheated coarse-grained HAZ regions, in which the previous microstructure is either tempered or partially retransformed and re-cooled.

The Coarse-grained Heat Affected Zone

Many studies (e.g. 4,5,7-9,11,14,15) have examined the interactive effects of the weld thermal cycle and the steel composition on the microstructure of the coarse-grained heat affected zone. In view of the complex and interacting effects of different alloy elements, it is hardly surprising that the reported effects are confusing and often conflicting. The key microstructural parameters so far as subsequent properties are concerned are as follows (see also Figure 2).

Austenite Grain Size

Austenite grain growth occurs rapidly at temperatures above 1000°C (1830°F), unless it is constrained by fine particles such as micro-alloy carbides or nitrides. Niobium has traditionally been used for this purpose (Figure 3); however the amount must be carefully controlled to ensure the particle dispersion is neither too coarse nor too fine. In recent years titanium, which

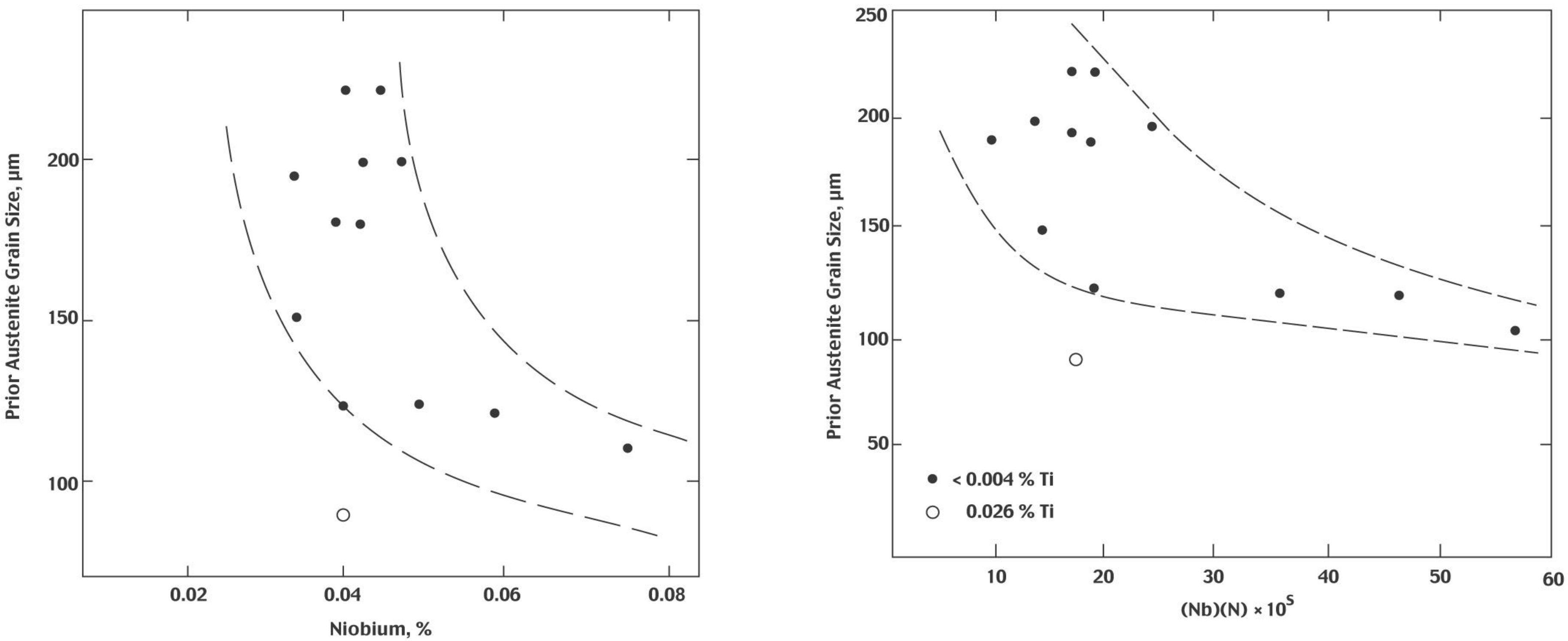

Figure 3: Effect of titanium, niobium and nitrogen on prior austenite grain size [from Kirkwood (1)].

forms a stable nitride even at the highest temperatures, has been used grain growth in a wide range of pipeline, structural (Figure 4) and shipbuilding steels.

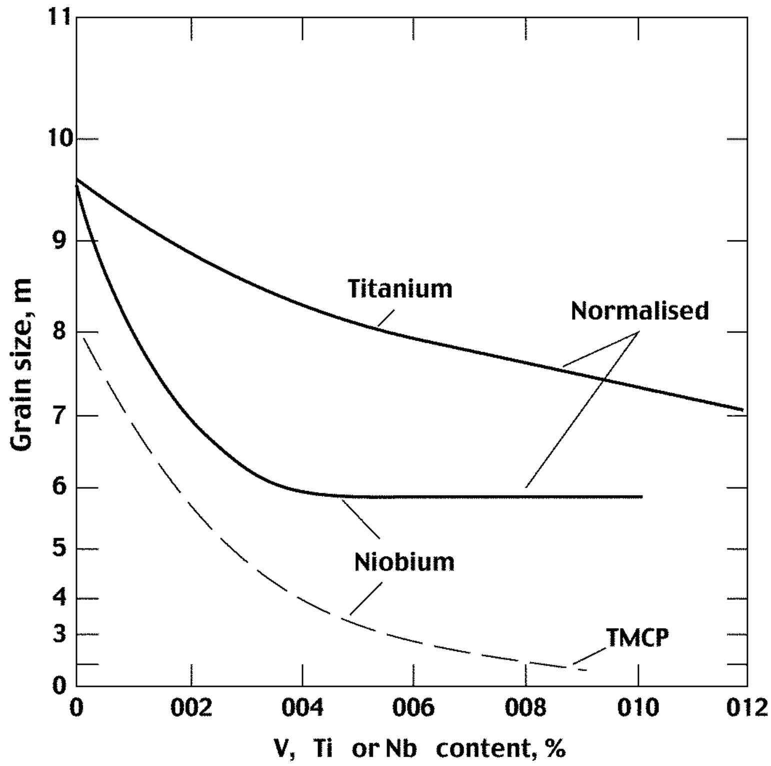

Figure 4: Influence of niobium and titanium on ferrite grain size [from Heisterkamp et al (12)], TMCP = Thermo-Mechanical Controlled Processing.

<u>Intragranular Microstructure</u>

The microstructure that forms on cooling depends on the hardenability of the material and therefore on the temperature range over which the transformation from austenite to ferrite and carbides takes place. This is influenced by the presence of alloying elements such as manganese, chromium, molybdenum, vanadium, copper and nickel. It is also influenced by the extent to which grain boundary nucleation of ferrite may be suppressed by elements such as boron, and the extent to which intra-granular ferrite nucleation may be promoted by the presence of fine particles such as titanium oxide. Finally, and most important, the microstructure is dependent on the carbon content, which determines the balance of ferrite and carbide phases in the final microstructure.

The relationships between transformation temperature and microstructure are illustrated in Figures 5 and 6. At the highest transformation temperatures the ferrite tends to form as equiaxed grains at the prior austenite boundaries and as randomly orientated grains in the interior. Carbon-rich areas transform as relatively coarse carbide aggregates. As the transformation temperature reduces, grain boundary ferrite is replaced by interior colonies of laths in parallel orientation or Widmanstätten ferrite; discrete carbide particles are formed at the lath boundaries at the higher transformation temperatures and are progressively replaced by islands of high-carbon martensite-austenite as the transformation temperature falls. At the lowest transformation temperatures, there is no longer sufficient time for long-range partition of carbon, and transformation to bainite or martensite occurs; if martensite forms, some auto-tempering may occur during subsequent cooling.

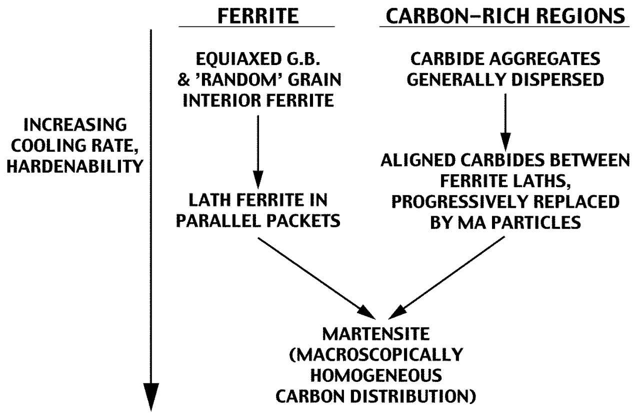

Figure 5: Effect of cooling rate and hardenability on HAZ microstructure [from Rothwell and Dorling (5)].

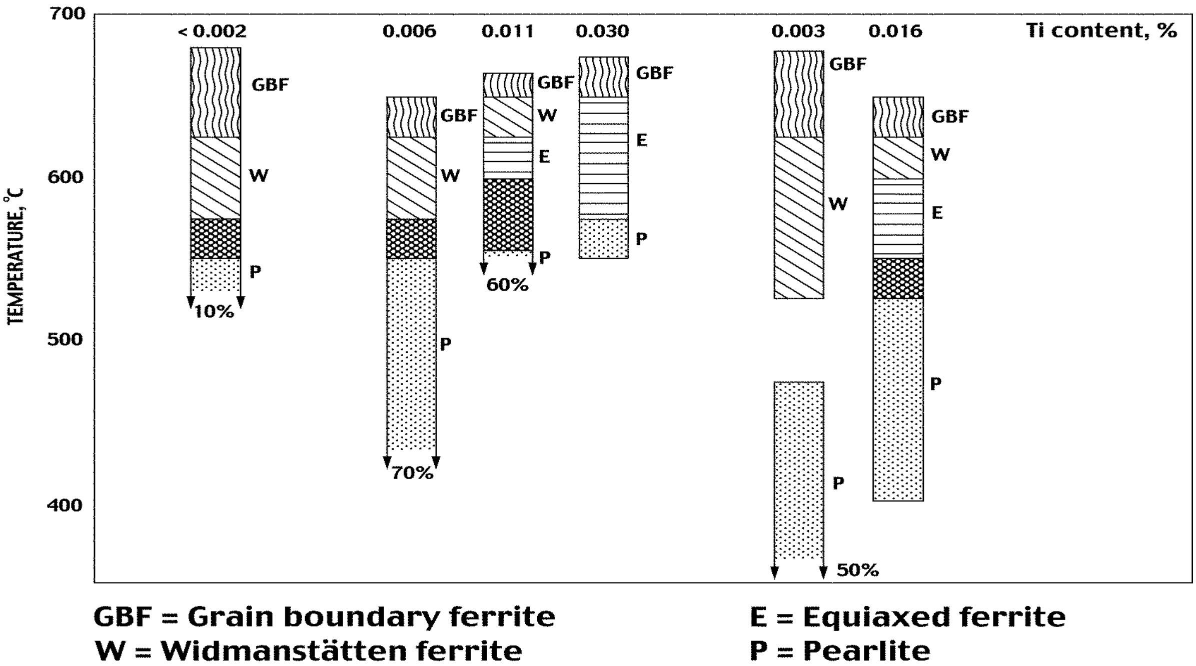

Figure 6: Effect of transformation temperature on HAZ microstructure of ESW welds in ship steels [from Cuddy et al (13)].

Microalloying elements can influence the coarse grained HAZ microstructure in several contrasting ways:

- Elements which form stable fine precipitate dispersions at high temperatures (e.g. Ti) can restrict grain growth, promote intragranular ferrite nucleation and raise the transformation temperature range;
- Elements which remain in solid solution (e.g. Mn, Ni) can reduce the transformation temperature range and contribute to solid solution hardening of the transformed product;
- Elements which are largely dissolved at the peak of the weld thermal cycle and which form fine carbides or nitrides on cooling can promote ferrite formation (e.g. V), raise the overall transformation temperature range and reduce intra-lath carbide formation; alternatively they may delay ferrite formation (e.g. Mo, Cr), lower the transformation temperature range and encourage martensite-austenite formation.

The unique role of niobium in determining HAZ microstructures is related to the solubility characteristics of its carbonitride; more soluble than titanium but less soluble than vanadium and molybdenum. This leads to the possibility that niobium, alone among the microalloying elements, can act in several of the above ways depending on the amount present, the carbon content, the 'competitive' interaction with other microalloying elements, the available nitrogen content, the previous manufacturing history and the weld thermal cycle.

Niobium carbonitride dissolves in austenite only at temperatures above around 1000ºC (1830ºF). For rapid weld thermal cycles, insufficient time is spent above this temperature for significant dissolution and the restricting influence of niobium on austenite grain growth is consequently linked to the preceding dispersion. At the slowest weld thermal cycles dissolution occurs and can be followed by reprecipitation in either austenite or ferrite. Precipitation in austenite can restrict grain growth and may promote early nucleation of ferrite at the boundaries. Precipitation may also occur as ultra-fine particles during transformation to bainite, reducing the width of bainite laths and encouraging carbide aggregate formation at the interstices. However at intermediate cooling rates the precipitation of ultrafine particles may be suppressed, promoting martensite-austenite islands in the interstices.

It is clear that when the effects of alloy composition and weld thermal cycle are superimposed, there are many different competing and interacting influences. As a result the development of a general quantitative description of the ways in which all these microstructural effects combine to determine HAZ properties is a problem of great complexity. Nevertheless, the key influencing factors described above provide useful general guidelines for predicting the probable outcome of specific changes in compositional or processing variables.

The Intercritically Reheated Coarse-grained Heat Affected Zone

In multi-pass welds, the superimposed HAZ from the second bead gives rise to a range of partially and fully retransformed zones within the original coarse-grained HAZ. The zone of most importance with regard to HAZ properties is the intercritically reheated coarse-grained HAZ, in which the previous microstructure is partially transformed to austenite (8,14). The partial transformation results in local areas of carbon-rich austenite, which are further enriched by dissolution of additional carbon from the surrounding area and subsequently retransformed to high-carbon twinned martensite on cooling. These brittle 'islands', up to 5 µm in size and up to 5% of the microstructure, can have a significant influence on toughness properties, as will be seen later.

Detailed studies of martensite islands in the inter-critically reheated zones have indicated that, as expected, their formation and distribution are influenced by chemical composition, heat input and the relative position of individual beads in the multi-pass weld. So far as compositional effects are concerned, increased alloy element content (for a given carbon content) increases the volume fraction of martensite islands and decreases the likelihood that the martensite will be decomposed by subsequent weld thermal cycles.

The Weld Metal

It is not the purpose of this paper to discuss in detail the microstructure and properties of the weld metal. Nevertheless it is important to note that the relationships between weld metal microstructure, composition and welding conditions are even more complex than in the HAZ.

This is because, while all the factors discussed above come into play, the chemical composition of the weld metal and its macrodistribution in the solidified weld pool are functions of the parent and consumable compositions, the flux activity and the welding process variables.

Many of the basic microstructural principles that apply to the HAZ are also valid with respect to the weld metal. However an additional important microstructure is acicular ferrite, consisting of fine interlocking grains that have a basket-weave appearance. Its formation depends not only on a suitable combination of chemical composition and cooling rate, but also on the existence of an appropriate distribution of fine non-metallic inclusions. The inclusion distribution is very dependent upon the flux or shielding gas composition. The development of acicular ferrite microstructures in weld metals in the 1970's and their associated good toughness led to the subsequent development of titanium-containing oxide-dispersion compositions for structural steels, with similarly improved HAZ properties.

Weldability - Resistance to Hydrogen-assisted Cold Cracking

Historical Development of Approaches to the Control of Cold Cracking

The history of approaches to cold cracking in structural and pipeline steels, and some of the associated testing techniques, have been reviewed and summarized by Yurioka (16) and by Glover and Rothwell (17,18). To a large extent, these developments mirror the advances in materials that have taken place over the same time-scale. In the fifties and early sixties, the steels that were used for line pipe and for other structural applications were rather simple, carbon-manganese formulations; if micro-alloying elements were used at all, it was solely to assist in grain-size control during the heat treatment of materials for which the specified toughness could not be achieved in the as-rolled condition.

It was already realized that hydrogen cracking occurred within a limited temperature range and depended on a critical conjunction of hydrogen content, tensile stress and susceptible microstructure. However, the transformation characteristics of these steels, and the susceptibility to cracking of the transformed microstructures, were such that an approach to crack prevention based primarily on HAZ hardness could be applied successfully. For the most part, hardened microstructures were so susceptible to cracking that their formation could not be tolerated, unless extremely low hydrogen contents and stress levels could be guaranteed. On the other hand, the variation of hardness (an approximate measure for cracking susceptibility, in these steels) with cooling rate through the transformation range was steep. The control of cooling rate, through restrictions on heat input and thermal severity (essentially, combined thickness of the weldment available to conduct heat away from the weld), could thus be used to avoid cracking. Critical hardness levels could be established, depending mainly on the hydrogen potential of the welding electrodes, and indexed to welding conditions that would avoid cracking. The traditional (“IIW”) carbon equivalent was an effective way of encapsulating the influence of chemical composition on hardenability. Methods such as the Controlled Thermal Severity (CTS) test, and under-bead cracking tests, for simple joint configurations such as pipeline girth welds were effective ways of determining weldability.

The practical application of controlled-rolled HSLA steels in the late sixties and seventies, coupled with developments in steelmaking and casting technology, made it possible to achieve equivalent, and even higher, strengths with lower levels of carbon and alloying elements. For these materials, if the carbon content was low enough, even the fully-hardened microstructure could have adequate resistance to cracking -see, for example (19). On the other hand, if considered in traditional terms, the critical hardness concepts that had been established for older steels were not appropriate. It was realized that hardness levels that would have been considered “safe”, for older-generation steels, were not necessarily so, under equivalent conditions, for these newer materials. Greater attention was paid to the role of hydrogen (generally expressed as the average hydrogen content of the weld metal, determined in conventional tests for hydrogen potential), as well as to the available time for diffusion of the hydrogen away from the weld zone prior to cooling into the cracking range (typically, less than 100°C). Approaches to the control of cracking in this period can be referred to as “average hydrogen” models, of which a number exist. Typically, they use a carbon equivalent such as P_{cm} that assigns greater importance to carbon itself, reflecting the importance of the hardness of the fully-hardened microstructure, rather than hardenability only.

A broader approach, that uses a carbon equivalent in which the relative importance of carbon and alloying elements varies with carbon content, was proposed by Yurioka in the early eighties (20), and has since been widely applied in the pipeline industry. In the application of these models, the cooling time was measured down to temperatures close to the cold cracking range (often 100°C), which is a better reflection of the potential for hydrogen diffusion, and is much more influenced by preheat than are cooling rates in the transformation range. The applied stress also began to be taken into account quantitatively, often determined in terms of reaction stress (that arising from thermal contraction against the restraint of the joint). Many specific equations for crack avoidance (on the basis of critical cooling time or preheat, or critical stress) were developed that involved these variables, or equivalents for them (see, for example, the approaches referenced by Yurioka in (16). Typically, these involved linear combinations of terms expressing the effects of stress, hydrogen content, cooling rate and chemical composition. Testing for weldability involved directly-applied stress (as in the Implant test) or externally or self-restrained specimens, for all of which the applied stress and cooling regime can be quantified.

As the HAZ cracking susceptibility of steels continued to fall, and strength levels rose, leading to wider use of higher-strength consumables (typically E8010 and E9010, in the pipeline industry), it was found that the cold cracking problem could be displaced from the HAZ to the weld metal. It thus became clear that an integrated approach to the management of hydrogen in welds needed to be adopted (21). The controlling variables for cracking will be the same in the weld metal as in the HAZ. However, the problem becomes more complex, since it is necessary to consider the evolution of hydrogen content and of stress in both weld metal and HAZ, as well as the effects of dilution and the differing inclusion populations in the two regions. The effects of ambient conditions on cooling in the low temperature range, in which cracking susceptibility changes and cracking occurs, have also needed to be considered, particularly for colder climates such as that of Canada.

While these approaches, which can be referred to as “distributed hydrogen” models, have not been fully developed or codified, substantial increases in understanding have been achieved that have delivered practical benefits (Ref 22). A complete analysis requires modeling of the coupled processes of weld metal solidification, transformation (both HAZ and weld metal), hydrogen diffusion and the development of stress, as well as accurate estimates of the critical conditions for cracking in both regions. Even today, the realistic achievement of such modeling is at the limit of computational capabilities. Practical success in the short term is thus still likely to be based on the empirical determination of the probable site and source of cracking and the application of test methods that realistically reproduce the specific field conditions. Examples of such approaches, drawn from the pipeline industry, will be briefly discussed in the following sections.

Root Pass Cracking Due to Lifting or Restraint

Initial approaches to the control of cold cracking were developed primarily for structural and ship-building applications, and it is natural that they would have concentrated on thermal contraction against restraint as the dominant source of stress. In fact, extremely comprehensive engineering approaches to the calculation of restraint stress were developed, notably by Japanese researchers in the Seventies (23), and applied in cracking models.

In the late seventies, however, pipeline research in Canada (24) returned to a more detailed analysis of the ideas put forward earlier (25) concerning the relative significance of directly applied lifting stresses, resulting from the manipulation of the pipe string. The principal concern, at that time, was for the cracking behavior of large diameter, relatively heavy-wall pipe, for which each root pass can undergo two significant stress cycles before the second (hot) pass is completed. Higdon and Weickert (24) carried out detailed analysis of the gross bending stresses arising from lifting, as well as those arising from thermal contraction and from restrained recovery of ovality after release of the line-up clamp. They also evaluated the effect of root pass geometry on the concentration of these stresses. For the "stiff" pipeline configurations of primary interest in North America at that time, they found that the overall stress level was dominated by lifting, and that the highest levels of stress (typically well beyond nominal yield) occurred in the 6 o'clock position, on the outside surface of the root bead. Full-scale testing and field experience supported this conclusion (26), which has obvious implications for the time-scale and magnitude of the stress cycles that need to be considered in evaluating weldability (18).

Later research, carried out in Australia (27,28), concluded that, for moderate lift heights and the smaller diameters and thinner walls typical of the Australian industry, lifting stresses would be much less significant than thermal contraction stresses, and that only extreme lifting would produce a second cycle of stress. This conclusion has different implications for the timing and magnitude of the stress cycles, and also for testing geometry, since the longitudinal bending associated with circumferential shrinkage concentrates tensile stresses at the inside, rather than at the outside surface.

The situation considered so far involves the potential for cracking within minutes of the root pass deposition, and it can be controlled by attention to the root-bead completion, the interval between root and hot pass, and preheat where required. In certain circumstances, however, true delayed cracking can occur, particularly in conditions of wind and low ambient temperature. Glover and Graville have presented results of research on this topic, and have evaluated the effects of procedural and atmospheric variables on the risk of cracking (22). In such cases, the local stress to be considered is the final residual stress in the completed weld, while the local hydrogen content must be tracked throughout the deposition of the successive weld passes and subsequent cooling. The appropriate testing protocol for such a situation will be quite different from that for "short-term" cracking.

Heat-affected Zone or Weld Metal Cracking

It has already been mentioned that, as the weldability of structural and pipeline steels improved and the strength level rose, the site of cracking could be displaced from the HAZ to the weld metal (WM). The issues involved are rather more complex than have been suggested, however. Matsuda et al (29) proposed and validated a framework for determining the site of cracking for high strength structural steels, based on the effect of hydrogen on the local stress for cracking as a function of time after weld deposition. Weld metal hydrogen content decreases monotonically with time after welding, while that in the HAZ rises from the low value typical of the base material to a peak, before declining. The time-scale of this process is minutes, the same order as that for stress application and the time between root and hot pass. By considering the magnitude and timing of the applied stress, together with the relative level of the critical cracking stress in the HAZ and WM, as influenced by instantaneous local hydrogen content, the site of cracking can be predicted. Since higher strength (harder) weld metal will be more susceptible to cracking, it will certainly promote WM cracking. However, other factors promoting WM cracking are seen to be high levels, and early application, of stress. Low strength weld metals and low levels of stress, or late application of stress, will promote HAZ cracking.

From the point of view of pipeline field welding, this framework can be applied to understand some of the changes that have taken place in the phenomenology of cracking, and also to examine some of the approaches to the control of cracking that have been developed, over the years. The application of stress to the root pass of a girth weld is always rapid, whether it is dominated by lifting or by restraint stresses, whose maximum levels develop as the heat is lost from the weld region after closure of the individual segments. The maximum level of stress is, of course, variable. If it arises mainly from restraint, as in thin-walled, small diameter pipe, it will be relatively low. If the pipe is stiff enough, on the other hand, very high stresses may be seen at the previous root-bead (rather than the root bead just completed) as a result of lifting. The difference in timing between the first and second lifting cycle does not appear to be a determinant of the cracking site, for the materials and welding consumables that are involved. In this sense, the timing of the stress, for crack initiation prior to hot-pass deposition, is always "early".

As regards the relative susceptibilities of the HAZ and WM, the historical trends can be easily understood. In the fifties and sixties, relatively high-carbon (though low strength) line pipe was typically welded with low-strength E6010 electrodes, and this combination resulted almost exclusively in HAZ cracking. In the seventies, as typical pipe grades for major pipelines increased to X70, most operators specified E8010 or even E9010 electrodes, in the interests of matching (or near-matching) strength. The higher strength of these electrodes was achieved through higher alloy element content, increasing their susceptibility. Since typical X70 compositions were of much lower carbon content and generally lower carbon equivalent than the older steels, their crack susceptibility was much lower, and accordingly the typical crack location moved into the weld metal. For lifting-dominated conditions, in fact, cracks were found to initiate precisely in the region of maximum stress concentration, on the outside surface adjacent to the fusion boundary (26).

Overall Implications for Weldability Testing

Over the last two decades, various approaches have been proposed for predicting procedural requirements for the avoidance of cracking, based on laboratory testing. Three of these approaches will be reviewed, to illustrate the differences in methodology, and, as importantly, in thought processes, which is required to address the range of behaviors that can be expected.

In the early eighties, Lorenz and Düren (Ref 30) proposed a method for the prediction of the necessary preheat for pipeline welding. They based their approach on a simple relationship between calculated HAZ hardness and critical preheat, for specific levels of average hydrogen content. The HAZ hardness was predicted on the basis of the cooling time from 800 to 500ºC and of a carbon equivalent whose coefficients themselves varied with cooling time, to account for the different relationships between hardness and chemical composition for martensitic, bainitic and mixed microstructures.

On this basis, they proposed criteria for field welding tied to the observed minimum preheat for the avoidance of cracking in the Implant test, at an applied stress equal to the nominal yield strength of the base metal, and with a heat input in the range 0.8-0.9 kJ/mm. Figure 7 shows an example of their results, in this case for cellulosic electrodes. The relationship shown is that between critical preheat temperature and the carbon equivalent for a cooling time of 4 s, an appropriate value for the welding conditions considered.

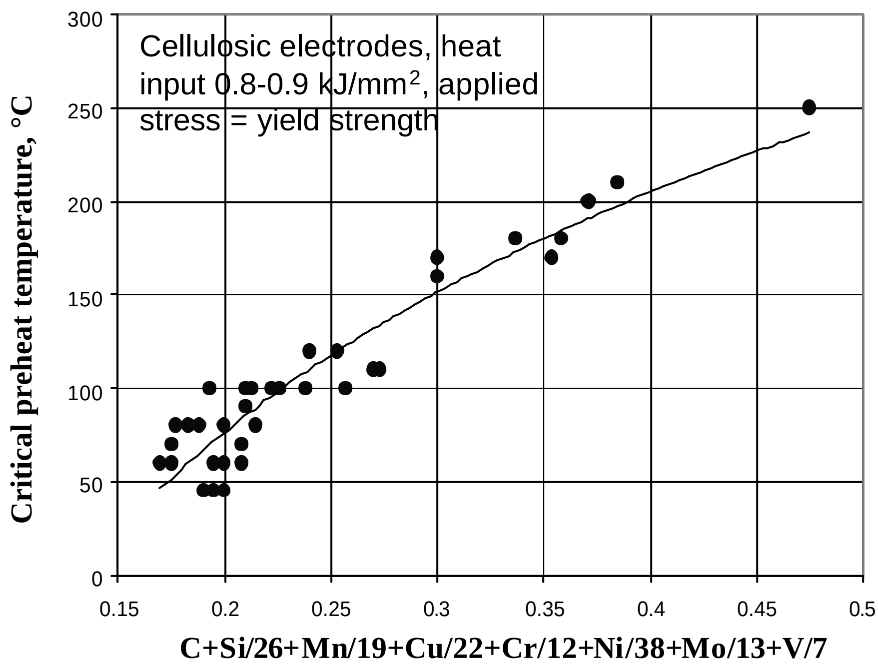

Figure 7: Relationship between critical preheat in the Implant test and carbon equivalent for $t_{800\text{-}500}$ = 4 s [after Lorenz and Düren (30)].

The correlation is clearly good over a significant range of chemical composition, and the critical preheat temperature range indicated is a reasonable one, for typical high strength pipeline steels in thicknesses up to 20 mm. Such materials would have values of the carbon equivalent shown up to about 0.25. However, the conceptual difficulty that must be faced is that in many instances, for such materials, cracking would in practice occur in the weld metal. Since the geometry of the Implant test, by placing a sharp notch in the HAZ, is designed to evaluate HAZ cracking susceptibility, its results provide little guidance relative to WM cracking.

A brief consideration of two other approaches to crack prevention illustrates the difficulty more clearly. Yurioka et al (20) proposed a form of carbon equivalent, usually called CEN, that assigned a variable weight to alloying elements as a function of carbon content; the aim was to accommodate the different hardening behaviour of both high and low carbon steels, and it is now quite widely used in the pipeline industry. They developed a cracking index that was a linear function of terms expressing the effect of carbon equivalent, local stress level and initial weld metal hydrogen content, and used it to predict critical cooling time to 100ºC. Using the Stout (Lehigh) slot test, they also demonstrated a linear relationship between critical preheat temperature and CEN for pipeline and other HSLA steels welded with E7010 electrodes

(Figure 8). Again, however, the geometry of the test and the relatively low-strength electrode promoted HAZ cracking.

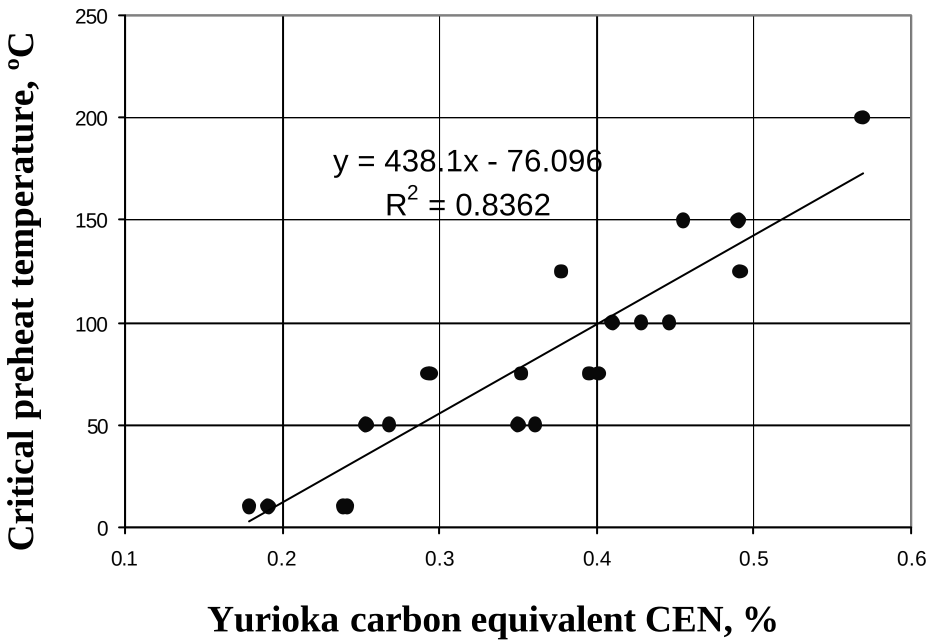

Figure 8. Critical preheat temperature in the Stout slot test as a function of carbon equivalent CEN [after Yurioka et al (20)].

At about the same time, the Welding Institute of Canada (WIC) test was developed as a simple and economical form of restraint cracking test that could accurately reproduce the root pass geometry in pipeline welding. Because of its short restraint length, it develops a high level of reaction stress in a short period of time. It has been widely used in the pipeline industry, and, for large-diameter, high-strength pipe, the results relate well to full-scale testing and field behavior (18). In particular, it reproduces accurately the crack initiation site associated with lifting-dominated, WM cracking, and as a result is particularly applicable to large, "stiff" pipelines welded with high-strength electrodes.

The results of WIC tests on a series of steels of constant thickness (12 mm) covering the full compositional range typically used in the pipeline industry, carried out at NOVA in the 1980's, can be compared with the trend-line of Yurioka et al. (Figure 9).

On the assumption that the factors influencing cracking are independent of each other (as implicit in the linear cracking index proposed by Yurioka et al.), the slopes of the variation of critical preheat temperature with CEN can be compared directly. The quality of the correlation is equivalent for both sets of data, but the slope of the WIC test results is much lower than that of the Yurioka results. A primary reason for this, of course, is that in the WIC test, weld metal cracking occurred. The influence of the elements considered in the base metal CEN can thus only be expressed through dilution into the weld metal (typically around 50% for a pipeline root pass preparation and heat input, though recoveries of specific elements vary). A plot of critical cracking stress in the WIC test against (calculated) weld metal CEN can be constructed (Figure 10) and again, a good relationship is obtained; the slope is steeper, but still only half that of the Yurioka results for the HAZ.

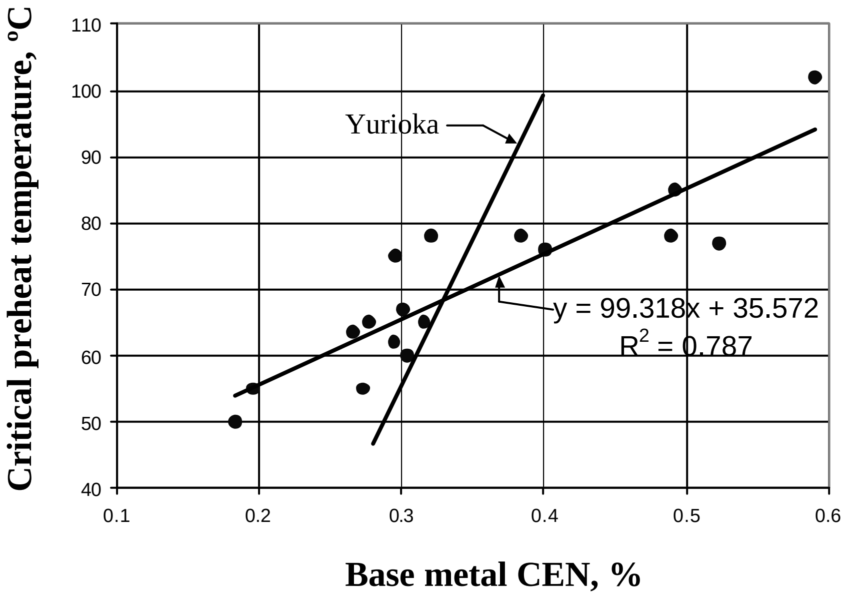

Figure 9: Variation of critical preheat temperature in the WIC test with base metal CEN.

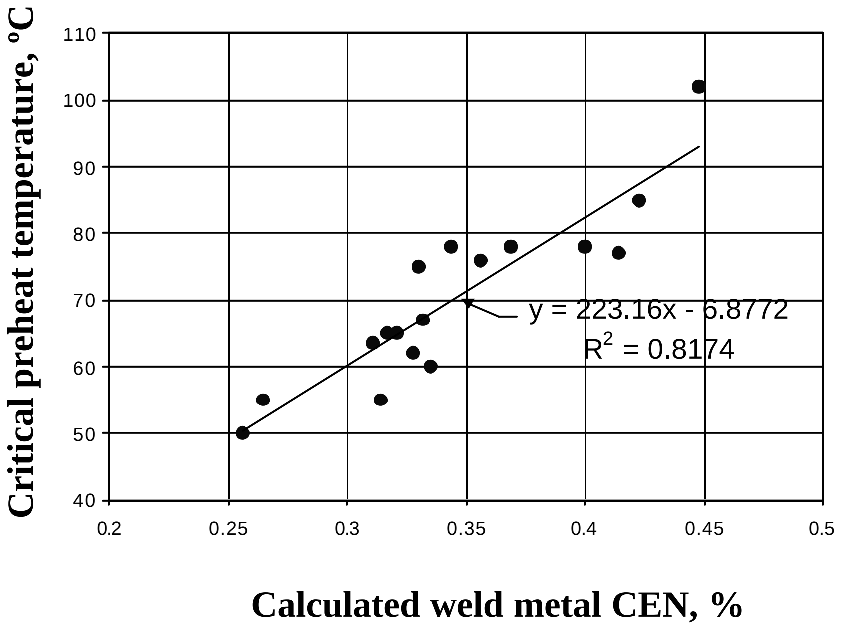

Figure 10: Variation of critical preheat temperature in the WIC test with calculated weld metal CEN.

This is, perhaps, easily understandable, since the relationship between hardenability and chemical composition will be different for WM and HAZ, as a result of the very different inclusion populations and austenite grain size and shape. In fact, while HAZ hardness in these WIC tests was well predicted by Lorenz and Düren's relationship, for the appropriate cooling rate, WM hardness was much lower than predicted. WM cracking also occurred at hardness levels for which it would not be anticipated in the HAZ (significantly less than 300 $HV_{0.5}$ in some instances). The radical differences in the relationships between cracking susceptibility and chemical composition, for HAZ and WM, means that it is difficult to apply a single predictive approach to both. In general terms, it is necessary to determine, experimentally or analytically, whether the dominant issue will be WM or HAZ cracking, and to select an approach and test method to match. For many existing cases, of course, this is already known on the basis of long experience, but as pipe materials and consumables continue to advance, empirical methods will be needed. Clearly, the development of appropriate specifications and viable field procedures for new combinations of materials depends on a full understanding of the issues and the availability of methods to address them.

HAZ Performance: Toughness

The coarse-grained HAZ

The overall toughness of the HAZ is dependent on the relative strengths and microstructures of the different regions in the weld zone and by their spatial relationship to one another. In a multi-pass weld the complex influences of HAZ microstructure described earlier are immediately apparent. The coarse grained HAZ is often dominant in controlling toughness.

Many studies of HAZ toughness have been conducted during the last 40 years (3-15) and it would be impossible to review them within this paper. Studies have included multi-pass welds, single beads and weld thermal simulations and have used test configurations ranging from tensile and Charpy specimens to CTOD and wide plate configurations. Several of these studies have addressed the relationship between microstructure and toughness, particularly for the coarse-grained HAZ, focusing on the austenite grain size and the intra-granular microstructure.

The significant influence of austenite grain size on toughness has been demonstrated on many occasions. Figure 11 illustrates the beneficial effects of reducing grain size for thermo-mechanically processed steels with a variety of base microstructures. In most instances the influence of grain size is probably indirect; the size of the 'covariant packet' or sub-grain that actually controls fracture initiation is related to the austenite grain size.

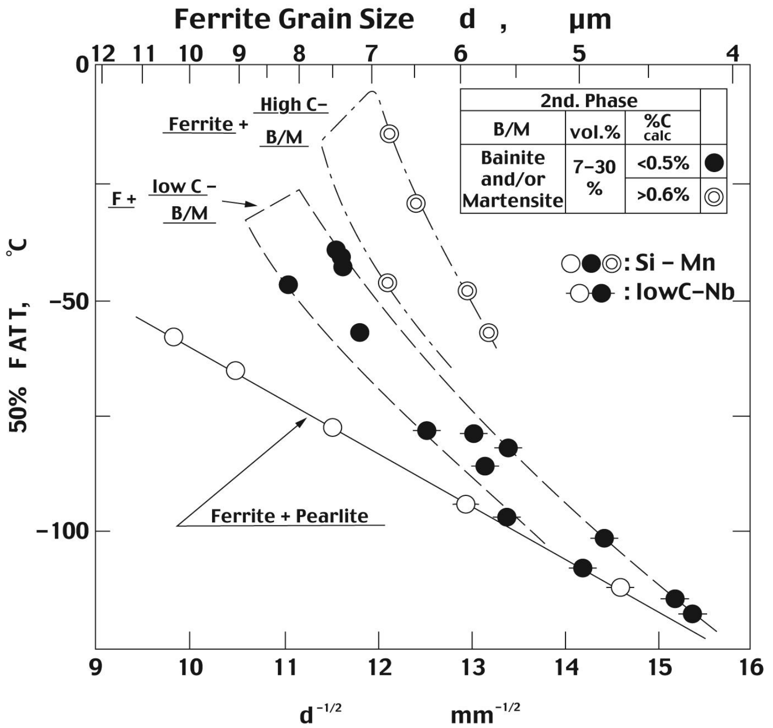

Figure 11: Effects of grain size and microstructure on toughness [from Shiga (8)].

The influence of intragranular microstructure on toughness is more difficult to discern, but can be illustrated by reference to the relationships defined earlier between cooling rate, transformation temperature, carbon content and microstructure (9,12). The influence of carbon content and transformation temperature on microstructure is illustrated in Figure 12a, and the corresponding influence on toughness in Figure 12b. The pair of diagrams should be considered as schematic rather than quantitative, although they have been extensively based on test results. Figure 12a delineates the zones within which different HAZ microstructural types occur. Figure 12b shows sections through Figure 12a at three carbon contents to illustrate more clearly the relationship between transformation temperature and toughness:

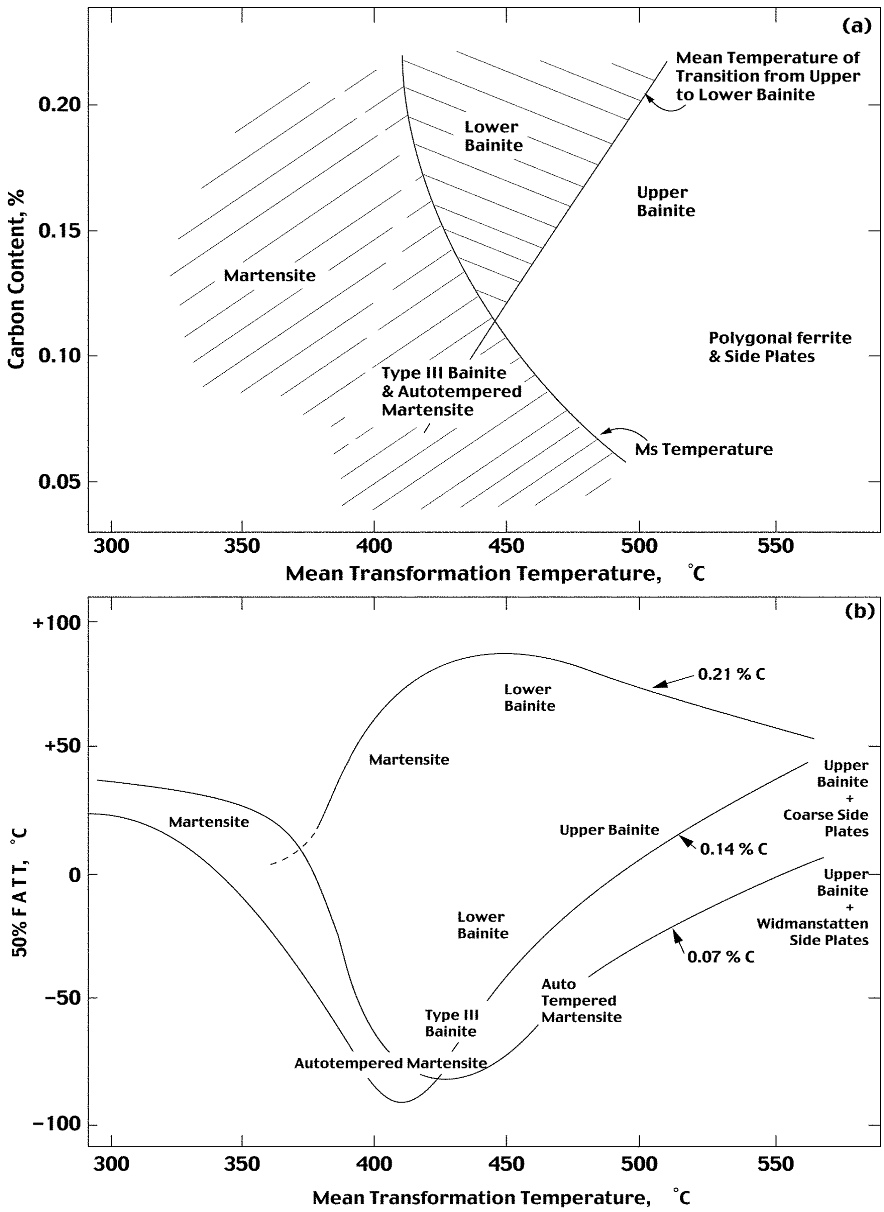

Figure 12: Influence of carbon content and microstructure on HAZ microstructure and toughness of structural steels [from Batte and Kirkwood (9)].

- At high carbon levels (0.21%) good toughness can rarely be achieved, since even lower bainite has poor toughness and martensite formed at low temperatures is twinned and inherently brittle in the untempered state; the Ms temperature is already so low that only limited auto-tempering can occur;

- At low carbon levels (0.07%) the poor microstructure occurring above 530°C is gradually replaced by lower bainite and auto-tempered martensite in the range 460-400°C. This results in the achievement of good toughness over a wide range of transformation temperatures, and hence over a wide range of weld heat inputs;
- For intermediate carbon levels (0.14%) a similar pattern applies with the exception that the range where good toughness can be achieved is narrowed.

The relationships described in Figure 12 were derived for basic C-Mn-Nb steels that were free from significant alloy additions. Nevertheless the basic principles have been shown to apply to a wide range of modern steels with microalloying additions, and are also evident in a wide range of HSLA pipeline (Figure 13), structural and shipbuilding steels. For these steels there is a general 'shift' of the toughness values depending on the extent to which the microalloying additions contribute to solid solution strengthening, grain size refinement or carbonitride precipitation.

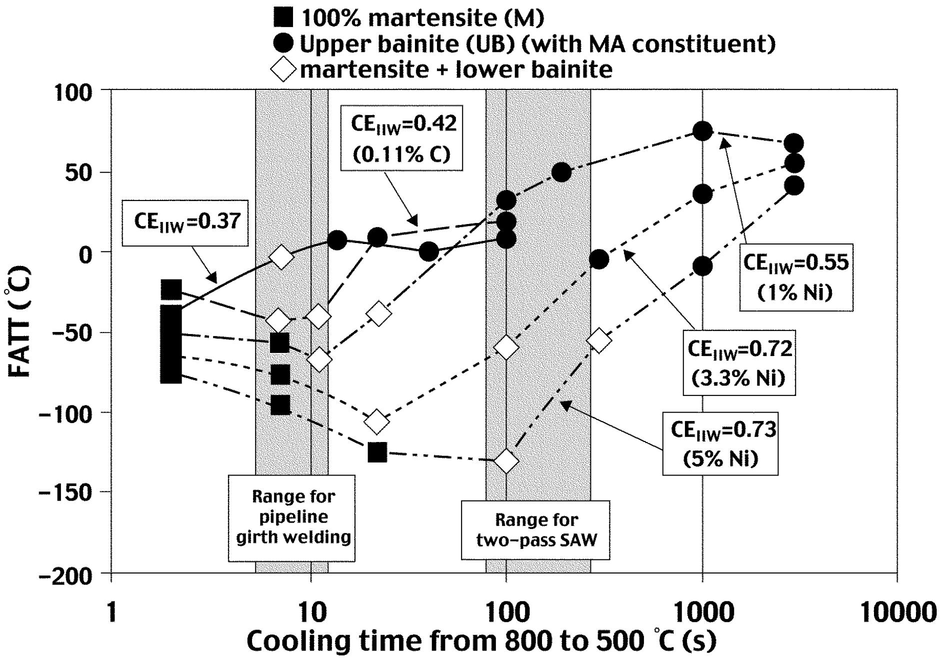

Figure 13: Correlation between FATT and cooling time for coarse-grained HAZ microstructures of pipeline steels [from Gräf and Niederhoff (3)].

The Intercritically Reheated Zone

As refinement of the chemical composition and the thermomechanical processes have led to increased toughness of the coarse grained HAZ, so the issue of the intercritically reheated coarse-grained region has attracted increased attention. As was indicated earlier the intercritically reheated region is known to be of possible poor toughness, particularly if islands of martensite-austenite occur.

Detailed studies of the occurrence and structural significance of martensite-austenite islands have demonstrated that in some steels the intercritically reheated region can have a toughness at least as low as the coarse-grained region (Figure 14), and that the reduction in toughness is dependent on the volume fraction of martensite islands present (Figure 15). While the adverse effect of martensite islands can be ameliorated to some extent in multi-pass welds by the influence of the subsequent weld passes, the only effective remedy is to select a chemical composition, in particular a balance of carbon contents and microalloying additions, that minimizes the extent to which martensite islands can develop (Ref 8). Recent studies have resulted in empirical formulae linking the volume fraction of martensite islands to alloy element concentrations.

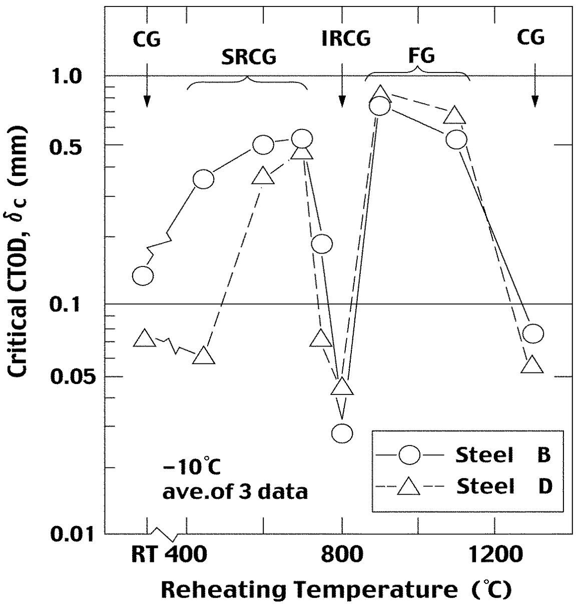

Figure 14: Toughness variations in simulated multipass weld HAZ microstructures [from Aihawa (15)].

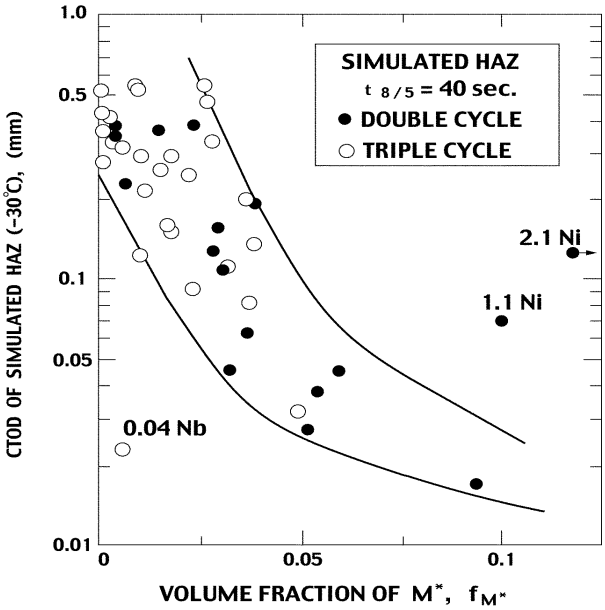

Figure 15: Dependence of intercritically reheated HAZ toughness on volume fraction of martensite islands [From Aihawa (15)].

The Influence of Local Brittle Zones

The issue of local areas containing martensite-austenite islands raises a more general concern regarding the toughness and structural integrity of multi-pass welds in which there are local zones of brittle material within the heat affected zones. The observations of low individual toughness test results are not new; however, they have become much more evident in CTOD tests, which can specifically sample the coarse-grained HAZ (Figure 16), than in Charpy tests have been found in tests where the pre-crack tip samples less than 1.0 mm depth of local brittle zone (LBZ) (Figure 17). Indeed, the issue of occasional low CTOD values has led to a standardized procedure for testing in the presence of local brittle zones, incorporated in API Standard RP 2Z (32).

The problem for the steel user is how to handle the occasional low CTOD values, some of which have been less than 0.05 mm (0.002 in). In an attempt to resolve this issue, attention has turned to the wide plate test. Several studies have demonstrated that low CTOD values do not correspond to low wide-plate test results. In one instance, for example, when CTOD and wide plate tests were tested in parallel (33), low CTOD values occurred when the pre-crack sampled coarse-grained material with grain size above 80 µm (0.003 in), whilst low strain (<0.5%) fractures occurred in the wide plate tests when the pre-crack sampled the coarse-grained microstructure in lengths of 30 mm (1.2 in) or more; however the failure loads in the wide plate tests were still equivalent to more than 95% of the yield strength of the parent plate.

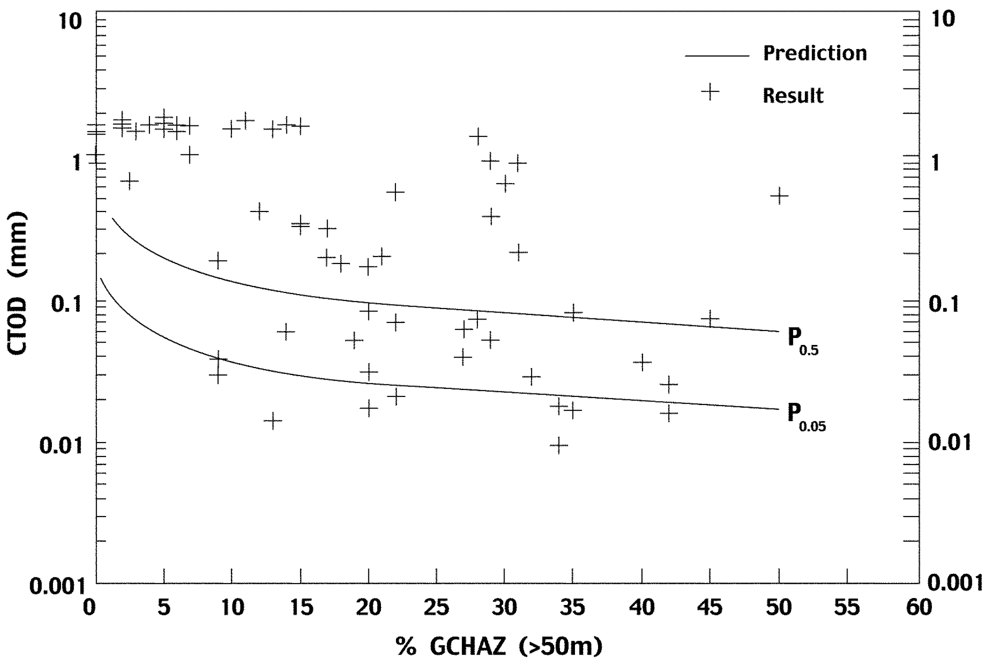

Figure 16: Relationship between measured CTOD and percent coarse grained HAZ sampled by the crack front [from Pisarski (31)].

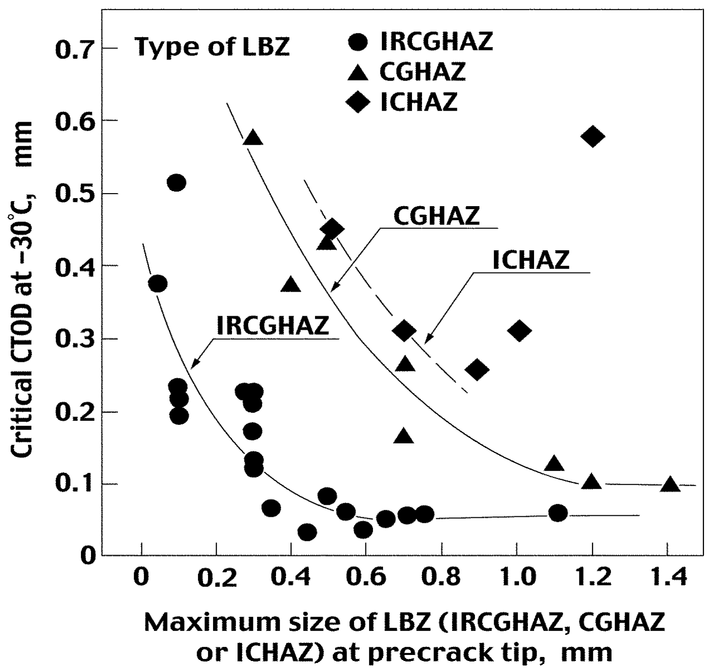

Figure 17: Effect of local brittle zone size at the precrack tip on measured CTOD [from Shiga (8)].

Clearly the size of the wide plate test, its ability to incorporate realistic levels of residual stress and realistic crack configurations, together with local strain redistribution during loading, make it a very attractive test that is generally regarded as being structurally more relevant than the CTOD test. However, there are difficulties in accurately locating the crack and for some weld configurations, notably in offshore structures where there are localized bending loads, care must be taken in interpreting the results. Nevertheless, for many applications the wide plate test adds confidence that occasional low CTOD test results can be accommodated.

For pipelines, low Charpy and CTOD test results have occasionally been obtained from the seam weld HAZ. In this situation the low toughness zone, usually the unrefined coarse-grained HAZ, can extend through a large proportion of the pipe thickness. To examine the structural significance of such results, attention has turned to pressurized burst tests of the full pipe section (34,35), supported in some instances by the use of constraint based fracture mechanics to provide an analytical understanding. These studies have demonstrated both experimentally and analytically that local low toughness within the localized zone adjacent to the (often overmatching) weld metal does not lead to premature fracture from simulated welding defects. The low constraint allows local strain and stress redistribution without unstable crack extension. Again, these results must be applied with caution because there are occasional situations where local stressing combined with particular defect configurations could still be at risk, but they nevertheless add confidence that occasional low seam weld toughness values can be accommodated.

The results of these studies are also supported by the general experience of weld zone performance in the structural and pipeline industries. Despite the circumstantial evidence that zones of low toughness within service welds are widespread, they have hardly ever been associated with failures unless significant defects are present.

New Developments

Higher Strength Steel Pipelines

The ongoing quest for improved combinations of strength, toughness and weldability continues to provide new challenges. During recent years one of these has been the emergence of new higher strength grades (X80 and X100) for pipeline steels (36-41). As with earlier developments, it has been necessary to establish that higher strength is achieved without sacrificing weldability or weld joint toughness.

Several of the leading pipe suppliers have developed X80 grade linepipe and have now delivered significant quantities on a 'regular production' basis. A smaller number have developed prototype and pre-production X100 grade pipes for evaluation and performance testing. It is evident from these developments that both X80 and X100 grade materials can be produced by further refinement of the thermomechanical processing and accelerated cooling manufacturing route, with low carbon contents and restricted addition of microalloying elements. The chemical compositions of pre-production materials show a slight increase in niobium contents compared to previous practice, but generally not above 0.05%. As a result, the toughness properties obtained for seam weld HAZs are not significantly different from those of the lower strength grades. The situation is slightly different for girth welds in that there has been a trend towards use of mechanized gas metal arc welding with lower heat input for mainline welds, while tie-in and other manual welds have used low hydrogen electrodes for fill and cap passes. Low heat input is beneficial for HAZ toughness, while intrinsically low hydrogen minimizes the risk of WM and HAZ cracking. The overall effect of these changes is that satisfactory girth weld performance can be achieved without additional palliative methods. Nevertheless, given all the previous experience regarding the pitfalls of comparatively small changes in compositional balance, it will be necessary to confirm experimentally, under the most appropriate testing conditions, that these higher strength pipeline steels do not pose any threat from the weldability and toughness viewpoint.

Improved Weld Productivity

A second area of development that has a significant potential effect on weldability and joint integrity is the challenge to improve weld productivity, whether in shipbuilding, offshore structure fabrication or pipe-laying. High heat input welding processes will require more rigorous control of the parent metal composition and HAZ microstructure to minimize the risk of hydrogen cracking and low toughness. Processes with reduced filler metal, such as narrow gap mechanized welds and autogenous or semi-autogenous laser welding, will not only require close control of all aspects of plate composition and the welding process; they will also require full understanding of the inter-related micro-mechanical behaviour of all the weld zones in order to set safe limits for defect sizes and toughness levels. For example, as plate and pipe wall thicknesses for high strength onshore pipelines are reduced, the structural significance of the individual weld zones, particularly those with low toughness or deformation tolerance, will become proportionally greater. Sophisticated analytical methods such as constraint based fracture mechanics may offer an analytical way of addressing these situations, but they are limited by the extent to which the deformation and fracture behaviour of each microstructural zone in the weld is understood. Large-scale testing, whether by wide-plate test or full-scale pipe burst test, will need to be increasingly precise about the location of the simulated defect within the weld zone.

Final Comments

The last four decades have seen considerable changes in the manufacturing and fabrication methods applied to HSLA steels for structural applications in the pipeline, offshore and shipbuilding industries.

The principle changes that have effected steel weldability and toughness during this time have been:

- the improved control of steelmaking and casting processes;
- the enhanced capacity of plate rolling equipment;
- the integration of thermal and mechanical treatment (including on-line accelerated cooling);
- improved weld process control and low hydrogen consumables.

These in turn have enabled beneficial improvements through:

- significant reductions in carbon content;
- more efficient use of grain size control;
- more efficient use of micro-alloying;

These have led to substantial microstructural refinement and hardness control in all regions of the HAZ. However, these improvements have only been possible through the close cooperation between physical metallurgists and welding process engineers, enabling a full understanding of the relationships between cooling rate, steel composition and transformation behavior within all parts of the various weld configurations used throughout the pipeline, structural and shipbuilding industries. Understanding of these relationships has been essential in delivering improved resistance to hydrogen cracking and improved toughness in the HAZ.

It is interesting to see how the apparently disparate results from research studies in all three industry sectors fit into the overall picture that has emerged, and how this in turn has enabled transfer of manufacturing technology between them. Thermomechanical processing for example has spread progressively, from the thinnest pipeline plates to the thicker offshore and shipbuilding plates. Similarly, the use of titanium to control HAZ grain size has spread to all sectors, giving benefits for the highest heat input electroslag welds as well as for the lowest heat input manual welds.

In many instances the understanding has been quantified in a series of empirically derived relationships between composition, welding process parameters and the resulting properties. These relationships have generally been very successful in enabling the safe and crack free construction of HSLA steel products. However the changes in manufacturing approach and plate composition have not always been straightforward in their effect on weldability and toughness.

The early understanding of weldability and toughness in the sixties and seventies was developed for C-Mn-Nb steels with carbon-dominated strengthening mechanisms. Consequently the original formulations for carbon equivalent have become increasingly uncertain, and sometimes misleading. Fortunately, reassessment of the underlying physical metallurgy accompanying the change in chemical balance has enabled a new understanding to be developed. Illustrative of this has been the issue of martensite-austenite islands, found originally in weld metal microstructures but appearing subsequently in the intercritically reheated zones of low-carbon microalloyed structured steels.

A wide range of test approaches has been developed over the years for HAZ hydrogen cracking. As experience has developed it is clear that some of the test techniques and the carbon equivalent formulations have been restricted in their applicability. In particular, the source and timing of the applied stress, the way in which bending stresses and constraint are accommodated, and the location of cracking within the WM or HAZ, all lead to concerns about the extent to which laboratory tests model the field welding conditions. Nevertheless, careful attention to the physical metallurgy and the micro-mechanics of the processes involved has provided a good overall understanding of the factors determining the occurrence and avoidance of cracking under field welding conditions.

The structural significance of local brittle zones in multi-pass welds, whether due to the unrefined or intercritically reheated coarse-grained HAZ, is still an ongoing concern to some sectors of the HSLA community. It is clear that local zones of low toughness can exist, and tests that sample these zones will generate low results. It is also clear that in many full-scale situations the micro and macro-mechanical behaviour of adjacent weld zones reduces their effect on the overall integrity of the structure. The application of techniques such as constraint based fracture mechanics has provided some of the answers; it may be that any remaining condition considered to be at-risk will prove to be highly improbable when practical weld joint configurations and defect locations are taken into account. Certainly the results for CTOD and wide plate tests indicate that significant proportions of the local brittle zone must be sampled by the crack before a structurally significant fracture problem occurs.

Conclusions

The introduction and widespread use of thermomechanical processing to achieve higher strength with leaner alloy compositions has had a significant impact on the weldability and toughness of niobium-bearing HSLA steels during the last 40 years.

Significant reductions in carbon content and more efficient use of micro-alloying elements have delivered major improvements in both weldability and HAZ toughness. However, the change in balance from carbon-dominated to microalloy-dominated strengthening mechanisms has necessitated a re-evaluation of the fundamental processes in order to gain a full understanding of the performance of modern steels.

The factors determining the size and the transformed microstructure of the coarse-grained HAZ are critical. Individual microalloying elements can act in several contrasting ways to control grain size, promote or delay transformation during cooling, depending on their interaction with other elements; such differences have a major effect on microstructure, hardness and toughness. Higher microalloying element contents can result in brittle martensite-austenite islands in the coarse-grained HAZ or the intercritically reheated zones in multipass welds.

Understanding of the effects of composition and weld process parameters on resistance to hydrogen cracking has been embodied in well-established empirical formulae. While resistance to HAZ hydrogen cracking has generally improved in line with reduced grain size and HAZ hardness, laboratory tests have not always accurately modeled field-welding behavior, particularly for modern HSLA steels. The situation has been complicated by the potential for change in the site of cracking from HAZ to weld metal. Nevertheless, for most situations the factors determining cracking resistance and the steps needed to avoid the problem through control of the welding process are well established.

The relationships between microstructure and coarse-grained HAZ toughness are also now well understood, but the multiplicity of interacting effects cannot be captured in simple quantitative relationships based on steel composition. The overlap of thermal zones in multi-pass welds can often result in local zones of brittle material in the HAZ. While tests that sample local areas can occasionally lead to low results, larger scale test and analyses indicate that in many situations such zones are not structurally significant.

As a result of the specific thermodynamic and kinetic attributes of its carbonitride precipitation in steel, niobium can justifiably be considered a unique enabler of modern HSLA steels with improved combinations of strength, toughness and weldability. Niobium interacts in a variety of ways with other microalloying and residual elements to influence the grain size and intragranular microstructure of the different regions of the multipass weld HAZ. Carefully controlled use of niobium is fundamental to optimization of HAZ properties when HSLA steels are used for pipelines and structural applications.

Acknowledgements

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