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Comparison of microstructure and mechanical properties of ultra-narrow gap laser and gas-metal-arc welded S960 high strength steel

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Abstract

The microstructural characteristics and mechanical properties, including micro-hardness, tensile properties, three-point bending properties and Charpy impact toughness at different test temperatures of 8 mm thick S960 high strength steel plates were investigated following their joining by multi-pass ultra-narrow gap laser welding (NGLW) and gas metal arc welding (GMAW) techniques. It was found that the microstructure in the fusion zone (FZ) for the ultra-NGLW joint was predominantly martensite mixed with some tempered martensite, while the FZ for the GMAW joint was mainly consisted of ferrite with some martensite. The strength of the ultra-NGLW specimens was comparable to that of the base material (BM), with all welded specimens failed in the BM in the tensile tests. The tensile strength of the GMAW specimens was reduced approximately by 100 MPa when compared with the base material by a broad and soft heat affected zone (HAZ) with failure located in the soft HAZ. Both the ultra-NGLW and GMAW specimens performed well in three-point bending tests. The GMAW joints exhibited better impact toughness than the ultra-NGLW joints.

Keywords: asAs welded condition; heatHeat input; coolingCooling rate; toughnessToughness; hardnessHardness; martensiteMartensite

1 Introduction

With the development of the thermo-mechanical rolling, quenching and tempering process technique, high strength low alloy (HSLA) steels have been developed for many years and achieve high tensile properties and good toughness [1]. HSLA steels are classified as low carbon steels and achieve their high strength levels from low percentage alloying additions that result in precipitation hardening, and/or grain refinement strengthening mechanisms through thermo-mechanical processing [2,3]. To achieve a combination of high strength and toughness, the microstructure of lower bainite or ferrite plus martensite has been designed for HSLA steels [4,5]. HSLA steels are used in a wide variety of applications as structural components, pressure vessel and linepipes, and in shipbuilding, offshore construction, automotive, industrial vehicles, lifting and cranes. The application of HSLA steels for such components enables lighter and more slender products and reduces construction costs without loss of structural integrity [6,7]. Hulka et al. [8] reported that under tensile stresses the plate thickness can be reduced by around 60% by using the steel grade S960 instead of S355.

In the manufacturing of structural components, welding is an indispensable manufacturing technique for joining the small components into the final complex products and the application of welding technique can mitigate the complexity of manufacturing large structures [9]. Conventional arc welding are the main welding technique applied in industry to join the high strength steels. However, these traditional arc welding procedures have relatively high heat inputs which result in slow cooling rates of the weldment, and this can produce softening in the heat affected zone (HAZ) [10], giving low strength of the weldment [11,12].

C	Mn	P	S	N	B	Ca	Si	Al	Cu
0.088	1.51	0.008	0.002	0.0089	0.0022	0.012	0.055	0.033	0.014
Sn	Cr	Ni	Mo	Ti	Nb	V	Fe	CE	
0.001	0.472	0.023	0.248	0.025	0.040	0.050	Bal.	0.496	

Table 2 Chemical composition of the Union X96 filler wire ~~(wt.%)~~ (wt%).

alt-text: Table 2

C	Si	Mn	Cr	Mo	Ni	Fe	CE
0.12	0.80	1.90	0.45	0.55	2.35	Bal.	0.79

Table 3 The mechanical properties of the Union X96 filler wire.

alt-text: Table 3

Material	Yield strength 0.2% (MPa)	Tensile strength (MPa)	Elongation (%)	Impact value (−50 °C) (J)
Union X96	930	980	14	50

A butt joint with a parallel groove configuration (groove width was 1.2 mm) was prepared for the multi-pass ultra-NGLW experiments, as shown in Fig. 1(a). To obtain full penetration of the GMA weld in the 8 mm thick S960 steel, a butt welded joint with a V-groove configuration with a 1 mm thick root face and 1 mm gap and 60° groove angle was prepared, as shown in Fig. 1(b). Because the conventional V-groove is generally applied in practice, the V-groove geometry was chosen for the GMAW in this study.

Fig. 1 Schematic of welding groove: (a) the groove shape for ultra-NGLW, (b) the groove shape for GMAW.

alt-text: Fig. 1

Prior to the welding experiment, the vicinity of the weld track on the base material was sand blasted to clean the surface oxide. The removal of the oxide can also make the groove surfaces rougher, which can have an effect on improving the absorption and reflections of the laser beam especially when using a 1 μm wavelength laser [29]. Acetone was used to clean the surface after sand blasting, and then the materials were clamped on the work table to ensure adequate restraint. Both the ultra-NGLW and GMAW were carried out perpendicular to the rolling direction of the base material.

A Miller Axxess 450 GMAW power source was used for the manual GMAW experiments. To protect the molten weld pool and obtain good penetration depth during welding process, the specimens were shielded using a mixture of 80% argon and 20% CO₂ with gas flow rate of 22 l/min. A continuous wave fibre laser (IPG YLS-16000) equipment with a maximum output power of 16 kW was used for the ultra-NGLW experiments. The detailed parameters of the fiber laser system and the schematic of the ultra-NGLW process setup are shown in [30].

The thermal cycles within both the ultra-NGLW and GMAW specimens were measured continuously throughout the welding processes using K-type thermocouples. The thermocouples were spot welded to the top surface of the plates, and they were located at different distances from the weld centreline, on a line perpendicular to the welding direction and half way along the length of the weld. A Squirrel-2040 series data

logger was used to record the thermal histories during the welding processes.

After welding, the welded specimens were sectioned transverse to the welding direction, then mounted ground, polished, and etched with a 2% Nital solution. The macrostructure and microstructure of the welded joints investigation were performed using a KEYENCE VHX-500F optical microscope and Philips XL 30 scanning electron microscope (SEM).

Microindentation hardness mapping profiles across the welded joints were measured using a 300 g test load and a dwell period of 15 s with a Vickers microhardness machine (Struers DuraScan 50) having a fully automated testing cycle. For the NGLW specimen, because of the ultra-narrow fusion zone and heat affected zone (HAZ), an indent interval of 0.2 mm was set in the area close to the fusion zone and HAZ and the indent interval for other areas further away from the fusion zone and HAZ was set as 0.4 mm. Whilst, because of the wide fusion zone and HAZ for the GMA welded specimen, the indent interval was set as 0.4 mm.

Tensile test specimens for the as-received BM and the welded samples were prepared according to the ASTM E8M-04. The gauge length for the tensile test was 50 mm. Three-point bending test samples were sliced in accordance with the BS EN ISO 15614-1. Both face and root bending tests were carried out with two test pieces for each type of test. Two samples of each were prepared for the bending test. Bend testing was executed until an angle of 180° was reached according to BS EN ISO 5173: 2010+A1:2011, 5173:2010+A1:2011. Sub-size Charpy impact test samples (8 mm thick) were prepared following BS EN 10045-1:1990, 10045-1:1990. Three replicates for tensile and Charpy impact testing were prepared to reduce experimental uncertainties. The mechanical properties test samples were sliced along the rolling direction of the material using wire electric discharge machining (EDM), and the configurations for each are shown in [16]. The weld reinforcements in the face and root regions were removed by manual grinding before the mechanical properties tests were carried out. The notches for the Charpy impact test coupons were located in the fusion zone, on the fusion line (FL) and in the HAZ, respectively to examine the impact toughness of the fusion zone (FZ), fusion line (FL) and HAZ for the welded samples. In order to fix the right position of Charpy V notch (CVN), the impact specimens were macro etched with 2% Nital solution to exhibit the outline of fusion line and HAZ. For the ultra-NGLW samples with notches located in the HAZ were sampled as FL ± 0.5 mm. Because the fusion zone for GMA welded specimens is a 'V' shaped area, there is an angle ($\sim 60^\circ$) between the fusion line and the bottom surface of the specimens. The usual procedure for the FL notch is to pass through the HAZ (real FL) at the mid-thickness point of the material (sample 50% weld metal and 50% HAZ) to delegate the nominated FL, while the HAZ notch is to sample the notch at FL ± 2 mm. The CVN position was located at the FL and HAZ for GMA welded samples schematically shown in Fig. 2. Tensile tests were performed at room temperature using an Instron model 8500 electronic universal test machine. Three-point bending tests were carried out on a Dartec bending test machine at room temperature. The diameter of the former and the gap distance for the three-point bending test are shown in [16]. The Charpy impact tests were carried out on a Zwick Roell Charpy impact test machine at -40 – -40 °C, -20 – -20 °C, 0 °C and at room temperature, respectively. Each test sample was kept at the related test temperature for half an hour to ensure that the temperature in each specimen was uniform. Following the tensile and Charpy impact tests, all of the fracture surfaces of the tested specimens were examined using Zeiss EVO 50 SEM installed with an Energy Dispersive X-ray Detector (EDX) to identify the fracture morphology and the fracture mechanisms.

Fig. 2 Schematic representation of the sample with notch at the FL and in the HAZ for GMAW samples.

alt-text: Fig. 2

3 Results and Discussion

3.1 Welding parameters

There are three main variable welding parameters (*i.e.* laser power, welding speed and wire feed rate) for ultra-NGLW process, statistical modeling was applied to optimise the welding parameters. The detailed optimising procedure was demonstrated in [30]. The welding parameters for both manual GMAW and ultra-NGLW of S960 steel are shown in Tables 4 and 5, respectively. From previous experience with single pass autogenous laser welding of 8 mm thick S960 steel, the optimum laser power was 6.5 kW [16]. In this investigation, moderate laser power (2 kW) was applied for the ultra-NGLW, which will have a benefit to save capital investment. Heat input for arc welding can be calculated by the following equation: heat input (J/mm)=efficiency×(voltage (V)×current (A))/welding speed (mm/s). The consumable electrode process, such as GMAW, exhibits an average arc efficiency of 80% [31]. Heat input for

laser welding is calculated by the equation: heat input (J/mm)=efficiency×laser power (W)/welding speed (mm/s). The welding efficiency considering the conduction heat loss only is about 80% for autogenous laser welding [32,33]. In addition, some beam reflection from the filler wire takes place in the ultra-NGLW process, which could result in 30% loss of energy applied into the workpieces [34].

Table 4 Optimised GMAW parameters for 8 mm thick S960 steel.

alt-text: Table 4

Welding pass	Voltage (V)	Current (A)	Welding speed (m/min)	Wire feeding rate (m/min)	Shielding gas flow (l/min)	Heat input (kJ/mm)
First	27	175	0.4	4.0	22	0.57
Second	27	165	0.46	4.0	22	0.46
Third (torch weave)	27	168	0.26	4.0	22	0.84

Table 5 Optimised ultra-NGLW parameters for 8 mm thick S960 steel.

alt-text: Table 5

Welding pass	Power (kW)	Welding speed (m/min)	Wire feeding rate (m/min)	Top shielding gas flow (l/min)	Back shielding gas flow (l/min)	Focal position (mm)	Heat input (kJ/mm)
Root pass	2	1.2	–	12	8	–6	0.08
Filling pass	2	0.6	3.3	12	8	0	0.11
Filling pass	2	0.6	3.3	12	8	0	0.11

The comparison of energy and consumed filler material between ultra-NGLW and GMAW techniques are summarised in Table 6. It can be found that the maximum power used for GMAW was 4.7 kW, while that for ultra-NGLW was 2 kW. The heat input for each welding pass for GMAW was approximately 5–8 times that of ultra-NGLW. The cumulative heat input (the heat input summation of all welding passes) for GMAW was ~~6~~~6 times that of ultra-NGLW. The welding speed for ultra-NGLW was ~~1.5~~~1.5 times that of GMAW. The consumed filler material for GMAW was ~~6~~~6 times that of ultra-NGLW.

Table 6 Comparison of energy and consumed filler material between ultra-NGLW and GMAW.

alt-text: Table 6

Welding technique	Maximum power used (kW)	No. of welding passes	welding speed (m/min)	Consumed filler material (g)	Cumulative heat input (kJ/mm)
GMAW	4.7	3	0.4	89	1.87
Ultra-NGLW	2	3	0.6	16	0.3

3.2 Macrostructure of the welded joints

It is seen that the cross-section of the welds (Fig. 3(a) and (b)) exhibited a significant microstructural change in the fusion zone (FZ) and heat affected zone (HAZ) for both the GMAW and ultra-NGLW joints. Both the welded joints are full penetration, and there are no macroscopic defects such as cracks, incomplete fusion, and porosity. The GMAW specimen presents a ‘V’ shaped fusion zone and ultra-NGLW specimen shows a nearly rectangle fusion zone. Three welding passes can be distinguished clearly for both the GMAW and ultra-NGLW joints. The various regions, ~~1~~e.i.e., base material (BM), fusion zone (FZ) and heat affected zone (HAZ) are visible. The top width of fusion zone and the width of HAZ for the GMAW specimen were approximate 12 and 3.5 mm, respectively. Whilst, the widths of the FZ and HAZ for the ultra-NGLW specimen were approximate 1.4 and 1 mm, respectively. Because different locations of the HAZ suffered from different thermal cycles, especially different peak temperatures. During the welding process, the grain size within the HAZ varied accordingly. In order of increasing peak temperature, the HAZ can be subdivided into the sub-critical HAZ (SCHAZ), the intercritical or partially austenitised HAZ (ICHAZ), the fine grained HAZ (FGHAZ), and finally the coarse grained HAZ (CGHAZ) which is adjacent to the fusion zone itself. Among these sub-HAZs, the CGHAZ experienced the highest peak temperature, which caused an increase in the grain size in this sub-zone.

Fig. 3 Cross-section profiles the welded joints, (a) GMAW, (b) ultra-NGLW.

alt-text: Fig. 3

3.3 Thermal cycles throughout the welding processes

The measured thermal cycles at the locations of the two thermocouples that were spot welded on the top surfaces of both the ultra-NGLW and GMAW specimens are shown in Fig. 4(a) and (b), respectively. Due to the damage by the spatters generated during the welding process, some thermocouples spot welded close to the weld centreline failed to monitor the thermal cycles during the welding processes. For the ultra-NGLW specimen, the points at distances of 1.7 mm fall within the ICHAZ, while the point at 4.2 mm approximately coincides with the BM. While for the GMAW specimen, the points at both 8.5 and 9.5 mm are outside the HAZ and expected to coincide with the BM. It can be seen from Fig. 4(a) and (b) that the peak temperatures for the test points on both the ultra-NGLW and GMAW specimens occurred for the third welding pass. This because the third pass was the last pass to fill up the prepared groove. The heat source was much closer to the thermocouples than other previous passes. The measured peak temperature at distance of 1.7 mm for the ultra-NGLW specimen is 785 °C, and the cooling rate is ~ 170 °C/s. While for the GMAW specimen, the measured peak temperature at distance of 8.5 mm is 520 °C, and the cooling rate is ~ 14 °C/s. it can be seen from here that the cooling rate for the ultra-NGLW process is much greater than that of the GMAW process.

Fig. 4 Comparison of measured temperature histories at different distance from the weld centrelines, (a) ultra-NGLW, (b) GMAW.

alt-text: Fig. 4

3.4 Microstructures in different sub zones

Both the GMAW and ultra-NGLW processes involve multi-welding passes to fill up the prepared welding groove. The weld bead in the upper part was formed in the last filling pass, so there was no more additional heat effect introduced to this part. However, the lower part suffered to some more additional thermal cycles from the subsequent filling passes, which would have a significant effect on the final microstructures. In addition, a reheated CGHAZ was generated in the GMAW joint. This sub-zone is located in the middle of the wide HAZ and surrounds the 'Vee' shaped FZ, as indicated in Fig. 3(a).

3.4.1 Microstructures in different sub zones of the ultra-NGLW joint

The microstructures of each different sub-zone for the upper and lower part of the ultra-NGLW sample are presented in Figs. 5 and 6, respectively. In the upper part, the SCHAZ retains the microstructure of the BM, consisting bainite and auto-tempered martensite. The ICHAZ consists of an over tempered BM microstructure and regions with partially transformed martensite and auto-tempered martensite. In the FGHAZ, the microstructure is mixed with equiaxed martensite and auto-tempered martensite, with a small prior-austenite grain size. The microstructure in the CGHAZ is a mixture of equiaxed martensite and auto-tempered martensite with a large prior-austenite grain size.

Martensite is predominated in the FZ with a small amount of auto-tempered martensite, and the prior-austenite grain shows a very large grain structure.

Fig. 5 SEM micrographs showing the microstructures in different sub-zones of the upper part of the ultra-NGLW joint, (a) SCHAZ, (b) ICHAZ, (c) FGHAZ, (d) CGHAZ, (e) FZ, (prior austenite grain boundaries are signified with arrows). The symbols M, TM and B represent martensite, tempered martensite and bainite, respectively.

alt-text: Fig. 5

Fig. 6 SEM micrographs showing the microstructures in different sub-zones of the lower part of the ultra-NGLW joint, (a) ICHAZ, (b) FGHAZ, (c) CGHAZ, (d) FZ, (prior austenite grain boundaries are signified with arrows). The symbols M, TM and B represent martensite, tempered martensite and bainite, respectively.

alt-text: Fig. 6

The microstructure in the lower part of SCHAZ, ICHAZ, FGHAZ, CGHAZ and FZ shows a similar microstructure to the corresponding sub-zones in the upper part of the ultra-NGLW joint, as shown in [Fig. 6](#). However, the microstructure of the sub-zones in the lower part presents more tempered martensite. In order to avoid duplicating the information in [Fig. 5\(a\)](#), the microstructure in the lower part of the SCHAZ is not presented in [Fig. 6](#).

The microstructures evolution within the different sub-zones of a welded joint is dependent on the local thermal cycles, and the peak temperature that is reached during the welding process. Grain size of HAZ are coarser near the fusion line, it becomes finer as it goes far. However, in the SCHAZ, during the welding process, the peak temperature does not quite reach the Ac_1 temperature, and so no material is re-austenized in this region [\[35\]](#). There is no metallographic colour change in this SCHAZ after etching using Nital solution. So this sub-zone cannot be distinguished under optical microscope. However, SEM examination of this region for both the ultra-NGLW and GMAW joints indicated that the phases apparent in this sub-zone retained the microstructure of the BM, mixed with tempered martensite and bainite, as is shown in [Fig. 5\(a\)](#) and [Fig. 7\(a\)](#).

Fig. 7 SEM micrographs showing the microstructures in different sub-zones of GMAW joint, (a) SCHAZ, (b) ICHAZ, (c) FGHAZ, (d) CGHAZ, (e) low magnification of reheated CGHAZ, (f) high magnification of reheated CGHAZ, (g) FZ, (prior austenite grain boundaries are signified with arrows). The symbols M, TM, B and F represent martensite, tempered martensite, bainite and ferrite, respectively.

alt-text: Fig. 7

The peak temperature experienced in the HAZ reduced gradually from the CGHAZ (adjacent to the fusion line) to the ICHAZ (close to SCHAZ and BM). In the CGHAZ, the peak temperatures were raised to 1300 °C or higher (but lower than the melting point) for a sufficient period to ensure complete re-austenization, and rapid grain growth [36,37]. The FGHAZ encountered temperatures slightly higher than the A_{c3} temperature. This resulted in complete re-austenization of the material, but there was limited austenite grain growth due to the relatively low peak temperatures and for short periods of time at these temperatures in the FGHAZ [38,39]. Whilst the ICHAZ experienced temperature between the A_{c1} and A_{c3} temperatures, the material in this sub-zone has a partial transformation into austenite. The FZ was melted during the heating process, and it subsequently solidified during the cooling process. In addition, very large grain size austenite was generated in the FZ.

The cooling rate is one of the most important factors influencing microstructure evolution in either fusion zone or HAZ regions [39]. The higher values of the heat input, the slower the cooling rate would be, and vice versa [40].

Because the FZ is located in the area directly worked by the heat source and the cooling rate in the FZ is relatively slower than the other sub-zones. Comparing the heat input per unit length between typical arc and laser welds, it is shown that the heat input in laser welds is typically almost an order of magnitude lower than in arc welds [41]. In this research, the heat input of each welding pass for GMAW is $\sim 0.5\text{--}0.8\text{--}0.5\text{--}0.8$ kJ/mm, while the heat input for ultra-NGLW is $\sim 0.1\text{--}0.1$ kJ/mm. Laser welding has lower line energy than arc welding processes and higher cooling rates [15]. The measured cooling rates for ultra-NGLW and GMAW of this S960 steel in Fig. 4 show the same results.

Owing to the fast cooling rate in the ultra-NGLW process, the transformed austenite in the ICHAZ, FGHAZ, CGHAZ and FZ was quenched. Martensite was formed in these sub-zones. Because of the high Ms temperature ($\sim 450\text{--}450$ °C) for S960 steel, some formed martensite has more chance to be auto-tempered in the following cooling process. These sub-zones included predominantly martensite in conjunction with some tempered martensite. The grain size varied gradually from the FZ to the FGHAZ because of the gradually reduced peak temperature, as seen in Fig. 5. In the ICHAZ, the partially transformed austenite was quenched into martensite, while the un-transformed material was tempered during the welding thermal cycles. The martensite within the ICHAZ reflects the original banded morphology of the BM, which is parallel to the rolling direction, as is shown in Fig. 5(b). In addition, more tempered martensite was observed in the lower part of the welded joint, especially the area between the two welding passes. This phenomenon can be attributed to the additional heat effect introduced from the subsequent welding pass to the prior welding pass. This newly generated thermal cycle has a tempering effect on the prior welding pass.

3.4.2 Microstructures in different sub zones of the GMAW joint

Both the upper and lower part of the GMAW sample were examined using SEM, the microstructures in the corresponding sub-zones for the upper and lower part did not show a big difference. Representative SEM micrographs illustrating microstructures of each sub-zone in the GMAW specimen are shown in Fig. 7. The microstructure in the SCHAZ is similar to the base material, comprising bainite and auto-tempered martensite. In the ICHAZ, some partially transformed martensite and auto-tempered martensite were found, and the other regions retained the BM microstructure. The microstructure in the FGHAZ is predominated of bainite with a mixture of small amount of martensite and tempered martensite, and with a small prior-austenite grain size. The microstructure in the CGHAZ is the same as that in the FGHAZ, but with a large prior-austenite grain size. In the reheated CGHAZ, some partially transformed martensite was observed encircled around the prior austenite grain boundaries and the generated bainite in the CGHAZ during the prior welding pass was retained inside of the partially transformed martensite circles, as shown in Fig. 7(e). Tempered martensite was found in the circular partially transformed sub-region (Fig. 7(f)) under higher magnification. In the FZ, the microstructure consists predominantly of ferrite with some martensite. Some micro-voids were found in the FZ because of the drop off of the inclusions during the polishing process.

For the GMAW joints, the arc has a relatively broad heating area and the higher heat input introduced into the joint makes its cooling rate much slower than the ultra-NGLW joint. In the slow cooling rate the newly formed austenite in the HAZ transformed into bainite, some small amount of martensite and tempered martensite were also formed in the HAZ. The grain size also reduced gradually with the increase of the distance to the FZ because of the gradually reduced peak temperature, as seen in Fig. 7. In the ICHAZ, the partially transformed austenite transformed into martensite and some tempered martensite. This sub-zone is close to the SCHAZ and BM, the cooling rate in the sub-zone may be relatively faster than in the FGHAZ and CGHAZ. Because the FZ cools from melt, the cooling rate in the FZ is relatively slower than in the HAZ. Ferrite and some martensite were formed in the cooling process. In addition, there is a reheated CGHAZ formed because of the multi-heat effect from the multi-welding passes. In the reheat process, some austenite nucleated on the prior austenite grain boundaries, but the growth of these newly generated austenite were suppressed because of the relatively low peak temperature from the reheat process. Martensite and tempered martensite were transformed from the austenite with very small grain size ($\sim 2\text{--}2$ μm), which are encircled around the prior austenite grain boundaries. While the generated bainite in the CGHAZ was retained, which were inside the partially transformed martensite circle, as seen in Fig. 7(e) and (f).

3.5 Microhardness

The hardness maps of the entire weld for both the ultra-NGLW and GMAW joints are shown in Fig. 8(a) and (b), respectively. The FZ, HAZ and BM can be distinguished clearly from the hardness contour map because of the hardness variation in different zones. Fig. 8(c) and (d) show hardness profiles extracted across the welded joints at 3 mm below the top surface for ultra-NGLW and GMAW samples, respectively. It is seen that the hardness of the base material for these two welded specimens shows some difference, approximately 320–345 HV_{0.3}. This may because the two plates for the welding experiments were extracted from different strips, which may not cool consistently during the steel strips manufacturing process. The hardness values of the FZ are higher in ultra-NGLW sample when compared to GMAW sample. The average hardness in the FZ for the ultra-NGLW sample is approximately 380 HV_{0.3}. The peak hardness is located in the un-tempered FZ and CGHAZ with hardness value of $\sim 390\text{--}390$ HV_{0.3}. The hardness rapidly drops from the CGHAZ toward the SCHAZ. The hardness value in the tempered FZ region was reduced to $\sim 330\text{--}330$ HV_{0.3}, which is located between the two welding passes, as shown in Fig. 8(a). The lowest Vickers hardness is located in the very narrow soft SCHAZ with hardness value of $\sim 310\text{--}310$ HV_{0.3}. However, there is a wide soft HAZ for the GMAW joint, as seen in Fig. 8(b). The lowest Vickers hardness is $\sim 280\text{--}280$ HV_{0.3}. In addition, a 'V' shaped relatively higher hardness contour can be observed in the wide soft HAZ, which surrounds the 'V' shaped FZ. This region is corresponding to the reheated CHGAZ. The peak hardness for the GMAW sample is located in the FZ with hardness value fluctuated around 350 HV_{0.3}. The hardness in the HAZ has a large variation from 285 to 325 HV_{0.3}.

Fig. 8 Microhardness measurements on the welded joint, (a) hardness map of the entire ultra-NGLW joint, (b) hardness map of the entire GMAW joint, (c) hardness profile across the ultra-NGLW joint, (d) hardness profile across the GMAW joint.

alt-text: Fig. 8

The ultra-NGLW introduced less heat input to the welded joint, which results in a rapid quench rate. The FZ and HAZ, which consist of martensite as the principal phase with some tempered martensite, would be of high cross-section microhardness and with a lack of toughness. In addition, the low heat input and rapid cooling rate caused the formation of narrow HAZ in the ultra-NGLW joint. The different peak temperatures and thermal cycles in this narrow transition region in the HAZ caused a sharp microhardness dropping from CGHAZ to SCHAZ, seen in Fig. 8. The SCHAZ endured a high temperature tempering (although lower than the A_{c1} temperature) during the welding process, which may result in a reduction in the dislocation density and softening of the microstructure in this very narrow region. This may result in a lower hardness than the BM, as seen in Fig. 8(c). However, GMAW introduced more heat input to the welded joint, which can lead to a slower cooling rate. Ferrite and some martensite were formed in the FZ of the GMAW joint, which results in the hardness in the FZ being comparable or a little higher than the BM. Due to the relatively higher cooling rate in the HAZ and the subsequent tempering effect from the multi-pass welding process, the HAZ was predominated of bainite with a mixture of small amount of martensite and tempered martensite, which results in a broader HAZ than ultra-NGLW joint and a much wider and softer HAZ than the ultra-NGLW joint.

3.6 Tensile and bending properties

The representative engineering stress versus engineering strain curves for the base S960 steel, the ultra-NGLW and GMAW S960 steel joints are illustrated in Fig. 9. All of these curves are smooth and continuous. The details of the transverse tensile test results are given in Table 7. It must be mentioned here that the welded specimens are not homogenous, especially in the FZ and HAZ. This means the values recorded for the yield stress and the elongation are not truly representative of any particular sub-zones. In addition, they will also vary with the choice of gauge length (50 mm in this case). All fractured samples after tensile test are shown in Fig. 10. It is apparent that the ultra-NGLW specimens present almost the same tensile properties as the BM even though laser welding results in a narrow soft SCHAZ. It suggested that the ultra-NGLW process did not deteriorate the tensile properties of the S960 HSLA steel. However, the GMAW specimens show inferior tensile properties when compared to the BM. The yield strength (YS), ultimate tensile strength (UTS) and apparent elongation for the ultra-NGLW joints were obtained to be 1019 MPa, 1040 MPa, and 8.3%, respectively, which were very close to those of the BM which had a YS of 1026 MPa, UTS of 1035 MPa and an apparent elongation of 8.5%, respectively. All ultra-NGLW specimens failed in the BM well away from the weld during the tensile test, as is shown in Fig. 10(b). Whilst GMAW specimens demonstrate a YS of 903 MPa, a UTS of 936 MPa and an apparent elongation of 7.1%, respectively, which present an approximately 100 MPa lower strength and 1.4% less elongation when compared to the BM. All tensile failures for the GMAW specimens were in the soft HAZ, as is shown in Fig. 10(c).

Fig. 9 Representative engineering stress versus engineering strain curves for base S960 steel, the ultra-NGLW and GMAW S960 steel joints.

alt-text: Fig. 9

Table 7 Tensile properties for the base material, the ultra-NGLW and the GMAW specimens.

alt-text: Table 7

Test specimens	Yield strength (MPa)	Tensile strength (MPa)	Elongation (%)
Base material (8 mm)	1026±7	1035±11	8.5±0.3
NGLW (8 mm)	1019±8	1040±11	8.3±0.3
GMAW (8 mm)	903±7	936±9	7.1±0.8

Fig. 10 Fracture location of the tensile test specimens, (a) base material, (b) ultra-NGLW samples, (c) GMAW samples.

alt-text: Fig. 10

It is well known that hardness is an indication of the strength. In general, an increase in hardness results in an increase in strength but decreasing formability or ductility [42]. Surface hardness mapping can be employed in general to depict the strength map. It can be deduced from the hardness map of entire weld for the ultra-NGLW joint in Fig. 8(a) that the FZ and HAZ (except the SCHAZ) were strengthened by the ultra-NGLW process. Some thin tempered layers between the two welding passes are observed, which are attributed the tempering effect from the tempering effect by the subsequent welding pass on the prior welding pass. These tempered thin layers may have the same or comparable strength as the BM. However, a very narrow SCHAZ can also be observed from the hardness mapping results, which would be the weakest region for the ultra-NGLW joint.

In addition, it can also be deduced from the hardness map of entire weld for the GMAW joint in Fig. 8(b) that the FZ has the same or a slightly higher strength when compared to the BM. The wide soft HAZ would make the welded joint lose strength. However, the very thin 'V' shaped reheated CHGAZ may strengthen the wide soft HAZ a little.

In the transverse tensile test, all the ultra-NGLW specimens failed in the BM well away from the FZ with almost the same tensile properties as the BM. However, the GMAW specimens failed in the soft HAZ show inferior tensile properties when compared to the BM. The strengthened FZ and HAZ for the ultra-NGLW joint present higher yield strengths than that of the BM. Conversely, both the yield and ultimate tensile strengths of the GMAW joint are considerably reduced by the softness of its HAZ. Compared to the base metal, the elongation to failure is reduced in all of the welded samples, albeit to a much greater extent by GMAW (1.4%) than with ultra-NGLW (0.2%). This loss of ductility is mainly attributed to a localization of strain during deformation; ~~i.e.~~ i.e., the non-welded base metal has a homogenous microstructure, whereas the welded joint exhibits variation in microstructure and hardness across the welded joint. The local strain generated during tension is therefore concentrated within a specific region, resulting in premature fracture compared to the uniform deformation that occurs in the BM [43]. Nevertheless, these results do demonstrate that the ultra-NGLW joint can actually retain its strength with only a relatively small loss of ductility which is attributed to the very narrow strengthened FZ and HAZ. However, due to the much wider FZ with a little strengthened region for GMAW specimens, the GMAW specimens show much reduction on the elongation than the ultra-NGLW specimens.

It is apparent that the ultra-NGLW specimens present almost the same tensile properties as the BM with failure in the BM not in the narrow soft SCHAZ, whereas the GMAW results in a substantial deterioration of all tensile properties with failure in the soft HAZ. In the case of the ultra-NGLW specimens, the FZ and HAZ (with the exception of the SCHAZ) had a much higher hardness than the BM, while the SCHAZ was a little softer than the BM, but this region is very narrow (~0.8 mm). The hard FZ and the hard regions of the HAZ can therefore act as a strong constraint on plastic deformation in the adjacent SCHAZ, thereby leading to the majority of the tensile plastic deformation accumulating in the BM and subsequent fracture in the BM [43–45], as seen in Fig. 10. Lee ~~et al.~~ et al. [43] reported that this concentration of deformation in the BM was confirmed by optical observation of the local strain distribution in their test on laser welded DP780 steel. However, it should be mentioned that this plastic deformation constrain by the surrounding strengthened material happens only in the case when the softened zone is narrow and the degree of softening is relative low. Otherwise, the wide softened zone as well as low hardness will result in the decreasing of the tensile strength of the welded joint [44]. Thereby, failure occurred in the soft HAZ for the GMAW samples.

Fracture surfaces of the tensile specimens were investigated to examine the fracture mechanisms of the steels. The fractured samples were examined under optical microscope and SEM. All ultra-NGLW specimens failed in the

BM well away from the weld, whilst all GMAW specimens fractured in the soft HAZ, as is shown in Fig. 10(b) and (c), respectively. All fractured surfaces show similar macro and micro morphology. Representative fracture surfaces of the failed tensile test samples are shown in Fig. 11. From the overall morphology of the fractured surface, it can be seen that the specimen splits into two segments (Fig. 11(a)), with the boundary between the segments coinciding with the mid-thickness position within the plate. It is speculated that this splitting is due to the specific rolling and fabrication process for the plate, which can lead to variation in the chemical composition in the through-thickness direction [46]. As plastic deformation accumulates, a crack may first develop parallel to the rolling direction and perpendicular to the fracture surface, before ultimately deviating to produce the final fracture. High magnification observations reveal that the fracture surface is composed of predominately of large voids and deep equiaxed dimples, as seen in Fig. 11(b). The dimples in the SEM photomicrograph of the fractured surfaces validate the ductile fracture behaviour of the tensile test samples. The macro and micro morphology of the fracture surfaces reflect the BM, ultra-NGLW and GMAW specimens have good ductility. There are many spherically-shaped inclusion particles located within the large voids. The chemical elements of these inclusion particles were identified by energy dispersive X-ray spectroscopy (EDX). The EDX spectrum (Fig. 11(c)) indicates that the inclusion particles are rich in Ca, O, Al, Mn, S and Fe.

Fig. 11 SEM micrographs showing the fracture surface morphology for the tensile test specimens, and the results of EDX analysis: (a) macro fracture surface, (b) high magnification fracture surface, (c) EDX spectrum for the inclusion particles.

alt-text: Fig. 11

Both the root and face bend tests for ultra-NGLW and GMAW joints were carried out on the as-welded joints, and the resulting specimens are shown immediately after testing (Fig. 12). The welds were subjected to significant plastic deformation and no surface cracks were visible after either root or face bend testing, which indicates that both the ultra-NGLW and GMAW joints exhibited good ductility and adequate bending strength.

Fig. 12 Three-point bend test results, (a) ultra-NGLW samples, (b) GMAW samples.

alt-text: Fig. 12

3.7 Charpy impact properties

Because the thickness of the base material used in this study was 8 mm, the Charpy impact test in this work was carried out using sub-size specimens (55 $\times$ mm $\times$ 10 $\times$ mm $\times$ 8 mm). Corrections were applied to account for the sub-size specimens to compare the results with those of standard samples (55 $\times$ mm $\times$ 10 $\times$ mm $\times$ 10 mm). There are many previous studies on corrections for sub-sized Charpy V-Notch (CVN) specimens [46–49]. According to previous investigation, it was postulated that the only correction that was required was for the thickness if the only non-standard dimension for the test specimens was the thickness [16,48,49]. In this work, a similar approach was followed in applying corrections to the energies on 8 mm thick specimens, i.e. by multiplying the absorbed energies by a factor of 1.25 (i.e. 10/8).

The original absorbed Charpy impact energy results for the sub-sized specimens, and the corresponding estimates for the absorbed energies in standard specimens, are shown at different test temperatures in Fig. 13(a) and (b) for the BM and GMAW and ultra-NGLW specimens with notches sampled at the weld centre line, at the FL and in the HAZ region. The proportion of the fracture surface that was subject to brittle fracture is plotted for each type of sample in Fig. 13(c) to exam the fracture mechanism.

Fig. 13 Absorbed energy results and proportion of brittle fracture on the fracture surface for the BM, FZ, FL and HAZ impact specimens in GMAW and ultra-NGLW S960 steel at different temperatures, (a) original absorbed energy results for sub-size specimens (55 $\times$ mm $\times$ 10 $\times$ mm $\times$ 8 mm), (b) corresponding estimates for absorbed energies in standard-sized specimens, (c) proportion of brittle fracture on the fracture surface.

alt-text: Fig. 13

Both the original absorbed energy results and the converted absorbed energy results for the BM, GMAW and ultra-NGLW specimens obey an overall trend: the absorbed energies increase with an increase in the test temperature, as shown in Fig. 13(a) and (b). The proportion of brittle fracture region for the tested specimens are summarised in Fig. 13(c), which presents consistent results with the absorbed energies in Fig. 13(a) and (b). A greater proportion of brittle fracture region on the fracture surface leads to a lower absorbed energy. There was no brittle fracture region for all the BM specimens, regardless of test temperature, which demonstrates all the BM specimens fractured in a ductile manner with good toughness.

All of the absorbed energy values for the GMAW and ultra-NGLW specimens with notches sampled at the FZ, at the FL and in the HAZ region are lower than the corresponding values for the BM from -40 to -40 °C to room temperature. It can be observed from Fig. 13 that the GMAW joints present better impact toughness than that for the ultra-NGLW joints.

The rapid cooling process characteristic of ultra-NGLW process makes hard martensite generated in the FZ and HAZ, which strengthen the welded joint, but also increased the brittleness of the joint. This is demonstrated by the results in Fig. 13: the absorbed energy values for the ultra-NGLW specimens with a notch sampled at the FZ, at the FL and in the HAZ are lower than the corresponding values for the BM from -40 to -40 °C to room temperature, especially for the FZ, which has a much lower absorbed energy values. It implies that the FZ of the ultra-NGLW specimens present lower toughness. However, the ultra-NGLW specimens with a notch placed in the HAZ presents high

absorbed energy values, which are just a slightly lower than those for the BM. For the specimens with notch in the HAZ, the notch centre was sampled at FL ± 0.5 mm. It was easy to offset the notch centre in the ICHAZ or even SCHAZ if there was a machining error because of the narrow HAZ. In addition, the cracks were easy to deviate into the SCHAZ and BM. The specimens with notch at the FL presents higher absorbed energy values than those with notch at the FZ at corresponding test temperatures changed from -40 – -40 °C to room temperature. Because prior austenite grain size in the FZ is very large, the grain boundaries have a weaker barriers prohibiting for the propagation of cracks.

The GMAW specimens with notch at the FZ and FL present higher absorbed energy values than those ultra-NGLW samples with notch at the FZ and FL at corresponding test temperatures changed from -40 – -40 °C to room temperature. This can be attributed to the relatively softer microstructure in the GMAW joint. The higher absorbed energy values for the GMAW specimens with notch at the FZ, FL and in the HAZ at corresponding test temperatures follow an order: $\text{HAZ} > \text{FZ} > \text{FL}$, $\text{HAZ} > \text{FZ} > \text{FL}$. In addition, the absorbed energy values for the GMAW specimens with notch at the FZ are very close or even a little higher than those for the FL. For the specimens with notch in the FZ, the microstructure in the FZ was predominated of ferrite with small amount of martensite, which had almost the same or a little higher hardness as the BM. However, there were a lot of inclusions in the FZ, which may cause the toughness in the FZ to be inferior to the BM. The FL notch was sampled to pass through the HAZ (real FL) at the mid-thickness point of the material (sample 50% weld metal and 50% HAZ) to delegate the nominated FL. So the machined FL notch centre would cover half CGHAZ and some reheated CGHAZ, which had very large prior austenite grain size. The grain boundaries would also have a weaker barriers prohibiting for the propagation of cracks for the FL samples and the relatively harder reheated CGHAZ also had some deteriorated effect on the toughness of the FL. While the HAZ notch was sampled at FL ± 2 mm, which would also cover some CGHAZ on the upper part of the welded joint and some reheated CGHAZ on the lower part of the welded joint. Even though the HAZ was much softer than the BM, the toughness of the HAZ was also inferior to the BM. This means that the coarse grains in the CGHAZ are having a more significant and detrimental influence on impact toughness than any benefits which may be attributed to the slightly reduced hardness. In addition, the relatively harder reheat CGHAZ may be also detrimental to the toughness of the HAZ.

The macroscopic fracture surfaces of the selected specimens tested at -20 – -20 °C are presented in Fig. 14. BM and ultra-NGLW specimen with notch in the HAZ fracture surfaces present completely ductile regions, as seen in Fig. 14(a) and (d). Brittle fracture region takes predominately part of the fracture surface for the ultra-NGLW specimen with notch at the FZ, as seen in Fig. 14(b). It can be observed that ultra-NGLW specimen with notch at the FL fractured with a brittle fracture region comprising 40% of the fracture surface, as shown in Fig. 14(c). A very small proportion of brittle fracture region ($\sim 8-10\%$)–($\sim 8-10\%$) can be observed on the fractured surfaces of GMAW specimens with notch at the FZ and in the HAZ, as shown in Fig. 14(e) and (g). Brittle fracture region takes approximately half of the fracture surface for the GMAW specimens with notch at the FL, as shown in Fig. 14(f).

Fig. 14 Selected macro-fractured surfaces at -20 °C, (a) BM, (b) ultra-NGLW specimen with notch at the FZ, (c) ultra-NGLW specimen with notch at the FL, (d) ultra-NGLW specimen with notch in the HAZ, (e) GMAW specimen with notch at the FZ, (f) GMAW specimen with notch at the FL, (g) GMAW specimen with notch in the HAZ.

alt-text: Fig. 14

Selected Charpy impact fracture surfaces tested at -20 °C were examined using SEM. All the ductile fracture regions present dimples and microvoids, with some of the dimples containing spherical inclusions. Selected ductile fracture region SEM image is shown in Fig. 15(a). The inclusions particle located in the microvoid were examined using EDX. The EDX results corresponding to the inclusion particle in the BM specimens shown in Fig. 15(a) is presented in Fig. 15(d). The inclusion particles are rich in Al, S, Ca, Mn and Fe. Cleavage fracture was confirmed in the ultra-NGLW specimens with the notch at the FZ, which were tested at -20 °C, and these specimens fractured with low absorbed energies. The brittle regions on the fracture surfaces revealed a cleavage dominated fracture, as is shown in Fig. 15(b). However, the GMAW specimens with notch at the FZ tested at -20 °C presented small equiaxed dimples, as seen in Fig. 15(c). The EDX results corresponding to the inclusion particles in the FZ of the ultra-NGLW and GMAW specimens are shown in Fig. 15(e) and (f), respectively. The inclusion particles are rich in C, Si and Fe.

Fig. 15 Fractured surfaces for the selected specimens after Charpy impact testing at ~~-20~~ ~~-20~~ °C, (a) BM, (b) ultra-NGLW specimen with notch at the FZ, (c) GMAW specimen with notch at the FZ, (d) EDX spectrum for an inclusion particle in (a), (e) EDX spectrum for an inclusion particle in (b), (f) EDX spectrum for an inclusion particle in (c).

alt-text: Fig. 15

4 Conclusions

From this investigation the following conclusions were derived:

- (1) The heat input for each welding pass, the cumulative heat input (the heat input summation of all welding passes) and the consumed filler material for the GMAW process are much higher than those for the ultra-NGLW process.
- (2) The tensile strength of the ultra-NGLW specimens were comparable to the strength of the BM, with all welded specimens failing in the BM. While the GMAW specimens failed in the soft HAZ with an approximately 100 MPa lower strength than the BM. The elongation of ultra-NGLW specimens is similar to that of the BM, while the GMAW specimens show much reduced elongation compared with that of the BM. Both the ultra-NGLW and GMAW specimens performed well in three-point bending tests.
- (3) Microstructure in the FZ of the ultra-NGLW was predominated of martensite mixed with some tempered martensite. The HAZ was also predominantly martensitic with some tempered martensite, with the prior austenite grain size reducing gradually moving from the CGHAZ to ICHAZ. However, in the FZ of the GMAW joint, the microstructure was predominated of ferrite with some martensite. While microstructure in the HAZ mainly consisted of bainite with a mixture of small amount of martensite and tempered martensite, with the prior austenite grain size reducing gradually moving from the CGHAZ to ICHAZ.
- (4) The peak hardness measured for the ultra-NGLW specimen was located in the un-tempered FZ and CGHAZ, which was ~~-40~~ ~~-40~~ HV_{0.3} than that of the FZ for the GMAW specimen. The hardness in the tempered FZ was ~~-60~~ ~~-60~~ HV_{0.3} lower than the un-tempered FZ for the ultra-NGLW specimen, which was located between the two welding passes. There was a very wider softening HAZ for the GMAW specimen, while the ultra-NGLW specimen had a very narrow softening region located in the SCHAZ.
- (5) The GMAW joint demonstrated better impact toughness than the ultra-NGLW joint. The generation of martensite in the FZ of the ultra-NGLW joint makes the FZ brittle.

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- Highlights**
- Narrow gap laser welding (NGLW) was carried out on the new S960 high strength steel.
 - The microstructures of NGLW and GMAW of S960 high strength steel were characterised.
 - The NGLW joints displayed better tensile properties than the GMAW joints.
 - The GMAW joints exhibited better impact toughness than the ultra-NGLW joints.
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Queries and Answers

Query:

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