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Effect of welding heat input on microstructure and mechanical properties of simulated HAZ in Cu containing microalloyed steel

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Abstract The influence of weld thermal simulation on ICGC HAZ microstructure and mechanical properties of Cu containing Nb-Ti-microalloyed steel has been investigated. Low heat input of 0.7 kJ/mm (simulated fast cooling of $\Delta t_{8/5} = 5$ s) and high heat input of 4.5 kJ/mm (simulated slow cooling of $\Delta t_{8/5} = 61$ s) were used to generate double-pass thermal cycles with peak temperatures of 1350 and 800 °C, respectively. The microstructure after high heat input mainly consisted of polygonal and quasi-polygonal ferrite (QF) grains with certain amount of acicular ferrite, whereas, after the low heat input, microstructure mainly consisted of lath or elongated bainite–ferrite, QF and M–A constituents. The size of ferrite grains decreased and volume of M/A constituents increased with fast cooling rate. The precipitation characteristics were found to be similar in both cooling rates. However, the precipitation of Cu-related phases was promoted by slow cooling rate. By fast cooling rate, the investigated steel exhibited an increase in hardness from 187_{HV} to 197_{HV}. Consequently higher yield strength with considerable loss in the (−10 °C) CTOD fracture toughness ($\delta_{fast\ cooling} = 0.86$ mm and $\delta_{slow\ cooling} = 1.12$ mm) were demonstrated.

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Introduction

Accelerated cooling of thermo-mechanically processed microalloyed steels is now widely recognized as a mean to obtain high strength in combination with superior toughness and formability [1–5]. The balance of high strength and good toughness in these steels, such as HSLA steels, can be upset by the thermal cycles experienced during welding producing poor toughness in the heat-affected zone (HAZ). In practice, the most essential properties of the steels used in Marine and offshore structures are good toughness characterized by CTOD and/or Charpy-V notch impact test, and tensile strength of the weld joints made by welding procedures. Nowadays, because of high heat inputs during the joining process, the coarse-grained heat-affected zone (CGHAZ) adjacent to the fusion line of this steel grade represents a region of pronounced low toughness. This is often revealed by fracture toughness tests, which are being increasingly used in marine structural applications. The CGHAZ regions are often the main reason for local brittle zone (LBZ) appearance. Although the structural significance of LBZs characterized by their low crack tip opening displacement (CTOD) value has recently been studied extensively and debated, it still remains as a controversial topic. Nevertheless, steel manufacturers have made improvements in conventional alloy design of this steel grade by aiming good weldability and high CGHAZ toughness in order to meet stringent requirements of marine constructions. To avoid brittle-fracture at low temperatures the weld joints, in particular their CGHAZ, should have adequate toughness. Evidently, the weld thermal cycle that results in a peak temperature of about 1300 °C being experienced by the microstructure adjacent to the molten weld metal can lead to pronounced precipitates dissolution. The consequences are austenite grain growth and the formation of hardened transformation products during cooling, which result in rather low toughness zones susceptible to brittle-fracture initiation LBZ. To prevent embrittlement, results from excessive grain coarsening in this region which has a bainitic microstructure containing different amounts of various martensitic/austenitic phases (M/A constituents), steel manufacturers have made an attempt to restrict the austenite grain growth by the introduction of finely dispersed stable particles such as Ti nitrides or Ti oxides into the different steel grades. Surveyed literatures [6–9] indicate that efforts in different countries have been made to achieve fine-grained HAZs in high heat input welds by using Ti-microalloyed steels for over a decade. In contrast, it was reported that [10–12], conventional Ti addition technique for improving HAZ toughness cannot be applicable to the high heat input (>4.0 KJ/mm) regime due to dissolution of Ti carbides/nitrides and subsequent austenite grain growth, which can be occurred easily during thermal cycle. Therefore, HSLA-100 and HSLA-80 steel grades are Cu bearing low carbon steels which are still being developed to adapt their weldability and toughness, besides their widely known high strength properties, in order to make them adequate for use in marine structural applications. During multi-pass welding of such steels, it has been found that the most degraded part in the HAZ is the intercritically reheated coarse grained HAZ (ICGC HAZ), which is the region of the GC HAZ heated to temperatures between the Ac_1 and Ac_3 by subsequent welding passes [13, 14]. Since the size of the ICGC HAZ (which results from the double-pass welding process) is very small, the study of this region is particularly difficult. Therefore, a thermal simulation was used to generate a relatively large region of ICGC HAZ which allowed the notch to be reliably located in the correct microstructure. Therefore, the steels were subjected to a double-pass welding thermal simulation. In view of the aforementioned, this investigation is aimed to focus on studying the influence of two different levels of heat inputs on the microstructure and mechanical properties of the simulated ICGC HAZ in low carbon Cu containing Nb-Ti micro alloyed steels.

Experimental procedures

In this investigation, an ultra-low carbon Cu bearing Nb-Ti-microalloyed steel was delivered by Pohang Iron and Steel Company (POSCO). The steel had already been simulated using a HAZ double-pass thermal cycle. During the experiments, thermal cycles with net heat inputs of 0.7 and 4.5 KJ/mm were applied, which is the most widely applicable range in welding of HSLA steel grades. The low heat input of 0.7 kJ/mm was simulated by fast cooling rate of

Table 1 Chemical composition (wt%) of Cu containing microalloyed steel

C	Si	Mn	Cu	Others	C_{eq}
0.04	0.15	1.46	1.01	Ni, Sol. AI, Ti, Nb, N	0.416

$\Delta t_{8/5} = 5$ s and high heat input of 4.5 kJ/mm was simulated by slow cooling rate of $\Delta t_{8/5} = 61$ s with thermal cycles.

The chemical compositions of the investigated steel are shown in Table 1 and the equivalent carbon content was calculated as

$$\#C_{eq} = C + (\mathrm{Mn} + \mathrm{Si})/6 + (\mathrm{Ni} + \mathrm{Cu})/15 + (\mathrm{Cr} + \mathrm{Mo} + \mathrm{V})/5 = 0.416.$$

Optical metallographic samples prepared by conventional grinding and polishing techniques and etched with 2% nital solution were observed in a light microscope. Scanning electron microscope (SEM) was used to reveal the presence of finer microconstituents as well as the morphology of fracture surfaces. Thin foils for transmission electron microscopy (TEM) were prepared by twin jet polishing in the electrolyte of 90% acetic acid and 10% perchloric acid. Thin foil samples were observed in a transmission electron microscope (Philips CM200 with EDAX) at 200 kV operating voltage. Hardness testing was carried out by Vickers hardness machine, AKASHI/Mitutoyo, at a load of 1 kg. The CTOD test was used to determine the fracture toughness properties of the investigated steel, which was carried out at −10 °C. To prepare a specimen for CTOD test according to ASTM E1920, a notch was machined in the center of the specimen and then an actual fatigue crack was carefully induced at the sharp point of the notch. Initially, average 2.5 KN with 2.0 KN amplitude was loaded from 3.2 mm to 3.5 mm and then average 1.1 KN with 0.9 KN amplitude was loaded about 1–2 mm as a fatigue crack. The actual test was performed by placing the specimen in three-point bending and accurately measuring the amount of the crack opens. For this purpose, a strain gauge is employed and mounted to a clip between two precisely placed knife edges at the mouth of the machined notch. The CTOD fractured samples' surfaces were investigated by using SEM to clarify the different morphologies and fracture aspects.

Results and discussion

Alloy chemistry

The chemical composition of the investigated sample (in Table 1) showed that the designed carbon content was in the permissible range (as per MIL-S-24645A standard) of HSLA-80 steel [15], and was lower (0.04 wt%) than the stipulated maximum value (0.06 wt%). This lower carbon content of about 0.04 wt% is nonetheless preferable for achieving a lower carbon equivalent (Ceq) needed for good weldability. On the other hand, because of the additions of Ti and Nb in the steel, the C/N ratio should be kept sufficiently low so that the formation of C-rich precipitates can be avoided. The higher solubility of such precipitates will reduce the pinning effect of grain boundaries, hence grain coarsening can occur. Moreover, the presence of Ti with the below 80 ppm of dissolved nitrogen will result in a Ti/N ratio in the range of 2–3 to get optimum HAZ toughness.

The achieved copper content was found to be 1.01 wt%, which was in the similar range (1.00–1.30 wt%) to HSLA-80 steel. In general, copper is added in marine structural steels primarily to achieve precipitation strengthening due to its higher solubility in the austenite phase and only limited ones in ferrite. Mostly, copper addition is associated with hot shortness of steel during welding process. Therefore, the production steel chemistry contains nickel to counteract the harmful effect of hot shortness. However, it is well known that copper raises the impact transition temperature (ITT), whereas nickel lowers it. Consequently, the additions of nickel in combination with copper might serve to nullify the harmful effect of the latter on the CVN (or CTOD) impact properties.

Microstructure

Figure 1 presents a comparison between microstructural evolutions of the investigated steel simulated ICGC HAZ at different $\Delta t_{8/5}$. The primary microstructural constituents at slow cooling rate ($\Delta t_{8/5} = 61$ s) were mainly polygonal ferrite (PF) and quasi-polygonal ferrite (QF) grains with a considerable amount of acicular ferrite (AF). Moreover, a small amount of martensite was observed which increased with increasing the cooling rate, as shown in Fig. 1a–d. In the case of a fast cooling rate, there was a strong tendency toward the formation of side plate type ferrite (ferrite/bainite) simultaneously with decreasing amounts of acicular as well as PF as shown in Fig. 1. The average grain size of the ICGC HAZ steel processed at those different cooling rates was found to be almost similar and ranged from 5 μm to 10 μm, as shown in Table 2. On the other hand, the microstructure processed at GC HAZ conventional cooling regime is known to be higher than 10 μm in grain size. Hence, the investigated microstructure was finer than that mentioned in the literature [11]. The grain refinement of the HAZ with high heat input was attributed to the effect of additions of microalloying elements Ti and Cu-Ni. The mechanisms of the improvement in HAZ microstructure, and hence toughness, due to Ti addition can be depicted as follows: (1) refinement of ferrite grains achieved by pinning effect of thermally stable Ti-nitrides

Table 2 Microstructural factors of Cu containing microalloyed steel with heat input conditions

Heat input condition	d_{Ferrite} (μm)	$V_{\mathrm{M/A\ constituents}}$ (%)
Heat input = 0.5 kJ/mm (fast cooling: $\Delta t_{8/5} = 5$ s)	5.3 ± 2.7	4.6 ± 1.5
Heat input = 4.5 kJ/mm (slow cooling: $\Delta t_{8/5} = 61$ s)	9.6 ± 3.4	2.3 ± 1.0

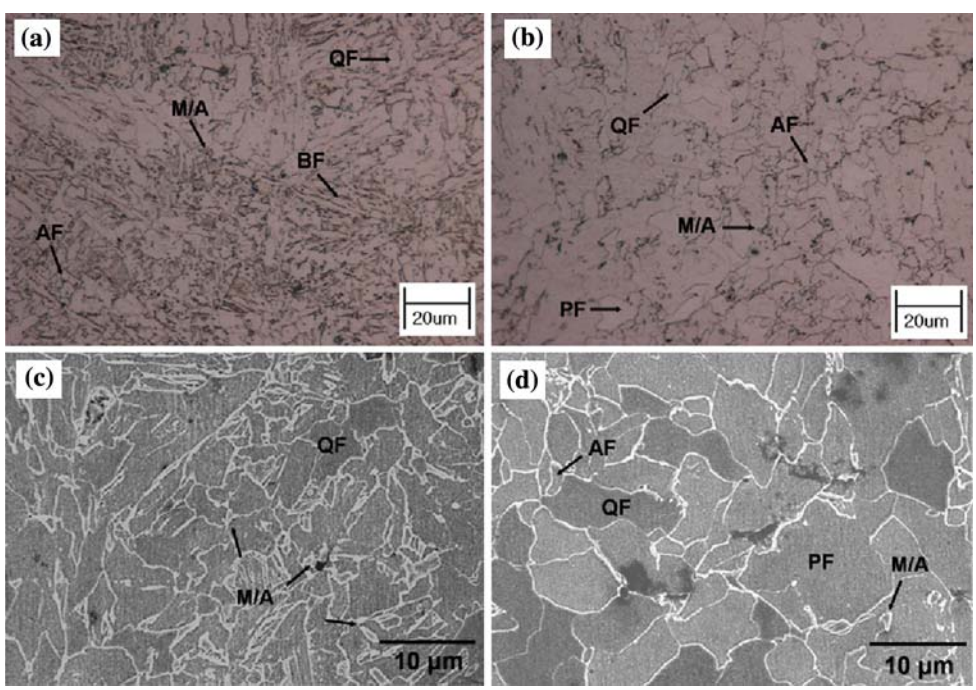

Fig. 1 OM and SEM micrographs of Cu containing microalloyed steel with heat input conditions of 0.5 kJ/mm; (**a**) and (**c**), and 4.5 kJ/mm; (**b**) and (**d**). *PF* polygonal ferrite, *QF* quasi-polygonal ferrite, *AF* acicular ferrite, *BF* bainitic ferrite, *M/A* martensite/austenite

particles distributed in austenite even at high temperatures (1350 °C), (2) formation of Ti-nitrides particles uniformly dispersed in austenite at high temperature, which can be considered as nucleation sites for acicular ferrite during the $\gamma \rightarrow \alpha$ transformation at the cooling part of simulated HAZ thermal cycle, and (3) decrease of the detrimental effect of soluble nitrogen <80 ppm in ferrite by formation of fine nitrides. Although TiN is thought to be comparatively stable even at high peak temperatures, partial or complete dissolution (depending on the size and composition of the precipitate) can be expected, because TiN precipitates can occur in various sizes. However, particles smaller than 0.1 μm have been found to be effective in grain boundary pinning [6, 16]. In agreement with this fact, the presence of TiN in a size of 50 nm was detected which enhanced the grain refinement and, consequently, might contribute in the improvement of HAZ toughness. Furthermore, microalloying with Nb was previously considered to be beneficial in this respect, but the austenite grain boundary pinning effect of Nb precipitates is restricted to peak temperatures below about 1100 °C [6]. The SEM microstructure indicated finer and uniform distribution of QF regions. However, it revealed that the pearlitic phases along the boundaries are almost absent in the microstructure at the slow cooling rate. This can be owed to the effect of Cu and Ni on the further lowering of austenite transformation start temperature. Additionally, in the current study, increasing a ferrite volume fraction in the microstructure of the steel was observed due to the simulated double-pass ICGC HAZ thermal cycle in both two different cooling rates ($\Delta t_{8/5} = 5$ and 61 s). In this thermal cycle regime, the resultant ferrite volume fraction during the $\gamma \rightarrow \alpha$ transformation was found to be a ratio of more than 90% with a mean grain size in the range of 5–10 μm. It was found that the ferrite grain size was approximately two times finer in comparison with the well-known traditional ones, HSLA-80 steel subjected only to GC HAZ, as mentioned above. The aforementioned reasons of the grain refinement and increased ferrite volume fraction might have an important role in improving the fracture toughness of the investigated steel. In contrast, the formation of M/A constituents leads to the deterioration of toughness in the weld HAZ [15]. Studies indicate that two types of martensite are involved in the M/A constituent. They are lath martensite with higher start temperature of martensite transformation and platelike martensite, in which twin can be identified [10, 17]. In addition, carbides in the M/A constituent can be precipitated during the self-tempering of martensite in the subsequent cooling process, or carbides can transform directly from carbon-rich austenite. Obviously, the decomposition of M/A constituents causes the improvement of HAZ toughness. With fast cooling rate ($\Delta t_{8/5} = 5$ s), the volume fraction of martensite is increased. In fact, the higher amounts of martensite in the simulated HAZ lower the toughness. Moreover, with fast cooling rate ($\Delta t_{8/5} = 5$ s), the morphology of M/A constituent will be transformed from bar to block. Typical morphology of M/A constituent is shown in Figs. 1 and 2 with different cooling time. Figure 2a and c shows the lath martensite and bainitic ferrite based on the TEM observation. The figures show that the martensite has clear lath character. Moreover, carbide-free bainitic ferrite can be found in the simulated HAZ, as shown in Fig. 2a.

The precipitation of Cu-related phases in the ferrite matrix at a low cooling rate was detected in a moderate amount and rarely in the high cooling one, as shown in Fig. 2b. TEM microstructure of slow cooling rate revealed homogenously distributed and very fine (10 nm) ε-Cu particles together with the presence of small amount of 9R phase. However, it was reported [12, 18] that the peak hardness in Cu containing steels can be obtained when both bcc Cu and 9R-precipitates are found. In contrast, the steel is usually lower in hardness when it has ε-Cu precipitates, which suggests that the mechanical property of only ferrite is not a major factor for higher strength. Therefore, the hardness values were found to be approximately the same in both steels, as mentioned in Table 3, due to the contribution of Cu-related phases. Furthermore, Carbo-Nitrides in both steels were formed by the addition of microalloying, Nb and Ti which in turn strengthened the steel matrix via solid solution strengthening mechanism. Interaction between dislocations and precipitates was also detected specially in the case of slow cooling rate (see Fig. 2b and d) which may present a sizable strengthening contribution even in PF microstructures.

In summary, among those strengthening contributions at the different cooling rate associated with their effects on hardness and CTOD fracture toughness, it can be concluded that the addition of Cu and Ni together with the controlling of cooling rates might have a further improvement on the hardness and CTOD fracture toughness of the investigated microalloyed steel.

Mechanical properties

The initial results of the tension tests showed that there is a further increase in yield and tensile strength by fast cooling rate. The CTOD at −10 °C and Vickers Hardness test results in high and low heat inputs with cooling rates of $\Delta t_{8/5} = 61$ s and 5 s, respectively, are shown in Table 3.

The results indicated that the CTOD values increased from 0.86 to 1.12 mm by decreasing the cooling rate from fast ($\Delta t_{8/5} = 5$ s) to slow ($\Delta t_{8/5} = 61$ s). However, the microhardness tests of the simulated ICGC HAZ revealed a slight decrease in the microhardness values from 197 Hv to 188 Hv with the same cooling manner. At such slight

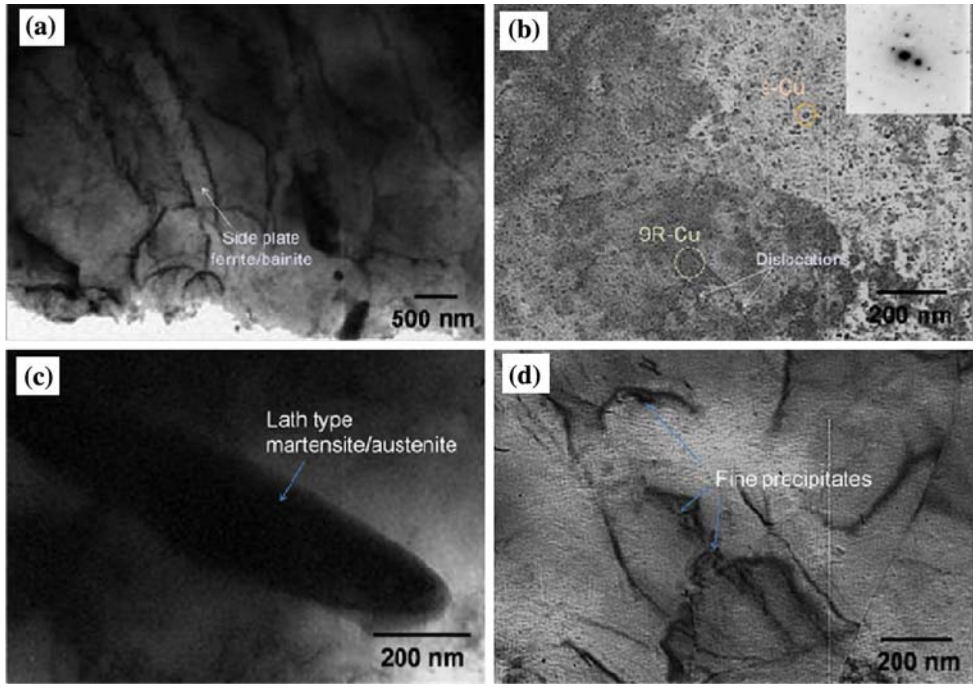

Fig. 2 TEM micrographs of Cu containing microalloyed steel with heat input conditions of 0.7 kJ/mm; (**a**) and (**c**), and 4.5 kJ/mm; (**b**) and (**d**). They show **a** lath type bainite-ferrite, **b** precipitates of ε-Cu and 9R-Cu, and interactions between Cu based precipitates and dislocations, **c** lath type martensite/austenite and **d** few interactions between other fine precipitates and dislocations

Table 3 Mechanical properties of Cu containing microalloyed steel with heat input conditions

Heat input condition	Hardness (HV)	E (GPa)	σ_{TS} (MPa)	σ_{YS} (MPa)	δ_m (mm)
Heat input = 0.5 kJ/mm (fast cooling: $\Delta t_{8/5} = 5$ s)	197	210	631	537	0.86
Heat input = 4.5 kJ/mm (slow cooling: $\Delta t_{8/5} = 61$ s)	188	210	600	510	1.12

change, hardness can be assumed constant. This stability of hardness might be attributed to the combined effects of Cu-phase precipitation hardening and the evolution of M/A constituent associated progressively with the different cooling rates. Furthermore, results showed that the toughness values were found to be increased by increasing $\Delta t_{8/5}$. This might occur due to the effect of acicular ferrite as well as variation of M/A decomposition due to the change in $\Delta t_{8/5}$. As reported by Xue et al. [19], the toughness of granular bainite is generally dependent on bainitic–ferrite and structures in islands such as martensite, M/A constituents, carbides or their combination. Therefore, the morphology, fraction, number density and size of M/A constituents had a great effect on the toughness of HAZ. In agreement with Aihara et al. [20], we can conclude that the difference in hardness between M/A constituents in HAZ and the neighboring base material is an important criterion of brittle-fracture trend. In this work, we find that with fast cooling rate of the welding thermal cycle, the area fraction and number density of austenite islands rich in carbon greatly decreased as a result of a large amount of transformed GBF on the early stage of $\gamma \rightarrow \alpha$ reaction. This occurs although a much higher hardness of M/A constituents could be expected after further transformation, which indicates a further decline in brittle-fracture of HAZ.

Fractography

Fracture surfaces of the ICGC HAZ specimens after CTOD testing were examined using SEM to determine the fracture morphology. Figure 3 shows an example of the fracture surface which was observed in the CTOD specimens. At the initiation site, a narrow dimpled region, with varying width, was observed on the fracture surface. A transition in the fracture mode from ductile to cleavage was evident during crack progression. The fracture mode in the propagation region was primarily cleavage associated with a small extent of ductile tearing. All the steels exhibited irregular facets in the cleavage regime. An existence of both ductile facets and voids/dimples regions were shown in the fracture surfaces of CTOD tested specimens. The fracture surface at the high cooling rate exhibits a high

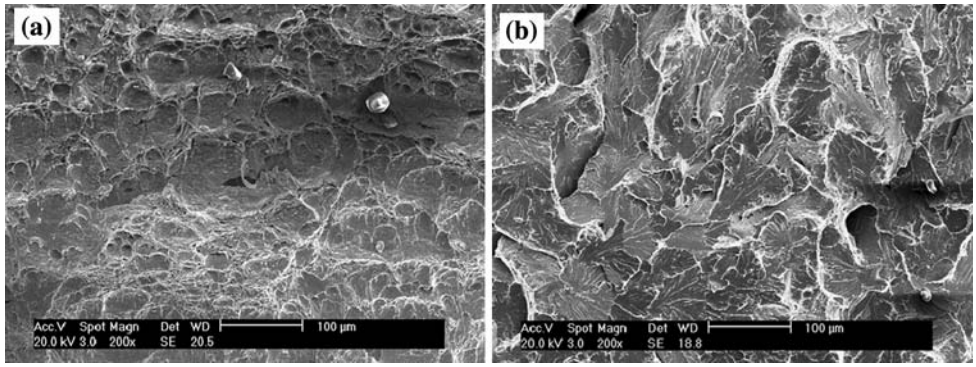

Fig. 3 SEM micrographs of fracture surfaces of specimens with heat input conditions of **a** 0.5 kJ/mm and **b** 4.5 kJ/mm tested after CTOD at −10 °C. They show **a** a high density of relatively uniform and small size ductile voids/dimples and **b** the low shallow and large size voids

density of relatively uniform and small size ductile voids/dimples, whereas, at slow cooling rate, shallow and large size voids are observed, as shown in Fig. 3. Indeed, it seems that there is a systematic increase in the density of small voids with increasing cooling rate. This suggests that the degree of microplasticity in steels is predominantly characterized by elongated and acicular ferrite/bainite. In contrast, dimple fractures of higher CTOD value ($\delta_m = 1.12$ mm) with slow cooling rate were observed instead of the lower value ($\delta_m = 0.86$ mm) with fast cooling rate. Furthermore, the decrease in CTOD value with fast cooling rate might be due to the presence of grain boundary precipitates, as shown in Fig. 3a, which behave as nucleation sites for the crack initiation.

Conclusions

The microstructure and mechanical properties of Cu containing Nb-Ti-microalloyed steels with different heat inputs were investigated. The Cu containing Nb-Ti-microalloyed steels have simulated ICGC HAZ double-pass thermal cycle. The major results are summarized as follows:

(1) At slow cooling rate, the microstructure of Cu containing Nb-Ti-microalloyed steel primarily consisted of PF with considerable amount of acicular ferrite, while at fast cooling rate, the microstructure was composed predominantly of elongated plates and lath type bainitic–ferrite.
(2) The change in the microstructure from ferrite to acicular ferrite with the increase in cooling rate was responsible for combination of good strength-toughness with the investigated steel.
(3) In the double thermal cycle regime the ferrite volume fraction produced during the $\gamma \rightarrow \alpha$ transformation was found to be more than 90% with the mean grain size of 5–10 μm contributed to the improvement of toughness property.
(4) The addition of Cu and Ni together resulted in relatively higher tensile properties combined with good toughness, due to the formation of a finer transformed product of Cu precipitation (10 nm) ε-Cu.
(5) By fast cooling rate, the investigated steel exhibited slight increase in hardness from 187_{HV} to 197_{HV}. Consequently, higher yield strength with considerable loss in the CTOD fracture toughness ($\delta_{\text{fast cooling}} = 0.86$ mm and $\delta_{\text{slow cooling}} = 1.12$ mm) were demonstrated.

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