IMPROVED WELDABILITY AND CRITERION FOR RELIABILITY OF HIGH STRENGTH PIPES STEELS

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Abstract

Welding criteria assessment has been investigated based on techniques for welding thermal cycle simulation and analysis of kinetics of austenite transformation within the heat affected zone. Results from an assessment of the heat affected zone (HAZ) low temperature toughness when exposed to increased heat input welding for heavy wall pipe, and susceptibility to cold cracking during decreased heat input welding are presented. The impact of alloying elements - nickel, chromium, vanadium and molybdenum in niobium microalloyed steels, on ductile-brittle failure parameters is demonstrated. Vanadium and molybdenum microalloying of niobium-based steel results in HAZ embrittlement. Optimum post welding cooling rates have been determined, for which the HAZ toughness is ensured at temperatures down to minus 30 ^{o}C in niobium-based steels with nickel or chromium additions.

Introduction

Control of the welding process of heavy-wall (23 – 34 mm) high-strength pipes is considered to be the relevant problem during DSAW welding (Double Submerged Arc Welding) in the pipe mill as well as in pipeline installations.

In the process of developing heavy-wall high-strength steel pipes, weldability evaluations are considered to be necessary. The criteria for weldability evaluation have been established based on simulations of welding thermal cycles and weld joint testing. Simulation of welding thermal cycles has been performed on samples under conditions as for the weld HAZ and provided simulation of the HAZ thermal fields within the temperature range from 1300 to 650 ^{o}C.

Techniques for weldability simulation and assessment have been developed by The Ferrous Metals Central Research Institute and have been actively used within the CBMM joint science programs. A considerable contribution into these investigations has been made by J. M. Gray, F. Heisterkamp and K. Hulka.

The simulation techniques used ensured application of multifactorial tests, and have been used to evaluate results for X70 steels (Tensile strength 580 MPa and yield strength 490 MPa). Chemical compositions are shown in Table I.

Table I. Chemical Compositions of Steels Investigated

Composition code	Chemical composition, %							
	C	Mn	Si	Nb	V	Mo	Cr	Ni
1	0.06	1.65	0.32	0.03	0.04	0.15	-	0.15
2	0.05	1.75	0.31	0.06	-	-	0.17	-
3	0.10	1.62	0.43	0.05	-	-	0.1	-

All supplier's rolled products have been in conformance with DET NORSKE VERITAS (DNV-OS-101) general requirements. Steel compositions varied in carbon, vanadium and molybdenum contents as well as chromium and nickel.

Heat Input – as Criterion of Welding Process Parameters Optimization

Nowadays there is no practical alternative to DSAW welding for the longitudinal seam weld with 4 or 5 wires. Modern wires and fluxes ensure adequate weld metal properties, however for the HAZ it is difficult to ensure properties similar to the base metal. Due to this, it is necessary to search for possible improvements in steel composition and optimisation of the welding process which is to be applied at welding stations in pipe mills.

One of the basic process parameters influencing weldability is heat input, adjustment of which allows reliable fusion and the required geometrical forming of the weld metal.

Figure 1 shows predictions of the rate of cooling at any given heat input depending on the temperature of the previous weld bead for pipe with a wall thickness of 25.4 mm. This data is important during specification of the pipe welding process cycle to maintain a continuous production flow because hot metal external welding (OD) possibly starts on the previous internal weld bead before it is fully cooled. In this case $W_{800/700}$ cooling rate can be varied within the range of 5 – 7 ^{o}C/sec. In the process of cold metal welding the cooling rate conforms to 10 - 12 ^{o}C/sec, Figure 1.

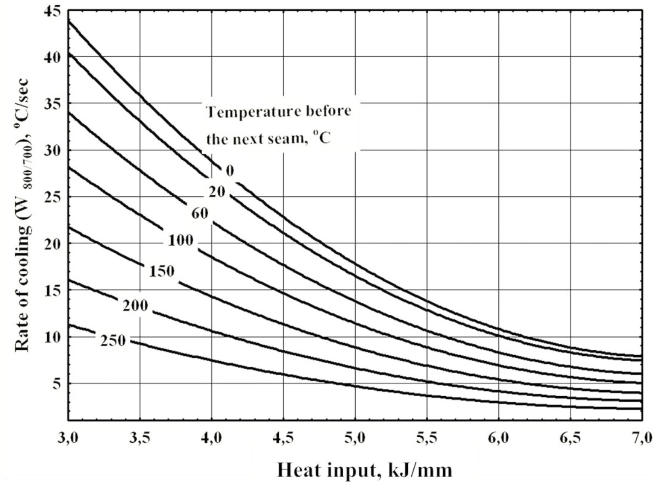

Figure 1. Determination of cooling rate after welding (pipes with thickness 25.4 mm) from heat input and temperatures before welding ($W_{800/700}$ = 1,92 x $W_{800/500}$).

Thermal Kinetic Analysis of Austenite Polymorphic Transformation – Assessment Criterion for Heat Affected Zone Microstructure Condition

This analysis refers to a physical metallurgical method for steel weldability assessment via construction of continuous cooling transformation diagrams for austenite transformation in the heat affected zone (HAZ).

Austenite transformation of the HAZ for samples heated above the temperature at which strong grain growth begins, has been assessed using dilatometer specimens rapidly heated to temperatures of 1300 – 1320 ^{o}C.

Transformation kinetics of austenite for the steels presented in Table I is characterized by an area of bainite transformations within a broad range of cooling rates, Figures 2 and 3. Martensitic transformations are observed at high, yet relatively realistic cooling rates for welding pipeline butt seams.

Figure 2 (Steel 1) shows that ferritic transformation, influenced by the effect of vanadium, molybdenum and nickel on diffusion kinetics, is shifted towards slower cooling rates typical of submerged arc welding of heavy-wall pipes. The area of ferritic transformation is limited to cooling rates of 2 – 3 ^{o}C/sec and below. Ferrite is formed in the temperature range of 720 – 620 ^{o}C. Transformation to cementite is complicated due to carbonitride formation. Martensitic transformations occur within the temperature range of 490 – 320 ^{o}C, the critical rate of local quenching is within the range of 50 – 120 ^{o}C/sec.

Transformation kinetics of austenite for the steel 2 composition studied (C-Mn-Nb-Cr) without molybdenum and vanadium, show significant differences connected with the formation of cementite carbide instead of carbonitride precipitation, both in the ferritic area and in the pearlite and bainite transformations, Figure 3. Ferrite is formed in the temperature range 740 °C - 630 ^{o}C, beginning from a cooling rate of 20 ^{o}C/sec. and below. Martensitic transformation occurs in the temperature range 500 – 360 ^{o}C, and the critical cooling rate of local quenching is within the range of 90 – 280 ^{o}C/sec.

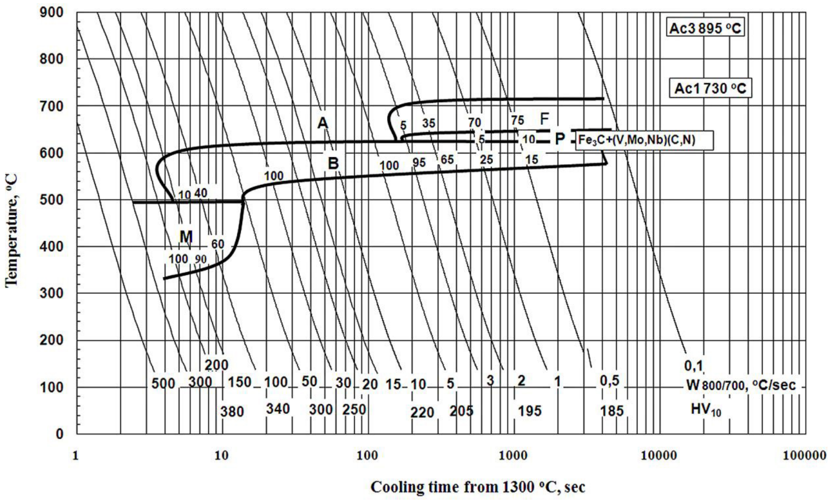

Figure 2. Austenite transformation diagram for the HAZ of steel N^{o} 1.

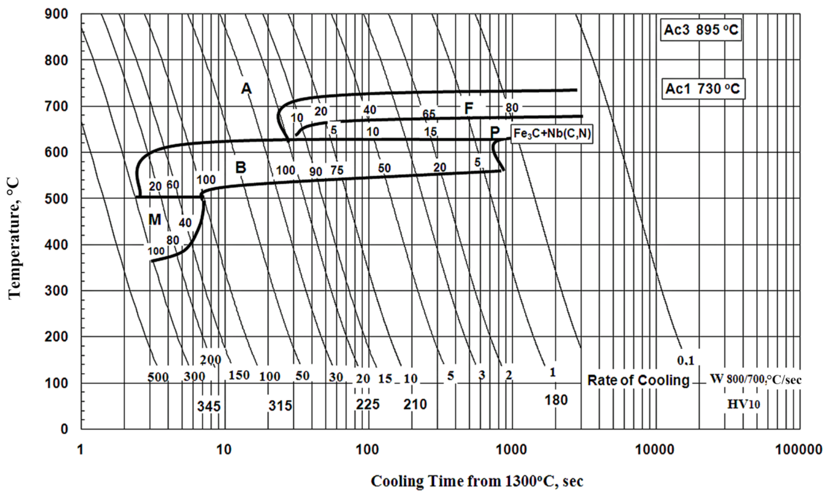

Figure 3. Austenite transformation diagram for the HAZ of steel N^{o} 2.

A carbon content up to 0.10% as in steel 3 increases austenite stability and initiates martensitic transformations especially at cooling rates typical for welding of field butt joints.

Bainitic microstructures formed in approximately similar temperature ranges, have differences in morphology of microstructure components (size, shape, presence of granular morphology and/or of lath morphology inside austenite grains) for the compositions studied.

Assessment Criteria for Tendency to Soften during Welding

Within the range of intercritical reheating temperatures, diffusional transformations occur at higher temperatures providing conditions for special carbide formation. This effect leads to softening of the HAZ.

While conducting experiments aimed at analysis of the influence of different heating temperatures in the HAZ, the cooling rate has been maintained at 12 ^{o}C/sec., i.e. at the cooling rate typical of submerged arc welding of pipes. Figure 4 shows austenite transformation temperature variation depending on the peak temperature in the HAZ.

The addition of V, Mo and Ni in composition 1 promotes austenite stability in the HAZ for reheat temperatures of 1150 ^{o}C to 1300 ^{o}C. Austenite transformation occurs between 610 – 530 ^{o}C which produces a bainitic microstructure.

In the heat affected zone, in the temperature range 1050 to 950 ^{o}C, the temperature of austenite transformation-start increases from 660 to 720 ^{o}C; along with bainite the ferritic phase, cementite and carbonitride phases also occur. In the HAZ reheating range from 950 to 890 ^{o}C (Ac_3) the temperature of austenite transformation increases up to a temperature of 760 ^{o}C; and ferrite containing cementite or carbonitrides is produced.

The intercritical temperature range in the heat-affected zone corresponds to 730 – 870 ^{o}C. Ferritic phases of various morphologies are formed in this temperature interval: consisting of the ferrite arising from austenite while cooling, and the ferrite obtained during heating in the area of partial transformation.

In the presence of vanadium and molybdenum the propensity for softening has slightly decreased in the heat-affected zone. The formation of carbonitride phases of (V, Mo) (C, N) type has to a certain degree led to an increase in hardness of the polygonal ferrite. In the steel without additions of vanadium and molybdenum the austenite transformation happens at higher temperatures and is mainly based on diffusion kinetics. An increased temperature of diffusional transformations in the HAZ intercritical heating range has been detected, which leads to a higher tendency for softening.

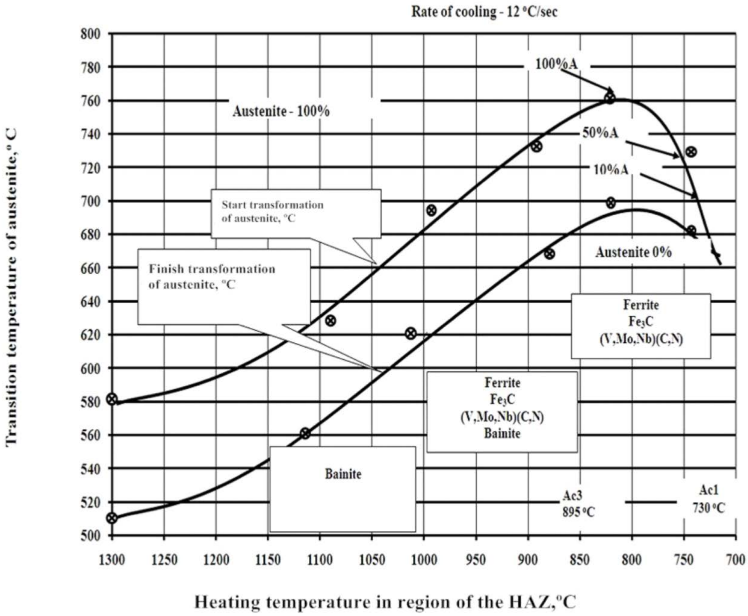

Figure 4. Change of austenite transformation temperature in the HAZ after heating to 1300 to 730 ^{o}C.

A hardness drop of 15% and more below the base metal level has been detected in the HAZ region when heated up to intercritical temperatures (Ac_1 - Ac_3).

The aggregate weld joint strength is a combination of the soft zone in combination with hardening phases. In order to define the aggregate strength the allowable soft interlayer width has been defined by the following equation:

$$\sigma_в = \beta\ \sigma_в^{HAZ} [d + (1 - d)*(a + (m/\chi + c)],\ \text{where} \qquad (1)$$

$\sigma_в$ - the aggregate strength of the weld joint;
$\sigma_в^{HAZ}$ - soft interlayer strength;
$\sigma_в^{bm}$ - base metal strength;
$d = \sigma_в^{HAZ} / \sigma_в^{bm}$ - ratio;
χ - soft interlayer width;
β, a, m, c - factors, depending on interlayer cross-section dimensions.

The softened layer strength has been defined by means of minimum hardness values converted to yield strength values at the set test temperature.

Toughness of the Weld Heat Affected Zone (HAZ) – Weldability Criteria for HAZ Resistance to Brittle Fracture

Impact toughness in relation to other mechanical properties is sensitive to changes of microstructural constituents. Assessment of the HAZ embrittlement effect after welding has been evaluated by means of testing impact specimens with a sharp notch (KCV), after heating – with different thermal welding cycles in combination with cooling rate variations.

For every cooling rate (type of welding) a series of KCV tests has been conducted at temperatures from +20 °C to -60 ^oC and the temperature of ductile-brittle transition has been defined as the average energy transition temperature or 50% of upper-shelf energy (50% USE). The upper-shelf energy corresponds to the initial stage of transition from a ductile to brittle fracture mechanism. The average energy (50% USE) from the maximum level of impact toughness to minimum at negative testing temperature corresponds to the ductile-brittle failure area. The average energy transition level for these steel compositions is 100 J/cm^2. The low energy transition temperature corresponds to brittle fracture as a result of undesired microstructures in the heat affected zone region. Figures 5, 6 and 7 show various transition curves for steels 1, 2 and 3 respectively. The HAZ cooling rates used for each simulation corresponding to different heat inputs are shown on the individual curves in the figures.

Relative characteristics of low temperature toughness for the HAZs of different micro alloyed steel compositions are measured, as a rule, under cooling conditions typical of welded joints made at various heat inputs. In our research, when welding pipes of large diameter, the outside seam cooling rate for the hot pass from the inside seam heat is 6 – 7 ^oC/sec., and the outside seam cooling rate after the inside seam is complete is 10 – 12 ^oC/sec.

Weldability evaluation according to the low temperature toughness criteria of 50% USE - KCV 100 J/cm^2 has shown that at a cooling rate of 6 – 7 ^oC/sec. none of the three alloying compositions studied exhibits a decrease in temperature durability of the welded joints. The temperature threshold for the HAZ of steel compositions N° 1 and 3 is plus 20 ^oC and above.

HAZ toughness deterioration as a result of the welding thermal effect and low cooling rates is connected with two interrelated factors adversely affecting the HAZ during welding. Embrittlement is related first to formation of ferrite micro-structural components of the "Widmanstatten" type with marginal distribution over grain boundaries during austenite transformation and secondly – to carbonitride phase formation related to vanadium and molybdenum. Figure 5 shows the embrittlement effect of complex micro-alloying additions of vanadium, molybdenum, and nickel and the reduction of HAZ toughness (steel 1).

Vanadium and molybdenum contribute to not only carbonitride phase formation at low cooling rates, but also affect transformation to bainite as well. Therefore bainitic morphology changes from granular to lath like. Bainite with lath morphology is formed as the cooling rate increases.

Weldability evaluation according to the criteria of 50% USE showed, Figure 8, that at a cooling rate of 10 – 12 ^{o}C/sec. for the outside seam, the HAZ meets the temperature threshold of ductile-brittle fracture (KCV 100 J/cm^2), down to -35 ^{o}C for steel 2. Alloy composition N^{o} 3 meets the threshold down to -10 ^{o}C while alloy composition N^{o} 1 meets the threshold down to 0 ^{o}C.

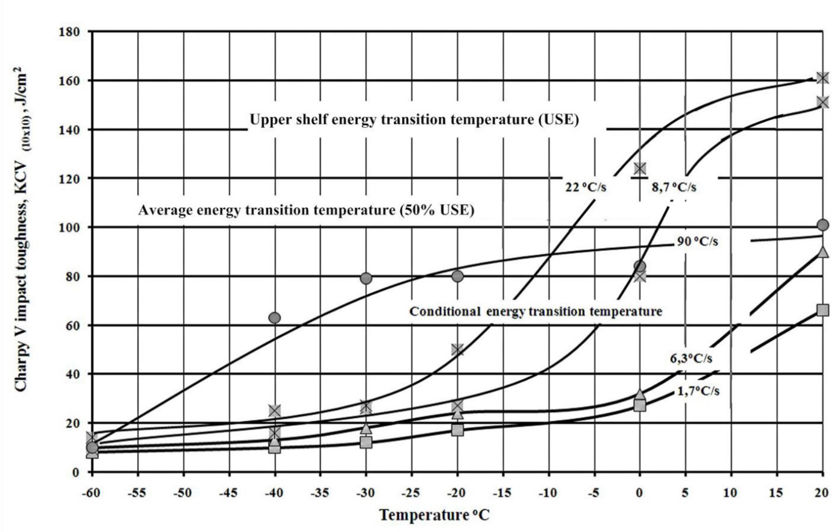

Figure 5. HAZ transition curves for different cooling rates, steel N° 1.

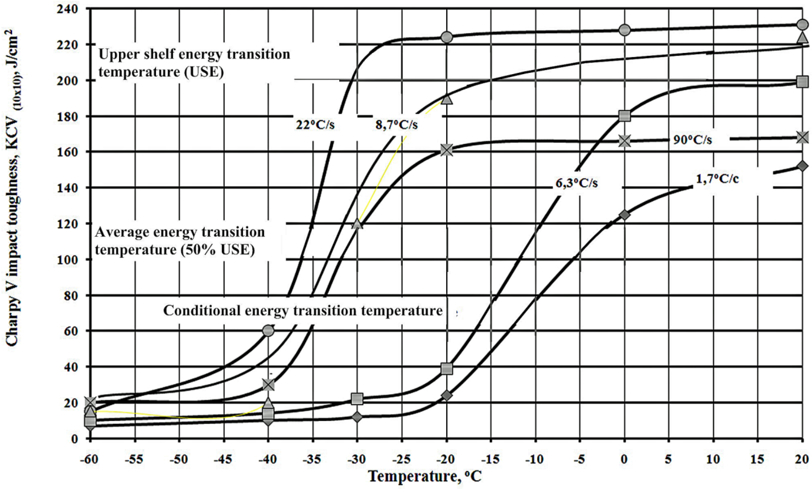

Figure 6. HAZ Charpy transition curves for different cooling rates, steel N° 2.

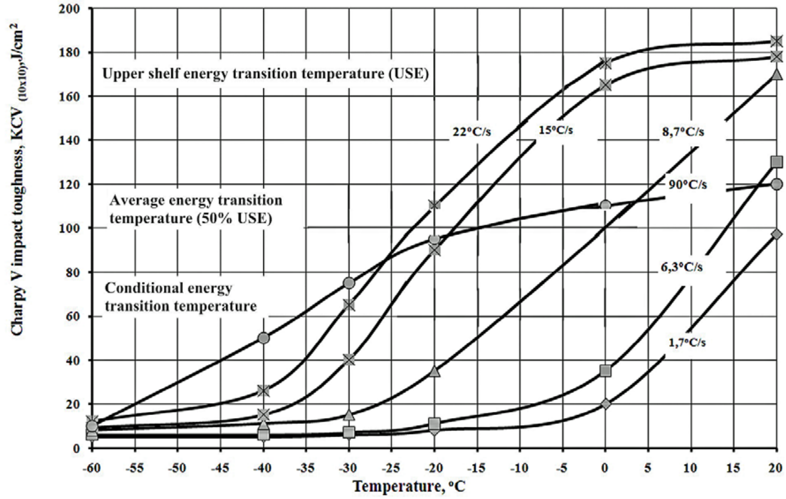

Figure 7. HAZ Charpy transition curves for different cooling rates, steel N° 3.

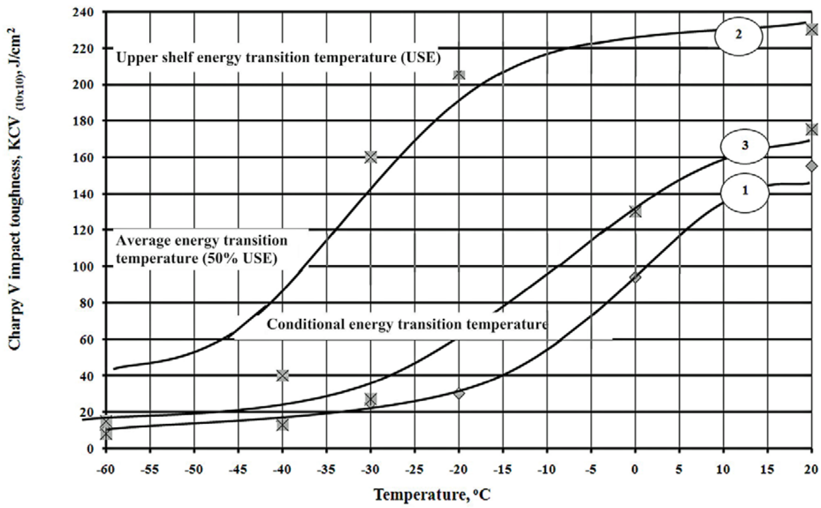

Figure 8. Comparison of HAZ Charpy transition curves for a cooling rate of 12 °C/s for steels 1, 2 and 3.

Acceptable Cooling Rate Range After Welding – Weldability Criterion According to the Set Level of Weld Zone Impact Toughness

One should consider the relative toughness threshold which is stipulated in the specification. In order to get absolutely reliable results during weldability research it is reasonable to select the temperature threshold T_{50} (100 J/cm^2). This is a high barrier that needs to be met by means of correct steel alloying and welding practice for pipes of large diameter.

Figure 9 shows curves of impact toughness change as a function of the cooling rate after welding, this enables one to determine acceptable cooling rates and heat input ranges during welding. Steel composition N$^{o.}$2 (Nb – Cr) has the widest range of acceptable cooling rates for both field welding and mill longitudinal submerged arc welding.

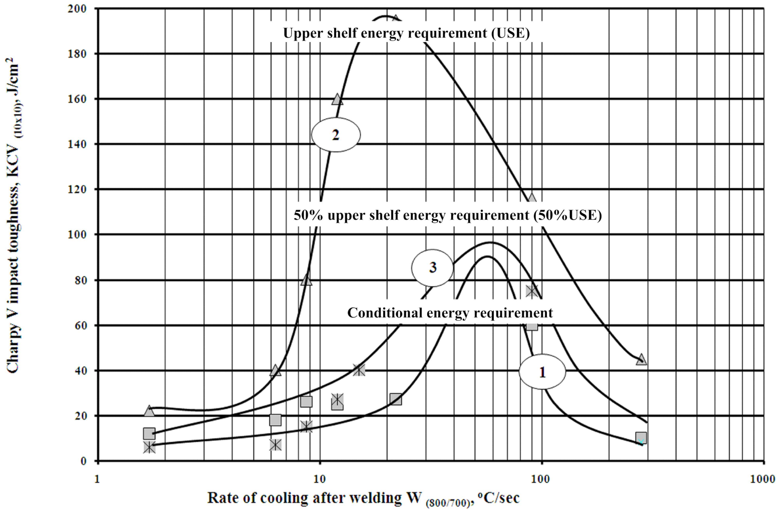

Figure 9. Determination of allowable cooling rate for maintenance of adequate HAZ impact toughness at -30 °C for steels 1, 2 and 3 depending on energy requirement (USE, 50% or conditional).

Crack Strength Assessment – Weldability Criterion Based on Ductile Fracture Resistance of the Weld Zone

At the present time, toughness requirements are based on fracture mechanics approaches that utilise engineering calculations for determination of allowable defect sizes in the pipe, including the HAZ and weld metal.

The analysis of the HAZ in a welded joint area with reduced toughness is of practical interest due to the necessity to ensure the working capability of welded structures under conditions of reduced temperatures, vibrational loads and other factors determining the demand for a particular level of welded joint toughness.

Methods of crack tolerance analysis involve testing of both base metal and full-size welded joints for determining the value of critical crack tip opening displacement – CTOD as per British Standard BS 7448 – 1.

In actual practice when, testing the HAZ of welded joints, a large number of samples are rejected due to the fact that the specimen fatigue crack, when grown, cannot always be located in the correct HAZ microstructure. Crack front and fusion line curvature results in control of the fatigue crack front location being very difficult. It often becomes impossible to estimate the value of critical crack tip opening displacement of actual welded joints.

The difficulties listed above are solved when δ_c (critical CTOD) value is determined on samples with simulated HAZ microstructures. The sample, pre-heat treated for a specified welding heating and cooling cycle along a complete section of test metal, ensures testing of microstructures relevant to the applied welding cycle.

The critical δ_c (CTOD) on samples of specified section thickness is determined according to the formula in British Standard BS 7448-1 fracture mechanics toughness tests - Part 1: Method for determination of KIC, critical CTOD and critical J values of metallic materials.

CTOD test results as well as impact test results can be correlated with microstructural changes that are dependent on post-welding cooling rates and steel chemical composition, Figure 10.

In contrast to general requirements for impact toughness, the δ_c parameter is a material property that can be used for “engineering” calculations.

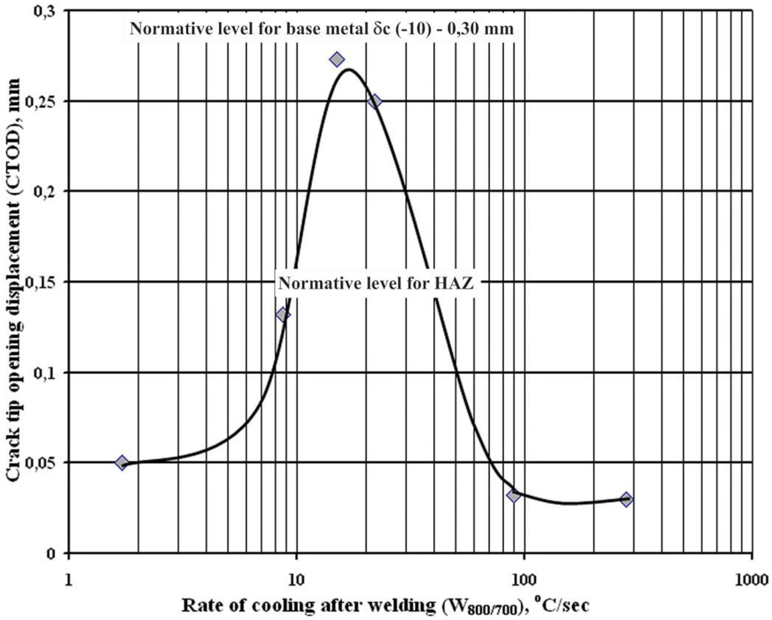

Figure 10. Influence of cooling rate on change of CTOD value, steel N^o 3.

The CTOD value shall be 0.30 mm at the minimum design temperature of -10 ^oC for the base metal and 0.15 mm for the HAZ.

Tendency for Cold Cracking – Weldability Criteria for Low Heat Inputs, Carbon Equivalent

The risk of cold cracking or delayed fracture occurrence depends on:

- Microstructure type and temperature during austenite transformation (less than 500 ^{o}C);
- Presence of diffusible hydrogen;
- The level and rate of strain increase occurring in welded joints.

All these factors are interrelated during the welding and cooling of welded joints. The study of each factor in the cracking process is rather complicated and regression analysis of test results sometimes leads to ambiguity.

Special techniques for determination of the tendency for cold cracking on simulation samples (Charpy-type) with HAZ microstructures have been developed for examination and quantitative assessment of each of these influences. A four-point bending specimen arrangement with varied loading speeds (1.6 x 10^{-4} … 1.6 x 10^{-8} m/s) has been developed. It was defined, that after welding thermal cycle simulation the average content of diffusion-active hydrogen was 1 cm^3/100g, but this hydrogen became concentrated in the tension stress area. Kinetic analysis of the effect indicated, that the hydrogen content in the top of the specimen can be more than an order of magnitude higher, than in the unloaded part of specimen. The higher the testing rate, the less hydrogen embrittling influence was exerted, because the time was not sufficient to allow hydrogen concentration.

The susceptibility of the HAZ to delayed fracture has been assessed on the basis of failure stress intensity factor value ($N/mm^{3/2}$). Specimens were tested at low loading rates (0.009 mm/min), i.e. in delayed fracture conditions. A method of constrained preliminary electrolytic hydrogenation has been used to examine the influence of high hydrogen concentration. After weld heating simulation up to 1320 ^{o}C, the content of diffusible hydrogen could reach 3 - 4 cm^3/100g.

The results from this type of test can be subjected to regression analysis, which allows specification of a critical (acceptable) HAZ hardness for resistance to delayed cracking.

Such tests are relevant to review the standard Carbon Equivalent (CE) and Pcm formulae for pipe steels with extremely low carbon content and the present-day practice of heavy wall pipe root field joint welding.

CE= C+Mn/6 + (…..)

Pcm = C+Mn/20 + (….)

Research results from I.P. Bardin TsNIIChermet indicate the expediency of adjusting the above formula for the manganese coefficient. In proportion to carbon content reduction, the effect of manganese influence decays as follows:

- Mn/25 for carbon 0.05 – 0.07%;
- Mn/30 for carbon less 0.05%.

These preliminary findings require further research and validation work before amendments to allowable pipe steel chemical composition and welding practice in regulatory specifications can be allowed.

Conclusions

A simulation method for weldability assessment based on HAZ thermal cycling, is represented in this paper, and displays a consistent approach to steel assessment. This method allows determination of weldability parameters for various steel compositions without the influence of processing factors often associated with tests on actual welded samples.

The simulation method is less metal-intensive, while allowing testing of sufficiently large numbers of specimens, and testing types. Long-term experience shows that defined correlations with the aid of welding thermal-cycle simulations, sufficiently reflects the behaviour of the HAZ during actual welding. Experimental results obtained by such a test method, can be used for developing regression models (predictive equations), which reflect the cooling rate after welding and the influence of chemical composition on specified “weldability” criteria. While using the simulation method, the cooling rate, closely matched with the value of heat input during welding, is used as the basis for weldability criteria.