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Investigation into the viability of K-TIG for joining armour grade quenched and tempered steel

Abstract

Motivated by significant loss of mechanical properties during conventional fusion arc welding processes owing to under matching filler materials used and low efficiency associated with multipass welding, the viability of keyhole tungsten inert gas (K-TIG) welding for joining armour grade quenched and tempered (Q & T) steel was presented. Single pass full penetration was achieved on 9 mm thick plates at a speed of 28 cm/min⁻¹ without using any filler materials and edge preparation. In-depth investigation into the weld was conducted by optical microscope, scanning electron microscope, electron back-scattered diffraction, microhardness and tensile test. The results show that the weld metal consists of dendritic structure and predominantly bainitic microstructure and is dominated by low angle grain boundaries. Hardness distribution across the weld is higher than current practice, which would lead to improved ballistic performance. Although the joint efficiency of the weld is 65% due to reduction in weld metal hardness, it is still much higher than that produced via conventional fusion welding, which is not surpassing 50%. It has been demonstrated that the K-TIG welding process offers a new way to weld medium thick armour grade Q & T steel with high efficiency and low cost, while maintaining the mechanical properties at a high level.

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Investigation into the viability of K-TIG for joining armour grade quenched and tempered steel

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Highlights

- The feasibility of K-TIG for joining medium thick (6-13mm) materials with relatively high thermal conductivity was first investigated.
- Single pass full penetration was achieved on 9mm thick armour grade quenched and tempered steel at a speed of up to $28\text{cm}/\text{min}^{-1}$ without using filler materials and edge preparation.
- Tensile properties and hardness distribution of weld were higher than current practice, which would lead to improved ballistic performance.

Graphical abstract

ABSTRACT

Motivated by significant loss of mechanical properties during conventional fusion arc welding processes owing to under matching filler materials used and low efficiency associated with multipass welding, the viability of keyhole tungsten inert gas (K-TIG) welding for joining armour grade quenched and tempered (Q&T) steel was presented. Single pass full penetration was achieved on 9mm thick plates at a speed of 28cm/min⁻¹ without using any filler materials and edge preparation. In-depth investigation into the weld was conducted by optical microscope, scanning electron microscope, electron back-scattered diffraction, microhardness and tensile test. The results show that the weld metal consists of dendritic structure and predominantly bainitic microstructure and is dominated by low angle grain boundaries. Hardness distribution across the weld is higher than current practice, which would lead to improved ballistic performance. Although the joint efficiency of the weld is 65% due to reduction in weld metal hardness, it is still much higher than that produced via conventional fusion welding, which is not surpassing 50%. It has been demonstrated that the K-TIG welding process offers a new way to weld medium thick armour grade Q&T steel with high efficiency and low cost, while maintaining the mechanical properties at a high level.

Keywords: K-TIG; Armour grade Q&T steel; Microstructure; Mechanical properties.

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1. Introduction

Quenched and tempered (Q&T) steels are much harder and stronger than normal carbon steel and they possess high ratio of strength to weight which improves their structure efficiency. This unique property is obtained through a Q&T heat treatment and from its chemical composition. In addition, they possess a particular combination of high

hardness, high strength with good weldability, toughness and ease of heat treatment, providing excellent resistance to shock and penetration. Therefore, they have been widely applied in situation where a combination of toughness, tensile strength, hardness as well as penetration resistance is required [1-4]. It is because the Q&T steels possess such unique property that they are mainly used for construction of hull and turret of military combat vehicles [5, 6].

Joining and welding of armour grade Q&T steels are certainly unavoidable in the fabrication process of armored vehicles. At current stage, welding techniques used to fabricate the weldment of this kind of steel are conventional fusion welding processes, such as Shielded Metal Arc Welding (SMAW), Flux Cored Arc Welding (FCAW) and Gas Tungsten Arc welding (GTAW) [7, 8]. However, due to the under matching filler materials used, such as austenitic stainless steel (ASS) filler materials and low hydrogen ferritic (LHF) filler materials, mechanical properties of the weldment deteriorates dramatically compared with base metal, especially the hardness in the weld metal region [9-11]. It is known that the harder the materials, the better the ballistic performance [12]. In a bid to find ways to increase hardness in weld metal region, a hardfacing technique was introduced in which a hardfaced interlayer made of either chromium rich carbide or tungsten carbide is deposited between capping and root pass of ASS or LHF weld metal [13-17]. Although improved hardness and correspondingly ballistic performance was reported due to the very hard interlayer materials, the extremely inhomogeneous microstructure along the thickness direction in the weld metal region may not produce satisfactory mechanical properties, such as UTS and elongation. Furthermore, groove preparation, multipass welding, along with the use of large amount of expensive hardfaced interlayer filler materials make this technique less efficient and cost-effective.

In order to fabricate the joint of armour steel with low cost and high efficiency, the feasibility of hybrid Laser-Gas Metal Arc welding process for welding armour steel was

studied by Kuzmikova et al. [18]. Besides improved hardness in weld metal region compared with conventional fusion welding process, it was also found that welding time and consumable usage can be reduced by 36% and 56% respectively due to the reduced deposition pass and fusion zone area. However, hybrid Laser-Gas Metal Arc welding process requires two welding equipment to operate at the same time, which intangibly increases the capital cost and complicates the control over welding parameters.

More recently, EI-Batahgy et al. [19] investigated the feasibility of friction stir welding (FSW) for welding 1640MPa high strength Q&T steel. A defect free and full penetration weld with thickness of 2.2 mm was successfully achieved by FSW at a travel speed up to 600mm/min⁻¹. In addition, the hardness in the weld metal region is similar or even higher than base metal when peak temperature experienced in weld metal region exceeds A_{c1} line. Nevertheless, the thickness of the high strength Q&T steel welded by FSW was merely 2.2 mm. Considering the critical durability and fatigue resistance of the FSW tool, FSW might not be suitable for welding armour grade Q&T steel with medium thickness. It is obvious that there is a need to find a process that is able to produce armour steel joint with high efficiency, low cost and superior mechanical properties, especially hardness.

Keyhole mode Tungsten Inert Gas Welding, also known as K-TIG, was invented by Commonwealth Scientific and Industrial Research Organisation (CSIRO). It is a process variant of the standard TIG welding, with the difference being that its functional mode leads to a keyhole shaped weld bath, just as the ones present in laser beam welding (LBW), electron beam welding (EBW) and plasma arc welding (PAW) process [20]. In K-TIG welding process, the cathode emission area is focused to a much smaller area compared with standard TIG, thus the current density and pinch effect is increased in the arc column. Once the welding current is increased to a high level, say more than 300A, the arc pressure is sufficient to overcome the surface tension and hydrostatic head and it will displace the

molten metal all the way through thickness to form a keyhole. Unlike EBW, LBW and PAW which rely on ablation pressure or stagnation pressure to create the keyhole, K-TIG uses a free-burning arc as the heat source and the pressure used to displace the metal aside is originated from the interaction between arc current and its self-induced magnetic field, which is completely electromagnetic in origin. Single pass full penetration weld with medium thickness (6-13mm) can be readily achieved without using any filler materials and edge preparation at a speed of up to 1000mm/min⁻¹, making it much more productive and efficient than conventional fusion welding. Furthermore, relatively simple torch configuration and high current capacity allows K-TIG welding process to be more simplified in the selection of welding parameters and much faster (3-folds) than PAW. K-TIG also has advantages over LBW and EBW in terms of equipment cost and being tolerant of joint preparation since the latter two tend to produce a slim weld seam, which leads to a poor gap bridging ability and requires high precision edge preparation and set-up [18, 21].

The successful operation of keyhole mode TIG welding relies on many factors, including physical properties of materials and welding parameters. For the K-TIG weld to be completed the keyhole has to close behind the arc, and this places a fundamental constraint: the molten metal must be held in place by surface tension while it is solidifying [22, 23]. This constraint in turn leads to a balance equation in the rear weld pool between surface tension and hydrostatic head as shown below:

$$wh \leq \gamma / \rho g \quad \text{Equation (1)}$$

Where γ is surface tension, w is the width of the root bead, ρ is the density of materials, g is acceleration due to gravity and h is the plate thickness. It is evident from Eq.(1) that the width of root bead will determine the maximum thickness that could be welded by K-TIG welding process. As the width of the root bead is going to increase with increasing

thermal conductivity [24], the possibility of destabilizing the keyhole is increased with increasing thickness and/or thermal conductivity. It is largely for this reason that K-TIG welding process has been confined to materials with low thermal conductivity, such as stainless steel and titanium alloy [23].

Due to the nature of high productivity and low cost, joining of mid-thickness materials with low thermal conductivity by K-TIG welding has already been adopted and demonstrated in industry, such as CP titanium by Lathabai et al. [23], zirconium by Lathabai et al. [25], stainless steel by Lohse et al. [26] and Feng et al. [27]. Motivated by applying K-TIG to weld materials with relatively high thermal conductivity, such as carbon steel, Fang et al. [28] and Liu et al. [29] tried to use K-TIG to weld Q345 and 16Mn structural steel respectively in the form of bead-on-plate. Although favorable mechanical properties were achieved on both cases, the thickness of materials that subject to investigation was less than 6mm (5.5mm). Overall, it is obvious that K-TIG technology is relative mature in terms of welding materials with lower thermal conductivity. At current stage, welding of carbon steel via K-TIG process is confined to bead-on-plate welding and thin plates which are less than 6mm thick. As mentioned above, the advantage of K-TIG is to weld medium thickness materials at high speed. Thus, in order to fulfill its maximum potential, the feasibility of K-TIG for joining carbon steels with medium thickness should be evaluated, especially for butt joint welding.

In this paper, K-TIG welding was conducted on 9 mm-thick armour grade Q&T steel in flat position. The effect of varying heat input on the surface morphology were analyzed, followed by the characterization of microstructure in both weld metal and heat-affected zone (HAZ) as well as mechanical properties of the weldment. The research results may pave ways for use of K-TIG to weld medium thick armour plates.

2. Experimental procedure

2.1 Materials

The armour grade Q&T steel subject to investigation in this paper conforms to the requirement of MIL-DTL-46100 and falls into the category of a family of Q&T steels for armour applications. This high hardness armour steel is manufactured by BluscopeSteel utilising the Basic oxygen steelmaking (BOP) process, including vacuum degassing and calcium treatment, followed by hot rolling to desired thickness and heat treatment involving quenching and low temperature tempering to achieve desired combination of hardness and toughness. Two plates, each with the dimension of 150mmx150mmx9mm, were applied to prepare for the butt joint plates. The chemical composition of the investigated armour steel is listed in Table 1. The microstructure of the parental metal is present in Fig. 1 and is mainly composed of tempered martensite with typical lath shape.

Table 1

Chemical composition of investigated armour steel (wt-%)

C	Si	Mn	P	S	Ni	Cr	Mo	Al	Cu	Nb	Ti	V	B	N	Ca	Fe
0.27	0.3	0.3	0.014	0.0025	0.19	1.05	0.25	0.052	0.018	0.0035	0.022	0.04	0.0012	0.0046	0.0008	Bal.

Fig. 1 SEM image of the parental metal

2.2 Welding set-up

The welding machine used in this study is manufactured by Keyhole TIG Limited (an Australian company). It consists mainly of K-TIG 1000 AMP power supply, control cabinet, water cooler, specially designed K-TIG torch and a control computer used to sense and display welding parameters, such as welding current, arc voltage. The torch was operated in DCEN mode. Both bead-on-plate and butt joint trials were conducted using a Bugo linear track to control the speed and movement of the torch as well as the arc length between the electrode tip and the workpiece. A clamping unit is used to fix the specimen on the worktable in order to prevent distortion resulting from thermal expansion. The schematic of the welding system is shown in Fig. 2. 6.4mm diameter tungsten electrode was used with small addition of lanthanum in order to provide suitable electrical characteristics. The electrode tip angle used is 45 degree. The surface of the sample near the edge on both sides was ground before welding in order to remove surface contaminants, such as oil, rust and oxide scales. The edges of the two plates were in close contact with each other with no joint gap. The square edges were milled till 90 degree and then degreased by acetone. The joint was manually aligned with the torch before welding.

Some bead-on-plate welding trials were first conducted on 9mm armour grade Q&T steel before identifying the optimal combination of welding parameters. Then, four different heat inputs were used via changing welding current to test its effect on the surface formation through butt joint welding trials. During the butt joint welding trials, other welding parameters, which are shown in Table 2, are all kept constant. Subsequent to the welding, detailed joint characterization was conducted on the joint with the best surface formation to evaluate the integrity and quality of the weld.

Welding parameters

9

2.4 Macro- and microstructural characterization

The photos of surface morphology of the four weldments were taken by a Nikon digital camera. Leica M205A stereomicroscope was utilized to capture the overall profile of weld cross section of the four weldments which were sectioned transverse to the welding direction. The extracted samples were hot mounted by Struers Citopress 20 hot mounting machine using Polyfast resin. Subsequently, the mounted samples were subjected to grinding with 220# SiC paper for 3.5 minutes, followed by polishing with different size of cloth from 9 μ m down to 1 μ m using Struers TegraPol 21 automatic grinder and polisher. The polished samples were etched in 2% nital to reveal the joint microstructure which was examined by both Leica DMRM Optical Microscope (OM) and JEOL JSM-7001F Field Emission Gun Scanning Electron Microscope (FEG-SEM). Energy dispersive spectroscopy line scanning analysis was undertaken using an Aztec 2.0 Oxford EDS system to reveal the precipitates in the HAZ area. Detailed microscopic analysis, such as misorientation angle distribution map, band contrast map and phase map were conducted using the same SEM equipped with an electron backscatter diffraction (EBSD) system. The EBSD sample was mechanically polished up to colloidal silica stage. The SEM equipment was operated at an accelerating voltage of 20 kV and probe current of 5 nA with the sample tilted 70 degree relative to horizontal in order to generate sufficient Kikuchi pattern intensity. The resolution of CCD camera was 1344x1024 pixels and 4x4 binning was used. In order to eliminate spurious boundaries caused by orientation noise and retain orientation contrast, boundaries with misorientation less than 2° were not taken into account. The data was analyzed by HKL Channel 5 software developed by Oxford Instruments HKL.

2.5 Mechanical test

Vickers microhardness test and transverse tensile test were carried out in accordance with ASTM E384-16 and ASTM E8/E8M-16a guidelines respectively. The hardness

survey was conducted using Struers DuraScan-70 automatic hardness tester at a depth 2mm below the weld top surface across all the weld region, including weld metal, HAZ and parental metal. The results were obtained at an interval of 0.5mm under a load of 1kg for 10s dwell time and then averaged to gain representative hardness value with lower error. The tensile specimens was extracted in a direction perpendicular to the welding direction, with the weld bead located in the middle of the specimen. The tensile tests were carried out at room temperature at a constant strain rate of $1\text{mm}/\text{min}^{-1}$ using Instron 8800 universal testing machine with capacity up to 500KN. An extensometer with a nominal gauge length of 50 mm was used to measure the deformation of the specimens during tensile test. The tensile samples of parental metal were sectioned in the same direction as those of weld joint. Leica M205A stereomicroscope and JEOL JSM-6490 SEM were used to observe macro and micro fracture surface morphology respectively to analyze the fracture mechanism. For tensile tests, three samples were tested and the average value were recorded in order to validate the reproducibility of the results. The tensile specimen dimension are shown in Fig. 3.

Fig. 3 Dimension of transverse tensile test specimen

3. Results and discussion

3.1 The effect of varying heat input on surface morphology

The front side and back side surface morphology of four different weld beams are shown in Fig. 4. When the applied welding current is 525A, the arc force and heat input are insufficient to fully penetrate the plate, as shown in Fig. 4(a,b). Furthermore, the weld bead width on the front side is inconsistent, ranging from 16.5mm at the beginning to

12mm at end. It is because at the beginning the plate is relatively cool, the heat is unable to dissipate through the backside due to insufficient penetration. Therefore, all the heat energy concentrates within the plate, which makes the top weld width become wide. As the welding proceeds, the thermal condition of the plate is improved and arc plasma can partially emit through the weld backside, which lowers the heat energy inside the plate as well as narrows the width on the weld top. Once the current is increased to more than 545A, the phenomenon of inconsistent top weld width disappears as shown in Fig. 4(c,e). This indicates that consistent keyhole was established with 545A and 565A welding current. When it comes to 575A, the excessive heat input and arc force disrupt force balance field inside keyhole channel, making the weld cool collapse in some area, as shown in Fig. 4(g,h). The results suggest that for K-TIG welding of 9mm armour grade Q&T steel, stable keyhole process can be achieved with current range of 545-565A at travel speed of $28\text{cm}/\text{min}^{-1}$, which is narrower compared with stainless steel as reported in [30].

Fig. 4 Weld surface morphology: (a,b) 525A, (c,d) 545A, (e,f) 565A, (g,h) 575A

The cross-sectional macrographs of the four joints are shown in Fig. 5. Fig. 5a clearly shows incomplete penetration on the backside of the weld, as indicated by red arrow. The front side melting width increases from 12mm to 13.2mm, while the root width increases from 1.9mm to 3.2mm with increasing welding current, as shown in Fig. 5e. The large error bar for 575A sample is a result of slight weld pool collapse, resulting in intermittent depression and inconsistent melting width. Furthermore, it is evident that the root bead reinforcement also increases with increasing welding current. It is interesting to note that the fusion zone outline changes from curve to straight line when the slight collapse occurs, especially at the keyhole lower region, suggesting that more heat is delivered to the plate lower region. This phenomenon is accompanied by the change in heat-affected zone outline from trapezoid to rectangle. As shown in Fig. 5(b,c), full penetration sound weld joint was produced without any visible defects, such as undercut, cracks, lack of fusion and porosities. The characterization of the weldment with the best surface morphology is presented in the following sections.

Fig. 5 Cross section macrographs of four plates produced by different welding current (a) 525A, (b) 545A, (c) 565A, (d) 575A and (e) Weld width variation

3.2 Microstructure

Fig. 6 is the cross-section image of various sub-zones of joint in as-welded condition. The bowl-like shape weld consists of a weld metal region (WM), coarse grain heat affected zone (CGHAZ), fine grain heat affected zone (FGHAZ), intercritical heat affected zone (ICHAZ), over tempered region (OT) and base metal (BM). It is well known that grains prefer to grow in the direction of heat gradient during welding solidification process [31]. The WM region in Fig. 7a presents typical solidification dendrites which grow from fusion boundary with lower temperature to the weld centerline with higher temperature in a direction perpendicular to the fusion line, as highlighted by the red arrow. The formation of dendrites in WM is attributed to the high heat input of K-TIG welding,

which leads to a weld pool with relatively large volume as well as low cooling and crystallization rate. Therefore, the cooling rate is not sufficient for the formation of equiaxed grains in the WM region [27]. High magnification OM image in Fig. 7b clearly depicts that the WM consists of plume-like microstructure with aligned carbide characteristic of upper bainite, along with lower bainite with carbide particles between as well as within subunits, both of which form by nucleation from prior austenite grain boundary and growth of bainitic ferrite sub-units and precipitation of cementite from residual carbon-enriched austenite in and between them [32]. In addition, sympathetic nucleation of bainite plates from existing sheaves is also observed, which is another common feature of bainite transformation [33]. High magnification SEM image in Fig. 7c also indicates the formation of small amount of granular bainite with the M-A constituents dispersed in the polygonal or bainitic ferrite matrix, which tends to form during continuous cooling. The M-A constituent is associated with carbon enrichment resulting from carbon partitioning during polygonal or bainitic ferrite transformation [34]. The upcoming partial transformation from austenite to martensite lowers the M_s temperature and drives M_f locally further and further down, eventually resulting in certain amount of untransformed austenite. In addition, accommodation of transformation strain via plastic deformation generates dislocations in austenite, which could also stabilize the austenite through mechanical stabilization [35]. Further investigation into the retained austenite (RA) in the fusion zone was carried out using EBSD obtained with 0.075 μm step size in order to detect small RA. The combined band contrast map and inverse pole figure of austenite was shown in Fig. 7d. It is noteworthy that only the austenite larger than 225nm can be detected by EBSD due to the 75nm step size utilized. The fraction of RA in fusion zone is around 1.5% and they are mainly distributed adjacent to prior austenite grain boundaries as circled by red color as well as between bainitic laths.

Fig. 6 Macrographs of joint cross section showing different regions

Fig.7 Microstructure of weld metal: (a) OM image of fusion boundary, (b) High magnification OM image of WM, (c) SEM image of WM, and (d) band contrast and inverse pole figure maps of austenite in WM (M:martensite, UB: upper bainite, LB: lower bainite, M-A: martensite and austenite constituents)

In full austenitic region which is composed of CGHAZ and FGHAZ, all of the tempered martensite has transformed to austenite which decomposes upon cooling and different kinds of products are formed, depending on the cooling rate and chemical composition of steels in question, or hardenability. In CGHAZ the temperature

experienced is relatively high so that the redissolution of large amount of stable carbides reduce the pinning effect of the second phase on dislocation movement and grain growth, leading to an increase in size of austenite grain [36, 37]. High carbon concentration and addition of austenite stabilizers significantly increase the hardenability, with the results being that large amount of martensite and small amount of bainite are present in this region even with air cooling as indicated in Fig. 8a. It was clearly shown in Fig. 8b that except for grain size difference, the fraction of martensite in FGHAZ increases compared with CGHAZ. In principle, smaller prior austenite grain size results in lower hardenability because the grain boundary area per unit volume increases with decreasing prior austenite grain boundary. The nucleation sites for pearlite and ferrite are increased in number, which speeds up these transformations and in turn leads to the decrease in hardenability [32]. This unusual phenomenon might be a result of faster cooling rate experienced due to lower peak temperature experienced in FGHAZ as well as high hardenability of the materials itself. This also indicates that the effect of cooling rate dominates over that of grain size in terms of hardenability. This situation was also reported in [38] during submerged arc welding of Q&T steel.

Fig. 8 SEM images of full austenitic region: (a) CGHAZ, and (b) FGHAZ. (M: martensite, B: bainite)

The ICHAZ, which refers to the region that has been heated to temperature between A_{c1} and A_{c3} , consists of freshly formed martensite, bainite and polygonal ferrite. The

increase in fraction of polygonal ferrite at the expense of martensite and bainite in ICHAZ is evident as depicted in Fig. 9a, which is attributed to lower hardenability resulting from smaller austenite grain size compared with CGHAZ and FGHAZ as well as smaller fraction of reformed austenite in the intercritical region. However, it is worth noting that the polygonal ferrite in this region may form via two different mechanisms. One comes from the nucleation of austenite at ' α ' grain boundaries when being reheated to the intercritical temperature regime that lies between A_{c1} to A_{c3} and transformation to martensite, bainite and polygonal ferrite upon cooling [37]. The other is a result of elimination of the fine martensitic crystals, say laths, and formation of equiaxed ferritic grains during high temperature tempering, a variation in microstructure ascribed to recovery and recrystallization mechanisms associated with the migration of the low angle boundaries that separated parallel martensite crystals with the same orientation as well as the rearrangement of the large angle parallel boundaries to form equiaxed ferritic grains in order to minimize grain boundary energy [32, 39]. The recovery process usually involves migration of boundaries with lower dislocation density, say lath boundaries, and dislocation cells towards higher dislocation density boundaries, i.e., block and prior austenite grain boundaries, which leads to complete disintegration of the laths [40]. High magnification of SEM in Fig. 9b clearly shows the equiaxial non-aligned carbides inside ferrite crystallites, which were further investigated using EDS line scanning. Fig. 9c reveal that the iron content decreases appreciably across the matrix-particle interface, whereas the carbon concentration saw an evident increasing trend, with the content of other alloy elements kept almost constant. This indicates that the carbides formed in the annealed martensite are cementite particles. It is because the diffusion rate of substitutional alloy elements is very slow, they are unable to partition into the cementite or to form alloy carbide due to very short holding time during welding process, although the peak temperature experienced is relatively high in the ICHAZ.

Fig. 9 Microstructure of ICHAZ: (a) low magnification SEM image, (b) High magnification SEM image of red rectangular area in (a), and (c) EDS line scanning across particles in (b) (F: Polygonal ferrite, C: Carbides)

The OT region presents typical tempered martensite microstructure similar to the base metal with the exception of large amount of cementite precipitated both along prior austenite grain boundaries and inside grains as shown in Fig. 10a. These cementite are the representative product associated with high temperature tempering during which cementite nucleates from transition carbide and then grows rapidly with increasing temperature [39]. High magnification image in Fig. 10b depicts the existence of both plate-like cementite and quasi-spherical cementite as highlighted by yellow arrow and red arrow respectively. Compared with very fine plate-like carbides in base metal shown in Fig. 10c, both the two kinds of cementite in OT region is much larger owing to the growth of cementite via ripening effect at high tempering temperature. It was reported that non-

isothermal tempering occurs generally during welding involving rapid heating, negligible holding time at tempering temperature. Unlike the coarse and spheroidized cementite in isothermal tempering, the formation of fine plate-like and quasi-spherical cementite was primarily ascribed to the synergistic effect of delay in cementite precipitation and insufficient holding time for carbon diffusion due to rapid heating that suppress the coarsening and spheridization of cementite [40]. What is more, it was also reported that the temperature at which cementite precipitation initiates is increased by increasing the heating rate to tempering temperature, which leads to a higher nucleation rate and a finer dispersion of cementite as a result [41]. This phenomenon was also present in the subcritical zone of laser beam welded joint and documented in [42]. As shown in Fig.10d, the EDS line scanning across the particle-matrix interface in the OT region presents similar trend as that in ICHAZ, further confirming that the particles formed in the OT region are also cementite and that the holding time is insufficient for the substitutional atom to rearrange.

Fig. 10 Microstructure of OT region: (a) SEM image of OT region, (b) high magnification SEM image of OT region, (c) high magnification SEM image of BM, and (d) EDS line scanning across particles in (b)

3.3 Microhardness

The Vickers microhardness profile of the weld is shown in Fig. 11. It is generally accepted that the hardness is closely related to grain size, phase balance and dislocation density. The hardness of WM ranges from 320HV to 360HV, which is in good agreement with the predominant presence of bainite in this region. It is evident that the highest hardness (458HV) is present at the full austenitic region where large fraction of untempered martensite containing large amount of dislocations forms with small amount of bainitic structure. The hardness gradually decreases while moving on to ICHAZ due to the lower fraction of austenite formed and subsequent decomposed product, such as martensite and bainite. It is worth noting that the freshly transformed martensite in

ICHAZ has higher carbon levels than nominal carbon concentration of the steel on account of the austenite grains richer in carbon than average carbon concentration as well as annealed martensite in the intercritical region (ferrite plus carbides) which is depleted in carbon with respect to average carbon concentration [43]. This means that the martensite in this area is supposed to have the highest hardness due to higher carbon concentration inherited from carbon-enriched austenite. However, this kind of strengthening effect is far less than formation of large amount of polygonal ferrite formed at the expense of martensite and bainite, which is the softening phenomenon resulting from phase balance. Sharp decrease in hardness was observed when moving from ICHAZ to OT region and the lowest hardness value (295HV) is located somewhere between ICHAZ and OT region due to complete loss of martensite body-centered tetragonal lattice, precipitation of coarsened carbides, as well as dislocation annihilation resulting from recovery at high temperature tempering [44, 45]. The hardness in OT region saw an increasing trend until it reaches the hardness value of base metal since temperature exposed is decreasing, which results in less softening effect.

Fig. 11 Hardness profile across weld

3.4 Tensile properties

The transverse tensile test was carried out using Instron 8800 universal test machine at room temperature. Fig. 12 depicts the engineering stress–strain curves of both welded joint and the parental metal. As with the other fusion arc welding, the tensile sample joined by K-TIG also fractured in the weld metal region. It is shown that the UTS and total elongation for base metal are 1680MPa and 12.6% respectively, whereas the UTS and total elongation of welded joint are 1080MPa and 4.9% respectively. It is well known that the strain hardening response is primarily governed by hardness distribution across the weld cross-section as well as microstructure [19]. Therefore, the formation of bainitic microstructure in weld metal region leads to significant decrease in hardness, which in turn results in the decrease in overall tensile strength. As shown in Table 3, although the UTS of the joint produced by K-TIG is 65% of that of the parental metal, it is still much higher than that obtained via conventional fusion welding because of under-matching filler materials used. The UTS of the latter is not exceeding 50% of the base metal. Moreover, the elongation (4.9%) of the welded joint is not appreciable less than that produced by conventional fusion welding technique with under-matching filler materials, and is more than twice that obtained via friction stir welding.

Fig. 12 Tensile curve for base metal and welded joint

Table 3

Comparison of tensile properties among different welding techniques

Joint type	UTS/MPa	Elongation (%)	Joint efficiency	Fracture location
^a SA	650	6.25	50%	Weld metal [46]
^b FA	571	5	44%	Weld metal [46]
^c FSW	1225	2.3	75%	HAZ [19]
K-TIG	1080	4.9	65%	Weld metal

^a SA: SMAW using austenitic stainless steel filler materials; ^b FA: FCAW using austenitic stainless steel filler materials; ^c FSW: Friction stir welded joint

The fracture surface in base metal shows a typical ductile fracture mode with evident fibrous zone and shear lip region, as shown in Fig. 13a. This kind of fracture is a result of microvoid nucleation as well as subsequent grow and coalescence. Once the crack approaches the edge, the stress condition changes to shear from tension, resulting in a fracture around 45 degree relative to the maximum stress direction [47]. In contrast, the fracture surface of welded joint presents brittle fracture mode, with very small area of crack initiation indicated by red rectangular and large proportion of radial zone which indicates rapid crack growth along the crystallographic planes, as depicted in Fig. 13b. It can be seen from Fig. 13c and d that the fibrous zone of both fractured samples show deep and equiaxial dimples, suggesting that there is plastic deformation occurring before fracturing. Compared with large and deep dimples in fibrous zone, the shear lip in fractured base metal sample (Fig. 13e) exhibits shallow and stretched dimples, indicating a combined shear and normal separation. The radial zone of fractured weld metal surface shows typical cleavage fracture with river pattern characteristics, as depicted in Fig. 13f, which is originated from the crack propagation through a number of cleavage planes of various levels [48].

Fig. 13 ensile fracture surface: (a) Macro-image of base metal; (b) Macro-image of weld metal; (c) SEM image of base metal fibrous zone; (d) SEM image of weld metal fibrous zone; (e) SEM image of base metal shear lip; and (f) SEM image of weld metal radial zone

The poorer fracture toughness of weld metal compared with base metal is a combined effect of bainitic microstructure with M-A constituent, misorientation angle distribution and grain size. It is well known that fracture occurs mainly through crack initiation and the upcoming crack propagation. As demonstrated by Lan et al. [49], the cleavage fracture mechanism is mainly controlled by crack initiation for coarse bainite microstructure. The M-A constituent, as shown in Fig. 8c is thought to have detrimental effect on fracture toughness since it is hard and brittle, and is considered to be crack initiation sites [50]. In addition, transformation induced residual tensile stress in the region adjacent to M-A constituents introduced by the freshly formed martensite transformed from retained

austenite, together with the decohesion of interface between M-A constituents and surrounding bainitic matrix assist and stimulate microcrack initiation, which increases the risk of cleavage fracture as a result [49]. Once a microcrack forms, the upcoming crack propagation is mainly governed by high misorientation grain boundaries [51-53]. The grain boundary maps superimposed on the orientation colour map of base metal and weld metal areas are shown in Fig. 14a and b respectively. The black lines represent the grain boundaries with misorientation angle larger than 15 degree, whereas white lines correspond to the grain boundaries with misorientation angle between 2 degree and 15 degree. It is well accepted that the HAGBs resulting from PAG grain and packet boundaries are able to efficiently deflect or even arrest the cleavage crack propagation, whereas the LAGBs never produce a remarkable deflection or arrest of the cleavage cracks [54-56]. It is obvious from Fig. 14c and d that the proportion of HAGB (>15 degree) in base metal (61.6%) is much higher than that in weld metal (33.1%), which in turn leads to more crack deflection and consumes larger energy during crack propagation. It was also demonstrated by Lan et al. [49] that high misorientation grain boundaries in coarse bainite matrix has very little positive impact on the crack deflection and arrest. In terms of grain size, it is known that increased number of grain boundaries helps to block crack propagation and increase the energy required for crack to propagate as a result [57]. It was also reported that finer grain size is conducive to minimizing stress resulting from dislocation pile-up and raise the resistance to cleavage fracture [37]. Therefore, The larger prior austenite grain size in weld metal and inherited larger packet size associated with high heat input compared with those in base metal increases the possibility of initiated cleavage crack propagation and deteriorate the toughness as a result.

Fig. 14 EBSD map; (a) Grain boundary map for Base metal; (b) Grain boundary map for weld metal; (c) Misorientation distribution map in base metal; and (d) Misorientation distribution map in weld metal

4. Conclusions

K-TIG welding has been used to weld 9mm thick armour grade Q&T steel in a flat position. The main conclusions are drawn as follows:

(1) Single pass full penetration can be achieved on 9mm armour grade Q&T steel with K-TIG at a speed of 28cm/min^{-1} without using any filler materials and edge preparation. Incomplete penetration and weld pool collapse occur when welding current is less than 545A and higher than 565A respectively.

(2) The highest hardness is obtained in FGHAZ due to the formation of untempered martensite and smaller grain size, while the lowest one is present somewhere between OT region and ICHAZ owing to over-tempered martensitic microstructure. The hardness distribution across the weld is higher than current practice, which would lead to improved ballistic performance.

(3) The weld metal region, which is composed of dendritic structure and predominantly bainitic microstructure, contains large proportion of low angle grain boundaries, resulting in poorer toughness compared with parental metal.

(4) Although joint efficiency of K-TIG joined weld is 65%, it is still much higher than that produced by conventional fusion arc welding processes. The elongation of K-TIG joined weld is not appreciably less than that obtained by conventional fusion welding process and is more than twice that produce by friction stir welding.

It has been demonstrated that the K-TIG welding process offers a new way to weld medium thick armour grade Q&T steel with high efficiency and low cost, while maintaining the mechanical properties at high level. Future work will focus on refinement of grain size in weld metal region in order to improve toughness and hardness as well as the expansion of operating window.

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