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Influence of κ-carbide precipitation on the microstructure and mechanical properties in the weld heat-affected zone in various FeMnAlC alloys

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A R T I C L E I N F O

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A B S T R A C T

This study aims to investigate the effects of κ-carbide precipitation behavior in the heat affected zone (HAZ) in FeMnAlC lightweight steels. Three alloys with different amounts of Al were prepared by vacuum induction melting and hot rolling. After solution treatment, the HAZ samples were simulated by a Gleeble simulator with two heat inputs of 10 and 30 kJ/cm. Microstructural analysis with XRD and TEM were carried out while sub-sized tensile test, hardness test, and V-notched Charpy impact test were performed for investigating the mechanical properties of the base steels and HAZ. The results showed that the mechanical properties and precipitation of κ-carbide within the HAZ were strongly related to the Al content and heat input; the tensile strength and hardness of the HAZ increased as the Al content and heat input increased while elongation decreased. On the other hand, in the Charpy impact test, fracture mode transitions in the HAZ (ranging from ductile fracture to brittle inter-granular fracture) occurred in accordance with the Al content and heat input. The different fracture behavior was explained by TEM results, which showed precipitation behavior of κ-carbide in HAZ. Coherent intra-granular κ-carbide was found to cause a transition from ductile fracture to trans-granular cleavage, and we observed that a severe drop of the impact toughness occurred when partially coherent inter-granular κ-carbide appeared. Therefore, our results illustrate that the HAZ of lightweight steels with the proper Al content can be strengthened with minimal loss of impact toughness due to κ-carbide precipitation during the welding process.

1. Introduction

As the importance of automotive energy efficiency has continued to increase, lightweight materials that can reduce the weight of vehicles have emerged as next-generation materials. Among these, high-strength Fe–Mn–Al–C lightweight steels, which were originally studied for cryogenic and corrosion resistance applications, have attracted considerable attention in the automotive industry because of their potential for density reduction [1–9]. Previous studies have shown that lightweight steel can be austenitic, ferritic, or duplex, depending on it's chemical composition and offer an attractive combination of high strength, hardness, and ductility. In addition, lattice expansion and atomic mass reduction cause density reduction compared to general ferrous alloys [9–12]. Moreover, precipitation of κ-carbide, which can be formed by spinodal decomposition mechanisms in perovskite crystal structures with (Fe, Mn)$_3$AlC, is known to dynamically change the mechanical properties in various Fe–Mn–Al–C lightweight steels [12–15]. Based on the literature, lightweight steels are believed to have promising applications for saving energy in automobiles while simultaneously maintaining the requisite mechanical properties of vehicles; this can be accomplished by decreasing the weight, thereby improving transportation efficiencies by reducing fuel consumption.

It is important to consider the influence of the welding process on lightweight steels. Welding processes involve rapid thermal transitions of heating and cooling, which affect the resultant microstructural and mechanical properties [16,17]. Many scientists have reported the characteristics of welding processes in various FeMnAlC alloys [18–20]. Chou et al. investigated duplex Fe-30Mn-10Al with different amounts of carbon and confirmed that a small amount of ferrite provides excellent hot-cracking resistance [21]. Moon et al. reported that the addition of refractory elements has a beneficial effect on the HAZ tensile properties due to grain refinement and precipitation hardening [22]. However, although many studies have investigated the welding process in lightweight steels, most research has primarily focused on the weld metal properties or transition of mechanical characteristics, while few studies have been directed specifically toward the behavior and effects of κ-carbide in the HAZ. Since we believe that the analysis of κ-carbide within the HAZ is necessary for further lightweight steel research and

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Table 1
Chemical composition of the lightweight steel samples.

Alloy (wt%)	C	Mn	Al	Fe
A	0.80	31.5	8.73	Bal.
B	0.93	30.0	10.43	
C	0.89	31.4	11.93	

development, this study focuses on understanding the changes in the mechanical properties within the HAZ in accordance with κ-carbide precipitation in various austenite-based lightweight steels that contain varying amounts of Al. To perform these analyses, Gleeble simulation was used to obtain simulated HAZ specimens and reveal the relationship between κ-carbide and the mechanical properties in the HAZ. The mechanical properties we tested include the tensile characteristics, hardness, and cryogenic impact toughness. Direct observations of κ-carbide precipitation were made using a transmission electron microscope (TEM), and a statistical approach was applied to evaluate the κ-carbide behavior.

2. Experimental procedures

The chemical composition of the different steel types used in this study and the measured effective austenite grain sizes are listed in Table 1. The present ingots were prepared in a vacuum induction melting furnace. The ingots were then hot rolled to a thickness of 13 mm at 1200 °C and water quenched. The hot-rolled steel was made into base steel by performing solution treatment at 1050 °C for 2 h prior to water quenching.

HAZ samples with different heat input conditions were obtained with a 1150 °C peak temperature via a Gleeble simulator (Dynamic Systems Inc., USA). Rectangular plate-shaped specimens were used for the HAZ simulations. The simulations were conducted using thermal cycles with 10 and 30 kJ/cm of heat input to analyze the effect of κ-carbide precipitation in the HAZ. Both thermal cycles were calculated based on Rosenthal's heat flow equation [16,23]. A schematic plot of the thermal cycles used in these experiments is illustrated in Fig. 1.

A sub-sized tensile specimen was used to obtain the tensile stress-strain characteristics of the base steel samples and the HAZ samples with a tensile test machine (Z100, Zwick Roell Group, Germany) using a strain rate of 2.4 mm/min, while full-sized V-notched Charpy impact testing was carried out according to ASTM A370 using a Charpy impact tester. Each V-notched sample was prepared along the rolling direction of the steel, and the impact direction was perpendicular to the direction of rolling. Impact tests were performed at a temperature of − 40 °C; specimens were immersed in a − 40 °C pure ethanol bath for 10 min and then tested within 5 s after removal from the bath. The tensile and Charpy tests were performed in triplicate for each condition. Vickers and nanoindentation hardness tests were carried out with a micro-indenter (HMV-2, Shimadzu, Japan) with a load of 1.96 N and a nanoindenter (MTS-XP, MTSC, USA) with a peak load of 5 mN. The hardness values reported for each sample in this paper were obtained by averaging 20 measurements. The types of samples utilized for measuring the mechanical properties are shown in Fig. 2.

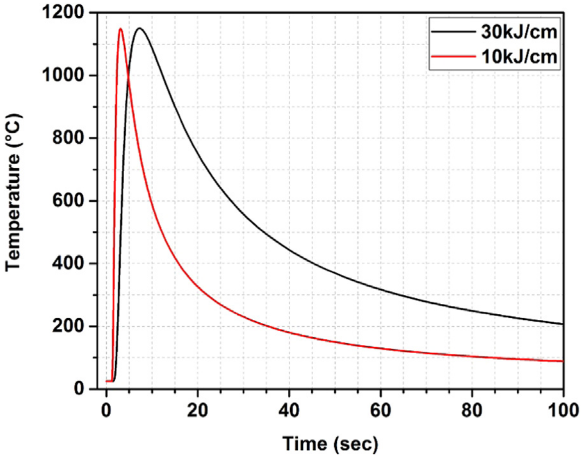

Fig. 1. An illustration of the thermal cycles for the simulated HAZ specimens with a peak temperature of 1150 °C with 10 and 30 kJ/cm of heat input.

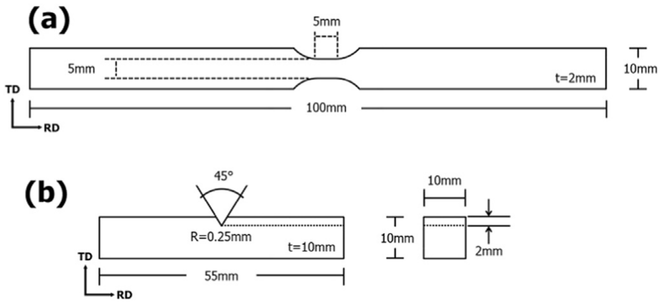

Fig. 2. A diagram of the experimental samples for (a) sub-sized tensile tests and (b) V-notched Charpy impact tests.

The specimens used for microstructural observations were mechanically polished by grinding with SiC paper up to 2000 grit, followed by micro polishing using a 1 µm diamond suspension. Then, each sample was etched with a 6% nitric acid solution at room temperature. The effective grain size of austenite was measured by analyzing SEM images (JSM-6360, JEOL, Japan) collected following the ASTM standard linear intercept method. XRD (SmartLab, Rigaku, Japan) investigations were performed for detailed phase analyses under 45 kV and 200 mA current conditions with analysis angles ranging from 20° to 80°. TEM samples for κ-carbide observations were prepared with a twin jet electro polisher (Tenupol-3; Struers, Denmark) in a solution of 5% perchloric acid in methanol at − 40 °C. Precipitation observation for the specimens were carried out using a transmission electron microscope (JEOL 2100, JEOL, Japan). The collected TEM images were used to quantitatively measure the particle size distributions and fractions with commercial image analysis software.

3. Results and discussion

Fig. 3 shows representative microstructural images of the experimental samples in solution treatment and HAZ conditions. The effective grain sizes for all samples are included in Table 2. The figures show that microstructural changes occurred for different amounts of Al. Steel A showed the typical microstructure of austenitic steel with coarse grains and annealing twins. On the other hand, as the Al content increased in steels B and C, secondary phases with elongated shapes were observed and austenite grain refinement occurred. The grain sizes in HAZ samples increased under both 10 and 30 kJ/cm heat input conditions, and the grain coarsening effect was more significant with less Al content.

XRD and TEM results of base steels are shown in Fig. 4. In the XRD results, ferrite phase and κ-carbide were detected in steel B and C, while steel A had only austenite phase. Furthermore, steel C also showed the DO3 peak, which is known that can be formed within the ferrite matrix via an ordering mechanism [24]. Researchers have long understood that the low strength and ductility of the DO3 phase lead to embrittlement (and should be avoided in Fe–Al alloys) [25]. It seems that the addition of Al, which is a ferrite stabilizer, caused the formation of the BCC phases in the microstructure. Liu et al. reported similar results after the addition of Si (i.e., a ferrite stabilizer), which promoted the formation of BCC phases [26]. However, the elongated secondary BCC phases were identified to be ferrite in TEM observation. It could be said that only little DO3 phase was formed in the steel C and had negligible