JMEPEG (2010) 19:963–969
DOI: 10.1007/s11665-009-9586-4

1059-9495/$19.00

Welding Characteristics of Ultrahigh Strength Steel in Annealed and Quench-Tempered Conditions

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(Submitted July 26, 2009; in revised form November 15, 2009)

In this research, the welding characteristics of a new UHSLA steel, 35NiCrMoV123, have been studied in two general conditions (annealed and quench-tempered). Carbon equivalent value of 35NiCrMoV123 steel is near 0.9 which classifies it as a "very difficult to weld" steel. The effects of welding heat treatment cycle (preheat, interpass, and postheat) on metallurgical and mechanical properties of weldments have been investigated by tensile, impact toughness, and hardness tests, as well as optical microscopy observations. It has been observed that by employing high-temperature stress relief (600 °C), welding could be performed in annealed condition successfully. Also, the results indicate that by applying precise welding heat treating cycle (preheat, interpass, and postheat temperature at 310 °C) in order to obtain lower bainitic microstructure in HAZ, also employing high-temperature stress relief (600 °C), welding in quench-tempered condition could be successfully performed.

Keywords cold cracking, 35NiCrMoV123, ultrahigh strength steel, welding

1. Introduction

Ultrahigh-strength, low alloy steels with medium carbon content and various amounts of chromium, molybdenum, nickel, and vanadium have been used for high-performance pressure vessels, rotors, etc. These steels can be successfully used at yield strengths equal or greater than 1400 MPa. Generally, quenching and tempering are well-established means to increase the strength of steel. This can be achieved mainly through the martensitic structure produced by quenching and the precipitation of a fine dispersion of alloy carbides during tempering (Ref 1).

Welding is a major step in the fabrication of most of the pressure vessels, structures, and equipments. Low alloy steels, such as AISI 4140, 4340, and 1045, generally have a carbon content of around 0.4% and carbon equivalent (CE) values between 0.6 and 0.9, which according to the literature (Ref 2, 3) are highly susceptible to cracking. These steels require care in selection of the filler metal and welding parameters. Also, they need preheating, interpass temperature control, and the use of postweld heat treatments (PWHT) (Ref 4).

In general, the main negative effects caused by welding of these steels are excessive grain growth and formation of nontempered martensite with a high level of hardness in the heat-affected zone, which when associated with the presence of hydrogen and tensile residual stresses can cause cold cracking.

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In fact, the amount of carbon and alloying elements promotes the formation of high carbon martensite in the weld metal (WM) and heat-affected zone (HAZ) of these steels. The high carbon martensite is very hard and brittle. The stresses caused by the contraction of the cooling metal, and the dimensional changes during martensite transformation lead to formation of cold cracks. Moreover, the presence of hydrogen will lower the stress level at which cracking may occur (Ref 2, 5, 6).

To avoid excessive grain growth, low-to-moderate levels of heat input are usually preferred in the welding of heat treatable low alloy (HTLA) steels (Ref 7). High heat input levels tend to promote wide HAZ with enlarged grain size and reduced notch ductility.

The new ultrahigh strength low alloy martensitic steel 35NiCrMoV123 has a carbon equivalent value of nearly 0.9 that classifies it as a "very difficult to weld" steel.

According to literature (Ref 2, 3, 7, 8), since these steels are weldable only in annealed condition, they are classified in "welding before heat treatment" category. But for some applications such as-weld repairs there is a need to establish the most efficient and the least risky procedure to weld in quench-tempered condition.

The aim of present investigation is to obtain the most efficient conditions for welding 35NiCrMoV123 which specifically includes introducing the best heat treatment cycle for welding (preheat, interpass, postheat and tempering temperatures) in two general working conditions (annealed and quench-tempered) of this steel. Specifications and procedures should be followed rigorously for difficult-to-weld materials.

2. Experimental Procedure

A 35NiCrMoV123 steel, supplied in the form of an annealed bar 270 mm in diameter, was used in this study. Its chemical composition is presented in Table 1. The as-received steel had a

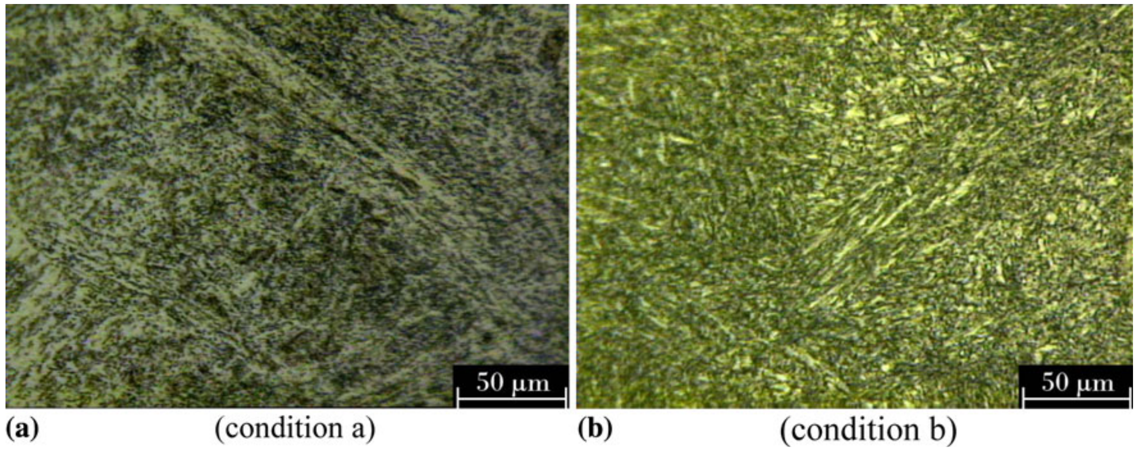

Fig. 1 Base metal microstructure in two conditions

Table 1 Chemical composition of materials (wt.%)

	C	Mn	Cr	Ni	Mo	Si	Ti	Cu	V
Base metal	0.4	0.33	1.0	3	0.61	0.37	…	0.14	0.12
Filler metal (a)	0.3	0.11	0.8	…	0.24	…	…	…	…
Filler metal (b)	0.1	0.7	3	…	3	0.4	0.4	…	…

bainitic microstructure with a hardness of about 290 HV (Fig. 1a).

Welding studies were carried out in two heat treatment conditions:

(a) welding the annealed plates and then heat treating the welded plates to obtain the desired properties, and
(b) welding the plates after they were given the desired quench-and-temper treatment.

In both of the cases, the weldments were characterized by evaluating the mechanical properties (impact fracture, tensile, and hardness) and microstructural observations by optical microscopy.

The toughness was characterized by the absorbed fracture energy of WM and HAZ regions at room temperature. Three specimens were used for each test condition.

Gas tungsten arc welding (GTAW) process was employed throughout the present investigation since the process is known to produce high-quality low hydrogen weld joints. The details of the weld set up are given in Fig. 2. The welding process was performed manually at constant welding parameters; current (165 A), voltage (13 V), speed (2.1 mm/s), under 99.999% pure argon protection.

2.1 Condition (a)

The annealed plate coupons were welded according to Table 2.

Isothermal transformation diagram for 35NiCrMoV123 shows a M_S temperature (the temperature at which martensite starts to form from austenite upon cooling) of approximately 290 °C (Ref 1).

Regarding to this diagram, five most common weld heat treatment cycles have been considered. The stress relief

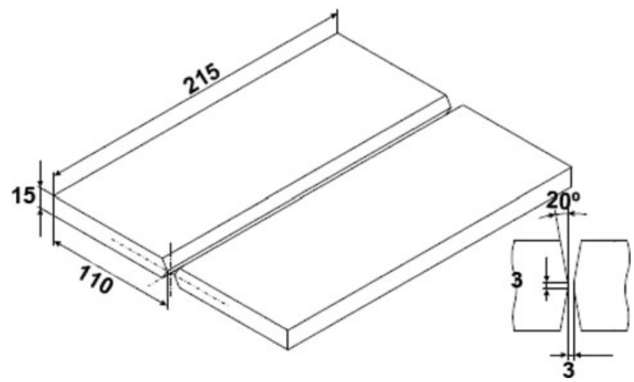

Fig. 2 Weld set up

treatment has also been carried out after welding in accordance with requirements of the ASME pressure vessel code (Ref 9).

Sections from each sample were mounted, polished and etched with 2% Nital solution for examination with optical microscopy.

Figure 3 presents the microstructure of HAZ in these samples. Different welding heat treatment cycles show significant influences on the microstructures of HAZ as shown.

Impact toughness results, obtained by an instrumented Charpy test at room temperature and ultimate tensile values of these specimens are also presented in Table 2. The hardness profiles of the traverse sections of the test samples, acquired from the connecting zone and extended through to the HAZ and base metal (before final heat treatment), are shown in Fig. 4. Likewise, Fig. 5 presents the hardness profile of these specimens after final heat treatment (quenched and tempered). Prior to mechanical tests and 48 h after welding, nondestructive tests were applied to determine the quality of the WM and HAZ.

2.1.1 Filler Metal. The chemical composition of the WM sometimes, but not always, matches that of the base metal for which they were intended. If this is not necessary or there is no matching filler metal available, then the use of a lower-strength WM is recommended to decrease the susceptibility to cracking (Ref 2, 3).

Table 2 Parameters of samples in condition (a)

Sample	Preheat, °C	Interpass, °C	Postheat, °C	Stress relief, °C	Cold cracking,	Impact energy, J HAZ	Impact energy, J WM	UTS, MPa
1	200	200	310	600	No	42.5	44	975
2	310	310	310	…	No	24	18.5	989
3	310	310	310	600	No	50	52.5	944
4	…	…	…	600	No	31	28	960
5	…	…	…	…	Yes	14	17	440

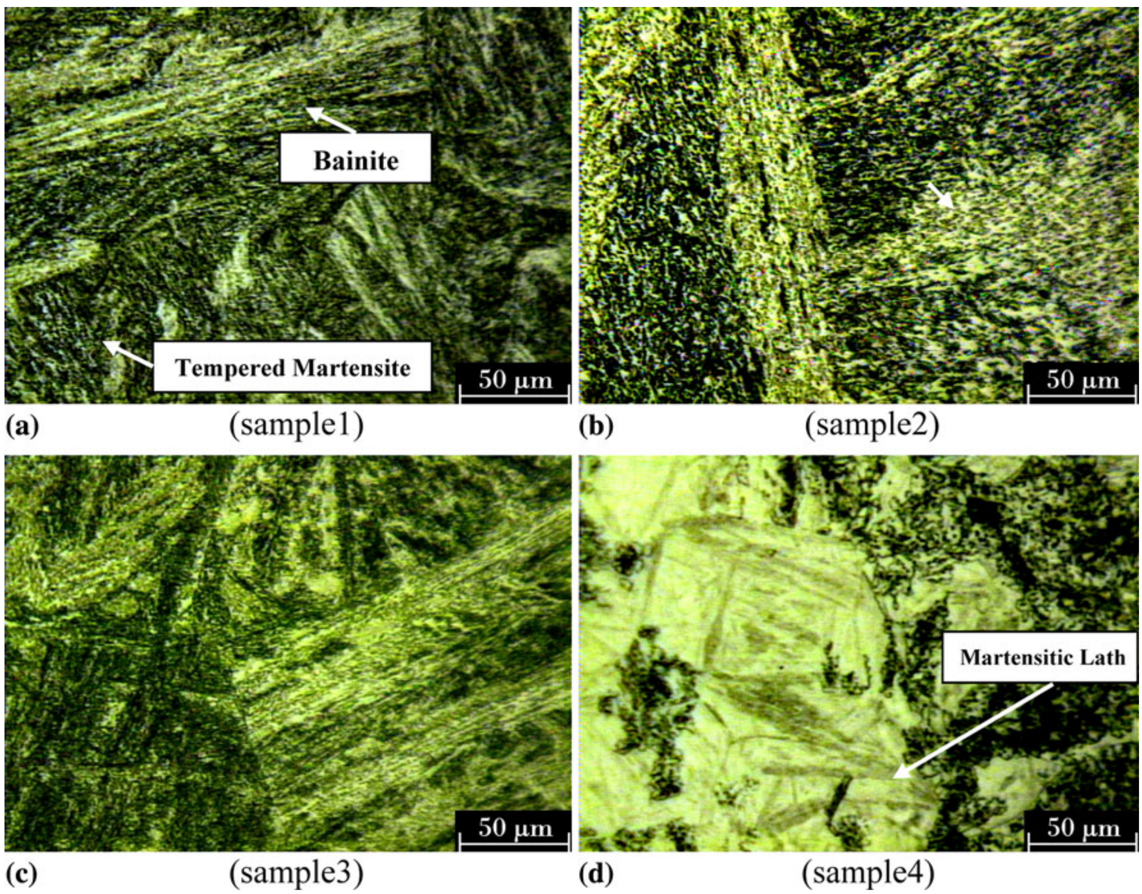

Fig. 3 Optical micrograph of HAZ microstructure in condition (a)

Filler metals that deposit WM which has a response to heat treatment matching that of the base metal are not available from weld wire manufactures for using in the welding of the new 35NiCrMoV123 steel that are quenched and tempered after welding.

To select the best filler metal, several compositions and preparation procedures of filler wire were studied, and the one giving the most similar mechanical properties to the base metal was selected. Its chemical composition (Filler metal (a)) is presented in Table 1.

2.2 Condition (b)

In this part, samples from the as-received steel were austentized at 875 °C for 2 h, followed by oil quenching to produce a quenched martensite structure (Fig. 1b). Specimens were then tempered at 620 °C for 1 h. The base metal hardness after heat treatment increased to 380 HV and its ultimate tensile strength improved to 1200 MPa. After welding no further heat treatment operation was carried out. The quenched and tempered plate specimens welded according to Table 3.

Uni-axial tensile results are shown in Table 3 as well as the results form the Charpy v-notched impact energy tests. Figure 6 presents the microstructure of HAZ in these samples. Figure 7 presents the hardness profiles of the transverse sections of the test samples, acquired from the connecting zone and extended through the HAZ and base metal. Nondestructive test results revealed the cold crack existence in the HAZ of samples 5, 6, and 7 (Fig. 8).

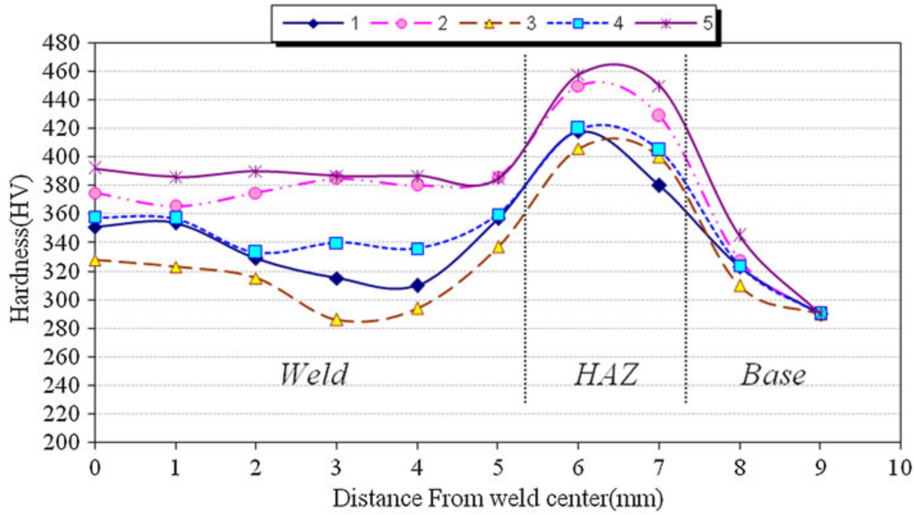

Fig. 4 Transverse hardness profile of weldments in condition (a)—before heat treatment

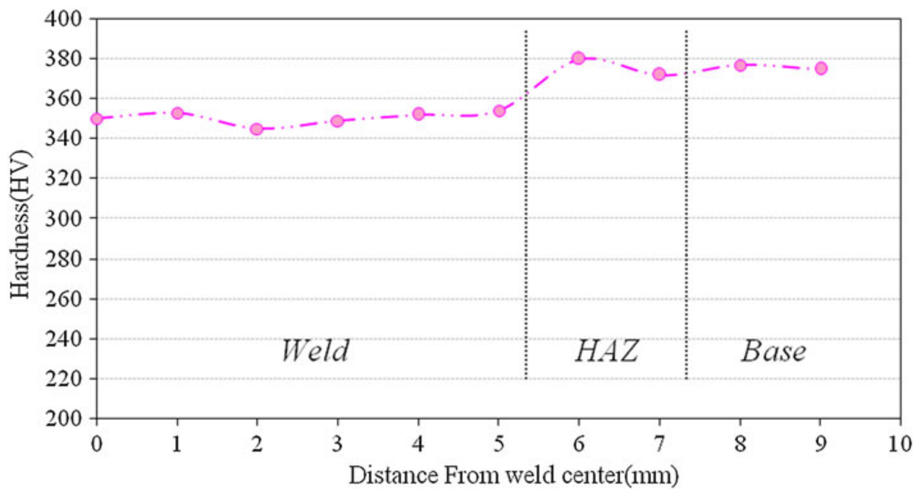

Fig. 5 Transverse hardness profile of weldments in condition (a)—after heat treatment

Table 3 Parameters of samples in condition (b)

						Impact energy, J		
Sample	**Preheat, °C**	**Interpass, °C**	**Postheat, °C**	**Stress relief, °C**	**Cold cracking**	**HAZ**	**WM**	**UTS, MPa**
6	…	…	…	…	Yes	6	10	200
7	…	…	…	600	Yes	10	19	380
8	200	200	200	600	Yes	16	23	1023
9	200	310	310	600	Yes	20.5	24	1162
10	310	310	310	600	No	38	30	1180
11	350	350	350	600	No	30.5	28	1148

2.2.1 Filler Metal. In condition (a), the samples started to yield near 920 MPa and fractured in the WM, indicating that strength undermatching of the welded joint was obtained. Thus, it is obvious that the filler metal (a) is not an appropriate choice for condition (b), where WM must have comparative strength, hardness, and wear resistance to the base metal without subsequent heat treatment.

Some types of filler metals that deposit nonheat-treatable WM can be used to a limited extent for the welding of HTLA steels in the hardened condition, as in repair welding, where it may not be feasible to anneal or over temper the steel prior to welding. The WM in this situation must gain a combination of strength, hardness, and toughness in as-weld condition (Ref 3, 10).

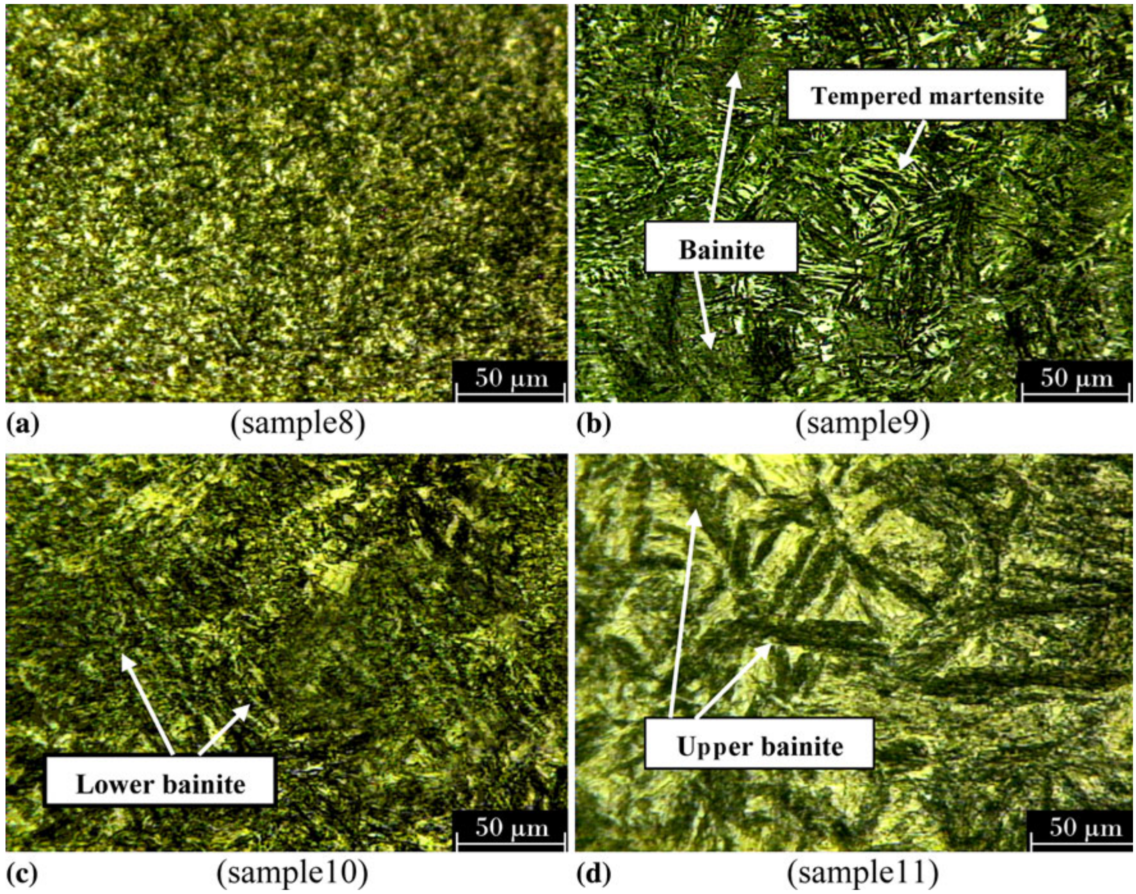

Fig. 6 Optical micrograph of HAZ microstructure in condition (b)

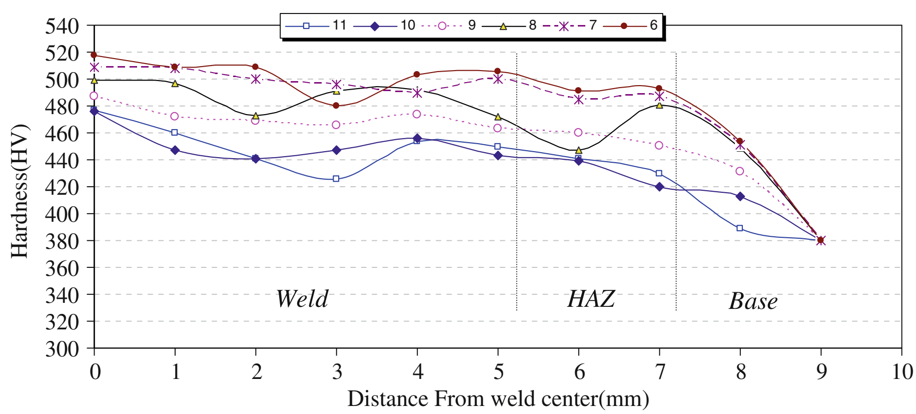

Fig. 7 Transverse hardness profile of weldments in condition (b)

It is sometimes possible to develop the same final hardness or strength in the weld using a weld deposit that has a lower carbon and higher alloy content than the base metal. Where possible lowering the carbon content of the WM generally minimizes the risk of weld cracking and improves the toughness of the weld deposit. A weld-deposit composition that offers greater resistance to weld cracking, higher ductility, and toughness is often a better choice (Ref 2, 3).

Hence, the low carbon, medium alloy filler metal was selected for condition (b). The chemical composition of filler metal (b) is presented in Table 1.

3. Discussion

3.1 Condition (a)

According to time-temperature-transformation (TTT) diagram, when 35NiCrMoV123 steel is welded by employing a preheat and interpass temperature ranging from 300 to 350 °C, and maintaining this temperature for a minimum of 1 h after completion of welding (postheat), following by slow cooling a bainitic microstructure can be obtained in the HAZ. The optical microscopy images of HAZ in samples 2 and 3 represent fully

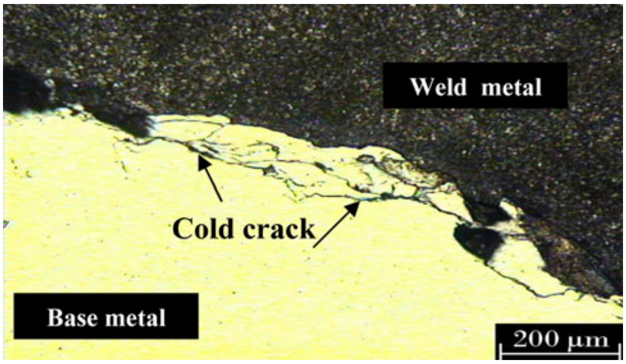

Fig. 8 Cold cracks in HAZ of sample 8

bainitic microstructure as shown in Fig. 3(b) and (c), respectively. According to Table 2, sample 2 in comparison with sample 3 shows a considerable loss in impact energy values, indicating the significance of the stress relieving or, in other words, high-temperature tempering on the toughness of WM and HAZ. Furthermore, it can be seen in Fig. 4 that there is a drop in the hardness values specifically in HAZ of sample 3, caused by tempering the bainitic microstructure.

Preheat and interpass temperature of 300 °C or higher is very harsh environment for welders because of the physical discomfort and because of the oxide layer which forms at the weld joint. Thus in samples 1 and 4, the influence of preheat temperature on weldment has been examined.

The welding procedure of sample 4 neither contains preheating nor interpass heating. After cooling to room temperature, the weldment only subjected to stress relieving treatment. Figure 3(d) shows the fully martensitic microstructure in HAZ of sample 4.

Sample 1 was subjected to the same condition as sample 4, although the weldment was not allowed to cool down to room temperature and after completion of welding, the sample was subjected to postheat treatment above M_s. As a result, a low-temperature tempering occurred in a fraction of HAZ that had already transformed from austenite to martensite, also the retained austenite part had been promoted to bainitic microstructure.

A combination of martensitic and bainitic microstructure in HAZ of sample 1 is shown in Fig. 3(a). The amount of martensite that forms depends on the preheat temperature and its relation to the martensite transformation temperature range.

No evidence of cracking was observed in these four samples. Therefore, the sample 5 is designated to specify whether the steel is sensitive to cold cracking in annealed condition or not.

According to the destructive and nondestructive results, also microscopic observations, without performing heat treating cycle on weldments, the brittle and hard martensitic laths form in WM and HAZ and cracking phenomenon is inevitable in this situation. HAZ of sample 5 reaches to high values of hardness (460 HV), that is too susceptible to cracking. As a result, the tensile specimens of sample 5 were fractured at low values of strength at HAZ region.

The tensile and hardness properties obtained from samples 1, 2, 3, and 4 after subsequent heat treatment fall in almost the same range, indicating that the prior heat treatment given to weldments had practically no effect on the final mechanical properties.

Thus, the crack free microstructure in condition (a) can be modified in the subsequent heat treatment to obtain the desired properties.

In conclusion, the results indicate that welding in annealed condition can be performed successfully by employing high-temperature stress relief at 600 °C.

3.2 Condition (b)

In the as-weld condition (sample 6) without performing any welding heat treatment cycles, the poor toughness (10 J) of the HAZ was due to the presence of nontempered martensitic microstructure. After stress relieving at 600 °C, the toughness rather improved due to the microstructural changes (sample 7) but in both cases cold cracks have been detected which are shown in Fig. 8.

By employing low-temperature heat treatment at martensite transformation temperature range (sample 8), slower cooling rate had beneficial effect on improving toughness, although microcracks have been detected.

Figure 6(a) presents the fully tempered martensitic microstructure in HAZ of sample 8.

Although the heat treatment cycle of samples 9 and 10 were similar to those of samples 1 and 3 at condition (a), but remarkably the microstructure of HAZ has been changed which corresponds to the different toughness of WM and HAZ. In sample 9, tempered martensitic structure prevailed in the matrix with some isolated tempered lower bainites (Fig. 6b).

In sample 10, WM and HAZ exhibited superior toughness due to the bainitic predominance in microstructure which assures the crack free region in HAZ (Fig. 6c).

The hardness profile of the sample 10 also confirms the formation of soft and ductile microstructure in WM and HAZ.

As a general rule, the harder the microstructure, the greater is the risk of cracking. Soft microstructure can tolerate more hydrogen than hard microstructures before the cracking occurs. Stresses are developed by thermal contraction of the WM during cooling and these stresses must be accommodated by strain in the WM. In rigid structures, the natural contraction stresses are intensified because of the restraint imposed on the weld by different parts of the joint (Ref 11).

In sample 11, applying very high temperature heat treatment has been examined. Thus, preheat, interpass, and postheat temperature were fixed at 350 °C (Table 3). According to the TTT diagram (Ref 1) in this thermal range, upper bainitic microstructure is predominant. Regarding to the impact test result and poor intrinsic toughness of upper bainitic microstructure (Ref 12), it is probable that this phase is formed mainly in HAZ of sample 10 which is presented in Fig. 6(d). Besides, according to literature (Ref 1, 5), if there were any martensite laths remained in the matrix, near 350 °C, a loss in toughness occurs due to tempered martensite embrittlement (TME). Thus, it seems that there is an optimum thermal range for heat treatment cycle around 310 °C, below that the brittle, high carbon martensite can be stabilized and cold cracks occurs and above that upper bainitic microstructure, TME, and excessive grain growth would be probable.

The slight undermatching in the strength of the WM in samples 10 and 11, did not significantly affect the tensile strength of the welded joint (Table 3). In other samples (6, 7, 8, and 9) fracture occurred in HAZ due to the propagation of cold cracking.

4. Conclusion

The welding characteristics of 35NiCrMoV123 in two general conditions (annealed and quench-tempered) have been studied throughout this research. The results can be summarized as follows:

- Without employing preheating and only by employing high-temperature stress relief at 600 °C, welding could be performed successfully in annealed condition.
- By applying precise welding heat treatment cycle (preheat, interpass, and postheat temperature near 310 °C) to obtain lower bainitic microstructure in HAZ and also by employing high-temperature stress relief at 600 °C, welding in quench-tempered condition could be successfully performed.

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