

WELDING OF ULTRA HIGH-STRENGTH STEEL (UHSS) WITH AUSTENITIC METAL POWDER WIRE (MPW)

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Abstract

The aim of this work was to evaluate the automatic welding of ultra high strength steel (UHSS) with austenitic metal powder wire (cored steel wire and metal powder inside with similar composition). Samples were welded using Automatic Shielded Metal Arc Welding (SMAW) process and two austenitic powders in protected (Ar-CO₂) atmosphere. Percentage of penetration/dilution in the base material, hardness measurements of both welded joints and welded deposits (in as-welded (AW) and post welding heat treatment (PWHT) conditions), influence of interpass temperature on the mechanical properties of weld deposits and metallographic analysis of welded joints and pure filler material were performed. Hardness measurements of fillet welds detected softened and hardened zone. Martensitic layer was found on dissimilar boundary ferrite/austenite; cracks were not detected in this region. Microstructure of weld deposit was austenitic with some amount of δ -ferrite.

Key words: UHSS, metal powder wire, dissimilar welding, microstructure, martensitic layer

Introduction

Production of welded structures in the shortest time is the basic task for any manufacturer essentially of economical reasons. Therefore, welding companies try to use mainly automatic welding processes and newly developed filler materials. On the other hand, the integrity and reliability of welded structure must be a primary task for all welding companies, because destroying of expensive and responsible welded structures means lost of money and in the most cases great number of human's victims and environmental pollution. Welding of newly developed ultra high strength steel (UHSS)

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is not so easy task. The problem is very complex, because many factors influence the quality of welded joint. The main problems are usually related to hydrogen induced cracks in a heat affected zone (HAZ) and hot cracks in a weld metal [1]. Automatic welding of UHS steel with austenitic metal powder wire (cored steel wire and metal powder inside the core has the same composition) is the process to be used. Benefits are expected using this type of welding process and filler material, are as follows: (i) Automatic Shielded Metal Arc Welding (GMAW) process enable faster welding production compared with other arc processes, (ii) metal powder wire, compared with solid wire and flux cored wire has ability to enable welding at lower heat settings, higher deposition rates at same current levels. Besides, using austenitic metal powder enable avoiding preheating which is not the case with low alloyed filler material. [1,2].

The development of modern UHS steels meets the pursuit towards the reduction of plate thicknesses used in order to cut costs of material and processing and to possible increase of the loading capacity. Compared with conventional steel plate grades, the ultrahigh strength material is able to offer at least the same performance but with significantly reduced weight. Thanks to sophisticated production methods, UHS steel plates offer uniform high quality with minimal residual stress. Generally these steel should have good weldability [3]. However, welded joints should always be regarded as sources of potential problems in any welded construction. Such difficulties especially could arise during welding of the newest class, UHS steels with yield strength higher than 1400 MPa mainly because of higher percentage of carbon and alloying elements, which lead to higher susceptibility towards martensite formation. During welding of UHS steels the utilization of conditions resulting in very slow cooling rates of the weld region must be avoided. Namely, in this case the obtained strength properties no longer correspond to those of the parent metal. Therefore, heat input must be restricted, and high heat input of welding process must be avoided. Excessively rapid cooling of the weld bed has detrimental effects on the joint deformation behaviour and also increases the risk of cold cracking [4,5].

According with formerly mentioned, the basic task in this research study is to perform sound welding of UHSS with austenitic metal powder wire as filler material. It means that appearing of cold cracks in the coarse grained heat affected zone and hot cracks in the weld metal must be avoided. Proper dilution of filler material in base mater must be obtained too.

Material and experimental

Experimental plates from UHSS with thickness 20 mm were prepared for automatic GMAW process. Mechanical properties of the experimental plates were as follows: $R_e = 1400$ MPa, $R_m = 1720$ MPa, $A_5 = 8-12\%$ [6]. Preparation for welding was automatic (executed with low tolerances). Fillet welding groove was prepared and welding was performed in PA welding position. Two types of austenitic metal powder wires with diameter of 1.2 mm, and with small difference in composition were used in this experiment. Chemical compositions of base and filler materials are given in Table 1. Fronius MIG/MAG welding machine was used for welding (pulse current, electrode (+) pole). As a shielded gas a mixture of Ar and CO_2 was used.

Table 1. Chemical composition of base metal and filler materials metal powder wire (MPW1 and MPW2)

Chem. comp.	Bas.mat	MPW1	MPW2
C	0.123	0.099	0.103
Si	0.261	0.62	0.55
Mn	0.863	jun.32	jun.21
P	0.008	0.017	0.014
S	0.000	0.016	0.011
Cr	0.518	18.86	18.75
Mo	0.324	0.03	<0.02
Ni	0.925	avg.19	sep.19
Nb	0.005	0.006	<0.005
N	-	0.013	0.024
V	0.026	0.08	0.05
Cu	0.197	0.04	0.02
Co	0.014	<0.05	<0.05

Based on previous requirements, first step in this experiment was to optimize welding parameters (welding current and arc voltage dependent on welding speed) for both welding wires [7]. It was possible to draw optimal welding rectangular, Figure 1. After preparation, experimental plates were welded together with welded deposits.

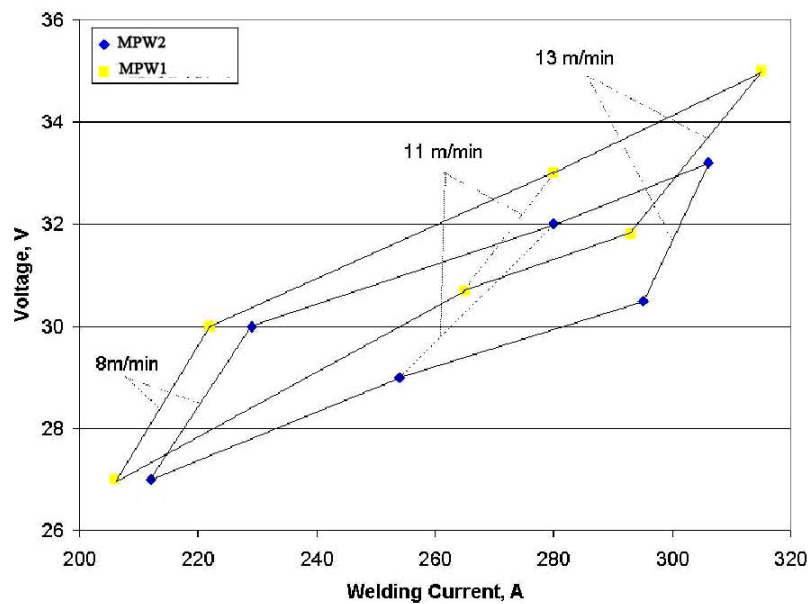


Figure 1 Optimisation of welding parameters for MPW1 and MPW2

After the welding it was possible to evaluate:

- percentage of penetration/dilution in base material (requirement is at least 25% penetration of weld metal in base material),

- hardness measurement of entire welded joints according to EN 287 and EN 288 for fillet welds, and hardness measurement of prepared welded deposits (pure filler material) AW and PWHT condition,
- analysis of mechanical properties of pure filler metals (samples with multilayers deposition were prepared),
- the influence of interpass temperature on the mechanical properties of weld deposits (MPW1),
- metallographic analysis of welded joints and pure filler material (to detect existence of soft and hardened zone in fillet weld and α -ferrite in welded deposits).

Standard metallographic preparation of specimens was performed. After automatic grinding and polishing chemical etching by immersion with Adler reagent (50 ml cold-saturated (in distilled water) sodium thiosulfate solution and 1 g potassium metabisulfite) was done.

For metallographic analysis was used "Reichert Jung mef3" optical microscope. Quantitative metallography was performed on macro specimens in order to determine percentage of dilution of filler metal into the base metal (fillet welds). Software "Axio Vision" was used for this investigation.

For microhardness testing was used Microhardness Tester MHT 4, (test load 5 g, indentation time 18 s). Hardness measurements were performed to fillet welds and weld deposits along the horizontal and vertical centerline. Weld deposits were investigated in the as welded AW and PWHT condition, 800 °C/6h. Vickers method (HV10) was implemented for all hardness measurements (test load 10 kg, indentation time 20 s).

Tensile testing was deduced to determine influence of interpass temperature to the mechanical properties of pure weld metal. From the deposited material round tensile specimens were machined. Dimensions of specimen and loading rate are in accordance with EN 10002-1 standard. "ZwickRoell RMC100" tensile machine was used for this experiment.

Results and discussion

Determination of percentage of penetration/dilution in base material

Determination of percentage of penetration was performed on metallographic specimens (macro) from fillet joints, Figure 2. A method of counting a number of pixels in area of molten parent plate (s) and in total area (S) of molten pool was used for quantitative determination percentage of penetration. Percent of penetration of weld deposit was determined using equation (1):

$$\% \text{ of penetration} = \frac{s}{s + S} \quad (1)$$

where:

s – area of molten parent plate

s+S – total area of molten pool

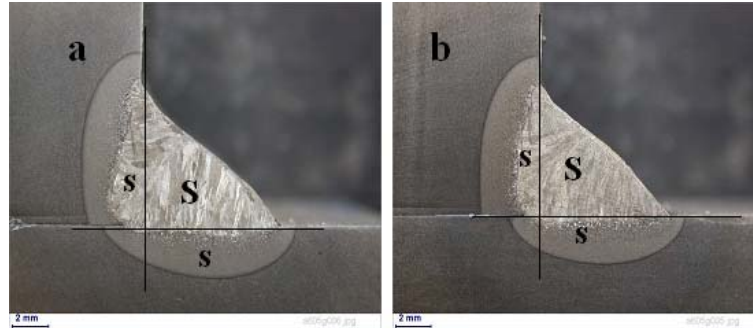


Figure 2 a, b Macrographs of fillet welds a) welding with MPW1; b) welding with MPW2

The obtained results from these measurements are presented in Table 2. It can be seen that both welds have good penetration, i.e. over required value of 25%. Besides, there is no significant difference in obtained values, slightly better penetration was obtained for welding with MPW1. From the measurements and macrographs it can be pointed out that although welding was performed under the equal conditions there is more deposited filler material in weld produced with MPW1, approximately about 13% (bigger area of molten parent metal). Note that the dendritic microstructure is coarser in weld obtained with MPW1.

Table 2 Penetration of weld deposit in base metal, %

	MPW1	MPW2
s	28429	22116
s+S	95451	81565
% of penetration	29,78%	27,11%

Hardness measurements

Measured hardness values can be significant predictor of the welded joints properties. Hardness measurements in this experiment were performed for investigations of fillet welded joints themselves and pure weld deposits. Welded joints (fillet welds) were created with MPW1 and MW2 in as-welded (AW) conditions. Measurements of the fillet welds were done according to EN 287 and EN 288 standards [8,9]. Results of hardness measurements of welded joint formed with MPW1 are presented in Figure 3. A significant difference between hardness values in different regions of welded joints is present, i.e. in austenitic weld metal, base material and HAZ. The highest hardness values were obtained in coarse grain HAZ near the fusion line (HV10 = 543). Such high hardness value is the result of the maximum peak temperature during heating and very high cooling rate. The lowest hardness values were detected in the weld metals. This is the crucial problem during dissimilar welding of ferritic UHSS base metal and austenitic filler material. In practice, lower mechanical properties of austenitic weld can be compensated with the proper design of welded structure, and the use of duplex filler material may be regarded as a possibility to improve mechanical properties.

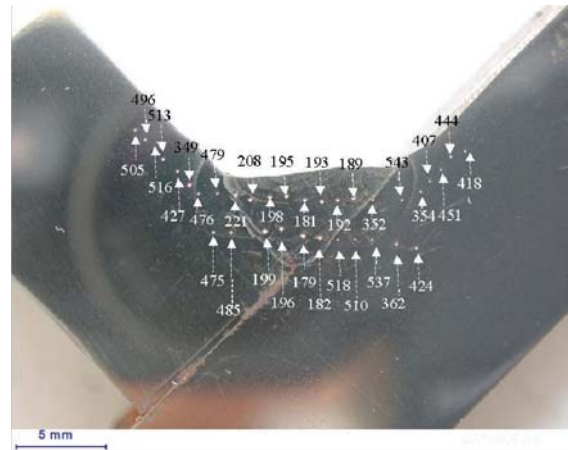


Figure 3 Hardness measurements of fillet welds (welding with MPW1)

Softened zone was found in the base material below A_1 temperature. This zone is the result of degradation of properties of base material which is heated over $200\text{ }^{\circ}\text{C}$ [7] for this steel grade. Actually, the presence of hardened and softened zones in HAZ can't be avoided but their width must be kept as thinner as possible with proper selection of welding parameters.

There is almost no difference in hardness between two filler welds (maximum 208 HV10 for MPW1 and maximum 212 HV10 for MPW2). Maximum measured hardness value in the base material is 528 HV10. It is assumed that this high hardness value is due to higher carbon content and higher and cooling rates.

Hardness measurement of weld deposits was also performed with both filler material. Deposits from MPW1 were investigated in the AW and PWHT condition and deposit from metal powder wire MPW2 only in AW condition. Hardness measurements of weld deposits with MPW1 in horizontal and vertical direction (AW condition) are presented in Figure 4.



(a)

(b)

Figure 4 Hardness measurements of weld deposits along horizontal line (a) and vertical line (b) - welding with MPW1, AW condition

Horizontal measuring line was 2mm beneath surface and the vertical line was parallel to the vertical axis of weldment. Highest hardness values of welded deposits are very similar (202 HV10MPW1 and 205 HV10 for MPW2). Also, measurement direction does not have any significant influence. Obtained values were very close to those of the weld metal (fillet welds). Comparison of hardness values between MPW1 welded deposits in AW condition and PWHT showed that PWHT contributes to lowering of hardness values from maximum 202 HV10 in AW to 180 HV10 in PWHT condition.

Influence of interpass temperature to the mechanical properties of weld deposits MPW1

In order to establish the influence of interpass temperature on the mechanical properties of filler metal deposit (MPW1), three samples with interpass temperature of 50, 150 and 250 °C were prepared by manual deposition. Cross sections of the deposited samples with interpass temperatures of 50 °C and 250 °C are shown in Figure 5. Tensile specimens were machined from these samples. Results of mechanical properties of the weld deposits are shown in Table 3 and Figure 6.

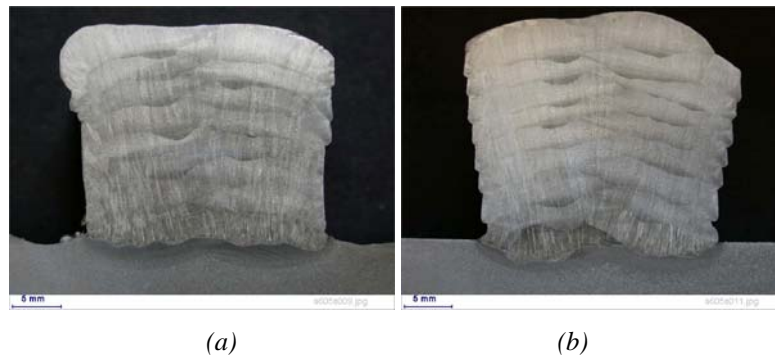


Figure 5 Cross section of filler metal deposit MPW1: interpass temperature (a) 50 °C, (b) 250 °C

Table 3 Results from mechanical tests of weld deposits

Weld Deposit MPW1, Welding Current 260A	Interpass temperature		
	Probe 01: 50°C	Probe 02: 150°C	Probe 03: 250°C
Apparent Yield Strength $R_{p0.2}$ [MPa]	376	343	323
Tensile Strength R_m [MPa]	623	615	618
Fracture elongation A5 [%]	44,0	47,2	49,3
Reduction of area RoA [%]*	54,0	59,9	58,8
No. of Hook cracks	0	0	0
E-Modulus [GPa]	137,2	150,0	143,6
* planimetric evaluation (elliptical necking)			

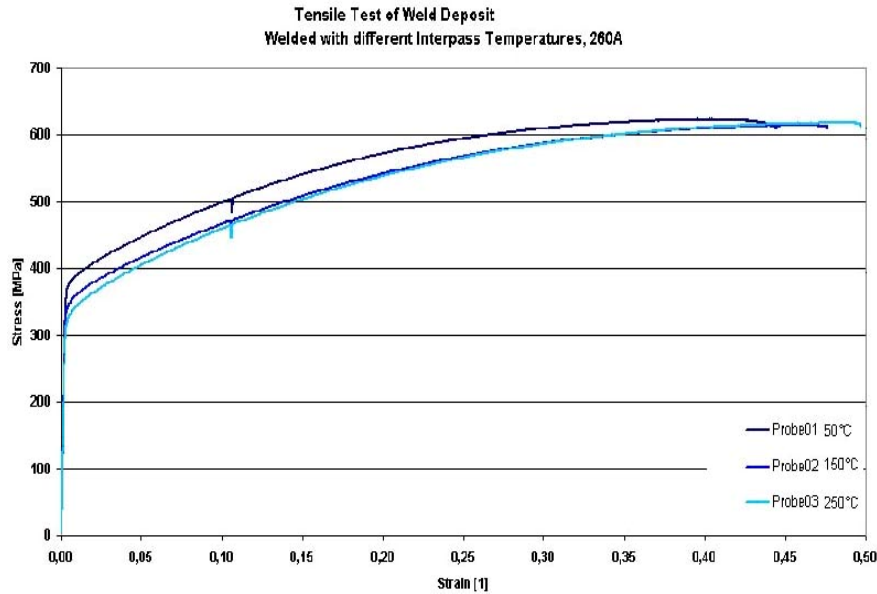


Fig 6 Stress/strain diagrams of weld deposits; interpass temperature 50, 150 and 250 °C

As can be seen from Table 3 and stress-strain diagram (Figure 6) mechanical properties of different weld deposits show similar values. The highest values for yield and tensile strength were obtained for specimen with interpass temperature of 50 °C (specimen 1). This behaviour is attributed to higher cooling rate in this case. Concerning higher mechanical properties, this specimen has lower percentage of elongation and reduction of area. From the Table 3 it can be seen that specimens 2 and 3 have very close values of mechanical properties, especially tensile strength.

Cross-sections of broken tensile specimens 1 and 3 are shown in Figure 7. There is no evident difference between fractured surfaces, but traces of deposited layers can be recognized on the both surfaces. In all cases, Hook or other type of cracks on fractured tensile specimens were not found.

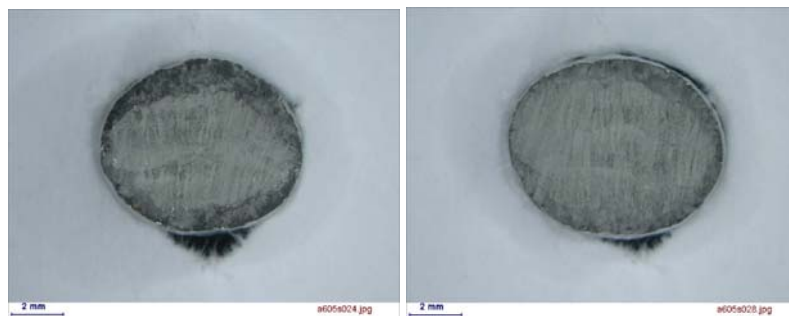


Figure 7 Cross sections of tensile specimens 1 (interpass temperature 50 °C) and 3 (interpass temperature 250 °C)

Metallography

Metallographic evaluation was performed in order to reveal microstructure of welded joints (fillet welds) and pure filler materials (weld deposits). Metallographic evaluation of fillet welds is related to base metal, HAZ and weld metals. Figure 8 shows microstructures obtained in welded joints with MPW1. As can be seen from Figure 8a, microstructure of the base metal is tempered martensite, enabling high strength and rather good impact toughness. It is clear from the Figure (8b and c) that there is a significant difference in microstructure between coarse grained and fine grained HAZ. Such difference is the result of heating to different peak temperature during the arc welding process. Fine grained HAZ, which is formed due to heating of the base metal to the peak temperature of about 900 °C, is supposed to give good mechanical properties especially impact toughness. In the coarse grained HAZ, coarse primary austenitic grains can be seen and coarse martensitic needles inside the grains were formed as a result of very high peak temperatures and cooling rates. Coarse grained HAZ is treated as the most unfavourable part of HAZ, because of possibility of cold cracks appearance.

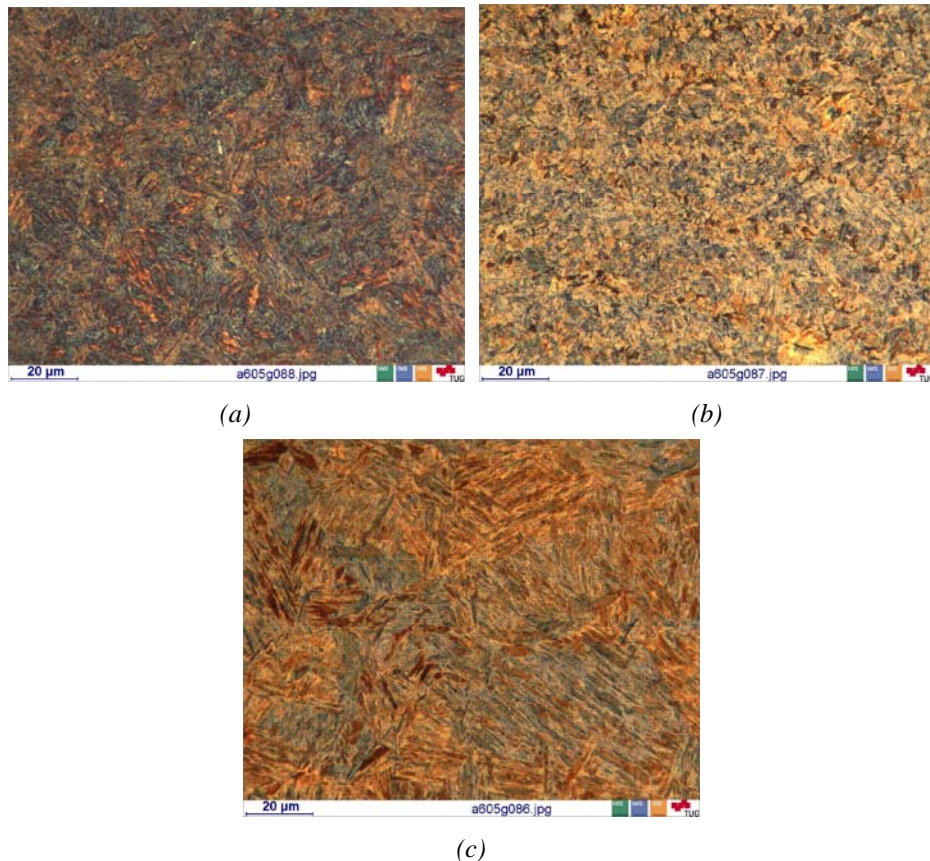


Figure 8 Microstructure base material and HAZ. (a) Base material; (b) fine grained HAZ; (c) coarse grained HAZ

On the boundary between weld metal and coarse grained HAZ (fusion line) a martensitic layer was found (denoted with black arrows on Figure 9), a usual feature in dissimilar weld joints (austenitic and ferritic). As expected, martensitic layer was found in both fillet welds and, as can be seen from the micrograph, this martensitic layer has different thickness. Many authors treated martensitic layer as the location where cold cracks appear, but as can be seen from Figures 9 and 10 no crack can be seen in this region. Formation of martensitic layer is usually at the temperature between solidus and liquidus, i.e. at the temperature of incomplete melting of the base material. Some authors connect existence of cracks with type II boundaries, but in this case such type of boundaries was not detected and growth of dendrites is perpendicular to martensitic layer as can be seen in Figure 9b [10-14].

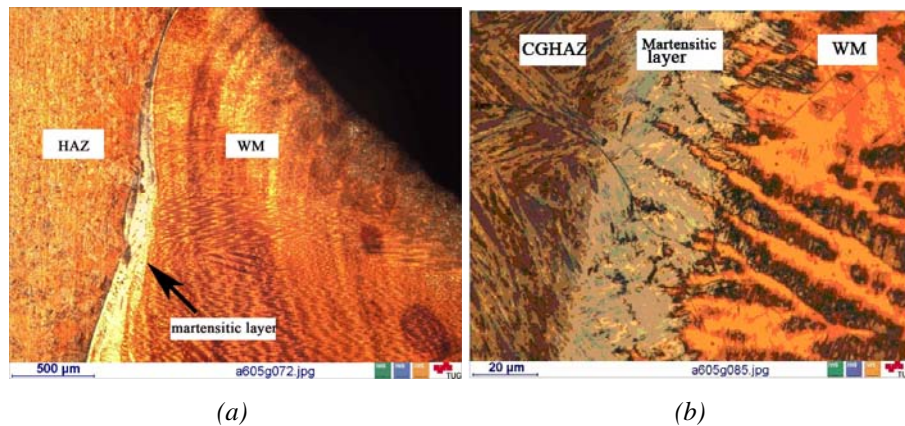


Figure 9 Boundary between weld metal and coarse grained HAZ. (a) Macrostructure of fillet weld; (b) microstructure near the fusion boundary

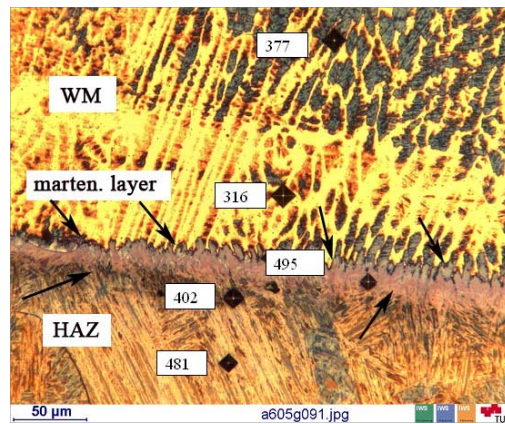


Figure 10 Microstructure and microhardness at dissimilar boundary

Detailed microstructural analysis and microhardness measurement were performed on dissimilar boundary. From measurements of microhardness (Figure 10) of martensitic layer and its neighborhood a rather big difference in hardness values was

found. For example, hardness value of martensitic layer is 495 HV, and beneath and over it measured values are 402 HV10 and 316 HV10. A number of factors support the existence of this transition region such as: different crystal structure between the ferritic base metal and the austenitic weld metal, diffusional mixing of alloying and impurity elements from the weld metal into a stagnant boundary layer adjacent to the fusion boundary resulting in the base metal dilution [10-14].

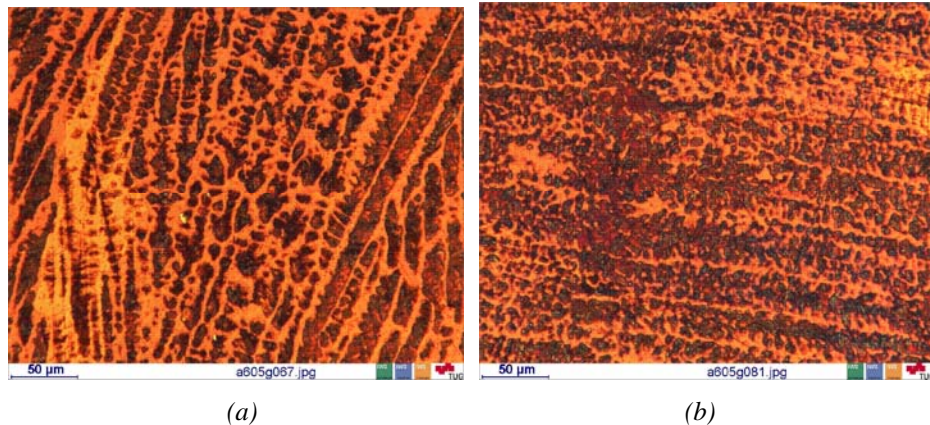


Figure 11 Microstructure of weld metal. (a) MPW1; (b) MPW2

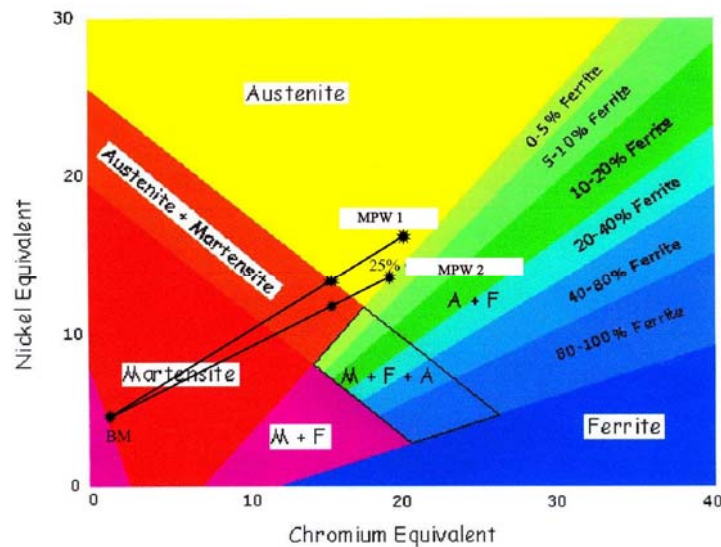


Figure 12. Schaeffler diagram

Microstructure of both fillet welds is dendritic and very heterogeneous, Figure 11. Hardness values of weld metal did not correspond to existence of martensite in weld metal in spite of martensite presence in microphotographs. This was confirmed by relatively low hardness values in weld metal, not higher than 208 HV10 MPW1 and 212 HV10 for MPW2. As an explanation may be used a fact that metastable austenite

transforms into martensite induced by deformation as a result of metallographic preparation of specimens (grinding). It should be pointed out that this transformation occurred only on the specimen surface. From Schaeffler diagram (Figure 12) it can be seen that in the case of 25% dilution, microstructure is near to the boundary austenite-austenite/martensite.

From the microstructural analysis of pure filler materials - deposits (Figures 13a and 14a) it could be concluded that in both cases primary type of crystallization is ferrite-austenite (FA). It means that microstructure contains primary δ -ferrite. Quantitative metallography showed that the amount of primary ferrite (black arrows on figures) in MPW1 deposit is 13.6%, and in MPW2 is 16.4%. It is obvious that there is no significant difference in content of δ -ferrite between two weld deposits. But in both cases measured values are higher compared with calculated values from WRC 92 [10] diagram for standard analysis (Figures 13b and 14b). This diagram provides information on the welding properties of the various types of microstructure, as a function of alloying elements they contain. The WRC diagram is today accepted as an improved version of the Shaeffler or the De-long diagram.

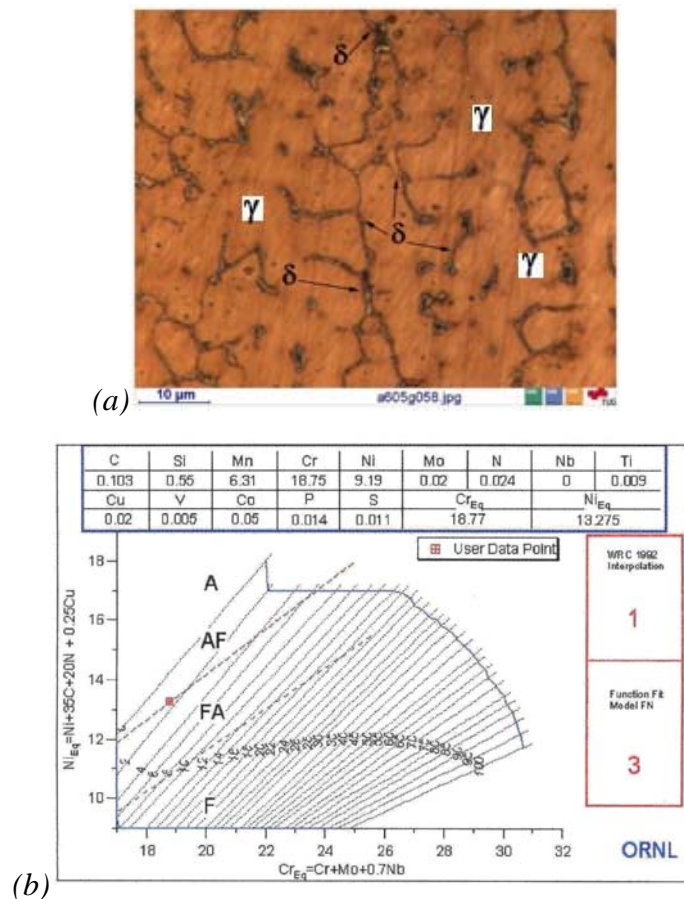


Figure 13 a, b Microstructure of pure filler material (deposit) MPW1

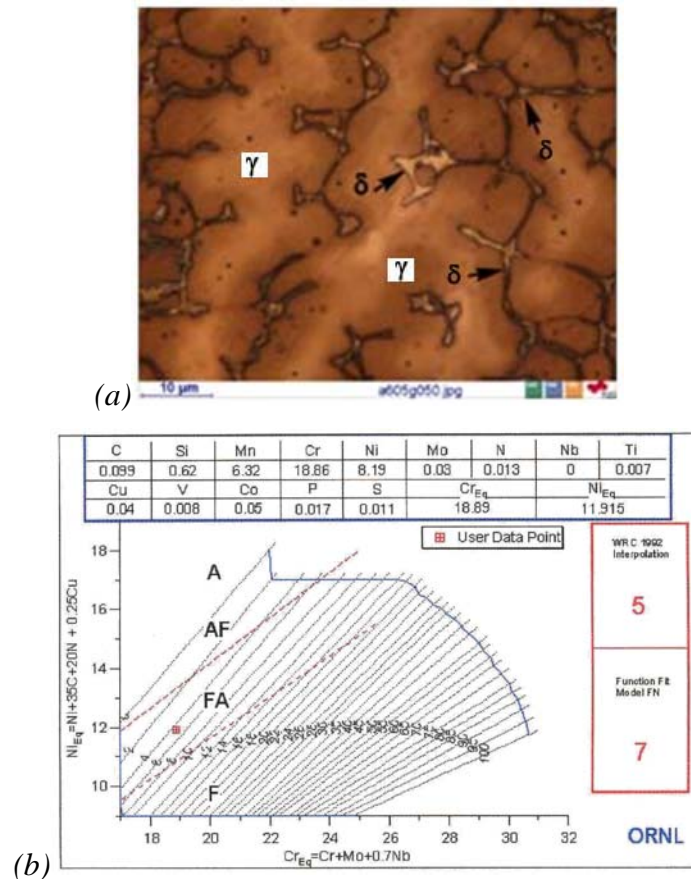


Figure14 a, b Microstructure of pure filler material (deposit) MPW2

Conclusion

- Hardness measurements of fillet welds detected existence of softened and hardened zone as results of applied welding thermal cycles (peak temperatures);
- Martensitic layer was found on dissimilar boundary ferrite/austenite, although cracks were not detected in this region;
- Microstructure of weld deposit is austenitic with some amount of α -ferrite (ferrite-austenite type of crystallisation);
- PWHT until 800 °C/6h contribute to significant lowering of weld deposit hardness values, from 202HV10 to 180HV10;
- Interpass temperature show significant influence on mechanical properties.

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