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Khemici Badri, Eng., M.Eng., IWE/EWE

Sodel Ltd.

khemici.badri@sodel.com

PORTEVIN LECTURE

Weldable High-Strength Steels: Challenges and Engineering Applications

Prof David Porter.

The presentation deals with the key features of HSLA steels and the challenges regarding their weldability and engineering applications.

The continuous use and development of steels is due to multiple reasons, namely abundance of raw materials, ease of production, recyclability and rigidity besides the fact that they are environmentally friendly materials. For all engineering applications, the tendency of the design is to use lighter constructions for saving fabrication costs and raw materials while meeting required performances. The reduction of weight in welded assemblies and constructions involves the use of thin parts and components, which in turn results in reducing materials consumption and maintenance costs. The development and use of high-strength steels have resulted in consistent energy savings and a reduction of carbon dioxide (CO_2) emissions.

Many strengthening techniques and mechanisms have been used for the development of high-strength steels. All of them are based on the combination of a refined chemical composition, e.g. micro-alloying, and processing methods. In fact, normalizing or controlled rolling, quenching and tempering are considered among conventional processes whereas thermo-mechanical control process TMCP is a modern process.

The yield strength range of high-strength-steels varies from 200 to up to 1300 MPa for weldable structural applications. Steels with the following yield strength are given as example: 355 MPa for pressure vessels, 550 MPa for gas pipes, up to 1300 MPa for lifting systems. The microstructure of low- strength grades is generally a polygonal ferritic while it is composed of martensitic, bainite or both for high-strength grades. High-strength ferritic steels are also subject to a change in their failure mode, from ductile mode to brittle mode, and their fracture toughness level decreases with increasing strength and thickness.

Features and characterization of high-strength steels

In welding, the heat affected zone “HAZ” remains the critical region since it undergoes microstructural changes under thermal cycles. However, welding thermal cycles associated with rapid cooling rates, namely 10 to 100 °C/s give rise to a HAZ with a microstructure comparable to that of the quenched base metal except the fact that the prior austenite grain size is larger in coarse-grained and inter-critical regions of the heat affected zone (CGHAZ, ICHAZ). As a consequence, this leads to an increase in the ductile-to-brittle transition temperature “DBTT” of the HAZ and a toughness drop of the welded joint. Some methods have been introduced to control the grain size of the CGHAZ in micro-alloyed HSLA steels, e.g. by titanium nitride TiN precipitates or oxide precipitates in the case of high heat input processes like electroslag welding (ESW).

In modern high-strength steels, the sub-critical heat-affected zone SCHAZ undergoes an increase of their DBTT as a result of strain ageing occurring in the temperature range of 250-450 °C. Because of their good cold formability, these steels are cold bent before being welded. The combination of both effects of bending and welding results in strain-ageing accompanied by a cracking risk in the SCHAZ of these steels.

The SCHAZ and ICHAZ are also prone to softening under the heat of welding. Direct quenched low-carbon steels are more prone to this phenomenon than high-carbon grades. In order to counter this softening and have a good tempering resistance, a small amount of elements like molybdenum and niobium are deliberately added to the steel chemistry. On the other hand, such elements increase the carbon equivalent and the susceptibility of the HAZ to cold cracking. Therefore, the addition of such alloying elements must be well adjusted, mainly to avoid the Zone III and possibly the Zone II on the Graville diagram.

It is noted that the susceptibility to hydrogen cracking increases as the strength of the steel increases. This concerns the heat affected zone as well as the weld metal in HSLA where the embrittlement is attributed to the presence of stress triaxialilty in thin sections. The sources of hydrogen are multiple: liquid steelmaking process, welding consumables, shielding gas, corrosion, cathode protection, etc.

As stated above, the low carbon content in quenched and tempered QT steels confers a good toughness in the soften HAZ but the welded joint cannot carry the full strength required by the design. Therefore, the increase of the cooling rate $t_{8/5}$, as a consequence of a low heat input, is recommended to overcome the HAZ softening and achieve the matching requirements of the welded joint. However, it is not yet possible to completely achieve a matching transverse strength in ultra-high-strength steels (UHSS).

Regarding high-temperature applications, the development of creep resistant ferritic steels e.g P91 and P92 grades, has contributed to increase the thermal efficiency and to reduce CO_2 emissions in modern power plants. The introduction of these grades has allowed to reach service temperatures above 600 °C. However, it appeared that the creep rupture strength in the welded joint of this type of steels is subject to a premature failure, referred as type IV cracking. The latter occurs in the outer edge of HAZ, namely in fine grain / intercritical region, during long-term service. Recent efforts have led to the development of new alloy steels such as MARBN (Japan) and NPM1 (Austria) for application temperatures of up to 650 °C. These modified grades, which are deemed to have a good resistance to Type IV failures, consist of boron micro alloying element with low nitrogen.

Contrary to non-welded structures whose fatigue strength increases with the material strength, the welded joints remain the critical region that limits the fatigue endurance in welded assemblies. Nevertheless, the application of some techniques such as TIG dressing, hammer peening and mechanical impact treatments to reduce residual stresses in the weld toe regions has led to an important improvements in fatigue strength of welded joints and structures. Also, other techniques like weaving or elongated bead welding have also proved to be effective for this purpose.

From the standpoint of design, there are no codes covering the design of structures made of high- strength steels having a minimum yield strength of 700 MPa (100 psi). In the current standards, the properties of steels is based on the combination of the chemical composition and steelmaking processes. Besides, the development of new materials is still subject to the modification of fabrication standards, which is a difficult and a long process. This condition even hinders the development of better steels that can be made with new and cost-effective processes. As consequences of these rules, the author of this article points out the fact that the standards should normally only focus on setting the technical performance levels that the new steels must meet while letting steelmakers look after the appropriate combinations between chemical composition, processing routes and parameters. The author makes such a suggestion after having analyzed, inter alia, the following findings:

- The limitation of the content of micro alloying elements (Nb+Ti+V) below 0.15 wt.% is not valid regarding the impact they may have on the HAZ toughness for structures, pressure vessels and pipelines. Indeed, it has also been demonstrated that:
    - The effect of titanium is non-linear on HAZ properties. This element has a beneficial effect at lower contents and a negative effect at higher contents.
    - The effect of niobium on HAZ properties depends on the amount of alloying elements and the thermal cycle experienced in HAZ. In fact, it was found that the combination of a high niobium content, namely 0.09 wt.%, with a low carbon content of 0.04 wt.%, constitutes a cost-effective approach for the production of modern TMCP steels with a yield strength of 420-550 MPa, e.g API 5L X70 & X80. Niobium rises the recrystallization temperature of the austenite which results in a thermomechanical rolling finition at high temperatures that gives rise to lower rolling loads and a better shape control.
- The content of manganese, limited to below 2.0 wt.% in TMCP steels should be higher as carbon content is reduced below its maximum limit.

Regarding the application limits, it is important to mention the higher ratio of yield strength to ultimate tensile strength (YS/UTS) for high-strength steels having a yield strength in the range of 500-690 MPa. This makes difficult and even risky to exploit their full strength. The Eurocode EN 13445 limits the design stress to YS/1.5 or to UTS/2.4 for pressure equipment.

Examples of some typical uses of high-strength steels include:

- Gas transportation where API 5L X80 is still widely used for high pressure gas transmission lines.
- Shipbuilding: TMCP plates with a yield strength of 460 MPa for deck structures in large ships.
- Offshore structures: QT steels, 550-800 Mpa.
- Mobile lifting system: highest yield strength grades, 900-1100 MPa.

PLENARY SESSION

Advanced pipe welding with gas metal arc welding

Dr Petteri Jernström.

The presentation relates to the advancements in GMAW process with the development of the possibility of welding root passes from one side in open grooves without weld backing system.

The new process, WiseRoot+ process, is a low energy cold welding process. It was particularly developed for improving the quality and the productivity of pipe girth welding. The operational mode is based on a short-circuit transfer mode that produces a smooth droplet transfer at low current levels without spatter. Besides, once the droplet is detached, the advanced power source quickly delivers a pulsed current to heat the weld pool and thus prevents the lack of fusion or cold lapping problems that are common in conventional MIG/MAG welding processes.

The RootWise+ process provides a good weld pool control and penetration characteristics. It is considered as a cold process and offers good capabilities for all-position welding besides vertical down position.

Furthermore, the process is tolerant for root gap variations and is highly productive for narrow grooves. Its welding speed is up to 4 times higher than that of GTAW process.

Degaussing of ferromagnetic materials

Mr Boyan Ivanov

The presentation of the company EWM Group deals with the phenomenon of magnetism in ferromagnetic materials and its consequences on arc welding operations. As well, a degaussing process with a specific device was presented with some recommendations.

The residual magnetism in the material e.g Fe, Ni, Co, causes the welding arc to deflect and to be unstable. This results in a bad seem quality, lot of spatters, lack of sidewall fusion, slag inclusion, porosity, etc. The magnetism in the metal can be induced by different ways like machining, hardening, welding, cutting, magnetic particle inspection, handling with magnetic cranes or lifter, metallic friction, etc. In such conditions, a degaussing operation must be carried out to eliminate the residual magnetism in the material prior to welding.

The degaussing device in question consists of a switch polarity system integrated in a manual metal arc welding machine with a power source EWM Pico 350 cel puls pws dgs. Other degaussing power sources like EWM degauss 600 are also available apart.

Degaussing is carried out by attaching the cables to the part or the pipe. At first, a degaussmeter is used to determine the magnetic flux density in the part. For degaussing, the grounding cable is winded around the workpiece or pipe to be degaussed with a sufficient number of turns to easily reduce the residual magnetism. The elimination of the residual magnetism in the material occurs by sending an alternating current (AC) through the windings around the material. The degaussing is achieved with changing the current flow direction and the reduction of its amplitude.

Applications in high-strength steels– possibilities on new segments

Mr Anders Ohlsson.

The topic of the presentation deals with the main features of high-strength steels and their different fields of application. For structural steels, the yield strength varies from 200 to 1300 MPa. For wear plates, the hardness level can reach up to 700 HV depending on the grade of the steel in question.

High-strength steels are used in several fields and engineering applications:

- transportation: cars and trains;
- energy: onshore and offshore platforms;
- cranes: loader cranes and mobile cranes with YS of 960-1100 MPa;
- earth moving: wear protection in buckets: Hardox HiTuf, Hardox 400, Hardox 600;
- agriculture: soil compaction;
- construction;
- automotive industry.

The driving force of the development of high-strength steels in the automotive industry is summarized in the following points: the increase in crash rates, weight reduction, reduction of fuel consumption with new CO_2 target emission, cost-effective manufacturing methods. The car structure (frame) and seat structure (seat back and seat track) are just few examples out of many. In train, examples include the safety cage to protect driver from collision, anti-climbers to absorb stress, etc.

The design of high-strength steels and their use in welded constructions must take into account the different welding issues related to these materials: risk of toughness drop in HAZ, mismatching joints, fatigue, etc. The choice of low hydrogen consumables, eg. H5 max, and also low transformation temperatures (LTT) filler metals are among the main factors controlling their weldability.

PARALLEL SESSIONS

SESSION 03 PRESSURE VESSEL AND PROCESS INDUSTRY APPLICATIONS

Neural network-based hardness and toughness prediction in HAZ of temper bead welding repair technology

Dr Lina Yu

The article deals with the prediction of the hardness and the toughness in HAZ of temper bead welding using Neural Network-based method.

Temper bead welding is applied in multi-pass welding. It is an effective and alternative welding method for PWHT. The bead tempering mainly aims to refine the coarse-grained heat affected zone (CGHAZ) of the base metal and weld beads. This tempering effect is caused by the heat arising from the consecutive layers of the deposited weld metal.

The study concerns a low-carbon alloy steel A533B. Samples were welded by GTAW using a similar filler metal. The Ac_1 and Ac_3 temperatures of the base metal are 670 and 837 °C respectively. 7 layers including 78 pass welds were deposited on the top of the sample. Gleeble 1500 apparatus was used to simulate the thermal cycles in HAZ of the welds. The most affecting parameters for hardness and toughness are peak temperature, cooling rate and the number of weld beads or thermal cycles. Three thermal cycles were used with different peak temperature ranges for four types of simulation. The 1st cycle has a peak temperature 400 (< Ac_1) to 1350 °C to simulate all HAZ regions from SCHAZ to CGHAZ. The 2nd cycle has a peak temperature from 670 (> Ac_1) to 1350 °C to simulate ICHAZ to CGHAZ. For the 3rd cycle, the peak temperature ranges from 400 to 650 °C (< Ac_1) to simulate the tempering in SCHAZ. The cooling rate was varied from 3 to 100 °C/s to cover the possible cooling rate ranges produced by GTAW process in HAZ.

The predictive neural network model is first built based on the measured results. It shows that the hardness in HAZ after 7 layers is lower than 350 HV and the higher hardness appears in the coarse-grained HAZ. Also, the toughness reveals that the lower toughness appears in the CGHAZ near the weld metal.

The predicted hardness in HAZ presents a good agreement with the experimentally measured values and the average of both cases exceeds 40 J/-20 °C.

Influence of the Weld Heat Input on the Electrochemical Behavior of the Q690 High-Strength Steel Welding Joints

Kai Wang

The study focuses on the role of the heat input in obtaining welded joints with similar electrochemical performances as the base metal for the high-strength steel Q690 (ASTM A514). Such a material is used, inter alia, for key parts of the drilling platform in offshore structures. The corrosion remains a serious concern in these environments even though anti-corrosive methods as painting and catholic protection are applied. The welded joint is the privileged site for a localized corrosion in the welded assemblies. The corrosion attack is initiated by the formation of holes and cracks in the welded joints before they propagate into the base metal and then affect the integrity and the service life of the structure.

The heat input affects the microstructure of the weld joint. Thus, its control in a narrow range is required to get the desired microstructure with a better behavior with regards to corrosion attack.

The base metal plate Q690 is a low carbon (≤ 0.09% C) quenched and tempered high-strength steel. 12 mm thick plates were welded by FCAW process using a rutile flux cored wire 1.2 mm ESAB OK Tubrod 15.09 (SFA/AWS A5.36. E111T1-M21A4-K3-H4.). Welding was carried out with the reverse polarity DCEP (DC+) under M21 shielding gas (80% Ar- 20% CO_2) with two heat inputs (10 and 20 kJ/cm). In both cases, the preheat temperature was kept constant, namely 150 °C. 5 weld layers were applied onto the base metal to simulate the all-weld metal of the welded joints. The welded samples were cooled in air until room temperature after welding.

The electrochemical methods used for determining the corrosion resistance are the linear polarization resistance (LPR) and the electrochemical impedance spectroscopy (EIS) in 3.5% NaCl water solution using the electrochemical workstation Autolab PGSTAT302N.

The corrosion tests show that the linear polarization curves are steadily linear and the curve related to the lower heat input shows the greatest slop. The LPR values of weld metals achieved with a heat input of 10 and 20 kJ/cm are 1532 and 1133 Ohm.cm^2 respectively compared to the base metal (1040 Ohm.cm^2). The high heat input-weld metal (20 kJ/cm) has a LPR closer to that of the base metal. Thus, it results that the low heat input-weld metal has a higher polarization resistance and then a better corrosion resistance than both the high-heat input weld metal and the base metal. Moreover, the electrochemical impedance spectroscopy (EIS) shows that the low heat input-weld metal has a higher charge-transfer resistance and then a better dissolution resistance into the solution. This confirms its better passivation behavior than both the high heat input-weld metal and the base metal.

The microstructure of the base metal consists of bainite and acicular ferrite. For the weld metal, it is shown that the high heat input decreases the cooling rate $t_{8/5}$ and, therefore, increases the austenitic isothermal time and stabilizes the austenite. This results in a coarse "lath and granular" bainite microstructure containing large laths of ferrite and martensite in addition to retained austenite islands. On the opposite, the low heat input produces a small and narrow "lath and granular" bainite.

The analysis of corrosion morphology by scanning probe microscopy JXA-8100 and EDX shows that the lower heat input welded sample shows a good corrosion behavior with the formation of a passive film on the surface which prevents its dissolution. The surface analysis of the base metal and the higher heat welded sample exhibits more pronounced corrosive attack than the lower heat input welded sample.

Narrow gap gas metal arc welding of S890QL steel

Dr.-Ing. Yaoyong YI

The study deals with the application of narrow gap GMAW with a novel automatic welding system for multi-pass butt-welding of 40 mm thick plate made of DILLIMMAX S890QL HSLA steel. The welding was carried out using a matching filler metal, Thyssen Uninion X90 (AWS A5.28 ER120S-G). The prepared joint is a square groove with a width varying from 10 to 13 mm. The preheat and interpass temperatures were kept constant at 170 °C. The heat input was increased from 1.6 to 2.0 kJ/mm with increasing groove width. Welding was done using a backing plate. Also, 10 layers were deposited to fill the joint.

It was concluded that the transverse tensile strength (YS & UTS) of the welded joints decreases marginally but the elongation increases a little bit with increasing heat input. For all samples, the fracture occurred in HAZ close to the base metal where the hardness level is the lowest; even lower than that of the weld metal. Nevertheless, the toughness of the HAZ is better than that of the weld metal

Regarding the weld metal, the mechanical properties at the lower half of the joint are better than that at the upper half. The measurement of the toughness follows the same trend as well. This is due to the tempering effect that subsequent weld passes or layers exert on the previous ones. This effect makes the lower region of the weld metal finer and tougher than the upper region for thick welded joints.

SESSION 07 MATERIALS PERFORMANCE SUBJECT TO WELDING

Effect of long-term post-weld heat treatment on the microstructure and mechanical properties of P91 weld metal

Dr Bhaduri Arun Kumar.

The investigation deals with the effect of Ni+Mn content on the microstructure and mechanical properties of the modified P91 (9Cr-1Mo) steel weld metal after long-term PWHT.

The experimental work concerns weld joints of a modified P91, 12 mm thick, welded with SMAW process using two non-synthetic P91 electrodes (W1, W2) with two different (Ni+Mn) contents, namely 1.3% (< 1.5%) and 2.3% (> 1.5%). The electrode with the lower (Ni+Mn) content has higher C, V, Nb and N contents than that of the higher (Ni+Mn) content. The preheat and interpass temperatures were kept within 200-250 °C during the welding time. Test specimens were first subjected to a post-heating at 350 °C for 15 mn before cooling to avoid cold cracking risks. Also, weld joints were X-Ray tested to find out if they are defect-free before PWHT. PWHT was done in the temperature range of 740-820 °C. Some samples were also post-weld heat treated at 760 °C for long-term, 3-100 hours, in order to understand the microstructural evolution during time. In addition, other samples were also normalized at 1050 °C for 30 mn before being subjected to PWHT in 740-820 °C for 3 h.

% wt	C	Mn	Si	S	P	Cr	Ni	Mo	V	Nb	N	Ni+Mn
W1	0.062	1.5	0.30	0.01	0.007	9.0	0.9	1.1	0.012	0.03	0.03	2.3
W2	0.1	0.75	0.32	0.008	0.010	9.0	0.52	1.0	0.21	0.065	0.06	1.3

For information, ASME code recommends that the temperature of PWHT is 730 °C minimum for this P91 creep-resisting steel whose the Ac_1 temperature is typically in the range of 790-820 °C for a (Ni+Mn) content lower than 1.5%. Practically, there is an agreement that the PWHT must be carried out at a temperature lower than Ac_1 transformation temperature in order to avoid the austenite reformation upon heating. However, it has been recently reported (*) that the PWHT should not only be determined by the Ac_1 temperature, but the mechanical properties must also be taken into account including creep rupture performance. In such a situation, it is noted that the fresh martensite formed after the reverse transformation of austenite when the weld metal is heated above Ac_1 would also contribute to improving the mechanical performances of the weld metal.

Thermocalc calculation showed that Ac_1 transformation temperature for the higher (Ni+ Mn) content Weld is estimated at 650 °C compared to that of the lower ((Ni+ Mn) content which is of 850 °C. This difference is obviously due to the high amount of austenite stabilizing element, Ni and Mn particularly.

The analysis of the weld root pass by SEM shows the formation of auto-tempered martensite with precipitates whose the amount is higher in the lower (Ni+Mn) weld metal than in the higher (Ni+Mn) weld metal. The tempering occurring in the root weld pass and not in the weld face is due to the heating effect from the succeeding weld passes. The SE images reveal that the lower (Ni+Mn) weld metal contains higher amount of precipitates than the lower (Ni+Mn) weld metal after a PWHT of 3 hours at 760 °C. In addition, the analysis made by transmission electron microscopy MET shows that the amount of retained austenite in the as-welded condition is higher in the higher (Ni+Mn) weld metal than in the lower (Ni+Mn) weld metal.

Concerning the higher (Ni+Mn) weld metal, it was reported that a decrease of the PWHT/3h from 760 to 740 °C results in finer precipitates in the microstructure. This explains that the optimal PWHT temperature for this weld metal is lower than that of the lower (Ni+Mn) content.

Similarly, an increase of the PWHT/3h temperature from 760 to 800 °C for the lower (Ni+Mn) weld metal results in a dissolution of precipitates. Furthermore, the PWHT at 760 °C for 100 hours results in a pronounced coarsening effect with a decrease of the precipitates density in both weld metal microstructures.

In the normalized condition, the lower (Ni+Mn) weld metal has a slightly higher hardness than the higher (Ni+Mn) weld metal. This is due to the fact that the latter has lower carbon content and precipitates fraction than the first.

Regarding the tempering effect after normalizing, the hardness of both weld metals decreases to reach a minimum of around 225 HV (0.5 kgf) at 740 and 760 °C for the higher and the lower (Ni+Mn) weld metals respectively. A continuous increase of the temperature up to 800 °C leads to an abrupt increase of the hardness which attain nearly 375 HV (0.5 kgf) for both cases.

The PWHT at 760 °C results in a decrease of hardness which gets its minimum, namely 220 $HV_{0.5}$ after a holding time of 3h. A further increase of the holding time up to 100 h did not produce significant change in hardness for both weldments.

The analysis of the toughness results reveals that the higher (Ni+Mn) weld metal has higher toughness than that of the lower (Ni+Mn) content for all PWHT/3h at temperatures of up to 800 °C. This is attributed to the difference of C, V and Nb contents between the weld metals. The highest levels of toughness are recorded at 760 and 780 °C respectively for higher and lower (Ni+Mn) weld metals. This difference of temperature is attributed to the difference in the amount of austenite stabilizing elements between the weld metals.

The effect of the duration time of PWHT realized at 760 °C on the toughness was also investigated at room temperature and at -20 °C. The results show that impact energy of the weldments post-heat treated at 760 °C remains higher than the requirements (47 J) of the European specification-BS EN 1599:1997- regardless of the duration of PWHT.

For sub-zero temperatures, the results show that the toughness of the higher (Ni+Mn) weld metal decreases noticeably after 10 h of holding time while that of the lower (Ni+Mn) undergoes a slight increase regardless of time. This increase is associated to the coarsening behavior of precipitates in the microstructure. However, whatever the variation of the results, the toughness level of the high (Ni+Mn) weld metal remains higher than that of the other for all PWHT duration time.

References:

BS EN 1599:1997. Welding consumables. Covered electrodes for manual metal arc welding of creep-resisting steels. Classification

(*****) Taniguchi, G. and Yamashita K., "Effect of post weld heat treatment (PWHT) temperature on mechanical properties of weld metals for high – Cr ferritic heat resistant steel", Kobelco Technical Review, No. 32 (Dec. 2013) pp 33-39

Investigation on practical application of low transformation temperature welding materials to ship hull structure made of high tensile strength steel plates for fatigue life improvement.

Prof Chiaki Shiga.

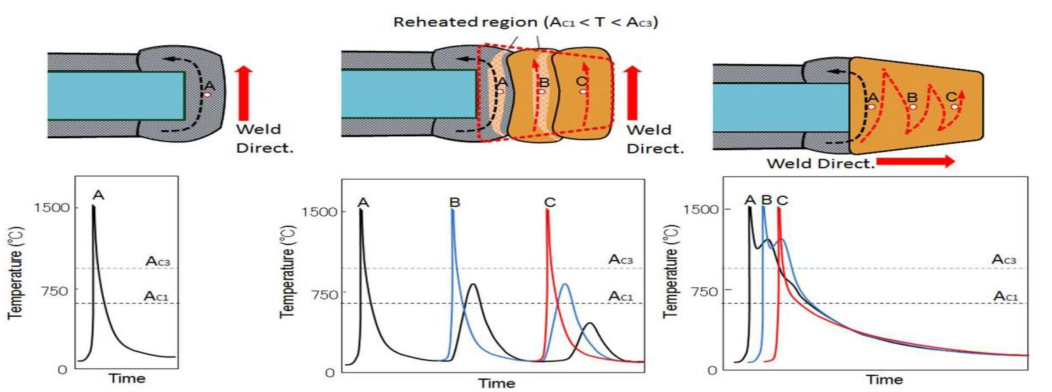
The article deals with a new welding technique consisting of the combination of elongated bead welding with the use of low transformation temperature LTT filler metals. This welding method was applied for out-of-plane gusset (stiffener) welded joints on ship hull structure with the objective of improving the fatigue strength of welded joints.

It is well known that the increase of the yield strength increases the fatigue strength of the unwelded base metal but not that of welded joints. On that basis, the fatigue strength in high- strength steels welded joint may be even lower than that of a mild steel plate. Also, the fatigue strength depends strongly on the type of weld joints. Thus, the boxing fillet welded joint presents the lowest fatigue resistance compared to other different types of joints. The presence of the tensile residual stresses in the welded joint remain the principal cause of the decrease in their fatigue strength.

In order to improve the fatigue performance in fillet welded joints, different methods are still being used. Among these are weld toe grinding and TIG/Plasma dressing, which consist of eliminating sharp discontinuities and then reducing the stress concentration at and near the weld toe regions. Other methods like impact treatment (air pressure or ultrasonic impact) consist of introducing compressive residual stresses at this critical region of the weld joint.

The concept of LTT elongated- bead weld method is based on the fact that the residual stresses in both directions, longitudinal and transversal stresses, become compressive and their amount increases with the increase of the weld length. Also, the intensity of the longitudinal compressive stress is largely higher that in the transverse direction. Hence, for the purpose of making maximum advantage of this concept, it is important to weld in the transverse direction to the expected crack direction in order to prevent the initiation and/or the propagation of the crack. This is contrary to the conventional weld metal that only intensifies tension residual stress in both longitudinal and transversal directions of the weld joint.

In the present study, the application of LTT elongated-bead welding concerns a boxing fillet welded joint of a gusset front. The welding procedure and the corresponding thermal cycle are illustrated below with comparison to a conventional boxing fillet weld (a) and a boxing fillet weld with two-pass appended welds (b).



(a) 1 pass boxing fillet weld (b) Boxing fillet weld and 2 pass appended welds (c) Elongated weld (weaving method)

The boxing fillet weld with two-pass appended welds *(b)* is performed with straight and distinct weld passes. Even both types of welds have the same weld length (extended weld bead after the boxing weld), the elongated-weld *(c)* is performed with a slight weaving ensuring that the weld bead is completed with only one weld pass. Thus, this mono-pass weld *(c)* results in one thermal cycle as compared to the multi-pass welds (b) that reheat and alter the fresh martensite. Thus, the advantage of using this type of mono-pass weld “elongated-weld bead” leads to retention of the martensite in its fresh condition which is accompanied with a volume contraction and compressive residual stresses in the welded joint.

For the experimental study, a conventional filler metal (CFM), a low transformation temperature 10Cr/10Ni filler metal (10Cr/10Ni- LTTFM) and also other low-cost low transformation temperature filler metals (LTTFM) were used for automatic welding of a 16mm-thick gusset plate on both sides of 20 mm-thick high-strength steel plate. The residual stresses were measured in the vicinity of the weld toe.

The conventional C-Mn filler wire (CFM), Kobelco MX-Z200, was used only for fillet and boxing weld joints. The M_S point of this weld metal is 703 °C and that of 10Cr/10Ni- LTTFM is 184 °C. The other LTTFM consist of 13Cr/5Ni, 3Mn/3Ni, 6.5 Mn with M_S points of 394, 472 and 408 °C respectively.

The four following welding procedures (WP1, 4) were tested:

- WP1: Fillet welds and boxing welds made with CFM.
- WP2: Fillet welds made with CFM and boxing welds with 10Cr/10Ni- LTTFM
- WP3: Fillet welds and boxing welds made with CFM, followed by elongated-weld metal with 10Cr/10Ni-LTTFM
- WP4: Fillet welds and boxing welds made with CFM, followed by elongated-weld metal with 10Cr/10Ni-LTTFM and other cheap LTTFM. The gusset was beveled all-around with an angle of 60 °C and a depth of 5 mm at the bottom.

The results showed that the elongated-bead welding after the boxing weld results in compressive or neutral residual stresses in the vicinity (~ 2 mm) of the weld toe. The compressive residual stresses are larger as M_S point is lower. The procedure WP4 results in a maximum compressive stress of 600 MPa at 2.5 mm from the weld toe.

For the residual stress distribution in the longitudinal direction around the front corner of out-of-plane gusset weld joint, the analytical study using FE simulation demonstrated the following:

- the use of CFM gives rise to high tensile residual stresses in the weld toe area for an usual boxing fillet welded joint. The tensile stress is even higher if the same wire CFM is also used for elongated-bead weld.
- the use of 10Cr/10Ni-LTTFM for only making the usual boxing fillet welded joint reduces the residual tensile stress just in the vicinity of the weld toe before it starts to increase again away from this region. However, the elongated-weld bead made with this wire gives rise to a completely compressive residual stress with an intensity of up to 600 MPa in a perimeter of 40 mm from the weld toe of the plate side.
- the residual stress is compressive at, near and after the weld toe for elongated-bead welding when using LTT filler metals: 10Cr/10Ni, 13Cr/5Ni and 3Mn/3Ni except 6.5Mn filler metal which generates low residual tensile stresses in the vicinity of the weld.

Regarding the fatigue strength, elongated-bead welding made with 10Cr/10Ni-LTTFM leads to an increase of 75% in mean fatigue strength and 6.5 times longer in fatigue life cycles compared to usual boxing fillet welds made either with CFM (WP1) or with CFM 10Cr/10Ni-LTTFM (WP2). This is due to the increase of compressive residual stresses at and beside the weld toe. It follows that the fatigue strength is improved not only by the type of the filler metal but also by the combination of the latter and the welding design.

The welded joints failed at the weld toe on the base plate for WP1 and WP2 joints whereas the fatigue cracks initiated at the root gap and propagated into the gusset plate in the case of the 10Cr/10Ni-LTTFM elongated-bead welded joints (WP3). It was also found that the presence of the bevel in the fillet joint has no effect on the fatigue strength.

Also, the fatigue strength is also improved for elongated low cost LTT beads but without the same extent as with the 10Cr-10Ni filler metal. The economic aspect should be taken into consideration for large constructions, e.g. shipbuilding.

It is concluded that the effect of elongated-bead welding on the enhancement of the fatigue strength is equivalent to or better than that achieved by impact treatment methods (air pressure or ultrasonic Impact).

The alform welding system- The world's first system for high-strength welded structures

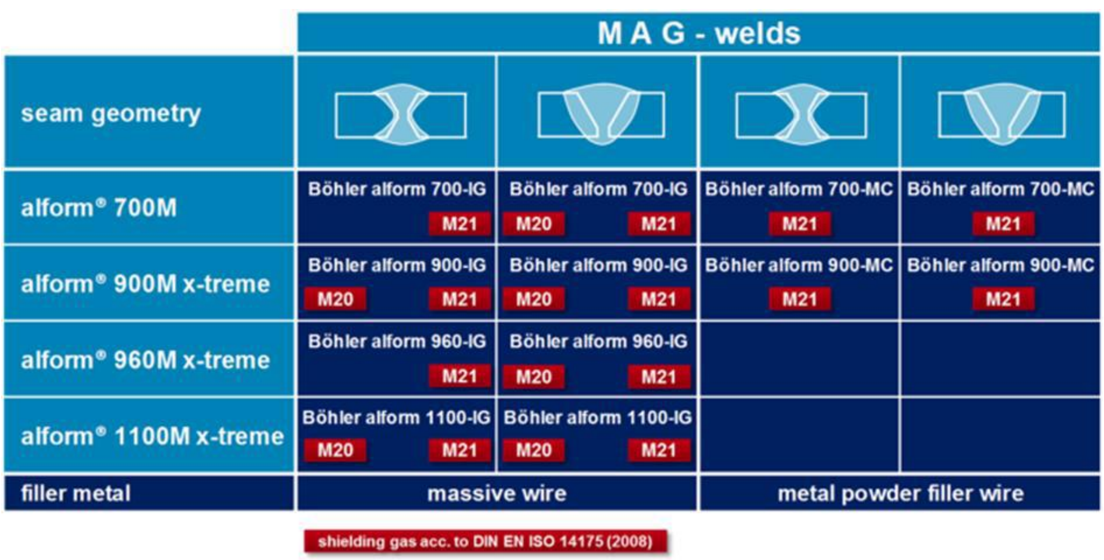
M. Fiedler

The altform welding system consists of a new optimized welding approach implemented by the producers of base metals and the welding consumables. The project involved a close collaboration of both parties, Vooestalpine and Voestalpine Bohler Welding, for the development of combination series of base metal and filler metals regarding high and ultra-high-strength steels with a yield strength ranging from 700 to 1100 MPa. The welding consumable in question include stick electrodes, solid wires, flux and metal cored wires and wire/flux combination.

The main objectives of this combined effort are intended to:

- adapt filler metals to base metal for each type of joining or application;
- decrease of carbon equivalent, as per Graville formula, of the base metal or the diluted weld metal;
- improve the resistance to cold cracking by reducing hardening in the coarse-grained heat affected zone CGHAZ;
- reduce the softening in the weld metal and the sub-critical heat affected zone SCHAZ;
- determine the alloy design of filler metals;
- extend the weldability range, e.g. increase $t_{8/5}$ cooling time with less loss in the weld strength;
- optimize the properties and increase the reliability of welded joints;
- widen the use of HSLA in different application fields;
- serve better the customer and reduce manufacturing difficulties;

An example of a qualified altform welding system for single-V and double-V butt joints in high and ultra-high-strength steels is shown below:



	M A G - welds			
seam geometry				
alform® 700M	Böhler alform 700-IG M21	Böhler alform 700-IG M20 M21	Böhler alform 700-MC M21	Böhler alform 700-MC M21
alform® 900M x-treme	Böhler alform 900-IG M20 M21	Böhler alform 900-IG M20 M21	Böhler alform 900-MC M21	Böhler alform 900-MC M21
alform® 960M x-treme	Böhler alform 960-IG M21	Böhler alform 960-IG M20 M21		
alform® 1100M x-treme	Böhler alform 1100-IG M20 M21	Böhler alform 1100-IG M20 M21		
filler metal	massive wire		metal powder filler wire	

shielding gas acc. to DIN EN ISO 14175 (2008)

Available system solutions qualified as per EN 15614-1

EN ISO 15614-1: Specification and qualification of welding procedures for metallic materials- Welding procedure test. Part 1: Arc and gas welding of steels and arc welding of nickel and nickel alloys

SESSION 10 ENERGY INDUSTRY APPLICATIONS A

High-strength steel application for welded stiffened plate structure of a fixed storage tank roof

Prof Karoly Jarmai.

The study is related to the optimization of the system design for the construction of a welded fixed roof of a vertical tank used for the storage of liquid kerosene. The objective consists of finding out an effective and economic assembling method for the construction of a roof tank using high-strength steel material (YS: 690 MPa). The roof is assembled from different plate elements fixed and stiffened circumferentially by stiffeners in the form of half-rolled I-section, and radially by I-section beams.

The different factors taken into account are: distance between stiffeners, thickness of the base plate, position, number and size of circumferential stiffeners, size of radial beams. The cost include material, welding and painting operations.

In comparison to an optimized system design for a tank roof made of mild steel (235 MPa) with same dimensions and service loads, it was concluded that the use of high-strength steel (YS: 690 MPa) results in considerable saving of 24 and 26% regarding the weight of material and the cost respectively.

Investigation of residual stresses and distortions produced in tubular K-joint

Gerhard Stix

The study relates to the influence of the joint design, welding parameters and welding sequences on residual stresses and distortion in the case of welded tubular K-joints formed by one cord and two braces. Materials used for the K-joint assembly are vanadium-microalloyed steel pipes 20MnV6 (A381) for the chord and low-alloy structural steel S355 tubes for braces. In the practical field, this design applies, inter alia, to lattice boom cranes.

The approach was carried out using FE welding simulation with software Simufact welding to assess distortion and residual stresses followed by a welding with a robot GMAW system (standard and pulsed) to verify the results of the welding simulation. Also, metallographic examination was made to characterize the weld penetration. The main design parameters consist of the following: angle between brace and chord, eccentricity of braces, gap between the braces.

The metallographic investigation shows a more homogenous penetration in welds made with pulsed GMAW than in GMAW standard welds. However, only standard welds are integrated in the simulation. Also, the pulsed-welding did not led to the HAZ hardening as the standard welding did.

The simulation of the standard welds shows that the distortion increases with increasing the gap between braces. This is due to the longer joint which results in more residual stresses. Furthermore, it was found that neither the angle between the brace and the cord nor the chord slenderness have an impact on the residual stress and distortion. It was also pointed out that the measured residual stress and the calculated one show the same trend even with some deviation. On the other hand, the variation of the Vickers hardness and that of the residual stress show a similar tendency.

Design of Triplex Stainless Steel Filler Metal for Ultra High-Strength Steel and Austenite Retention in Multipass Weld Metal

Prof Kazuyoshi Saida

The study deals with the analysis and the characterization of a new welding filler metal for welding ultra-high-strength HT980 steel (YS: 980 MPa). Among the target applications for which the HT980 steel was developed are large-scale penstocks and aqueducts. This research has been motivated by the following considerations:

- Improving the weldability of such high-strength materials while getting high-strength in symbiosis with a good ductility in the welded joint;
- Increasing and broadening their use in different welded structures;
- Dispense with or reducing the need of preheat and PWHT which are inevitable when using high-strength steel filler metals to produce crack-free welded joints

The study was based on the mechanical and microstructural characterization of different filler metals having a triplex microstructure: martensite-delta ferrite-retained austenite. The weld metals were obtained by autogenous GTAW welding of base materials having a typical chemistry of 13.5Cr-Ni-0.5Mo with varying Ni% (6.93- 9.83%) and C% (0.014- 0.049 %) contents in order to optimize the content of the retained austenite. To simulate the practical welding conditions, the oxygen content was approximately of 200 ppm in the weld metal by deliberately using a shielding gas of Ar-10 % O_2 for the autogenous GTAW.

The mechanical and metallurgical characterization were carried out using tensile testing, impact testing at -30 °C, hydrogen-induced cracking test, hot-cracking test, microstructure analysis by optical microscope and electro back-Scattered diffraction (EBSD).

It was demonstrated that:

- the strength significantly increases with the increase of the amount of the martensite phase in the microstructure;
- the tensile strength decreases whereas the impact energy of the weld metal increases with the increase of the retained austenite. If the required impact energy of the filler metal must be higher than 35 J at -30 °C, the amount of retained austenite in the microstructure shall be of at least 15%.
- the risk of hydrogen cracking (HICC) was evaluated by U-groove weld cracking test using PAW process with a shielding gas mixture of Ar-3% H_2. Cracks investigation was done after 72 hours from the completion of welding. It was shown that the risk of cold cracking is negligible with a retained austenite fraction of 20 to 40%, which serves both to dissolve hydrogen and to improve the toughness of the weld metal.
- the hot cracking decreases markedly with a δ-ferrite content of 40-50% in the microstructure. This susceptibility was assessed by the transverse-Varestraint test, under a varied strain from 0.66 to 1.96%, and characterized by the BTR (Brittle Temperature Range). Also, it was confirmed that the lowest tendency to solidification cracking is associated to alloys solidifying with primary ferrite solidification mode (FA).

The retention of austenite is caused by a local microsegregation of alloying elements prior to the martensitic transformation. The amount of the retained austenite is assessed based on the distribution of the martensitic transformation temperatures M_s and M_f in dendritic cells, which in turn is controlled by the distribution of solute elements (C, Cr, Ni, Mn and Mo). Such a transformation progresses then from the cell core towards the cell boundary. The retained austenite starts to form when the starting temperature of the martensitic transformation (M_s) drops below the room temperature. The amount of the retained austenite increases with increasing nickel and carbon contents. The retention of austenite in multipass weld metal is reduced in the first passes due to the effect of dilution. At the end, a great agreement was established between the experimental (measured) and the calculated contents of the retained austenite.

It was concluded that the triplex filler metal which meets the weldability requirements of the ultra-high-strength HT980 steel with the possibility of preheating-free welding must produce a weld metal with a volume fraction of 15 to 35% of retained austenite. To conclude, two triplex stainless steel weld metals are retained to fulfill the requirements listed above. Their chemical composition (wt %) are the following:

1) 0.028% C- 0.31% Si- 0.79% Mn- 8.45% Ni- 13.40% Cr- 0.51% Mo- 0.0056% N- 0.017%Al
2) 0.028% C- 0.27% Si- 0.80% Mn- 9.07% Ni- 13.43% Cr- 0.47% Mo- 0.0057% N- 0.017%Al

The amount of the retained austenite is 14.9 and 34.8% for the 1st and the 2nd alloy respectively. Even though both weld metals are insensitive to hot cracking, the first alloy has the advantage to have a primary ferrite solidification mode (FA) while the second alloy is assigned a primary austenite solidification mode (AF).

SESSION 14 INNOVATIVE JOINING METHODS FOR HIGH-STRENGTH MATERIALS A

Effect of Welding Temperature on Microstructure and Mechanical Properties of Friction Stir Welded Joints of Middle Carbon Steel

Dr Miyano Yasuyuki.

The investigation deals with the friction stir welding FSW of a 1.5 mm plate thickness of carbon steel JIS-SC45 (0.45 %C). The single pass butt-welding tests were carried out at several welding speeds, 100 to 400 mm/mn, with tool rotation speeds of 100 to 400 rpm under the pressure of a WC based material tool. Argon shielding gas was used to prevent oxidation of the plate. The temperature on the plate's top surface and at the center of the joint were monitored using infrared thermography system.

The objective of the study is to understand the effect of welding conditions, namely the temperature, on the microstructure morphology and the mechanical properties of the FSW joint. The microstructure was characterized by optical microscopy OM and a field emission-scanning electron microscopy (FE-SEM) equipped by electron back-scattering diffraction (EBSD). Mechanical properties were evaluated by Vickers hardness test through the transverse cross-section of the weld.

The research study has been motivated by the fact that middle-carbon steels are still largely used in engineering industry and construction due to their low cost. Also, it has been revealed important to look for an alternative and reliable welding technique to conventional fusion welding processes which have several drawbacks for the case of middle and high carbon plain steels: hardening tendency of HAZ, risk of cold cracking, etc.

The results show that the hardness increases near the center of the joint and the highest values are observed close to the top surface in that area for all the welds. Also, the hardness increases proportionally with increasing the rotation speed of the shoulder and the welding travel speed. Besides, it was outlined that the difference between the hardness at the top and at the bottom of the weld surface increases with the increase of the rotation speed.

In fact, it turns out that the increase of the rotation speed induces more heat in the material from the joint surface. A peak temperature above A_3 temperature point was recorded for the welding condition corresponding to the fastest speed rotation of 400 rpm. This condition leads to the martensite transformation upon rapid cooling. This effect is more marked as the travel speed increases since this parameter directly affects the cooling rate of the weld. On the other hand, the lowest temperature measured was below the A_1 transformation temperature, which corresponds to the slowest speed rotation of the tool (100 rpm). In such a condition, no phase transformation does occur.

From this study, it is concluded that the optimized welding conditions are those which lead to a peak temperature below the A_1 transformation temperature in the stir zone to avoid any phase transformation in the weld joint. As a solid state joining process, the friction stir welding offers the possibility of controlling the heat input in the material. The thermal controllability of the process allows to overcome the difficulties met in the conventional welding of middle and high carbon steels.

Dissimilar Welding of High-Strength Steels

Eric Martial Mvola Belinga

The approach aims to develop a basis data for dissimilar metal welding (DMW) of different categories of high-strength steels and to provide suitable welding procedures for satisfactory welded joints required by the design. This theoretical study deals with the case of similar and, more particularly, dissimilar welding of this class of materials. It relies on the compilation and the analysis of research data regarding the weldability and the behavior of these materials under fusion welding conditions.

Different welding processes have been used for welding dissimilar high-strength steels like laser beam welding (LBW), hybrid laser/GMAW welding, resistance spot welding (RSW), GTAW, GMAW, etc. Depending on the process, the welding is carried out with or without filler metal. From different similar and dissimilar welding experiments involving different HSS and using diode laser welding process, it was outlined that the weld microstructure can be predicted from the measured hardness which increases with increasing the carbon equivalent (CE) in the fusion zone. In the case of autogenous welds, the hardness level depends on the fusion level of both materials and thus the involved welding process. In such a case, the diffusion of alloying elements, namely from the low-alloy steel side towards the high alloy side must be taken into account. This is because this welding situation may lead to the softening of the HAZ of the low-alloy side and, at the same time, to a hard microstructure in the joint.

The use of Graville Diagram serves as a helpful indicator to assess the degree of difficulty for welding similar or dissimilar high-strength steels. Thus, it determines the risk of cracking in the welded materials in function of both the carbon content and the carbon equivalent (CE_{IIW}). For joining high-strength steels, whatever dissimilar or dissimilar joints, it is advised to avoid incomplete joint penetration since the root of the joint serves as a stress raiser for cracks propagation.

For arc welding dissimilar HSS, the austenitic grade 307 (Cr-Ni-Mn) was satisfactorily tested among others. The reason for that is to achieve a good toughness in the weld joint while preventing the risk of hydrogen cracking (HICC) in the heat affected zone (HAZ).

The development of matching filler metals is still facing challenges since the performances of HSS are continuously increasing as per the application requirements. For that, it is essential to evaluate the need to apply a matching or an overmatching filler metals while taking into account the direction and the value of stresses in the welded assembly. For dissimilar metal welding, the mismatching between the filler metal and the base metal has been the object of numerous researches. In fact, it appears that high-strength steels accommodate well the undermatchnig weld joints, which consequently result in decreasing the susceptibility to cold cracking, the lamellar tearing and also the distortion. Also, it is to mention that the use of undermatchnig weld joints is less demanding for preheat requirements than matching or overmatching weld joints. In such a case, more time could be saved for the welding operations. In addition, the risk of the toughness drop in the HAZ becomes also lower especially for advanced or ultra- HSS.

The overmatching weld metal is normally applied for full penetration butt welds undergoing transverse tensile loads. For Fillet welded joints such as tee joints, there is no need for matching filler metals and only the use of under-matching filler metals is sufficient for the performances of this type of joints. In general, the matching degree of the filler metal depends on the type of the joint and the service conditions of the weldment.

Regarding the weld joint integrity, it is suggested to select filler metals having an improved ductility even with a slightly lower strength than the base metals. In this context, it was emphasized that nickel, as alloying element in the composition of ferritic welding consumables, plays a beneficial role in determining the weld metal features. In fact, this element promotes the formation of acicular ferrite (AF): a phase which enhances the mechanical properties of the weld microstructure. Others studies mentioned in this article corroborate the fact that the presence of 2-3 %Ni in the ferritic weld metal leads to the formation of acicular ferrite, bainite and martensite in the microstructure.

The welding of high-strength steels requires an optimum control of the heat input and thus the cooling rate $t_{8/5}$ in order to reduce the HAZ extent and to prevent its softening. The welding process has, in turn, an effect on the control of heat input and consequently the weld joint features. For example, laser beam welding (LBW) leads to a narrow HAZ with a predominantly hard martensitic microstructure in the fusion zone.

Because of the sensibility of the HSS microstructure to change under the welding heat conditions, it is recommended to strictly control the interpasse temperature and the heat input in narrow ranges during welding. The preheat decreases the cooling rate and allows the hydrogen to escape from the weld metal and thus reduce the risk of cold cracking. Preheat can also limit the holding time of the metal at high or critical temperatures. The interpass temperature and the heat input are controlled within a narrow range in order to limit the extent of the HAZ and to prevent grain coarsening in this area. The post-weld heat treatment (PWHT) could be recommended in some cases, depending on the chemistry of the base metal and the filler metal. This treatments exerts a tempering effect which improves the toughness of the welded joints.

References

- Eurocode 3: Design of steel structures - Part 1-8: Design of Joints
- ISO EN 16834: Welding consumables -- Wire electrodes, wires, rods and deposits for gas shielded arc welding of high-strength steels – Classification

Improving Weld Strength with HF Post-heating Process in Resistance Spot Welding of Ultra-High-Strength Steel Sheets

Akira Terajima

The research aims to develop an advanced spot welding process (RSW) in order to improve the cross-strength of the welded joints. This deals with the application of a high-frequency (HF/25 kHz) post-heating for the tempering of spot-welded joints (weld nugget and HAZ) in plain carbon steels.

In the as-welded condition, spot welds suffer from a reduction of the fracture strength due to the hardening effect of the weld metal. This phenomenon is more pronounced with increasing the carbon content or the strength of the steel. The use of RSW is increasingly widespread in joining thin, light weight and rigid ultra-high-strength steels (UHSS) sheets in automotive industry. The current demand of this industry consists of providing a frame assembly with an improved fracture strength in the welded joints and thus a better impact resistance.

Tests were carried out on normalized plain steels sheets, 1.2 mm thickness, with different carbon levels, from 0.15 to 0.55% C, which correspond to a carbon equivalent (Ceq*) values of 0.23 to 0.69%. It is mentioned that the post-heating was done under the same applied force as for welding. A comparison is also made between the tempering effect obtained by this high frequency (HF) post-heating and the conventional commercial-frequency mode (CF/ 50 Hz). The high-frequency post-heating, through the skin effect, allows the electrical current to flow under and around the electrode contrary to the conventional frequency mode where the current only flows under the electrode. This advantage permits to realize a direct and uniform tempering of the weld nugget and HAZ.

It is noticed that the microstructure of the spot welds prior to post-heating consists of untemeperd martensite in the nugget and HAZ. The HF post-heating exerts a softening effect on the microstructure with a significant decrease in the hardness of the spot-welds from the as-spot welded condition. Thus, the tempering effect is more pronounced as the electrical power and heating time increase. For the fracture modes, the analysis shows that the fracture mode in the as-spot welded condition consists of interfacial failure or partial plug failure whereas it transforms to a plug failure mode after HF post-heating.

The cross-tension strength (CTS) decreases markedly in the as-spot welded condition for the steel sheets having a carbon content higher than 0.35% wt. The HF post-heating results in increasing the weld strength as compared to that in the as-spot welded condition. This increase becomes also more pronounced as the carbon content exceeds 0.35%. In fact, the strength increases from 1.5 to 2 times and from 3 to 5 times in welds with 0.25% C and ≥ 0.35% respectively. The degree of this increase depends on the heat condition (electric power and time) in each type of steel.

The improvement of the fracture mode and the cross-strength of the welds is due the tempering effect of the HF post-heating around the weld nugget. This combination of spot welding with the HF-post heating can then be considered for the use of sheet steels with a carbon content over 0.35%, which have lower mechanical performances in the as-spot welded condition.

The effect of the tensile strength of the steel on that of the weld was also investigated for the case of the 45C (0.45% C) steel plain sheet that is considered as a difficult-to-weld material. For that, the sheet steel was spot-welded in three conditions of tensile strength, namely 0.77, 1.33 and 1.95 GPa, obtained by normalizing or quenching and tempering before welding.

The results show a drastic increase of the hardness in the weld nugget (~ 720 HV) in the as-spot condition regardless of the initial tensile strength of the sheet steel. A remarkable softening of the microstructure was also noticed in the heat affected zone in the case of the sheet steels hardened to 1.33 and 1.95 GPa.

The HF-post heating produced the following effects: the hardness of the weld nugget drops to around one-half and the steep drop in the hardness disappears in the HAZ. The-cross tensile strength increased by a factor of 3 to 5 times in the HF post-heated condition. Also, all specimens were fractured by plug failure and not by interfacial failure as in the as-spot welded condition.

The simulation results show a good agreement between the measured and simulated results for the HF post-heating of the 0.45% C sheet steel

Notes:

- Ceq calculated as per the formula JIS A5523 - Weldable Hot Rolled Steel Sheet Piles- Ceq= C+Si/24+Mn/6+Ni/40+Cr/5+Mo/4+V/14

SESSION 15 TRANSPORTATION INDUSTRY APPLICATIONS B

Strength and Impact Toughness of High-Strength Steel Weld Metals- Influence of Welding Method, Dilution and Cooling Rate

Prof Leif Karlsson.

The study deals with the effect of the welding method on the weld metal mechanical properties while emphasizing the influence of the dilution and the cooling rate. Test data consists in the following:

- Base metals: two 12 mm high-strength steel plates, namely Weldox 700 and Weldox 1100 with a yield strength of 777 and 1193 MPa respectively.
- Welding processes used: MMA, GMAW (solid and core wire), Rapid Arc (RA) Welding, SAW, Hybrid-Laser/GMAW
- Different filler metals with a minimum yield strength of 810 to 1006 MPa, which overmatch Weldox 700 and undermatch Weldox 1100.
- Preheat: 50 °C (Weldox 700), 75 °C (Weldox 1100).
- Cooling time $t_{8/5}$: 5 and 15 s for each process.
- Type of joint: V-groove butt weld
- Joint angle: 60° (MMA, SAW, GMAW, MCAW), 15° (Rapid Arc), 7° (Laser-Hybrid)
- Backing system: plate of the same grade of the base metal except Laser-Hybrid process (narrow joint).

The different selected filler metals used are considered as matching the base metal Weldox 700 but undermatchnig the Weldox 1100. It is to mention that the chemical analysis of the different welding consumables have lower carbon, slightly higher Cr, Mo and Ni contents than the base metal. Also, it is underlined that the most influencing alloying element in the filler is the nickel, which is not present in the base metals.

At first, it appears that the cooling rate is in part related to the type of the welding process. Laser-Hybrid and Rapid Arc (RA) which produce single-pass welds are associated to high dilution levels: namely 71-73 % and 43-46 % respectively. This is in partially due to the narrowness of the groove joint. In this condition, the-single-weld bead does not undergo a bead tempering like in multi-pass welds. The resulting single-pass weld is thus expected to develop higher mechanical properties than multi-pass welds.

The dispersion of the dilution results is explained by the fact that the calculated dilution does not reflect the actual dilution level for each alloying element in the same weld for the different welding processes.

The cooling rate obtained with different welding processes varies from 5 to 15 s depending upon the type of the welding process and welding parameters. It was concluded that it is impractical to get the same predetermined cooling times of 5 and 15 s in each welding process for both base metals. Also, it is to mention that the chemical analysis of the all-weld metal did not undergo a change with the change of the cooling rate for all welding processes.

The combined results of the weld strength (YS & UTS) for both base metals show an overall increasing trend, with scattered values, as the carbon equivalent (Pcm) of the weld metal increases. The use of this parameter is a useful indicator allowing to take into account the effect of dilution on the mechanical properties of the weld metal for any welding process.

It is outlined that the tensile strength associated either to weld metals undergoing high cooling rates or single-pass welds show high values versus other weld metals. Indeed, a rapid cooling promotes the formation of hard microstructural constituents such as martensite in the weld metal.

Impact tests outline an apparent decreasing trend of impact energy with increasing tensile strength. In the same way, the results highlight that the dilution has also a negative effect on the toughness of the weld metal.

For the transverse tensile strength, the fracture occurs in the base metal for the Weldox 700 and in the soft region in the case of Weldox 1100. All welding consumables have lower yield strength levels than that of Weldox 1100 (1100 MPa). However MMA and Laser-Hybrid processes, both used with low heat input with $t_{8/5}$ of 5 s, and GMAW-Rapid Arc resulting in $t_{8/5}$ of 12 s produced weld metals with transverse tensile strengths that only reach the nominal yield strength of Weldox 1100.

The single-pass GMAW-Rapid Arc is well suited for welding high-strength steels since it produces weld metal with good tensile properties combined with a high impact toughness. However, the Laser-Hybrid, even though it gives an acceptable weld strength, leads to a poor impact toughness in the weld metal.

From the study results, it appears possible to use undermatchnig filler metals or electrodes for welding ultra-high-strength steels since there is no matching welding consumable for them. However, this possibility must be limited by the design conditions.

Remark

- Table 5: there may be an error related to the type of the electrode OK 73.15. This electrode is like E8018-G (AWS A5.5) with a YS and UTS of 460 MPa and 550 MPa minimum instead 996 and 1128 MPa as indicated in the table 5.

Notes:

- Rapid Arc (RA) Welding is a high travel speed GMAW-P process with short-arc length
- Pcm: Critical metal parameter adopted as Ceq formula by JWES (The Japanese Welding Engineering Society) to assess hydrogen cracking risk for low carbon alloy steels.
    - *Pcm = C + Si/30 + Mn/20 + Cu/20 + Ni/60 + Cr/20 + Mo/15 +V/10 + 5B (wt%)
- MMA: OK 75.78 & OK 73.15, GMAW (Solid wire, Laser-Hybrid, Rapid-Arc RA): OK AristoRod 79 &89, GMAW (cored wire): Coreweld 89 and H184, SAW: OK Flux 10.63/DA090 & OK Flux 10.63/DA091.

Hydrogen Management in High-Strength Steel Welds

Prof Huijun Li.

The study deals with the analysis and the prediction of the susceptibility to hydrogen-assisted cold cracking "HACC" of the weld metal in high-strength-steels. This HACC issue is generally less pronounced in HAZ than in the weld metal in high strength steels. This is because of the low carbon content of this type of base metal. As known, the responsible factors of HAZ HACC consist of the combination of a hard microstructure, tension stress and diffusible hydrogen.

In conventional high carbon steels, the austenite-to-ferrite phase transformation in HAZ occurs at relatively low temperatures, i.e. after the weld metal transformation. The weld metal having low carbon content transforms first as carbon lowers the phase transformation temperature austenite-ferrite. The insoluble hydrogen rejected from the transformed weld metal diffuses towards the HAZ still in the austenitic state. However, for modern low alloy steels, the base metal generally has lower carbon content compared with the weld metal. Thus, the weld metal transforms last after HAZ and becomes a sink for the hydrogen generated during the welding operation.

This study approach focalizes on the distribution of hydrogen in the weld metal for different combinations of welding. The different filler metals used consist of austenitic and ferritic wires, namely AWS A5.22 E308MoT1 & EC308Mo, A5.9 ER307, A5.20 E71T-1. The base metal is a micro-alloyed high-strength ferritic steel, 0.35% C (0.72% CE_{IIW}), having a low austenite-to-ferrite temperature transformation. The welding parameters are kept constant for all welding combinations and experiments. The preheat temperatures are 7, 23 and 80 °C to give rise to average cooling rates $t_{8/5}$ of 32, 31 and 24 °C/s respectively. Following the CCT of this type of steel, these critical cooling rates correspond to the higher levels of its hardenability.

Welded samples are quenched into cold iced water immediately after completion of welding and then transferred into a bath of liquid nitrogen to prevent any hydrogen diffusion from taking place in the weldment at cryogenic temperatures. Samples are kept in this condition prior to being transferred into Y-tube filled with mercury to measure the diffusible hydrogen by collection at room temperature in accordance with the Australian Standard AS/NZ 3752: 2006. For that, test samples are kept 7 days to allow the majority of hydrogen to diffuse before measuring the diffused volume of hydrogen. Regarding the retained or residual hydrogen, the control testing was conducted after the completion of the diffusible hydrogen testing. Retained hydrogen was determined by fusion using ELTRA's ONH-2000 analyzer which consists of an inert gas melt extraction method according to Australian Standard AS/NZ 1050.

The austenite-to-martensite transformation of the investigated base metal starts at roughly 410 °C and ends at 205 °C. The hydrogen starts to diffuse from high temperatures to up to 100 °C and HACC usually occurs when the temperature of the weld starts to drop below 200 °C, after the end of the austenite transformation temperature (M_f).

From the results, it is confirmed that the diffusible hydrogen is negligible (1.3 ml/100 gr-WM) in austenitic weld metals. However, these ones contain high residual (retained) hydrogen (5.8 ml/100 gr-WM), which is liberated and determined by fusion method. On the other side, the ferritic weld related to the E71T-1 wire retained just a very small hydrogen content (0.9 ml/100 gr -WM) while diffusing a high amount of hydrogen (23 ml/100 gr-WM). However, this high content of diffusible hydrogen is abnormal, which is presumably due to a contamination of the weld pool.

To ensure that the diffusivity of hydrogen in austenitic metals and alloys is negligible if not nil, another welding test was carried out, using an austenitic filler metal and an austenitic base metal. The result indicates that the diffusible hydrogen is nil and the hydrogen is completely retained in the austenitic weld (8.0 ml/100 gr WM).

The comparison between the diffusible hydrogen levels of the different filler metals (rutile flux cored, metal cored and solid wire) is made by depositing each of them onto a ferritic base metal. The results indicate that the rutile wire E308MoT1 produces more diffusible hydrogen, namely 0.8 ml/100 gr-WM, followed by the metal cored wire EC308Mo and then the solid wire ER307 with 0.5 and 0 ml/100 gr-WM respectively. This difference is explained by the hydroscopic nature of the rutile flux and the little amount of flux enclosed in the metal cored wire.

The effect of preheat temperature, from 7 to 80 °C, of the ferritic base metal on the hydrogen distribution was investigated. For that, the E308MoT1 weld metal was quenched into ice brine when it reached the temperature of 150 °C upon cooling. This temperature was chosen because it is below the end transformation temperature (M_f) of the HAZ. Thus, the objective is to understand how the diffused hydrogen in HAZ will be acting. The results show a decrease in the diffusible hydrogen level in the HAZ, from 1.3 to 0.55 ml/100 gr–WM, and an increase of the retained hydrogen in the weld metal, from 6.1 to 8.0 ml/100 gr-WM, when the cooling time increases before quenching. In fact, this increase in the retained hydrogen in the weld metal is therefore due to the back diffusion of the diffused hydrogen from the transformed HAZ to the weld metal. Also, a decrease in cooling time to reach 150 °C caused by lower preheat temperature results in lower levels of retained hydrogen in the weld metal.

It is concluded that the use of austenitic filler metals is beneficial for improving resistance to hydrogen-assisted cracking in high-strength steels in both the HAZ and the weld metal. This is due to the high capability of austenite phase to dissolve or to retain hydrogen. Ferritic weld metals that transform after the HAZ causes the hydrogen to diffuse back from the HAZ to the weld metal. This moves the risk of hydrogen cold cracking from HAZ to the weld metal.

Ref.

Australian Standard AS/NZ 3752: 2006/ Determination of hydrogen content in ferritic steel arc weld metal

SESSION 19 ENERGY INDUSTRY APPLICATIONS B

Effect of pre-heat temperature and alloy design on the structure and properties of self-shielded arc welded hardfacing deposits

Prof. Druce. Dunne

The study deals with the effect of the preheat temperature on the cooling rate and the microstructure of the self-shielded arc weld deposit intended for metal-to-metal wear applications. It deals also with the design and the optimization of the chemical analysis of the gasless wire, especially in terms of aluminum content that may prevent a full martensitic transformation from taking place in the metal deposit upon cooling.

In fact, the deliberate addition of aluminum in the self-shielded welding wire aims essentially to kill and control the excess of nitrogen and oxygen in the weld pool and thus prevents its atmospheric contamination or the porosity formation. However, an excessive residual content of this solute element in the molten pool, e.g above 1.0% Al, is considered as detrimental to the microstructure of the weld metal. This is because aluminum is a strong ferrite former which promotes delta-ferrite phase at the expense of austenite at high temperatures.

The investigated low alloy steel wire, Alloy 14 as per AS/NZS 2576 classification (~ AWS A5.13 EFe3 / Fe5), has the following chemistry (wt.%): 0.18% C- 1.37% Mn- 0.63% Si- 3.08% Cr- 0.41% Mo- 1.18% Al- 0.07% N- 0.0038 O. Its CE_{IIW} is 1.10 %. For information, the aluminum content of the ferritic gasless wire E70T-4 is of 1.7 wt%. The undiluted self-shielded weld metal consists of 5 layers deposited by FCAW process with preheat temperatures of 100, 200, 300 and 400 °C, resulting in different cooling rates $t_{8/5}$. The estimated Ms point of the weld deposit is 385 °C. Dilatometry was used to simulate the thermal welding cycle of the solidified weld metal subjected to reheating from subsequent welds. EDS was also used to examine the residual aluminum content in the weld metal. Furthermore, the solidified weld metal was reheated to 1100 and 1300 °C to simulate the grain refined and the grain coarsened regions (GRHAZ, GCHAZ) of the heat affected zone.

The microstructure of the deposited weld metal corresponding to a preheat of up to 300 °C consists mainly of tempered martensite/bainite and also residual delta-ferrite. The high preheat to 420 °C, which is above the M_S temperature, leads to a partial transformation of austenite to bainite during the isothermal holding time before the residual fraction of austenite transforms to martensite upon continuous cooling to room temperature. Despite the slow cooling related to this high preheat temperature, the same type of microstructure was revealed but with a slightly lower hardness level than the maximum values related to lower preheat temperatures.

The analysis of aluminum by EDS reveals concentrations of 0.9 and 1.3 % in the transformed austenite (bainite/martensite) and delta-ferrite respectively. Thus, for a volume fraction of 25 % of delta-ferrite phase in the structure, the average of the residual aluminum content in the weld metal is then 1.0% compared to 1.18 % in the wire alloy. It follows that the molten weld metal undergoes a cooling through the austenite/δ-ferrite field which then results in a final microstructure comprised of martensite and/or bainite with certain amount of residual δ-ferrite.

The increase in preheat temperature causes a decrease in the cooling rate. The simulation results reveal that the decrease in cooling rate increases the delta-ferrite in the final microstructure. This is because there is more time spent at high temperatures for the reversion of austenite to ferrite. Moreover, the highest cooling rate corresponds to the highest fraction of transformed austenite and then to the lowest delta-ferrite in the microstructure.

It arises also that the M_s temperature increases as the cooling rate increases. This leads to an increase in the amount of martensite at the expense of delta-ferrite as a consequence of the reduction in the time transformation of the austenite to delta ferrite at high temperatures. The increase in M_s temperature can also be caused by some partitioning of carbon between austenite and delta-ferrite.

Regarding the hardness tendency, the results of the dilatometer simulation show that there is a good correlation of the hardness increase with the increase of both the martensite fraction and the corresponding decrease in delta-ferrite in the microstructure.

Ref.

AS/NZS 2576 Welding consumables for build-up and wear resistance, equivalent AWS A5.13

Fatigue Strength of Ferritic Stainless Steel Rectangular Hollow Section T-Joints As-welded and Post Treated Conditions.

Niko. Tuominen

The study highlights the effect of post-weld treatment on the fatigue strength of fillet welded T-joints between two rectangular hollow section (RHS) members, 2 mm wall thickness, made of ferritic stainless steel type 409 (1.4003 EN 10088-2).

The fatigue endurance strength was assessed in the as-welded and post-treated conditions. Post-welding treatment consists of two conditions, namely TIG dressing and high frequency mechanical impact (HFMI). The latter is carried out by two means: ultrasound (UP) and Air Pressure (HiFIT): High Frequency Impact Treatment). Fatigue tests were performed in reverse plane bending mode (R= -1) with 105 to 106 cycles. The residual stresses were determined by X-ray diffractometer near the weld, namely at the corner and the middle of the side walls on chord and brace. Measures were taken with a pitch of 1 mm from the weld toe line in all welding conditions.

From the results, it turns out that the geometry of the weld toe angle is the critical point that affects the fatigue strength of the welded joints in the as-welded condition. Moreover, high fatigue performances are attributed to welded joints presenting lower weld toe angles. Also, the manual welding, contrary to automatic or robotic welding, leads to create stop/starts points often at the corners, which serve as fatigue crack initiating points.

Residual stresses decrease slightly in the as-welded condition from the weld toe line towards the base metals (chord or brace). In the post-treated condition, UP and HFMI, residual stresses are completely transformed into compressive stresses at and in the vicinity of the weld toe, which improves the fatigue behavior of the welded joints.

The test results show that the post-treated welded joints by TIG dressing or by high frequency mechanical impact (HFMI) have up to three times better fatigue life than as-welded joints. Besides, TIG dressed joints show more stable results with lower variations than HFMI.

Note:

- HiFIT: High Frequency Impact Treatment is a high frequency peening process functioning with air pressure.