The effects of PWHT on the toughness of weld HAZ in Cu-Containing HSLA-100 steel

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A study was conducted to examine the effects of postweld heat treatment(PWHT) on the impact toughness and microstructures in the weld heat affected zone(HAZ) of Cu-containing HSLA-100 steel. The Gleeble, thermal/mechanical simulator, was used to simulate the weld HAZ. The details between toughness and PWHT of HAZ were studied by impact test, optical microscopy(O.M), scanning electron microscopy(SEM), transmission electron microscopy(TEM) and differential scanning calorimetry(DSC). The decrease of HAZ toughness in the single thermal cycle simulation relative to the base metal's toughness is ascribed to the coarsened-grain formed by heating to 1350°C. While the increase of HAZ toughness in the double thermal cycle simulation relative to that in the single thermal cycle is due to the fine ferrite(α) grain transformed from austenite(γ) by heating to α/γ two phase region. Cu precipitated during aging, which increases the strength of the base metal, is dissolved during the single thermal cycle to 1350°C and remains in the matrix during subsequent thermal cycle simulations. It precipitates by introducing PWHT suggesting that the decrease of toughness in the triple thermal cycle simulation ($T_{P1} = 1350^\circ\text{C}$, $T_{P2} = 800^\circ\text{C}$ and $T_{P3} = 500^\circ\text{C}$) does not occur from the Cu precipitation. The behaviors of Cu-precipitates in the weld HAZ are similar to those in the base metal. The PWHT at 550°C shows the highest hardness and the lowest toughness in the HAZ, whereas PWHT at 650°C shows a reasonable toughness and strength combination suggesting the possibility of improving the toughness of HAZ.

KEY WORDS: High Strength Low Alloy Steel, Postweld treatment, Aging, Precipitates, Impact toughness

1. Introduction

Cu-bearing high strength low alloy(HSLA) steels have shown good combination of high strength, toughness and superior weldability due to the low carbon contents and precipitation strengthening by copper precipitate. The addition of copper has been used to compensate the strength loss from reducing carbon contents in HSLA steels with few deleterious effects. However, in welding, Cu-bearing HSLA steels are known to sensitive to the stress relief cracking(SRC) due to the re-precipitated ϵ -Cu during post weld heat treatment(PWHT)¹⁾. Lundin reported that the re-precipitation of ϵ -Cu after PWHT at 538°C increase the hardness of weld heat affected zone(HAZ)²⁾. Also Bolliger³⁾ showed a minor toughness improvement after PWHT for 10hours at 620°C, and suggested that the Cu-bearing A710 can be used in the as-welded condition.

In the multipass welding, coarsened-HAZ formed from the previous pass can be heat affected similar to PWHT by subsequent welding passes leading to form ϵ -Cu which is known to deteriorate the HAZ toughness⁴⁾. However, there is a little doubt regarding the effect of the PWHT on the HAZ toughness. It is imperative to clear whether the

PWHT can improve HAZ toughness, if so, mechanical properties of the weld joint in the Cu-bearing steel will be greatly enhanced.

This paper was to suggest the possibility of improving the impact toughness of weld HAZ through the investigation of microstructures and mechanical properties of base metal and simulated weld HAZ.

2. Experiments

The HSLA-100 steel used in this study was produced in the vacuum induction melting furnace. Table 1 shows chemical composition of HSLA-100.

The plates rolled to 13mm were austenitized for 1hour at 900°C and quenched, and followed by aging at 300-670°C. Impact and tensile tests were conducted in aged specimens. Weld HAZ simulation tests were performed to the specimens aged at 650°C using a gleeble 1500 thermomechanical testing machine.

Table 1. Chemical composition of HSLA-100 (wt%)

C	Si	Mn	Cu	Sol-Al	Ni	Cr	Mo	Nb	Fe
0.035	0.4	0.9	1.5	0.027	3.44	0.6	0.36	0.042	Bal

The HAZ simulations were categorized into three groups according to their thermal history. Firstly, in the single thermal cycle treatment, specimen was heated up to 1350°C cooled down to room temperature. Secondly, in the double thermal cycle, specimen was reheated up to 800°C after single thermal cycle treatment. Finally, reheating the double thermal cycled specimen at 500°C, was the triple thermal cycle treatment. Heating rate was controlled as 135°C/sec. and holding time was 5sec. at respective peak temperatures. The cooling rate between 800 and 500°C was controlled as 12sec equivalent to heat input of 20kJ/cm when welding 25mm thick plate by GMAW process at room temperature. The weld simulated specimens were post weld heat treated at 500°C, 600°C and 650°C, respectively and hardness and impact toughness tests on those samples were also performed.

Transmission electron microscopy(TEM) and differential scanning calorimetry(DSC) were used to investigate the microstructures and ϵ -Cu precipitation behaviors in base metal, HAZ simulated, and 650°C PWHTed specimens.

3. Results and discussion

3.1 Characteristics of base metal

Fig. 1 shows the yield strength variations with respect to aging temperatures. The strength of specimens aged between 300°C and 600°C developed greater than that of as-quenched specimen. Also it is notable that the aging curve has two peaks at 300°C and 500°C. We reported in elsewhere that the first peak strength at 300°C results from the formation of Cu-clusters coherent with matrix, and the strength drop at 400°C, tempering effect. The second peak at 500°C suggested that the strengthening effect from the ϵ -Cu precipitation overwhelmed the tempering effect. Beyond 500°C, strength decrease with aging due to the precipitate coarsening and tempering effect^{5,6}.

TEM micrographs of the specimens aged at 500°C and 650°C (Photo.1) show ϵ -Cu precipitates which are known to increase yield strength. It can be seen that ϵ -Cu precipitates are finely and homogeneously distributed (Photo 1(a)) by aging at 500°C while they are coarsened, at 650°C(Photo. 1(b)).

Fig.2 shows the variation of impact toughness in LT direction with respect to aging temperature. The impact toughness increases with increasing aging temperature until 400°C and reaches its minimum at 500°C. It increases again in much steeper ways beyond this temperature. 650°C was selected as a proper aging temperature based upon the Fig. 1 and Fig. 2's results

promising the reasonable combination of yield strength and impact toughness.

Fig. 1 Yield strength with aging temperature.

Photo. 1 TEM micrographs steel aged for 1 hr. at 500°C(a) and 650°C(b).

Fig. 2 Impact energies tested at R.T with aging temperatures in base metal.

3.2 Effects of PWHT on HAZ

3.2.1 Microstructures

The microstructures obtained from each thermal cycles are shown in Photo. 2. The grain size of the base metal aged at 650°C is about 14μm and that of the single thermal cycled specimen heated up to 1350°C, 80μm.

Photo. 2(c) shows partially smaller grains in the double thermal cycled specimen than those in the base metal. The microstructure of triple thermal cycled specimen in Photo. 2(d) is similar to that of Photo 2(c).

Photo. 2 Optical micrographs with thermal cycle.

3.2.2 Hardness

Fig.3 shows the effect of PWHT on the hardness of the double thermal cycled specimen. The hardening trend of the specimen is similar to that of the base metal in Fig.1, because the ϵ -Cu dissolved by the single thermal cycling did not precipitate during double thermal cycling but precipitated by PWHT. The hardness of specimens PWHTed at 550°C, 600°C is higher than that of as-double thermal cycled specimen, whereas the hardness(Hv260) of 650°C PWHTed specimen is lower than that of either as-double thermal cycled specimen(Hv277) or base metal (Hv280). ϵ -Cu is considered to be major strengthening factor in the 550°C and 600°C PWHTed specimens as well as in the base metal. Considering a reasonable combination of hardness and toughness of 650°C PWHTed specimen, this temperature can be suggested as an optimum PWHT temperature.

Fig. 3 Variation of hardness with PWHT temperature after double thermal cycle.

3.2.3 Impact toughness

Fig.4 shows the effect of PWHT on the impact toughness of the double thermal cycled specimen. The impact toughness value of as-welded state(double thermal cycle) is 180J, and those of PWHTed for 1hr. at 550°C and 600°C, about 100J in both cases, and that of 650°C, 220J which is close to the impact toughness of the base metal and higher than that of double thermal cycled specimen.

It is interesting to note that the decrease of impact toughness by aging at 600°C where AWS.D.1.1 specification⁷⁾ recommends as the PWHT temperature, which is below the tempering temperature(660°C) of high strength steels such as HY-100steel.

Lundin⁴⁾ reported that the high sensitivity to PWHT cracking in Cu-containing HSLA steels is ascribed to the coarsened precipitates at grain boundary which precipitate from the supersaturated solid solution and grow under the stress and elevated temperature, inducing the stress concentration at grain boundary and reducing rupture time in the C-ring test.

Fig. 4 Impact energies tested at -50°C with PWHT temperature of the base metal and double cycled specimens.

SEM fractographs of the double thermal cycled and the double thermal cycled plus PWHTed at 550°C, 600°C and 650°C specimens are shown in Photo. 3 (a), (b), (c) and (d), respectively. The brittle transgranular cleavage fracture mode was developed in Photo. 3(a), (b) and (c) except Photo. 3(d) containing the ductile transgranular dimple mode.

Fig.5 shows the variations of impact toughness of the base metal and the double thermal cycled plus PWHTed at 600°C and 650°C with respect to impact test that of PWHTed at 600°C and 650°C drastically decreased with decreasing temperature. It can be said that the transition temperature of impact toughness in the base metal is around -120°C, however, it is not clear in the case of

PWHTed specimens. The brittle transgranular cleavage fracture mode(Photo. 3(c)) was observed in the 600°C PWHTed specimens tested at -50°C, while the ductile dimple fracture mode, in the (photo. 3(d)) 650°C PWHTed.

Photo. 3 SEM fractographs with PWHT temperature of HSLA-100 steel.

Fig. 5 Impact energies with test temperature of base metal and PWHTed specimens for 1 hr. at 600°C and 650°C after double thermal cycle.

Fig.6 shows the variations of impact energies at -50°C with thermal cycle. The impact energies of specimens PWHTed for 1 hr. at 650°C. were higher than those of as-welded specimens. The lowest impact toughness in the single thermal cycled specimen was ascribed to the grain coarsening as shown in Photo. 2(b). However, the impact toughness of the double thermal cycled specimen showed 225J approaching that of the base metal due to the fine grains formed from austenite to ferrite transformation during the second heating cycle. Lastly, the impact toughness decreases again in the triple thermal cycling. Although most researchers^{1,4)} reported that re-precipitation of Cu during the HAZ simulating was the major reason for toughness decrease, it is more likely that

Cu-precipitates are not formed from simulating thermal cycle but PWHT considering the age hardening curve of the HAZ (Fig.3) is similar to that of the base metal(Fig.1). The discussion of the precipitation behaviors of Cu-precipitates reserved in Fig.7 and Photo.4

Grange et al reported that the Fe-C alloy containing Mo, Si, Cr, Ni and Mn has higher hardness than the binary Fe-C alloy has when it is tempered at 260°C and 540°C⁸⁾. This indicates that the decrease of impact toughness in the triple thermal cycling results from the hardening effects relating to the Mo, Si, Cr, Ni and Mn.

Fig. 6 Impact energies tested at -50°C with thermal cycle.

3.2.4 Precipitation behaviors

It is generally known that the impact toughness of Cu-containing HSLA steels decrease when the base metal is exposed under the A_{c1} temperature during the welding process due to the precipitation of Cu-precipitates.

Fig. 7 shows DSC curves obtained from different thermal cycled specimens. Two heat evolution peaks at near 280°C and 500°C appeared in Fig. 7(a), and they disappeared in the aged specimen (Fig. 7(b)). Also the same type of peaks re-appeared in the triple thermal cycled specimen(Fig. 7(c)) and disappeared again in the PWHTed specimen (Fig. 7(d)). We reported previously that the heat evolution peak at 280°C is formed due to the formation of Cu-cluster which is coherent to the matrix and the peak at 500°C, ϵ -Cu phase^{5,6)}. The fact that the heat evolution peak did not appear in the aged plus PWHTed specimens is that the precipitates are already formed by again and PWHT.

The triple thermal cycled specimen showing the heat evolution peaks indicates that the precipitates dissolved by heating to 1350°C in the first thermal cycle stage did not re-precipitates during the DSC. This means that the precipitates does not occur during thermal cycle and have no relation to the deterioration of impact toughness.

TEM micrographs show only the martensite lath 0.1 μ m in the triple thermal cycled specimen(Photo. 4(a)) and fine ϵ -Cu precipitates in the 650°C PWHTed specimen(Photo.

4(b)). This is consistent with the DSC analysis results of Fig. 7(c),(d). From these results, it is concluded that Cu precipitation occurs from PWHT, not from the simulating thermal cycles, therefore, the decrease of impact toughness in thermally simulated specimens is nothing to do with precipitation of Cu as long as keeping the temperature below A_1 .

Fig. 7 DSC curves with thermal cycle.

Photo. 4 TEM micrographs of HSLA-100 steel.

(a) Triple cycle (b) Triple cycle + 650°C × 1hr. PWHT

4. Conclusion

The characteristics of the base metal and the simulated weld HAZ have been investigated as a function of aging temperature. The effect of PWHT on the impact toughness of HAZ was also studied. The results can be summarized as followings.

1) The base metal aged at 500°C shows the highest yield strength, but the lowest impact toughness. Above 500°C aging temperature, the yield strength decreases while the impact toughness increases.

2) The decrease of the impact toughness in the single thermal cycle simulation results from its high temperature (1350°C), which contributed to grain coarsening, on the other hand, the increase of the impact toughness in the double thermal cycle simulation is caused by the fine ferrite grains, where transformation of an austenite to ferrite occurred during the second thermal cycling.

3) The form of ϵ -Cu precipitates and its morphology strongly affect the mechanical properties of the simulated weld HAZ as well as the base metal; i.e. the impact toughness decreases while the yield strength increases with the ϵ -Cu formation in the case of 500°C and 600°C PWHT, however, in 650°C PWHT vice versa, with the ϵ -Cu precipitates coarsening.

4) Dissolved Cu into the matrix in the first thermal cycle stage does not precipitate during the following second and third thermal cycling, but precipitates by the PWHT. Therefore, the decrease of the impact toughness in the triple thermal cycled specimen results from effect of the secondary hardening by other minor alloying elements, rather than from the Cu precipitation.

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