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Improvement of impact toughness by microstructure refinement of simulated CGHAZ through enhancing welding heat input of low carbon Mo-V-Ti-N-B steel

Huibing Fan\(^1,2\), Genhao Shi\(^1,2\), Qiuming Wang\(^1,2\), Leping Wang\(^1,2\), Qingfeng Wang\(^{1,2,*}\) and Fucheng Zhang\(^1,2\)

1 State Key Laboratory of Metastable Materials Science and Technology, Yanshan University, 066004 Qinhuangdao, People’s Republic of China
2 National Engineering Research Center for Equipment and Technology of Cold Strip Rolling, Yanshan University, 066004 Qinhuangdao, People’s Republic of China
* Author to whom any correspondence should be addressed.
E-mail: wqf67@ysu.edu.cn

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**Abstract**

Welding heat input greatly influences the microstructure and impacts the affected zone’s toughness. To interpret the relationships between the welding heat input, microstructure, and low-temperature toughness of the coarse-grained heat-affected zone (CGHAZ) of Mo-V-Ti-N-B steels, welding heat cycles with different heat inputs (25–75 kJ cm\(^{-1}\)) were performed on a Gleeble 3500 simulator. Intragranular ferrite in the simulated samples subjected to different thermal cycles was characterized and quantified, and the impact energies of simulated samples were evaluated at \(-20^\circ\)C. Upon increasing the heat input, the intragranular ferrite content rose sharply from 4.3% to 76.0%. The V(C, N) enrichment on the precipitate surface increased the size of precipitates, providing favourable nucleation conditions for intragranular ferrite. The prior austenite grain (PAG) and martensite/austenite (M/A) constituents became rough, and the content of the M/A constituent increased while the impact energy of the CGHAZ increased. This behaviour occurred due to the formation of intragranular acicular ferrite (IGAF), which refined the microstructure of the CGHAZ. Grain refinement eliminated the negative influence of higher M/A content on the impact toughness of the CGHAZ.

**1. Introduction**

Due to its toughness and high strength, low-carbon micro-alloyed steel has been widely used in large ships, heavy-duty steel bridges, and thick-walled products. In low-carbon micro-alloyed steel, although columnar crystals appear in the weld seam, there is fine acicular ferrite (AF) inside the columnar crystals, which ensure the excellent impact toughness of the weld seam. However, during thermal welding cycles, a coarse-grained heat-affected zone (CGHAZ) with low toughness is frequently generated [1]. Coarsened austenite grains and a brittle microstructure that includes coarse grain boundary ferrite and coarse granular bainite form at high heat input, deteriorating the toughness of the CGHAZ badly.

Amount of studies indicated that the impact toughness of CGHAZ decreases with the increase of heat input [2–4]. At high heat input, the prior austenite grain (PAG) was coarsened in the CGHAZ of micro-alloyed steel [5]. In the CGHAZ of V-Ti micro-alloyed steel, the PAG size increases from 81 \(\mu\)m and 133 \(\mu\)m as the heat input increases from 50 kJ cm\(^{-1}\) to 125 kJ cm\(^{-1}\) [6]. Side-plate ferrite microstructures and large martensite/austenite (M/A) constituents that decrease the impact toughness easily form in coarsened PAGs at high heat input [7].

Due to the C content of the M/A component being higher than that of the matrix, the hardness is higher than that of the matrix. The large size complex M/A constituent can lead to local embrittlement and crack initiation deteriorating the toughness of CGHAZ [8, 9]. However, recent studies have shown that the relationship between
the impact toughness of the CGHAZ and the welding heat input does not decrease monotonically with the increase of the heat input \([7, 10]\). Vanadium, carbon and nitrogen in steel can form VN and V\((C, N)\) particles, which exhibit a small lattice mismatch degree with ferrite and effectively facilitate the heterogeneously-nucleated ferrite \([11]\). As the heat input increases, the small VN may dissolve and then precipitate on the dissolved particles during cooling, coarsening the particles and reducing the barrier for the particles to promote the formation of acicular ferrite. Zhang et al\([12]\) revealed that the optimum impact toughness is obtained at a high heat input \(t_{8/5} = 180\) s for the increased formation of intragranular ferrite, especially AF, and high angle grain boundry. The interesting phenomenon that AF increases as the increase of heat input in the low-carbon micro-alloyed steel remains unclear in detail. Shi et al\([6]\) studied the effect of heat input on the impact toughness of the CGHAZ of low carbon V-N micro-alloyed steel. The results showed that with the increase of heat input, the acicular ferrite content in the CGHAZ increased and the microstructure was refined, thus improving the impact toughness of the CGHAZ. While, grain boundary ferrite appeared at \(t_{8/5} = 180\) s \([6]\). With the further increase of \(t_{8/5}\), grain boundary ferrite may undergo growth-deteriorating impact toughness of the CGHAZ. Fortunately, proper amount of element B is added to the steel, it can converge to the grain boundary and inhibit the formation of grain boundary ferrite \([13]\). Moreover, B can combine with N to form BN particles, and BN particles promote the acicular ferrite nucleus in the grain \([14]\), which further refine the CGHAZ structure. However, the effect of heat input on the microstructure and impact toughness of CGHAZ in V-N micro-alloyed steel containing B is rarely reported.

In this study, the low-carbon Mo-V-Ti-N-B steel was created for welding simulations. This study aims to explain the link between heat input, microstructure, and toughness of the CGHAZ in low-carbon Mo-V-N-Ti-B steel. The particle and PAG size changes in the CGHAZ as a function of heat input was examined. Additionally, the micro-mechanisms of fracture were carefully investigated by assessing the appearance of the original fracture and secondary microcrack propagation path. The relationship between impact toughness and microstructure was investigated and discussed in detail.

2. Experimental

2.1. Materials and Experimental procedure

Low-carbon Mo-V-Ti-N-B steel was created through a vacuum induction furnace to melt the steel. The chemical composition of the low-carbon-Mo-V-Ti-N-B steel is given below in table 1.

\[
\begin{array}{cccccccc}
& C & Si & Mn & S & P & Mo & V & Ti & N & B \\
0.06 & 0.27 & 1.53 & 0.001 & 0.004 & 0.28 & 0.064 & 0.018 & 0.0144 & 0.0012 \\
\end{array}
\]

Table 1. Chemical composition of the base steel (wt.%).

The influence of heat input on the impact characteristics and microstructure of a simulated CGHAZ in low-C Mo-V-Ti-B micro-alloyed steel was evaluated using a Gleeble-3800 thermomechanical simulator. Cuboid samples and rod bars were obtained from the plate (along the rolling direction) and used for simulations.

After austenitizing at 1200 °C for 2 h, the 200 kg ingot was smelted in a vacuum induction furnace and then underwent a controlled rolling and cooling process. The starting and finishing temperatures of rough rolling were 1108 °C and 1016 °C, and the starting and ending temperatures of end rolling were 953 °C and 857 °C, respectively. After reaching a final thickness of 18 mm, the plate was cooled from 798 °C to 416 °C at a rate of 15 °C s\(^{-1}\) and then air-cooled to room temperature. Table 2 shows the mechanical properties of the steel produced by the procedure described above.

| Yield stress | Tensile stress | Elongation rate | Impact energy |
|--------------|---------------|----------------|--------------|
| 503 MPa      | 625 MPa       | 20.36%         | 237 J        |

Table 2. Mechanical properties of the base steel.

2.2. Welding simulation procedure

Because of the coarse-grained heat-affected zone’s (HAZ) limited size and irregular shape, the impact sample with only the CGHAZ microstructure can’t be processed in the actual welded joint. As a result, the Gleeble thermal simulator was used to mimic the CGHAZ’s thermal cycling process to obtain a sample with a uniform microstructure and investigate the influence of heat input on the microstructure and impact properties of the CGHAZ.

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Figure 1 illustrates the thermal process curves for welding 20 mm thick steel plates with various heat inputs (25, 35, 50, 65, and 75 kJ cm$^{-1}$). The samples were first heated to 1320°C at a rate of 100°C s$^{-1}$ and held for 1 s. Then, the samples were cooled to room temperature along the curves. Thermal welding cycles were performed on the cuboid samples to determine the impact energy of the CGHAZ at various heat inputs. At the same time, the rod bars were used to determine the starting temperature of austenite transformation into ferrite during welding thermal cycles using a thermal dilatometer. The heating process was paused at 800°C by water quenching to obtain the properties of PAG and particles in the CGHAZ before austenite was changed into ferrite.

2.3. Microstructural characterization
Metallographic specimens of the CGHAZ thin cylinders were taken from the area adjacent to the thermocouple and then mechanically ground, polished, etched with 4 vol% Nital, and examined by optical microscopy (OM) and scanning electron microscopy (SEM, SU5000). The chemical composition of the precipitate particles, which induced intra-crystalline acicular ferrite nucleation, was characterized by energy-dispersive spectrometry (EDS). The metallographic samples were re-polished and etched with Lepera etchant for 120 s before being viewed using an Olympus BX51M OM to observe the M/A constituents in the simulated CGHAZ. The size and area fraction of the M/A constituent were calculated by Image-Pro Plus. The samples for prior austenite observations were prepared by mechanically grinding, polishing, and then etching them with supersaturated picric acid. The standard line-intercept method was employed to determine the PAG sizes.

The fine particles of the CGHAZ were characterized via transmission electron microscopy (TEM) using carbon replica samples. EDS determined the chemical composition of the precipitated particles. For carbon replica samples, the surface of the metallographic sample was polished, and a carbon layer was deposited on it. After that, a blade was used to cut the carbon layer into squares. Finally, 4 vol% Nital was used to extract them for TEM analysis.

The orientation relationships of grains in the simulated CGHAZ were determined by electron backscatter diffraction (EBSD). The EBSD samples were mechanically polished and then electropolished for 30 s in a solution of 5 vol% perchloric acids in alcohol (AES, ULVAC PHI710). The EBSD maps (140 μm × 140 μm) were scanned on a SU-5000 with a step size of 0.30 μm and then cleaned and analyzed by TSL OIM Analysis-5 software.

2.4. Impact properties
Charpy impact tests were conducted using standard specimens (10 mm × 10 mm × 55 mm) machined from simulated Cuboid samples at −20°C on a pendulum impact tester (JB-300B). To overcome the temperature rise of the sample in the air, we reduce the sample to −21°C. The specimens were placed in a cooling tank, the sample temperature was dropped to −21°C, and the specimens were kept warm for 5 min. The specimens were taken off and immediately put on the impact tester for impact test one by one. (so that the temperature of the sample during impact is close to −20°C). Where the thermocouple was positioned, a V-notch was created. The impact energy reported was the average of three simulated samples. SEM observed the impact fracture surfaces, and the secondary microcracks in the impact fracture cross-section were examined via SEM, as shown in figure 2.
3. Results

3.1. Microstructures

The representative OM micrographs of the simulated CGHAZ are shown in figure 3. The microstructure in the CGHAZ was granular bainite (GB), and some acicular ferrite (AF) at the heat input of 25 and 50 kJ cm\(^{-1}\). When the heat input increased from 50 to 60 kJ cm\(^{-1}\), the AF increased, and the GB decreased. With the heat input further increased to 75 kJ cm\(^{-1}\), the main microstructures constituents in the simulated CGHAZ were Polygonal ferrite (PF) and AF, as shown in figure 3(d). The area percentage of the microstructure in the CGHAZ obtained at various heat inputs was calculated using the image-Pro Plus software. Figure 4 shows that when the heat input rose from 25 to 75 kJ cm\(^{-1}\), the AF content in the simulated CGHAZ increased sharply.

The detailed morphology of the M/A constituent in the simulated CGHAZ at heat inputs of 25 kJ cm\(^{-1}\) (figure 5(a)) and 50 kJ cm\(^{-1}\) (figure 5(b)) was small and spindly. Figure 5(c) indicates that the M/A constituent was lumpy at 75 kJ cm\(^{-1}\) heat input. The amount of M/A constituent increased, and the M/A morphology changed from slender needles into blocks, which decreased the length-width ratio of the M/A constituent as the heat input increased.
3.2. Impact properties

Figure 6 presents the impact energies of simulated samples with different heat inputs. The impact value of the simulated samples first gradually increased from 9 to 37 J at heat inputs of 25 – 50 kJ cm$^{-1}$ and then sharply increased to 193 J at 75 kJ cm$^{-1}$.

The macro and micro-fracture appearances of the simulated CGHAZ at heat inputs of 25, 50, and 75 kJ cm$^{-1}$ are depicted in figure 7. The ductile zones size followed the order 25 $< 50 < 75$ kJ cm$^{-1}$. At a 25 kJ cm$^{-1}$ heat input, the whole fracture surface of the impact samples presented cleavage fracture characteristics, with small cleavage facets and some tear ridges near the notch zone. Large river-patterned cleavage facets characterized the
fracture surface in the brittle zone. At a 50 kJ cm\(^{-1}\) heat input, the ductile zone contained a few large, shallow dimples and parabolic dimples, while the brittle zone included cleavage facets, some of which were joined together by the tear ridges. The dimples were larger and deeper in the ductile zone of the sample obtained at 75 kJ cm\(^{-1}\). The brittle zone’s cleavage facets were the smallest, and there were tearing edges along the borders of these cleavage planes. The dimples on the fracture surface of the simulated samples obtained at different heat inputs displayed different sizes and depths. This indicates that ductile fracture possessed different characteristics—the greater the depth and size of the dimple, the larger the specimen’s impact toughness. The optimum impact energy was obtained when the heat input was 75 kJ cm\(^{-1}\).

4. Discussion

4.1. Effect of heat input on the CGHAZ microstructures

Figure 3 shows the microstructure of the simulated CGHAZ with different heat input. The microstructure of the simulated CGHAZ changed from GB and a small amount of AF to PF and AF as the heat input increased. The type and content of microstructures varied greatly when the samples were subjected to different heat cycles. To better understand the effect of heat input on microstructure evolution of CGHAZ, the dilatometric curves of bar samples were measured during thermal cycling welding, as shown in figure 9. The phase transformation temperature of CGHAZ increased with the heat input increasing.

Figure 8(a) indicates that the starting and ending temperatures of the phase transformation of the sample under a heat input of 25 kJ cm\(^{-1}\) were 566 °C and 440 °C, respectively. Zhou et al [15] and Ravi et al [16] revealed that bainitic ferrite forms in the carbon-poor region at PAG borders and that some carbon from bainitic ferrite enters the surrounding austenitic matrix to form a carbon-rich austenitic matrix. Carbon-rich austenite transforms into martensite or remains as austenite during rapid cooling. Compared with other simulated thermal processes, the phase transformation time at a 25 kJ cm\(^{-1}\) heat input was the lowest, as presented in figure 8(b). Under these circumstances, a low phase transformation temperature (566 °C–440 °C) and a very short time (10 s) prevented the carbon atoms from diffusing, which facilitated the bainite transformation. In addition, a small, local carbon-rich austenite zone was formed due to the insufficient diffusion of carbon, which was converted to small M/A constituents at a lower temperature (figure 5(a)).

The starting temperature of the phase transformation increased from 566 °C and 712 °C as the heat input increased from 25 to 75 kJ cm\(^{-1}\). As seen in figure 8(b), the phase transition period increased as the heat input
increased. An elevated phase transformation temperature (712 °C–548 °C) and prolonged high-temperature residence time (150 s) promoted carbon diffusion, which was favourable for forming ferrite by diffusive phase transition. In addition, the PAG size grew from 48 to 53 μm as the heat input increased from 25 to 75 kJ cm⁻¹ (figure 9), which provided favourable conditions for AF transformation [17].

Jhs et al [18] stated that particles with a size of 0.2 ～ 2 μm could effectively promote ferrite nucleation, as shown in figure 10(a). However, the SEM images in figure 10(b) and figure 10(c) indicate that particle with sizes of ～ 100 nm can also act as ferrite nucleation sites; therefore, the micrographs of fine precipitates of the samples obtained at different heat inputs were examined (figure 11). The results illustrate that precipitated particles’ content increased and grew in size. The corresponding energy spectra prove that the V content in particles increased upon increasing the heat input (figure 11). V precipitated on the Ti(C, N) particles in the matrix so that the particle size became coarsened and the surface became V-rich. The surface V-rich particles can effectively promote ferrite nucleation because the mismatch between V(C, N) and ferrite was smaller than Ti(C, N). Furthermore, the Baker-Nutting orientation relationship ([001]α//[001]V(C, N)), [110]α//[100]V(C, N)) between ferrite and V(C, N) endowed the interphase boundary with low energy [11], which promoted the
nucleation of ferrite on V(C, N) precipitates. The growth of precipitated particles and the precipitation of V on Ti(C, N) particles improved particle nucleation by providing more favourable positions for ferrite nucleation.

The density of fine precipitates with the size around 80 nm increased sharply as the heat input increased from 25 to 75 kJ cm\(^{-1}\). The increased V-rich coarse particles provided favourable conditions for transforming intragranular massive ferrite and acicular ferrite. Therefore, the transformation of massive ferrite and acicular ferrite occurred at 75 kJ cm\(^{-1}\) (figure 10). Concurrently, the diffusion of carbon atoms was sufficient when the heat input was 75 kJ cm\(^{-1}\), so large carbon-rich austenite was formed. This carbon-rich austenite was transformed into M/A constituents at ultra-cold conditions, resulting in several blocky M/A components in the sample.

4.2. Effect of heat input on the CGHAZ impact toughness
Extensive studies have demonstrated that the impact toughness is associated with the M/A constituents [19], grain size [20], and microstructural components [21]. Due to the different thermal expansion coefficients between M/A and ferrite in the CGHAZ, there will be residual stress around and inside the M/A constituent during cooling. Furthermore, the hardness and elastic modulus of the M/A constituent differ from the matrix’s, resulting in uncoordinated displacement during the impact test deformation process. This will further increase the stress around the M/A constituents [19, 22]; therefore, the M/A constituents in the CGHAZ tend to be the crack sources. Previous studies have indicated that toughness deteriorates upon increasing the M/A constituent content and size. In this study, although the content and size of M/A constituents increased as the heat input increased (figure 5), the impact toughness of the sample increased instead of decreasing.

Microcracks in the fibre region near the notch and the propagation region were detected and compared at heat inputs of 25 kJ cm\(^{-1}\) and 75 kJ cm\(^{-1}\) to further understand the influence of M/A constituents on cracks. Figure 12(a) shows that the size of M/A constituents in the sample with a 25 kJ cm\(^{-1}\) heat input was small and secondary cracks initiated around the M/A constituents. At 75 kJ cm\(^{-1}\), micropores formed around the M/A constituents rather than microcracks. The M/A elements in the 25 kJ cm\(^{-1}\) sample, on the other hand, were modest and were dispersed in a chain shape. As a result, they are prone to the formation of fine microcracks. The M/A constituents of the 75 kJ cm\(^{-1}\) sample were distributed as isolated islands (figure 12(c)). Thus, micropores are formed by debonding from the matrix rather than slight cracks in the sample.

Figure 12(b) shows that the microcracks in the bainite structure were thin and long, while the micropores in the sample were wide and short. A more significant amount of plastic deformation occurred around the M/A constituents during the formation of micropores. More strain storage energy was consumed, resulting in stress release around the M/A constituent. In the crack propagation zone, crack initiation was easy at the M/A in both samples obtained at 25 and 75 kJ cm\(^{-1}\), but the crack in the 25 kJ cm\(^{-1}\) sample penetrated directly through the PAG and stopped at the PAG boundary (figure 12(b)). The crack in the 75 kJ cm\(^{-1}\) sample deflected the
intragranular ferrite (figure 12(d)); therefore, even if M/A grew under a 75 kJ cm$^{-1}$ heat input, it did not seriously deteriorate the impact toughness of the CGHAZ.

Grain refinement improved the toughness of the heat-affected zone. The formation of an IGAF in CGHAZ divided the primary austenitic grain into several parts, which resulted in grain refinement [6, 23], as shown in

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**Figure 12.** SEM micrographs showing the crack propagation paths in the simulated CGHAZ with heat inputs of 25 kJ cm$^{-1}$ (a) (b) and 75 kJ cm$^{-1}$ (c) (d).

**Figure 13.** The EBSD orientation images of simulated CGHAZ were obtained at heat inputs of (a) 25 kJ cm$^{-1}$, (b) 50 kJ cm$^{-1}$, and (c) 75 kJ cm$^{-1}$.
In addition, Hu et al. [24] confirmed that interlocked AF plates could hinder crack propagation, thus improving the impact toughness. The grain size and distribution of high-angle grain boundaries varied significantly with the heat input (Figure 13), which is the main reason for the different impact energies.

Figure 13 and Figure 14 illustrate the inverse pole and image quality figures associated with the boundary distribution and the accompanying histograms for a fraction of grain boundaries with various misorientations. The image quality maps display the morphology of different microstructures, in which the red lines and yellow lines represent grain boundary misorientations above and below 15°, respectively.

Previous studies showed that high-angle grain boundary (HAGB) had an important effect on crack propagation [23]. As the heat input increased from 25 to 75 kJ cm$^{-1}$, the effective grain size defined by 15° decreased from 10.6 to 5.6 μm, and the distribution of grain boundaries higher than 15° became more uniform. Figure 14 shows that the proportion of high-angle grain boundaries increased, and the fraction of low-angle grain boundaries decreased, as the heat input increased. The sample with a heat input of 75 kJ cm$^{-1}$ had the highest impact energy because it contained the most intragranular acicular ferrite. The presence of intragranular acicular ferrite refined the microstructure of the CGHAZ and made the high-angle grain boundaries distributed more uniformly, thus, enhancing impact toughness of CGHAZ.

5. Conclusion

This paper investigates the microstructures and impact properties of the simulated CGHAZ in low-carbon Mo-V-Ti-N-B steel with different heat inputs. The main conclusions are summarized as follows:

1. The cooling rate decreased as the heat input increased from 25 to 75 kJ cm$^{-1}$, and the PAG grew in size. Moreover, due to V precipitation, the size of precipitated particles increased, reaching the critical size for serving as the site of ferrite nucleation; this reduced the LB and GB and increased the AF and PF.

2. The CGHAZ microstructure changed from LB + GB to AF + PF as the heat increased from 25 to 75 kJ cm$^{-1}$. In addition, the MED of the ferrite grains decreased from 10.6 μm to 5.6 μm, and the fraction of boundaries at MTAs higher than 15° increased from 19.8% to 46.1%.

3. As the heat input increased from 25 to 75 kJ cm$^{-1}$, the impact properties of the CGHAZ increased, which is attributed to the IGAF content. The IGAF refined the microstructure of the CGHAZ and made the high-angle grain boundaries evenly distributed, which improved the toughness.

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