

Article

Effect of Heat Input on Weld Formation and Tensile Properties in Keyhole Mode TIG Welding Process

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Abstract: Keyhole mode Tungsten Inert Gas (K-TIG) welding is a novel advanced deep penetration welding technology which provides an alternative to high power density welding in terms of achieving keyhole mode welding. In order to facilitate welding procedure optimisation in this newly developed welding technology, the relationship among welding parameters, weld formation and tensile properties during the K-TIG welding was investigated in detail. Results show that except for travel speed, the heat input level also plays an important role in forming undercut defect by changing the plasma jet trajectory inside keyhole channel, leading to the formation of hump in the weld centre and exacerbation of undercut formation. Both undercut defect and root side fusion boundary can act as a stress concentration point, which affects the fracture mode and tensile properties considerably. The research results provide a practical guidance of process parameter optimisation and quality assurance for the K-TIG welding process.

Keywords: K-TIG; heat input; weld formation; tensile properties; welding procedure optimisation

1. Introduction

Tungsten Inert Gas welding, also known as TIG or GTAW, is a clean and widely used welding technology for metals. Nevertheless, a main disadvantage of this welding process is the limited penetration ability. It may require multipasses to complete the weld for medium thickness materials, which results in very low productivity. In order to overcome this drawback, K-TIG was developed as a new TIG variant to increase the penetration capability and production efficiency. In K-TIG welding, a free burning arc is applied as the heat source to melt the weldment by forming a cylindrical keyhole [1], which provides a nice alternative to high power density welding. This type of keyhole welding is characterized by its dependence on arc force, as opposed to ablation pressure in laser-based and electron beam-based welding, and a combination of stagnation and recoil pressure in plasma arc welding (PAW), as reported by Jarvis and Ahmed [2]. Compared with standard melt-in mode TIG welding, K-TIG process possesses higher energy density, process efficiency and greater penetration ability. Although heat density of K-TIG is far lower compared to high power density welding, it is easier to operate and more cost-effective, as stated by Liu et al. [3]. In addition, Liu et al. [4] reported that K-TIG was able to achieve continuous open keyhole and was more stable than PAW. It was also demonstrated by Liu et al. [5] that K-TIG had much wider operating window compared with PAW owing to the fact that it has much smaller arc pressure/arc current ratio.

To date, K-TIG has found applications in industries for joining medium thickness materials, say titanium [6], zirconium [7], stainless steel [8], low carbon steel [9] and dissimilar metal welding [10]. It was shown in these studies that K-TIG can complete the medium thickness weld in a single pass with a V-shaped morphology. It was also shown that K-TIG produced a welded joint with smaller fusion zone and larger heat-affected zone compared with conventional TIG welding process. In addition, it has been demonstrated that K-TIG has the potential to increase the productivity while maintaining the mechanical properties at a level similar to the joints produced by conventional TIG welding. Recently, it was reported by Cui et al. [11] that K-TIG can even be used in underwater condition and can produce a duplex stainless-steel weld that meets underwater welding standards. In addition, because of the fact that the material properties of some welded joints produced by K-TIG are not satisfactory, several optimisation technologies have been introduced into the K-TIG welding process. For example, Fei et al. [12] introduced filler materials into K-TIG to improve the weld microstructure and mechanical properties of armour steel weld. Fei et al. [13] also applied specially designed interlayer to tailor the microstructure and solidification sequence of K-TIG-welded armour steel joint. Post-weld heat treatment and ultrahigh frequency were utilised by Xie et al. [14] to improve both the mechanical properties and corrosion resistance of 430 ferritic stainless steel welded joint. On the other hand, since K-TIG has a narrower operating window for carbon steel because of its higher thermal conductivity, several studies have focused on the keyhole behaviour and stability improvement, such as the application of back purging unit [15], one pulse one keyhole technique [16] and high frequency pulse current waveform [17]. Furthermore, in order to develop on-line control strategy for the K-TIG welding process, several penetration monitoring systems have been developed, such as arc voltage-based oscillation frequency sensing [18] and combination of acoustics and vision-based sensing [19].

Overall, the current research directions of the K-TIG welding technology can be divided into four parts, namely metallurgical qualification of K-TIG welded joints, optimisation technology for joint microstructure and properties, process dynamics and stability improvement as well as development of on-line monitoring system. Up to date, there has been little research on the relationship among process parameters, weld formation and mechanical properties in the K-TIG welding process. More importantly, unlike conventional TIG welding for which the weld formation can be partially controlled or compensated by the operators, the K-TIG operates in a completely automated mode, with the weld formation being entirely dependent on the welding parameter combination. Although in some studies, small amount of welding trials were conducted to test the influence of welding parameters in the K-TIG welding, the main purpose was to find a parameter combination to achieve full penetration, as did by Cui et al. [8] and Feng et al. [20]. The mechanism for defect formation and the correlation between weld formation and mechanical properties in the K-TIG welding process have not been addressed. To a practical view, the mechanical properties, especially tensile properties, are easily affected by weld formation which is a function of process parameters used during welding. Figuring out the relationship among welding parameters, weld formation, and mechanical properties is helpful to the process optimisation and is crucial for the avoidance of defect formation and premature failure during service.

In this study, K-TIG welding was performed on 6.2 mm thickness high hardness armour (HHA) plates using different welding parameter combination, aiming at evaluating the effect of heat input on weld formation and tensile properties. The mechanism by which various behaviours occurred has been discussed. The results would provide a practical guidance of parameter optimisation and quality assurance for the K-TIG welding process.

2. Materials and Methods

The material used for investigation is 500 grade armour steel, also known as HHA. Detailed chemical composition and mechanical properties of the base plate are listed in Tables 1 and 2, respectively. The chemical composition was measured by atomic emission spectroscopy. The base

metal shows tempered martensite microstructure, as shown in Figure 1. It is worth noting here that as this paper is intended to analyse the relationship among welding parameters, weld formation and tensile properties, the material-dependent aspect will not be discussed in detail.

Table 1. Chemical composition of HHA (wt%).

C	Si	Mn	P	S	Ni	Cr	Mo	B	Fe
0.27	0.3	0.3	0.014	0.0025	0.19	1.05	0.25	0.0012	Bal.

Table 2. Mechanical properties of HHA.

Materials	Ultimate Tensile Strength (MPa)	Elongation (%)	Hardness (HV)
HHA	1775	14	495



Figure 1. Microstructure of the HHA base metal.

As depicted in Figure 2, the experimental equipment designed to conduct the welding trials was composed of a K-TIG power supply, a data acquisition device and a high-speed camera. The welding output was provided by a special power source with capability up to 1000 A. A high-speed camera in combination with a 10% transparency neutral filter was placed beside the worktable in order to record the arc behaviour. An inert gas back purge unit was designed to provide shielding from the atmospheric contamination on the root side of the weld pool. The argon shielding gas was transferred into the back purging unit from the left-hand side hose hole and allowed to be ejected through the switch on the right-hand side to avoid accumulation of pressure inside the unit. The image capturing was conducted after the process reached stable. During welding, the torch and high-speed camera were kept stationary while the worktable moved at the pre-set speed.

To test the influence of process parameters on weld formation and tensile properties, seven butt-joint welding experiments were performed on 6.2 mm thickness HHA plates by using control variable method. Process window identification tests were conducted to find the appropriate parameter combination leading to full penetration with the absence of both incomplete penetration because of insufficient heat input and weld pool collapse because of excessive heat input. Detailed parameter combination is depicted in Table 3. The heat input was calculated by the product of welding current and voltage divided by the travel speed plus a thermal coefficient of 0.6. The arc length was measured by the stacked thickness measurement device before welding. The dimension of the plates used in this study is 250 mm × 75 mm × 6.2 mm. The welding current and travel speed were chosen as variables. The fixed parameters for all experimental groups are depicted in Table 4. After welding, both transverse and longitudinal cross sections of the welds were cut from the weldments. The longitudinal cross section was cut in the arc extinguishment area in a bid to capture the in-situ profile during welding. The surface morphology and cross section macrographs were captured by Nikon digital camera and

Leica M205A stereomicroscope, respectively. The mounted samples were etched with 2 wt% nital for both macro and microstructure observation. Microhardness test was undertaken using Struers DuraScan-70 automatic hardness tester 2 mm below the surface with 0.5 mm interval under a load of 1 kg for 10 s dwell time. The preparation of tensile samples and the execution of tensile tests were in accordance with ASTM E8/E8M-16a guideline. The tensile samples were cut perpendicular to the welding direction using Wire Electrical Discharge Machining. Tensile tests were performed in as-cut condition without any finishing on the surface of tensile samples with 1 mm/min crosshead speed and 25 mm gauge length (Figure 3). The thickness of all tensile test specimens is 6.2 mm. A CCD camera was used to measure the total elongation by capturing the movement of two white spots with 25-mm interval on the tensile specimens. In order to ensure the reproducibility of tensile results, three samples were tested and the average value was considered.

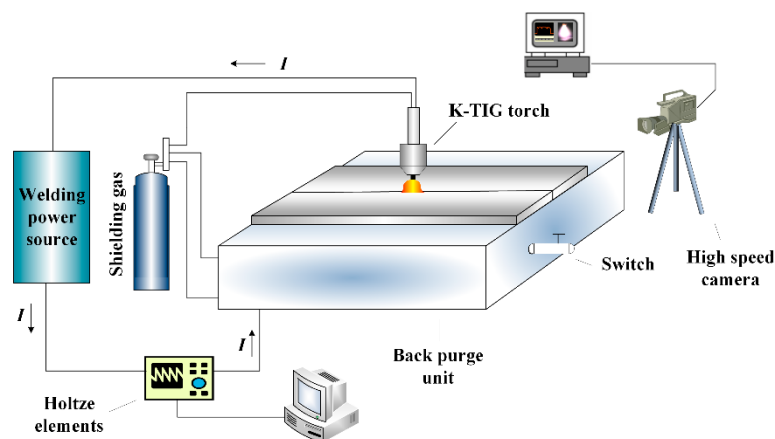


Figure 2. Schematic of experimental system.

Table 3. Welding parameter combination.

Test No.	Welding Current (A)	Arc Voltage (V)	Travel Speed (cm/min)	Heat Input (kJ/cm)
1	450	16.41	34.2	7.78
2	465	16.52	34.2	8.1
3	480	16.69	34.2	8.42
4	510	17.11	34.2	9.18
5	450	16.59	30	8.96
6	450	16.44	26	10.22
7	450	16.57	24	11.16

Table 4. Fixed welding parameters (all experiments).

Process Parameters	Details
Electrode material	Lanthanated tungsten
Electrode diameter	6.4 mm
Electrode tip angle	45 degree
Shielding gas	99.95% Ar
Shielding gas flow rate	20 l/min
Back purging gas	99.95% Ar
Back purging gas flow rate	10 l/min
Post flow shielding time	10 s
Arc length	3 mm

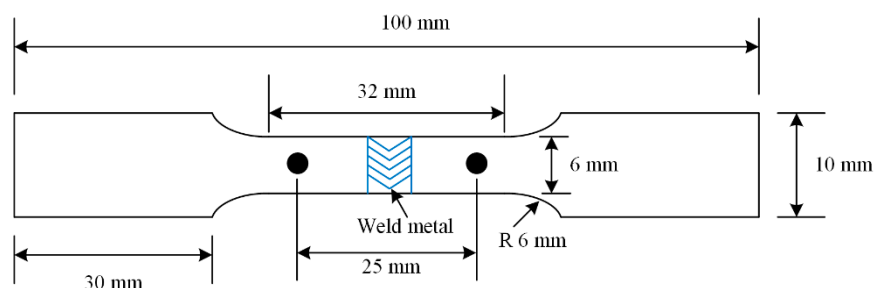


Figure 3. Dimension of tensile specimen.

3. Results and Discussion

3.1. Surface Formation

Tables 5 and 6 illustrate both the face side and root side surface formation for the weldments with different parameters. It can be observed that when the travel speed was fixed at 34.2 cm/min, undercut was experienced for all the weldments, as marked by red arrows and shown in Table 5. When the travel speed was reduced to 30 cm/min, the undercut defect disappeared, together with smooth surface formation on both sides, as shown in Table 6. A further decrease in travel speed led to wavy ripples along the face side weld edges, as highlighted by red circles, although undercut was not experienced. Also, even irregular bead face was produced on the root side when the travel speed was reduced to 24 cm/min in Test 7 sample.

Table 5. Weld surface formation with different arc current.

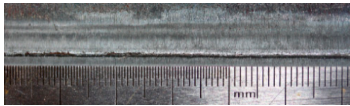
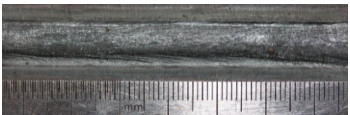
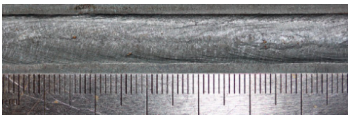
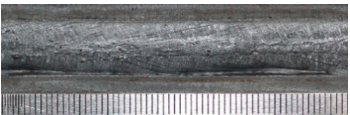
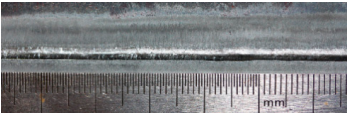
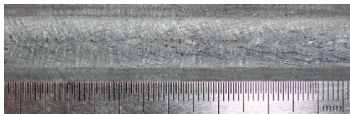
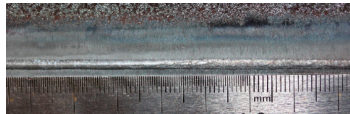
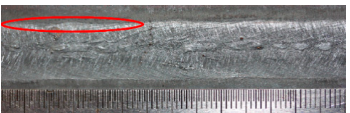
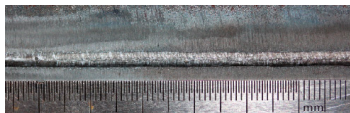
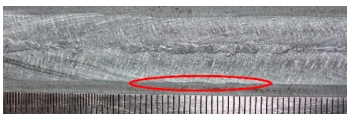
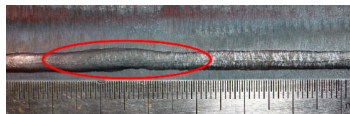
Test No.	Defect Type	Fixed parameters: Speed = 34.2 cm/min, Arc length = 3 mm		Variable Current
		Face side	Root side	
1	Undercut			450 A
2	Undercut			465 A
3	Undercut			480 A
4	Undercut			510 A

Table 6. Weld surface formation with different travel speed.

Test No.	Defect Type	Fixed parameters: Arc length = 3 mm, Welding current = 450 A		Variable
		Face side	Root side	Speed
5	N/A			30cm/min
6	Ripple			26 cm/min
7	Ripple			24 mm/min

It is known that high welding speed is easy to cause undercut formation. This is because the molten metal experiences much stronger backwards momentum at higher travel speed. The mass flow rate is the highest along the centreline of the weld, which tends to draw molten metal towards the centreline of the flow and induces lack of melt on the side walls on the face side [21]. Thus, undercut was consistently observed when high travel speed was used. Considering that the Tests 4 and 5 samples possess similar heat input and different travel speed, the presence of undercut in Test 4 may further indicate that travel speed plays a part in the formation of undercut. In addition, as reported by Tomsic and Jackson [22], high heat input makes the keyhole widen, more molten metal displaced by the keyhole flows back to the trailing region for a longer distance and a uniform bead is hard to obtain. This could be the reason for the formation of wavy ripples on the face side and irregular bead face on the root side. As the mitigation of undercut defect was observed with increasing arc current, it is reasonable to believe that the heat input also plays certain role in the formation of undercut in the K-TIG welding process. The effect of heat input on the undercut formation is discussed in following section.

3.2. Weld Longitudinal Cross-Section

The weld longitudinal cross-section and arc behaviour are shown in Figure 4. After arc was extinguished, the weld pool was suddenly frozen. The profile inside the keyhole channel was retained. When lower heat input was used, as in Test 1 sample, strong arc plasma was experienced on the face side, together with very high deflection angle relative to the horizontal line, as shown in Figure 4a. As the heat input increased, the amount of arc plasma on the face side decreased significantly, along with the increase in the slope of keyhole leading wall, as shown in Figure 4b. What is more, the deflection angle of plasma cloud relative to the horizontal line decreased with increasing heat input.

As can be seen from the schematic in Figure 4, when lower heat input was used, there was insufficient heat deposited on the lower half of the keyhole leading wall. Thus, the keyhole leading wall exhibited a severely curved shape. Only small amount of the arc plasma can be ejected via the keyhole exit, leaving majority of them guided towards the face side. As the heat input was increased, more heat was deposited on the keyhole bottom region. The bottom region (A zone) of the keyhole leading wall in Figure 4a was melted away into liquid and displaced to the rear part to form the weld, which makes the keyhole exit deviate less from the welding torch. Therefore, larger amount of the arc plasma could be easily ejected through the keyhole exit, leaving very small amount of them guided towards the face side. In addition, the increase in heat input further decreased the slope of keyhole rear edge, which allowed the plasma cloud to be deflected at a very small angle.

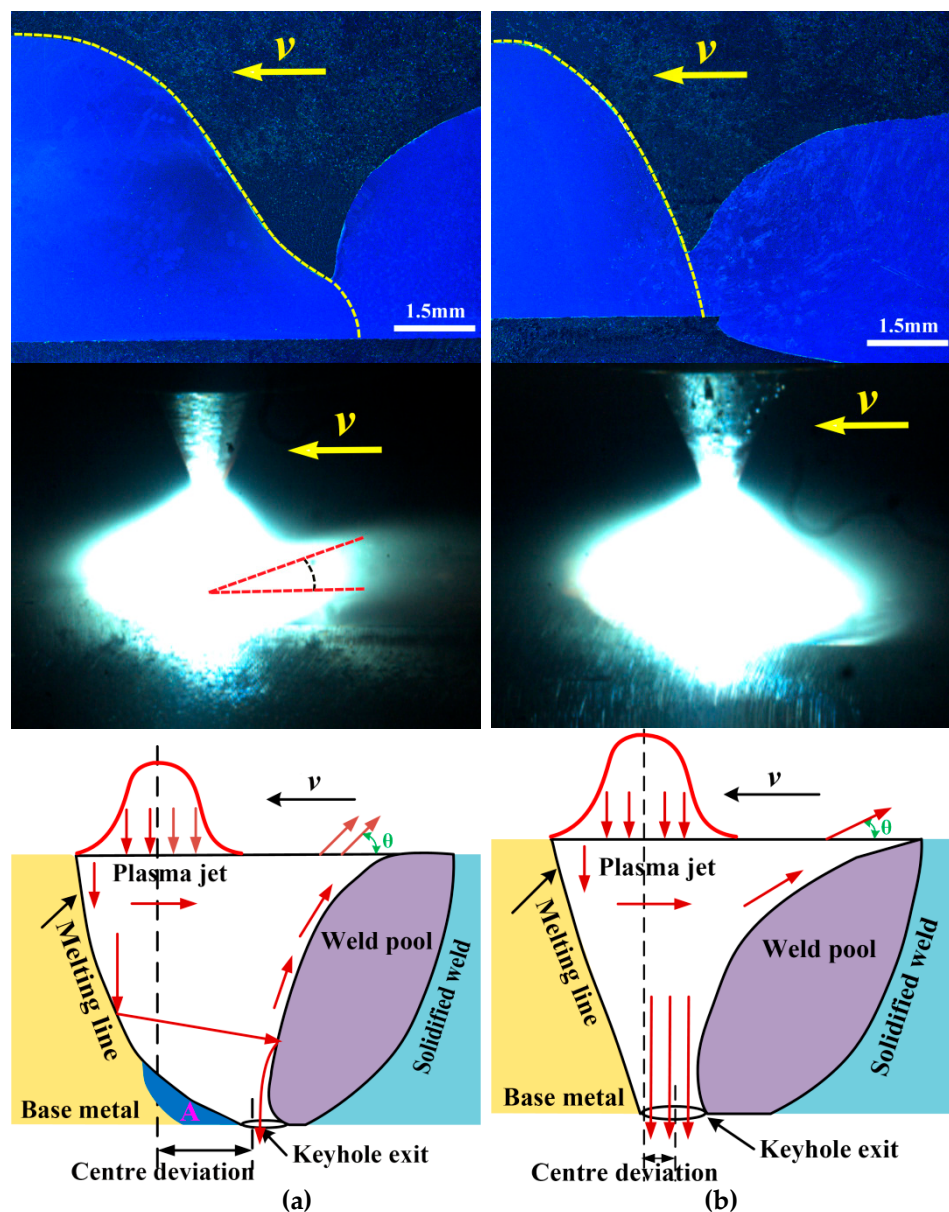


Figure 4. Weld longitudinal cross section and arc shape; (a) Test 1; (b) Test 7.

3.3. Weld Transverse Cross-Section

The weld macrographs are depicted in Figure 5. It can be seen that both the face side and root side width increased with increasing heat input. More importantly, there existed slight hump in Tests 1 and 2 samples, as shown in Figure 5a,b. The hump decreased and even disappeared with increasing heat input, as shown in Figure 5c,d. This is consistent with the mitigation or elimination of undercut defect.

As mentioned above, the heat input is also believed to be a controlling factor in the formation of undercut defect. Also, the variation trend of heat input, hump and undercut formation are consistent. Thus, it is necessary to consider the interaction among them. As mentioned above, when a high heat input was used, the keyhole front wall was less curved, majority of the plasma gas jet spurted out via the keyhole exit and very small amount of them was reflected by the keyhole front wall towards the rear keyhole wall, as indicated in Figure 6a. Therefore, the face side weld formation was smooth with the absence of humping and evident undercut. While if the heat input was reduced, the keyhole wall on the front side bent against the travel direction and the distance between the welding torch and the keyhole exit would increase, as demonstrated previously in Figure 4a. In this case, the plasma

gas flow field changed dramatically as depicted in Figure 6b. Almost all of the plasma jet was guided towards the lower region of the keyhole wall on the backside. As the welding current used in K-TIG welding process was relatively high, the pressure associated with the plasma jet flow could overwhelm the surface tension coming from the weld pool on the backside. Therefore, the trailing weld pool was severely deformed. Certain amount of the weld pool was pushed towards the face side, and hump appeared in the weld centre. This kind of hump would reduce the amount of liquid metal on the edges of the weld pool on the face side, which may facilitate the formation of undercut. This implies that the heat input level is also an important parameter in determining the undercut formation.

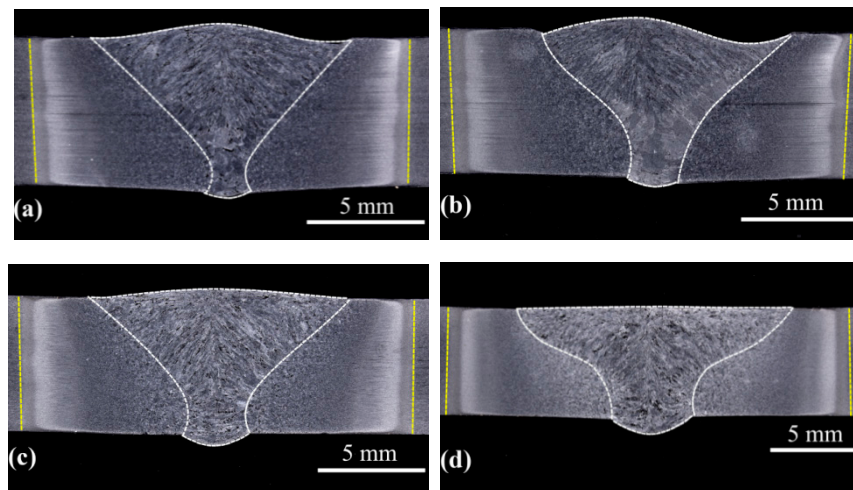


Figure 5. Macrographs of four typical welds with different heat input; (a) Test 1 (7.78 kJ/cm); (b) Test 2 (8.1 kJ/cm); (c) Test 4 (9.18 kJ/cm); (d) Test 7 (11.16 kJ/cm).

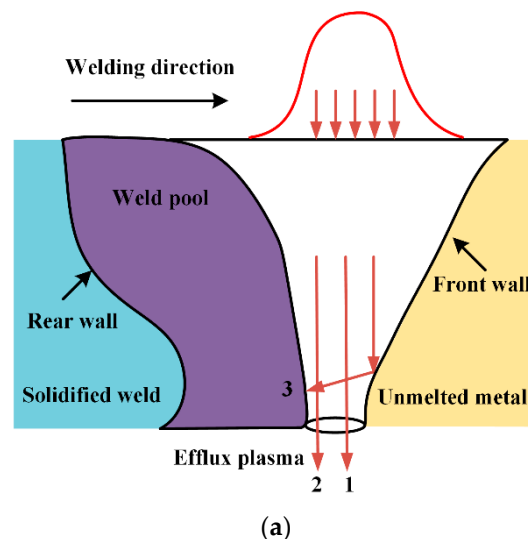


Figure 6. Cont.

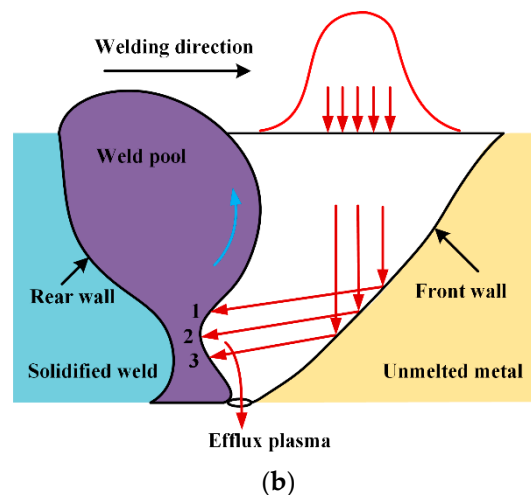


Figure 6. Schematic of plasma gas flow inside keyhole; (a) high heat input; (b) low heat input.

3.4. Weld Microstructure

The weld microstructure with different heat input is depicted in Figure 7. The weld metal in Test 1 consists of martensite and small amount of bainite, whereas only bainite with larger grain size is observed in the weld metal of Test 7. The same is true for the microstructure in the coarse grain heat-affected zone (CGHAZ) where a combination of martensite with bainite is present in Test 1 and only bainitic microstructure is found in Test 7. When it comes to fine grain heat-affected zone (FGHAZ), the microstructure consists predominantly of martensite. However, it is evident that the fraction of bainite in FGHAZ of Test 7 is higher than that of Test 1.

It is known that microstructure is closely linked to welding thermal cycle and hardenability. During welding thermal cycle, CGHAZ and FGHAZ were heated to austenite phase region. Upon cooling, the reformed austenite would transform to martensite, bainite or ferrite, depending on the cooling rate and hardenability of the base metal. Since the base metal contains large amount of carbon and chromium, the hardenability is relatively high, which favours the formation of martensite and/or bainite instead of ferrite. As the heat input increases, the cooling rate decreases, which favours the formation of bainite rather than martensite. The same is true for the weld metal as it still experiences phase transformation from austenite to various transformation products upon cooling, although it was melted during welding thermal cycle. That is why higher fraction of martensite was formed in the weld and HAZ with lower heat input.

3.5. Hardness

The hardness distribution of various welds is depicted in Figure 8. Lower hardness is observed in the weld metal region. The hardness increases when moving from the weld metal to the CGHAZ and FGHAZ. After that, the hardness decreases again to the lowest point when it comes to the over-tempered region. Finally, the hardness would increase continuously until it reaches the hardness value of the base metal (495 Hv). The hardness variation trend is very similar to that found in the armour steel weld produced with conventional fusion welding, as shown by Reddy et al. [23]. In addition, the overall hardness decreases with increasing heat input. Once the heat input reaches 10.22 kJ/cm and beyond, the overall hardness decrease becomes evident.

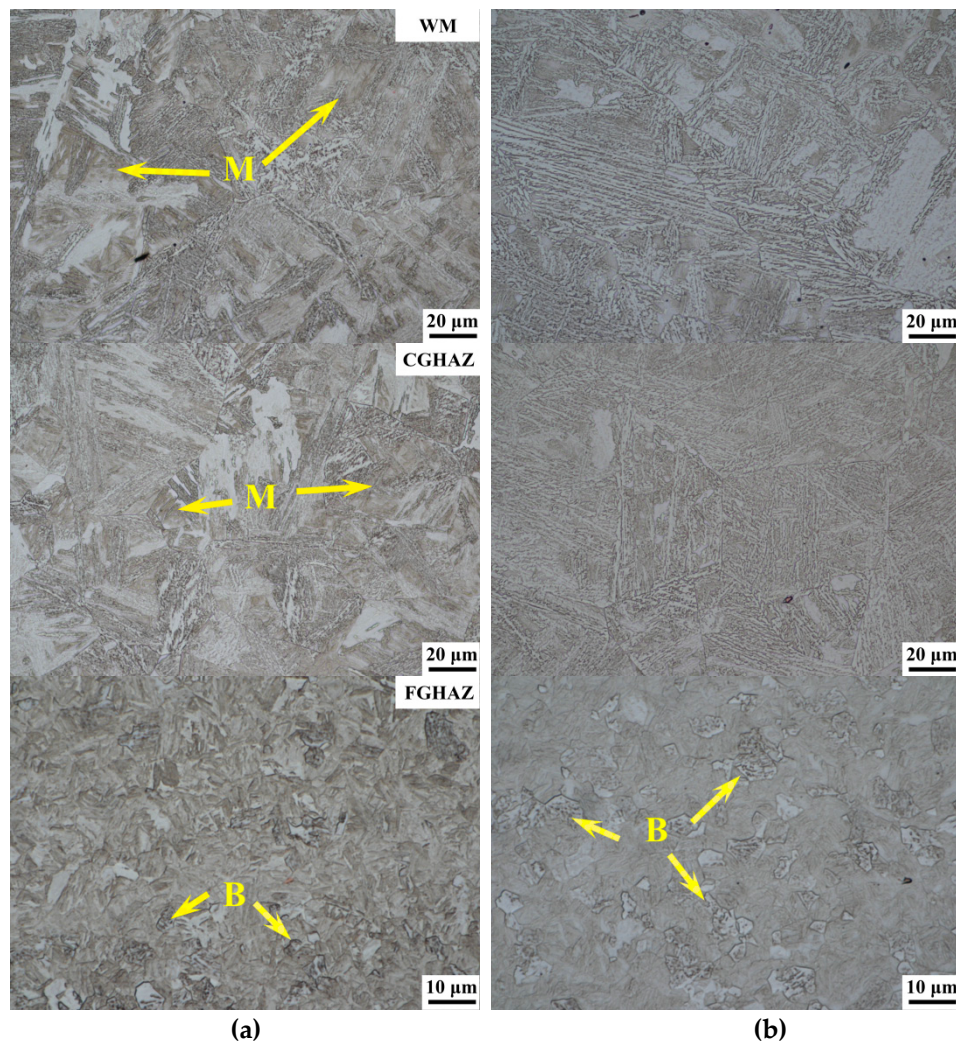


Figure 7. Microstructure in weld metal and heat-affected zone (HAZ); (a) Test 1 (HI = 7.78 kJ/cm); (b) Test 7 (HI = 11.16 kJ/cm) Note: M and B represent martensite and bainite respectively.

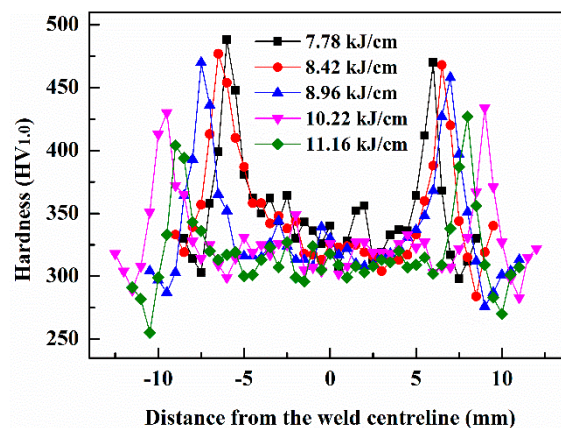


Figure 8. Hardness distribution across weld with varying heat input.

The variation trend of hardness across the weld is in good agreement with the microstructure change. As martensite and smaller grain size favours higher hardness, the increase in hardness from weld metal to FGHAZ is unexpected because of the increased fraction of martensite and decreased grain size from weld metal to FGHAZ, as shown in Figure 7. In the over-tempered region, the martensite

was heavily tempered, which led to the reduction in dislocation density, precipitation of carbides and reduction in solid solution strengthening in the matrix. Thus, the hardness in the over-tempered region decreased dramatically compared with the HHA base metal (495Hv).

When the heat input was increased, the fraction of martensite decreased in both the weld metal and HAZ, as shown in Figure 7, because of the slower cooling rate induced from higher heat input. Furthermore, the softening behaviour in the over-tempered region became severer because of prolonged holding time in this region with increasing heat input, leading to more significant reduction in dislocation density, precipitation of carbides and reduction in solid solution strengthening. That is why the overall hardness of the welded joints decreased with increasing heat input.

3.6. Tensile Properties

The tensile results of the welded joints with different heat input are shown in Figure 9, while the fracture location of each welded joint is shown in Figure 10. Two tensile samples for each setting were used to present the fracture location on both face and root side, with the dotted yellow lines representing the weld metal region on both face and root sides. Figure 9b shows that the elongation of Test 1 (6.47%) and Test 3 (7.65%) is much lower than the other four samples although the ultimate tensile strength (UTS) remains at a high level. The fracture behaviour of these two samples is depicted in Figure 10a,b respectively. They both fractured from the face side undercut. Once the heat input was increased to more than 8.96 kJ/cm, as is the case in Tests 4–7 samples, the elongation reaches more than 10%. Although the Tests 4–7 tensile samples all fractured in the weld metal, the Tests 4 and 5 samples fractured in the weld metal from the root side fusion boundary (Figure 10c,d), while the fracture location of Tests 6 and 7 samples were randomly located in the weld metal (Figure 10e,f). It is worth noting here that the wavy ripples seem to have no evident effect on the tensile properties in Tests 6 and 7 samples.

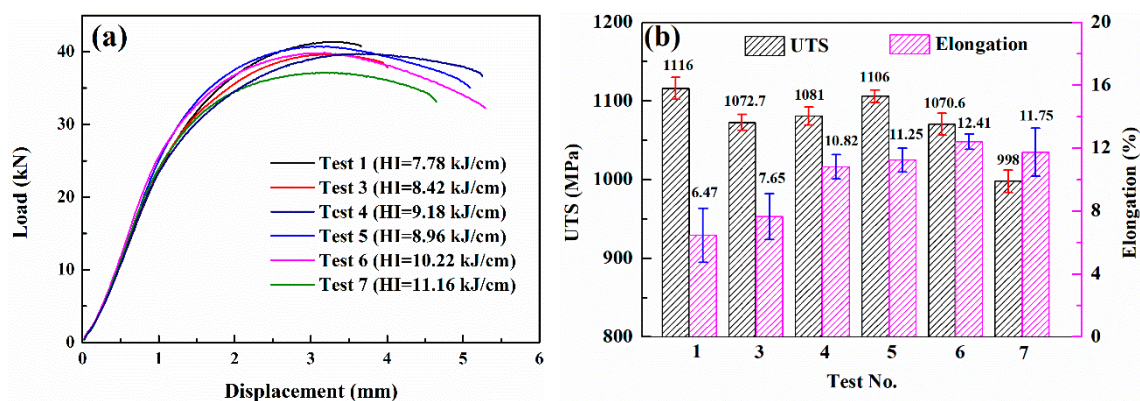


Figure 9. Tensile properties; (a) tensile curves for welded joints with different heat input; (b) tensile results for welded joints with different heat input.

It is well accepted that the tensile strength is closely linked to the overall hardness of the welded joint. With the increase in heat input from 8.96 kJ/cm to 11.16 kJ/cm, the overall hardness decreases, as shown in Figure 8, which is probably a result of the change in cooling rate and microstructure. In addition, higher heat input also tends to produce wider weld metal and HAZ, which could increase the width of reduced hardness region, as is the case in this study. Both these two factors can contribute to the decrease in tensile strength with increasing heat input. The significant reduction in elongation in Tests 1 and 3 is a result of undercut defect, which is a well-known stress concentration point. This can also be confirmed by fractography, as shown in Figure 11a,c and e. Note that the left-hand side on the fracture surface is the face side of the tensile sample, whereas the right-hand side is the root side of the tensile sample. The fracture surface in Test 1 sample consists predominantly of delamination both across two planes and on a single plane, as indicated by yellow and red arrows, respectively in

Figure 11a, as well as small amount of radial zone on the left. This means that after quick fracture initiation and propagation, the tensile sample experienced delamination fracture mode along the vertical direction down to the root side, as shown in Figure 11b. The recovery of elongation in Tests 4–7 samples indicates that the mitigation or elimination of undercut effect is an effective way to improve the tensile properties by reducing or eliminating stress concentration. However, it is worth mentioning that although the tensile properties of Test 4 were not affected by slight undercut in this study, it is still highly desirable to eliminate it in real fabrication. Even in the absence of undercut, the fracture still presents two different routes. As can be seen in Figure 11c, after crack initiated from the root side fusion boundary (dark grey area on the right-hand side), quick crack propagation occurred until reaching the face side, as evidenced by the completely radial zone in the fracture surface. The schematic is shown in Figure 11d. This indicates that the fusion boundary on the root side can also act as a stress concentration point in the absence of appreciable backside reinforcement, although it is less harmful than undercut. Thus, it is recommended that post-weld surface treatment be carried out for the K-TIG welded joint if the workpiece root side is accessible. With the formation of appreciable reinforcement on the root side, as is the case in Test 6, evident fibrous zone is present in the centre region, as indicated by the dotted yellow circle in Figure 11e. This is a result of plastic deformation occurring before crack initiated. In addition, there is larger amount of shear lip formed on the fracture surface, indicating that this sample followed a regular fracture route in which the stress condition changed to shear from tension and resulted in a fracture around 45 degree relative to the maximum stress direction after crack reached the edge. The schematic of fracture path is shown in Figure 11f.

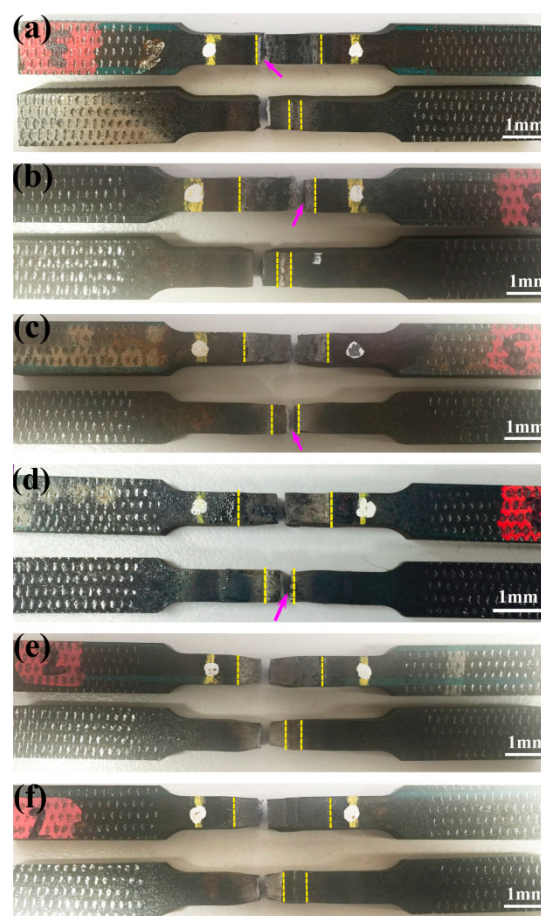


Figure 10. Fracture behaviour of welded joints; (a) Test 1 (HI = 7.78 kJ/cm); (b) Test 3 (HI = 8.42 kJ/cm); (c) Test 5 (HI = 8.96 kJ/cm); (d) Test 4 (HI = 9.18 kJ/cm); (e) Test 6 (HI = 10.22 kJ/cm); (f) Test 7 (HI = 11.16 kJ/cm).

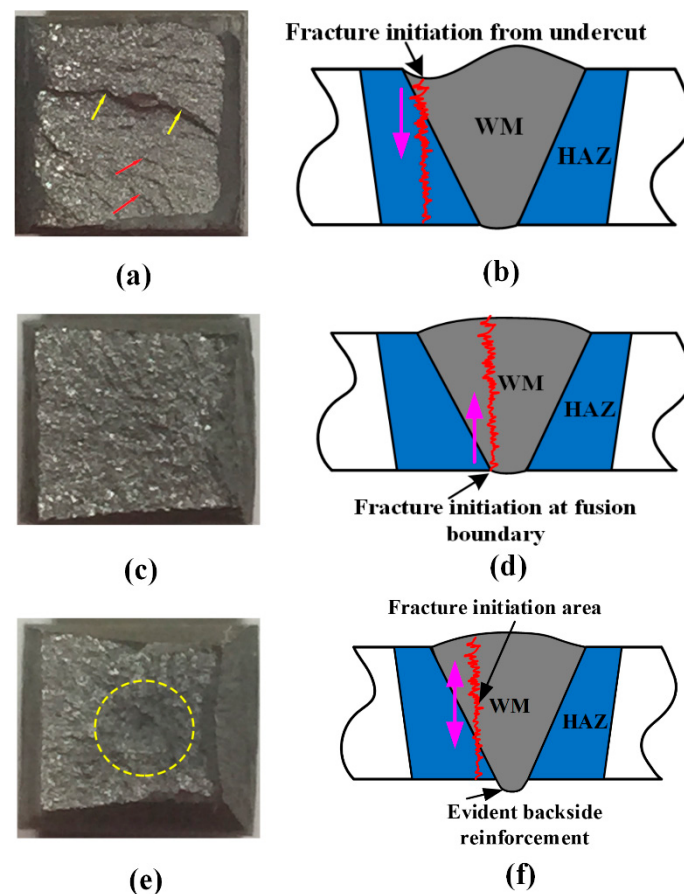


Figure 11. Schematic of fracture mechanism; (a,c,e): Fractography of Test 1 (HI = 7.78 kJ/cm), Test 5 (HI = 8.96 kJ/cm) and Test 6 (HI = 10.22 kJ/cm) respectively. (b,d,f) The corresponding schematic of fracture mechanism for these three samples.

It is worth noting that the much-reduced elongation would make the prediction of service life ineffective and lead to premature failure. In addition, as the stress concentration in the root side fusion boundary may also lead to premature failure in highly stressed or fatigue conditions, it is highly desirable and necessary to eliminate any stress concentration and avoid the formation of undercut defect in the K-TIG welded joint. It is suggested that the heat input level be carefully monitored during the welding process through either arc behaviour or other feedback signals described in the literature, together with appropriate use of welding speed. The results presented in this study can be useful in the real fabrication process and could be used as a guidance of parameter optimisation and quality assurance for the K-TIG welding process.

4. Conclusions

1. Undercuts were easily produced when the travel speed was fixed at 34.2 cm/min, while wavy ripples were experienced when the heat input was too high. The formation of hump in the weld centre is a result of low heat input, which changes the profile of weld longitudinal cross section and plasma trajectory inside the keyhole channel and leads to the exacerbation of undercut formation. This indicates that both travel speed and heat input are the contributing factors for undercut formation.
2. The tensile properties of the K-TIG welded joint were closely correlated to the weld formation. The undercut defect decreased the elongation considerably by imposing stress concentration and inducing delamination fracture mode. In addition, although the tensile properties were not appreciably affected by the height of root side reinforcement, stress concentration tended to occur

near the root side fusion boundary in the absence of appreciable reinforcement, which would also affect the regular fracture route. The wavy ripples produced with high heat input had no evident effect on the tensile properties. It is required that a combination of appropriate heat input level and travel speed be selected, leading to the absence of undercut defect, either appreciable root side reinforcement or a combination of post-weld surface treatment with the absence of appreciable root side reinforcement. It has been demonstrated that the welding parameters should be carefully controlled to avoid defect formation and maintain tensile properties in the K-TIG-welded joint.

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