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Effect of Tool Centreline Deviation on the Mechanical Properties of Friction Stir Welded DH36 Steel

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Abstract

Friction stir welding of steel has gone through recent tool and optimisation developments allowing the process to be considered as a technically superior alternative to fusion welding. This study expanded the scientific foundation of friction stir welding of DH36 steel to analyse the effect on weld quality when the rotating tool increasingly deviates away from the weld centreline. A centreline defect was deliberately but gradually introduced along the length of the weld seam. The tolerance to tool deviation towards both the advancing side and the retreating side of the weld was measured in terms of the transverse yield strength. Three discrete fracture modes were observed in transverse tensile specimen. Up to a tool deviation of 2.5 mm, ductile fracture in the parent material was observed and there was not a significant reduction in the yield strength of the weldment. The critical tool deviation occurred at 4 mm, where transverse tensile specimens fractured in a high strength ductile mode in the weld metal. Brittle behaviour in specimens above the 4 mm tolerance level resulted in a significant decrease in the transverse yield strength. Fracture within the weld metal was directed along the boundary between the heat-affected zone and thermo-mechanically affected zone, attributable to an abrupt change in the grain size and complexity of the two weld zones at this boundary. Friction stir welding of DH36 was found to be a tolerant joining process to the centreline deviation of the rotating tool.

Keywords: Friction stir welding; Low alloy steel; Tool deviation; Centreline defect; Mechanical properties

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1. Introduction

Friction stir welding (FSW) is an established joining process [1], predominantly applicable to light metal alloys such as aluminium and magnesium. Furthermore, it has been demonstrated that FSW of these light metals has many benefits over more commonly applied fusion joining techniques in terms of weld quality [2], durability [3, 4] and corrosion resistance [5]. FSW has also had success in joining low weldability materials, such as 2XXX and 7XXX aluminium alloys for aerospace application, eliminating hot cracking that prohibits the use of fusion welding for these alloys [6]. Rajakumar *et al.* [2] reported that the ultimate tensile strength and yield strength of FSW AZ 61A magnesium alloy was respectively 12 % and 18 % higher than a joint formed using pulsed current gas tungsten arc welding (P-GTAW). Superior fatigue crack growth resistance was observed by Balasubramanian *et al.* [3] in FSW of AA2219 aluminium alloy compared to both gas tungsten arc welding (GTAW) and electron beam welding (EBW). The same conclusion was reported in another comparator study where FSW of Al-Mg-Si alloy 6082 exhibited better fatigue performance than equivalent metal inert gas (MIG) and tungsten inert gas (TIG) welds [4]. FSW has been found to produce welds with higher grain refinement [7], overmatching of the parent material combined with lower defect levels [8] and lower distortion [9] than fusion welded aluminium alloys.

There has been increasing interest, particularly in the shipbuilding industry [10 - 13], in examining the viability of FSW of structural steels to realise the same technical advantages exhibited in friction stir welding of light metal alloys. Lienert *et al.* [10] performed an initial feasibility study on 4.5 mm thick DH36 steel friction stir welded at two different traverse and rotational speeds. In both cases, significant grain refinement of the weld compared to the parent material was observed. More, overmatching of the parent material occurred, as in previous aluminium alloy investigations; the ultimate tensile strength and yield strength was 16 % and 36 % greater in the weld than the parent material respectively [10]. Superior mechanical performance in the weld was also confirmed in a later study of 8 mm thick FSW DH36 steel [11]. McPherson *et al.* [11] additionally noted that low distortion was present in single and double-sided variants of the FSW process. Mechanical and microstructural assessments exhibited similar characteristics in FSW HSLA-65 steel [12] and 409M ferritic stainless steel [13], showing FSW to be a technically viable joining technique for steel.

However, tool durability dictates the feasibility of the process in the current market of FSW of steel [14, 15]. In particular, Meshram *et al.* [15] stated the need for advancement in tool materials if FSW of maraging steel (grade 250) is to become a feasible joining technique for aerospace application, despite the high mechanical performance of the welds. Recent developments in tool technology have allowed the process to compete with fusion welding methods, exhibiting comparable welding

speeds and improved tool life [16], with improved weld quality and reduced distortion [17]. Polycrystalline cubic Boron Nitride (pcBN) based composite tools have proven to possess excellent high-temperature strength and abrasion resistance [18], capable of welding A36 steel up to 80 m before failure [17], and consistently exceeding 45 m in weld length [19]. In concurrence with these developments, research has been conducted to investigate the process parameter window for DH36 shipbuilding steel, whereby post weld mechanical properties were optimised for a range of welding speeds [20]. Toumpis *et al.* [20] reported that the correct balance of rotational speed and traverse speed produced excellent weld mechanical properties. A high performance friction stir weld was produced at a traverse speed of 500 mm/min. Microstructural heterogeneity was observed in this weld but ductile fracture in the adjacent parent material indicated to a high transverse weld strength. All transverse tensile samples produced at welding traverse speeds between 100 mm/min and 400 mm/min fractured in a ductile mode in the parent material, the expected fracture mode for quality welds. In relation to the work by Reynolds *et al.* [21], it was concluded that high performance welds in DH36 steel can now be friction stir welded at traverse speeds up to five times faster than the earlier adopted rates of 100 mm/min, making the process a technically viable contender in the shipbuilding sector.

A comparator study between FSW and Submerged Arc Welding (SAW) of DH36 highlighted the potential benefits of friction stir welding over fusion welding [22]. A series of 4 mm, 6 mm and 8 mm thick plates were friction stir welded in single-sided and double-sided configuration and were compared against SAW. McPherson *et al.* [22] showed that all FSW variations were superior in mechanical performance than their SAW counterparts. FSW of 8 mm thick DH36 plate exhibited a maximum longitudinal distortion six times less in magnitude than the SAW equivalent and no evidence of torsional bending, unlike the SAW variant. Double sided 8 mm thick FSW plate showed the lowest maximum distortion of 10 mm over a 2000 mm long plate; the SAW equivalent was distorted by a peak value of 80 mm. In terms of fatigue performance, both low cycle and high cycle fatigue regimes performed better in FSW compared to SAW. Toughness and hardness were also of the required standard for FSW to be considered a technically viable industrial process [22].

The present study aims to broaden the scientific foundation of friction stir welding of DH36 by investigating the impact of processing defects on the mechanical properties of a butt-welded joint. For all joining processes, weld misalignment or inadvertent root gaps associated with poor fit-up, are likely to introduce intrinsic process related defects in the welded joint. It is essential to understand the tolerance to the aforementioned fit-up conditions for any joining process. In the case of FSW, the effect of increasing tool centreline deviation on the transverse yield strength of DH36 steel plate was examined, along with related microstructural effects. This highly novel study was conducted to define the tolerance level of FSW when the rotating tool increasingly deviated away from the weld centreline such that a centreline weld defect was deliberately but gradually introduced along the length of the weld seam.

Similar studies were previously performed on FSW of aluminium alloys. Widener *et al.* [23] studied the impact of tool centreline deviation on the ultimate tensile strength (UTS) of friction stir welded, 3.175 mm thick AA7075-T73 in the butt configuration. Acceptable UTS was averaged to 479 ± 1.24 MPa, with a total tolerance zone of 1.68 mm across the weld. The advancing side of the weld was two times more tolerant to tool deviation than the retreating side of the weld. A lack of consolidation at the weld root within the thermo-mechanically affected zone led to brittle fracture in the weld metal and a significant reduction in the mechanical properties [23]. The tolerance to mating variations of robotic friction stir welded, 5 mm thick AA50583-H111 was researched [24]. Cole *et al.* [24] found the UTS and yield strength of the alloy critically decreased beyond a tool deviation of 2 mm from the weld centreline, for both the advancing side and retreating side of the weld. Weld misalignment, caused by the deviation of the tool away from the weld centreline, was the principal contributor to a decrease in the mechanical properties of the weld that was induced by processing defects [23, 24].

The current study shall solely focus on the effect of tool deviation from the weld centreline on the transverse yield strength of friction stir welded DH36 steel.

2. Experimental Details

Four single-sided friction stir weldments (6 mm thick DH36 plates) were produced in the butt configuration, using a PowerStir FSW machine. Post weld plate dimensions were 400 mm x 2000 mm and each plate was denoted by the following reference numbers: W01, W02, W03 and W04. The weld on plates W01 and W02 deviated to the advancing side, where the rotating tool pushed plasticised metal towards the traverse direction, i.e. forwards. The weld on plates W03 and W04 deviated to the retreating side, where the rotating tool pushed plasticised metal in the opposite direction to the traverse direction, i.e. backwards. The PowerStir FSW machine is a moving gantry design with a large operational bed of dimensions 6000 x 4000 mm. Plates were securely clamped on the machine bed in both the vertical and horizontal direction. All plates were welded in the 'as received' condition, perpendicular to the direction of rolling, using the hybrid composite WRe-pcBN tool manufactured by MegaStir. The tool consisted of a scrolled shoulder with a stepped spiral pin of length 5.7 mm and was mounted to the FSW machine via a welding head. The basic dimensions of the tool employed in this study are provided in Figure 1. The tool rotated in an anti-clockwise direction and was protected during welding by an inert gas environment to prevent oxidation at the high operational temperatures of the FSW process for steel. The plates were welded using position control whereby the tool was set to maintain a constant plunge depth during welding irrespective of the forces that act upon it. The FSW machine was equipped with data recording capability that ensured real time monitoring of the welding operation. Sensors recorded both primary process parameters (weld traverse and rotational speeds) and

secondary response parameters (plunge and traverse forces and tool spindle torque). This data was plotted on a force summary chart, as shown in Figure 2 for welded plate W02.

Maximum tool centreline deviation did not exceed 6 mm, either side of the weld centreline. X-Ray inspection of all four plates showed no additional defects or flaws post welding. Consistent weld parameters were used: traverse speed of 250 mm/min and rotational speed of 450 rpm. Compared to other researchers' work [20], such speeds lay within an intermediate set of process parameters producing acceptable quality welds. The same grade of DH36 was used as that of previous studies [11, 19]; the composition of which is shown in Table 1.

The steady-state process region, the area in which the applied forces have stabilised, defined the starting point of weld analysis. Steady-state conditions were reached after approx. 120 mm of weld traverse and marked the initial point from which transverse tensile specimens were sectioned from all four plates. Given no tool centreline deviation, the mechanical properties at any point of the steady-state region would be indicative of the expected performance over the entire length of the weld and would be therefore used as a benchmark. The onset of the steady-state region was visually identified whereby a good quality weld surface without excessive flash formation, surface voids or cracks was observed. Further validation of the steady-state region was performed through analysis of the force summary plot for each plate, as applied in a prior investigation [20] (Figure 2). From Figure 2, both the longitudinal 'traverse' force and the vertical 'plunging' or 'Z' force have stabilised after approximately 120 mm of welding. Eighteen equidistant increments, denoted by the reference lines 1 – 18, were marked for sample extraction on the remaining welded plate lengths.

Figure 3 shows the referencing and sample extraction convention for transverse tensile specimens. Three tensile specimens and one microstructural sample were extracted from each reference line for tool deviation towards the advancing side of the weld. The three tensile specimens from each reference line verified the yield strength data calculated for advancing side tool deviation. Additionally, verification of the yield strength would concurrently confirm process parameter repeatability across all four plates. The same process was adopted for specimens with tool deviation towards the retreating side of the weld. Transverse tensile specimen dimensions adhered to ISO Standards [25,26], as shown in Figure 4, and followed the testing procedures therein. All transverse tensile tests were assessed using an Instron Servo-hydraulic 8802 250 kN uniaxial tensile testing machine. The strain rate was consistent for all tests: 0.5 mm/min up to 1.25 mm elongation; 5 mm/min thereafter until fracture. The transverse yield strength of each specimen was calculated from the elastic limit of the resultant stress-strain curves, and then expressed as a function of the tool centreline deviation towards the advancing side and retreating side of the weld.

The extraction convention for microstructural samples is shown in Figure 3. Microstructural samples aided both microstructure characterisation and tool deviation measurements, examined in ImageJ software. Tool deviation was measured from the original plate interface to the local centreline of the deviated weld path. Standard metallographic preparation techniques were used: hot mounting, grinding, polishing and etching using Nital 2%. Macrographic investigation defined key features of each weldment, allowing for further detailed analysis using optical microscopy. Optical microscopy was performed using an Olympus GX51. Metallurgical features of the weld were discussed to aid the explanation of the fracture modes of the transverse tensile specimens.

Micro-hardness testing was performed on a Mitutoyo MVK-G1 Hardness Tester, operating at a load of 200 gf. Three hardness profiles were taken from the top of the weld cross-section (near the tool shoulder location), to the bottom (near the weld root). Indentation spacing was 225 µm. Results spanned the parent material towards the advancing side of the weld to the parent material towards the retreating side of the weld.

3. Results

Macrographic and micrographic images used the following naming convention, as adopted in a prior publication [20]:

AD: advancing side of the weld, located on the left side of all macro/micrographic images.
RT: retreating side of the weld, located on the right side of all macro/micrographic images.
TMAZ: thermo-mechanically affected zone consisting of weld metal stirred during welding.
HAZ: heat-affected zone that was not directly stirred by tool assembly but subjected to heat energy from TMAZ.
PM: parent material unaffected by the FSW process.

Tool deviation towards the AD side of the weld resulted in the centreline defect, herein after referred to as the original plate interface, appearing on the RT side of the weld, and vice versa. Figure 5 shows an arbitrary macrograph displaying the important weld zones of a sample with tool deviation towards the advancing side of the weld.

A datum was defined for yield: the transverse yield strength at zero tool deviation (perfect weld alignment) using an earlier study [20], at similar process parameters. The transverse yield strength at zero tool deviation was in the range of 380 – 405 MPa. Specimens that failed in the parent material with yield strength in the specified range were characteristic of the mechanical properties expected from high quality weldments. Process parameter repeatability was confirmed across all four plates, as

shown by Figures 6 and 7. The two plots showed the transverse yield strength against increasing tool deviation towards the advancing and retreating side of the weld respectively. The right hand axes displayed the percentage strength of each specimen, normalised to the datum yield strength. The datum yield strength, hereafter referred to as the average parent material yield strength, was taken as 392.5 MPa. This value was lower than the disclosed yield strength in transverse tensile testing of DH36 in an earlier study [22] but above the specified minimum for this grade of steel. It can be seen from Figures 6 and 7 that there was little change in the yield strength from plate to plate up to approximately 4 mm tool deviation. This behaviour was consistent for all tensile test results throughout the study. Beyond 4 mm tool deviation, poor mechanical performance was consistently observed in that the transverse yield strength significantly decreased.

All data points from Figures 6 and 7 were consolidated onto a single curve, shown in Figure 8. A best fitting trend line was attached to the transverse yield strength data, with 95% confidence bounds, using the "Curve Fitting Toolbox" in Matlab. The transverse yield strength appeared to significantly decrease below 90% of the average parent material yield strength. Tolerances to tool centreline deviation were suggested at the points in which the two intersection lines, at 95% and 90%, crossed the trend line, as shown in Table 2. Figure 9 was derived from Figure 8, overlaying the three discrete fracture modes of the transverse tensile specimens relative to the increasing tool deviation. Tensile fractures were characterised as ductile parent material fracture, ductile weld metal fracture, and brittle weld metal fracture. Each fracture mode can be seen in the representative fracture surfaces and the associated stress-strain curves of W02 in Figure 10. Transverse tensile specimens that fractured in the parent material exhibited typical ductile behaviour, i.e. necking and strain hardening (Figure 10a), up to a tool deviation of 2.5 mm. This alluded to a stronger weld metal, as observed in previous studies [10, 21]. Strain-to-failure ranged from 15% - 30% elongation, approx. 2% greater than the maximum elongation measured by Reynolds *et al.* [21]. Ductile fracture within the weld metal of tensile specimens generally occurred between 2.5 mm and 4 mm for tool deviation towards both the advancing and retreating side of the weld. Fracture within this range was characterised by high yield strength, as shown in the stress-strain curve of Figure 10b. Strain-to-failure did not exceed 12% elongation. Fracture occurred in the weld, at the original plate interface. This fracture mode defined the region from which the critical tool deviation for advancing and retreating side tool deviation would be discussed, at 90% of the average PM yield strength. Brittle weld metal fracture occurred after 4 mm tool deviation, shown in the stress-strain curve of Figure 10c. These were low yield strength specimens with no identifiable yield point. Fracture in all of these samples occurred in the weld, at the original plate interface.

The microstructure of 6 mm thick friction stir welded DH36 was examined to highlight potential characteristics that influenced the tensile behaviour as tool deviation increased. Process parameter repeatability allowed the assumption that all metallurgical samples were indicative of the expected microstructure across all four

plates. The macrograph of W02.9 highlights the key areas for microscopic investigation in Figure 11.

The parent material, (Figure 12a), exhibited features common to hot rolled steel. A banded structure of proeutectoid ferrite and pearlite was evident. The coarse, equiaxed ferrite typically had a grain size in the region of 15-25 µm. The heat-affected zone (Figure 12b) consisted mainly of equiaxed grains of ferrite. Heat dissipating from the TMAZ began degeneration of the banded pearlite. Finer, dispersed colonies of pearlite were formed closer to the weld TMAZ, aligned in the direction of rolling. Ferrite grains in the HAZ were more refined than in the PM, with sizes ranging from 10-15 µm. The microstructure across the weld was generally homogeneous. Highly refined, randomly mixed grains of acicular-shaped bainitic ferrite were formed within the TMAZ, as shown in the upper RT TMAZ in Figure 12c. Prior austenite boundaries were detected throughout the TMAZ. The microstructure within the TMAZ differed from that reported by Reynolds *et al.* [21], where a mixture of bainite and martensite formed in a specimen welded at approximately 450 mm/min and 780 rpm. The presence of martensite in this study may explain the lower percentage elongation and indicated unbalanced process parameters. The microstructure of the weld was, however, in line with the findings of a previous publication [20] in which refined acicular-shaped bainitic ferrite grains were observed in a specimen welded at 350 mm/min and 450 rpm. This weld was reported to have an excellent balance of parameters. The study additionally concluded that weld microstructure homogeneity was dependant on well balanced process parameters. The upper AD TMAZ, marked in Figure 11, was a localised region (area ~ 0.5 mm^2) exhibiting a different microstructure. It appeared to contain poorly mixed bands of acicular ferrite and acicular-shaped bainitic ferrite shown in Figure 12d. Prior austenite grain boundaries were observed only in the acicular-shaped bainitic ferrite regions. The heterogeneous upper AD TMAZ region appeared to have negligible effect on the transverse yield strength of each weld specimen; fracture was in the parent material or at the location of the original plate interface.

Micro-hardness profiles were taken across the microstructural sample of W02.1, indicated in Figure 13. In general, the hardness was consistent across the TMAZ, suggesting that weld homogeneity had been achieved. Previous studies also stated that the weld micro-hardness remained relatively constant from the advancing to the retreating side [20, 21]. Moreover, one publication [20] showed that there was a variation of approx. 50 HV along a 10 mm profile for the same specimen highlighted in the microstructural results. This was compared to a specimen with up to 150 HV variation within the same set of intermediate process parameters. The upper AD TMAZ appeared to be the anomalous region, as identified by the microstructural study. The hardness peaked at 450 HV in that region, over two times the hardness of the parent material. Figure 13 also seemed to indicate a severe drop in hardness at the transitional point between the TMAZ and HAZ. The decrease was more dramatic on the advancing side of the weld. The decrease in hardness occurred at approximately 4mm away from the centreline in the upper and mid-plane, mirroring

the apparent drop in the transverse yield strength of specimens in Figure 8 at 4 mm tool deviation. This is additionally in line with the change in fracture mode identified in Figure 9. The microstructure at the boundary between the HAZ and TMAZ may be a contributing factor to the unacceptable decrease in yield strength of the weld, close to 4 mm tool deviation. The hardness profiles appeared to show that the weld metal had superior mechanical properties compared to the parent material.

4. Discussion

The intersection at 95% of the average parent material yield strength (Figure 8) coincided with the change from ductile PM fracture to ductile weld metal fracture of the transverse tensile specimens (Figure 9). Below a tool deviation of 2.5 mm, transverse tensile specimens fractured with a yield strength that was generally recorded between 380 – 405 MPa, the expected yield strength of the parent material. The unwelded original plate interface was therefore considered to be a weld root flaw, also reported elsewhere [10,21], with minimal detrimental effect on the mechanical properties of the welded joint. Acknowledging the presence of the weld root flaw, the original plate interface required to be approx. 1 mm in length through the plate thickness (see Table 3) before fracture initiated from this flaw. The intersection point at 90% of the average parent material yield strength (Figure 8) coincided with the change in fracture mode from ductile to brittle weld metal fracture, shown in Figure 9. This was taken to be the tolerance level to tool centreline deviation. The tolerances to tool deviation towards the advancing side and retreating side were recorded at 3.8 mm and 4.3 mm respectively. There was therefore a comparable tolerance level to tool centreline deviation towards both sides of the weld. Decreasing from 95% to 90% of the average parent material yield strength, the vertical length of the root flaw increased to over one quarter of the plate thickness, 1.5 – 2 mm, towards both the AD and RT side (Table 3). Beyond a tool deviation of 4 mm from the weld centreline, the tensile specimens exhibited brittle weld metal fracture with a significant reduction in the transverse yield strength thereafter.

Fracture within the weld metal may be characterised with reference to the micrograph of W03.11 (RT side tool deviation of 4.3 mm), shown in Figure 14. The micrograph highlighted the abrupt change in microstructure at the boundary between the HAZ and TMAZ, and shows that fracture initiated at the root flaw and propagated along the original plate interface as the tensile load increased. The highly refined grains of the TMAZ, compared to coarse, equiaxed HAZ grains, may have acted as a barrier to deflect propagation of the centreline weld defect away from the TMAZ. The propagation of the weld defect no longer followed the vertical path of the original plate interface. Fewer grain boundaries at the interface between the HAZ and TMAZ meant that the activation energy required for the growth of the defect was lower in this region compared to within the complex TMAZ microstructure. The path of least resistance was therefore directed along the HAZ – TMAZ boundary shown in Figure

14. Fracture along this plane was exhibited in the fracture face of Figure 10b, where the failure path was skewed to follow the curvature of the weld cross-section on the advancing side of the weld. As suggested by the micro-hardness results, and reported by previous studies [20, 21], the mechanical properties of the weld, particularly tensile strength, were superior to those of the parent material. As such, when tensile specimens fractured in the weld metal, the transverse yield strength decreased to levels comparable to fracture in the parent material, despite the increasing length of the root flaw. This can be seen in Figure 8, on the retreating side of the weld, where the transverse yield strength formed a plateau within the boundaries of the ductile weld metal fracture mode.

The transition from ductile to brittle weld metal fracture at a tool centreline deviation of 4 mm, and the subsequent decrease in yield strength, was likely to be associated with a reduction in impact toughness across the weld. Researchers [20] found that the impact toughness of high quality welds, at similar process parameters, significantly decreased when measured at a distance of 4 mm away from the weld centreline. The impact toughness dropped by approximately one third of the peak value, on both the AD and RT side. At 4 mm tool deviation, the propagation of the centreline defect would follow along the original plate interface up to a region of the weld metal that contained poor impact toughness. This induced brittle behaviour into the weld that was exacerbated by the length of the root flaw at large levels of tool deviation. Similarly, the hardness of the weld metal decreased around 4 mm away from the weld centreline, showing a drop of as much as 100 – 150 HV from the peak in the upper AD TMAZ to hardness close to the HAZ. The combination of these three factors resulted in the significant reduction in the transverse yield strength beyond a tool deviation of 4 mm.

5. Conclusions

The limits of the FSW process were identified when 6 mm thick butt welded DH36 steel was subjected to an increasing tool deviation from the weld centreline. The tolerance to a centreline weld defect was found to be 4 mm of tool deviation, at a level of 90% of the average parent material yield strength. Despite the asymmetric nature of the weld there was no recognisable difference in the tolerances levels between tool deviation towards the advancing side and the retreating side of the weld. The critical ratio between the vertical length of the root flaw (original plate interface) and the level of tool centreline deviation was found to be approximately 1:2. Friction stir welding can therefore be viewed as a tolerant joining technique, in terms of transverse yield strength, to tool centreline deviation when welded using a traverse speed of 250 mm/min and a rotational speed of 450 rpm.

Ductile fracture within the parent material indicated that high strength, quality welds were still attainable up to a tool centreline deviation of 2.5 mm. Fractures within the weld metal were predominantly reliant on the complex microstructural interactions at

the boundary of the HAZ and TMAZ. Ductile weld metal fracture between 2.5 mm and 4 mm tool deviation exhibited high transverse yield strength, at a comparable level to that of the parent material. The significant decrease in transverse yield strength, above 4 mm tool deviation, correlated to a reduction in the weld impact toughness from the weld centreline, recorded in a prior study [20]. The increasing length of the root flaw and a reduction in weld hardness towards the HAZ additionally contributed to the reduction in transverse yield strength. It was recognised that the allowable tolerance for the deterioration of the transverse yield strength would vary depending on the application of the welded joint and the operating environment therein. Fatigue testing was beyond the scope of this study but it is likely the determination of the tolerances levels to tool centreline deviation would be influenced by fatigue.

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Table Captions

Table 1 – Chemical composition of 6 mm thick DH36 steel plate.

Table 2 – Centreline tool deviation given to 95% confidence level for AD and RT tool deviation for 95% and 90% of the average PM yield strength.

Table 3 – Length measurements of the original plate interface (root flaw) through the thickness of the parent material for critical centreline tool deviations, given at percentage average PM yield strength.

Figure Captions

Figure 1 – Basic dimensions of the WRe-pcBN FSW tool.

Figure 2 – Force summary plot of W02. Steady-state region reached after 120 mm of weld traverse.

Figure 3 – Schematic of the referencing convention, using centreline tool deviation towards the advancing side. Samples with a tool deviation towards the retreating side were applied the same numeric reference markers and spacing.

Figure 4 – Transverse tensile specimen dimensions with machining tolerances. Plate thickness: 6 mm.

Figure 5 – Typical micrograph of the metallurgical zones on a friction stir welded sample with centreline tool deviation to the advancing side of the weld.

Figure 6 – Transverse yield strength variation with centreline tool deviation towards the advancing side of the weld.

Figure 7 – Transverse yield strength variation with centreline tool deviation towards the retreating side of the weld.

Figure 8 – Transverse yield strength against centreline tool deviation, with attached trend line. Additionally measured as a function of the average PM yield strength with lines of intersection at 95% and 90%.

Figure 9 – Fracture mode boundaries relative to centreline tool deviation, derived from Figure 8 transverse yield strength data.

Figure 10 – Examples of three fractured modes from tensile specimens W02, with associated stress-strain curves.

a – ductile PM fracture at 2.1 mm centreline tool deviation.

b – ductile weld metal fracture 3.7 mm centreline tool deviation.

c – brittle weld metal fracture at 4.8 mm centreline tool deviation.

Figure 11 – W02.9 macrograph with key areas labelled for microscopic study.

Figure 12 – Micrographs of W02.9 at x500 magnification, etched with Nital 2%.

a – PM.

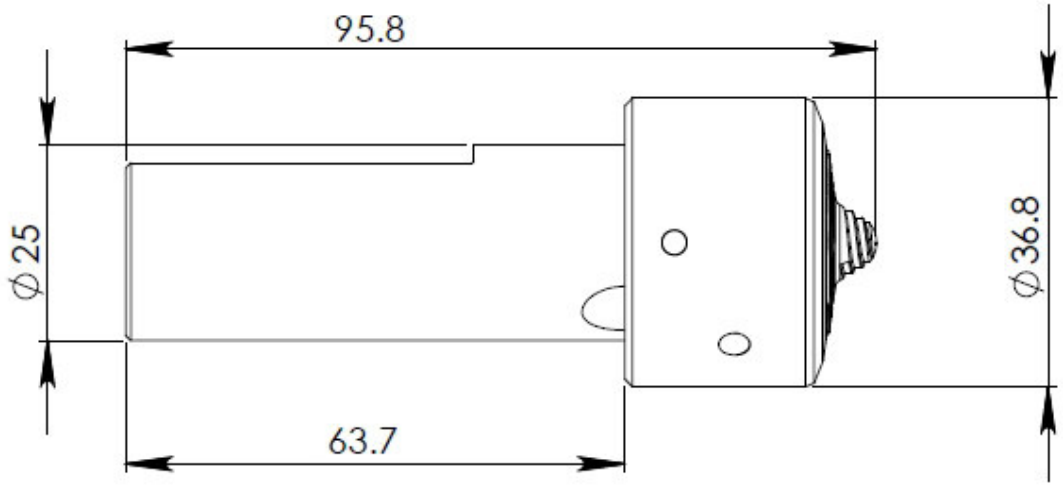
b – HAZ.

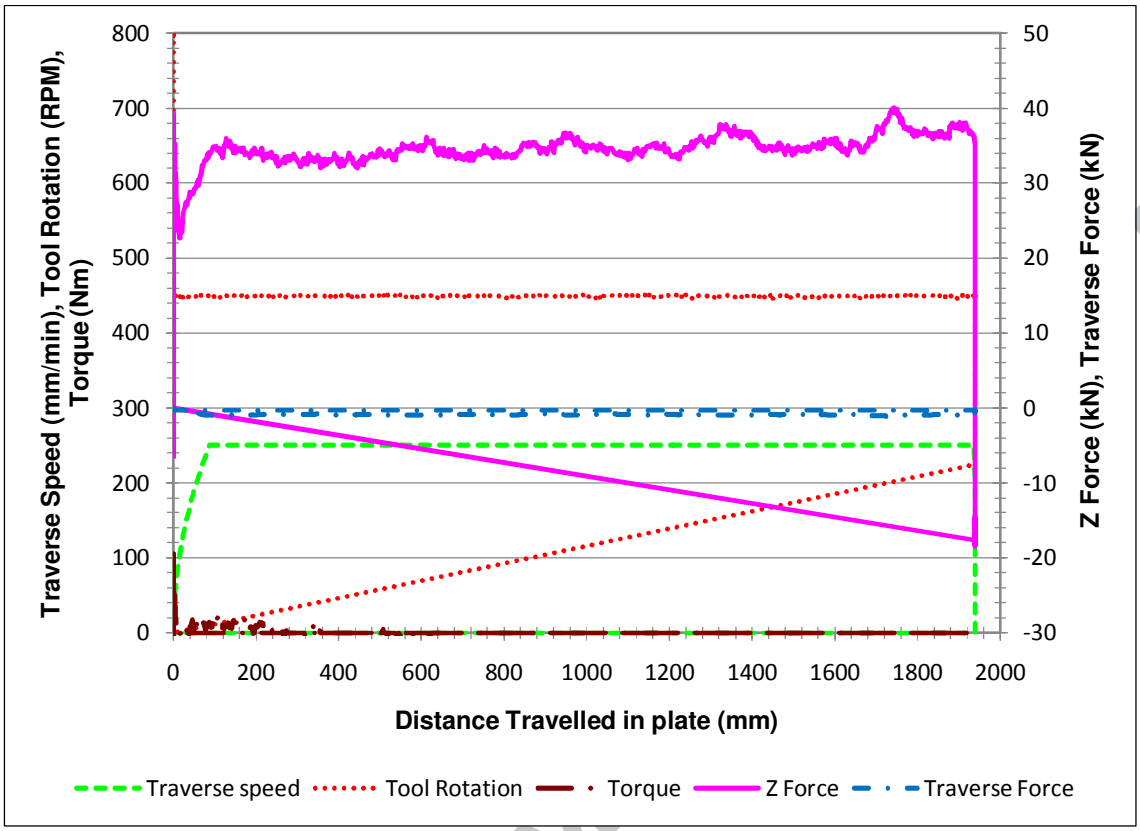
c – upper RT TMAZ.

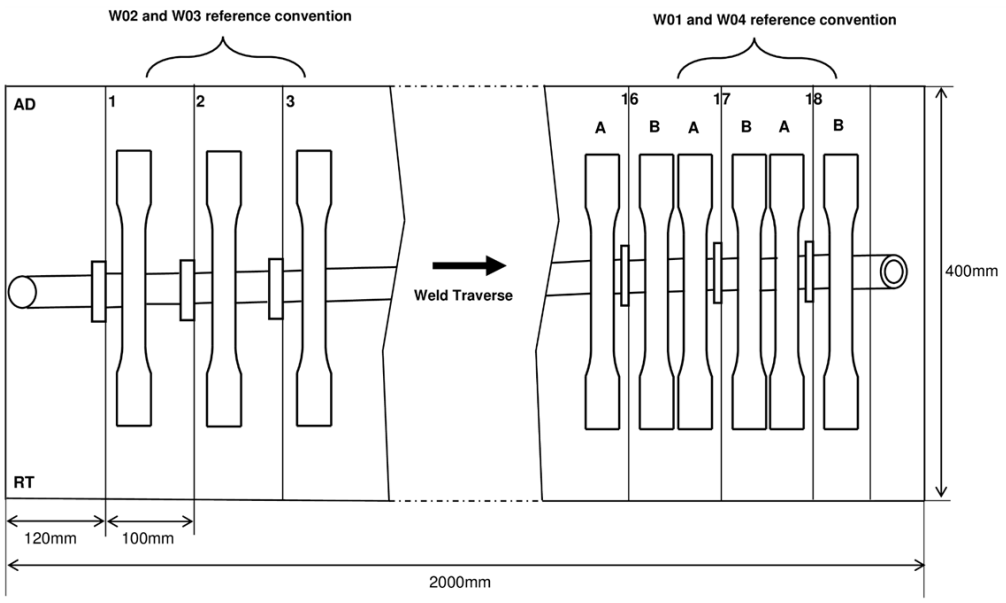
d – upper AD TMAZ.

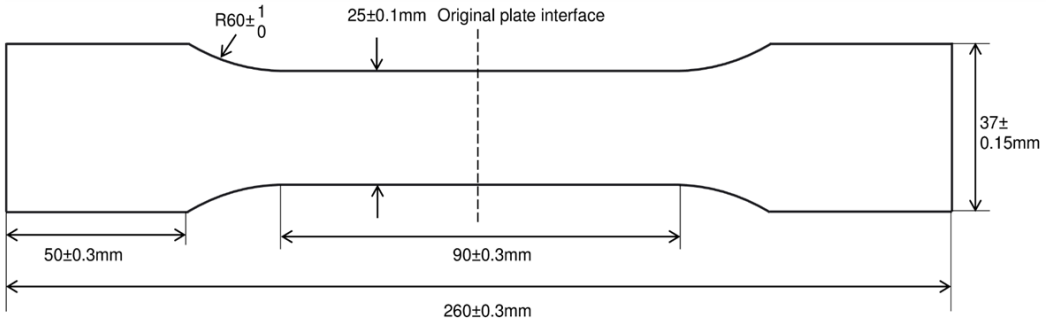
Figure 13 – Micro-hardness variations across sample W02.1. Macrograph indicates hardness locations on metallurgical sample.

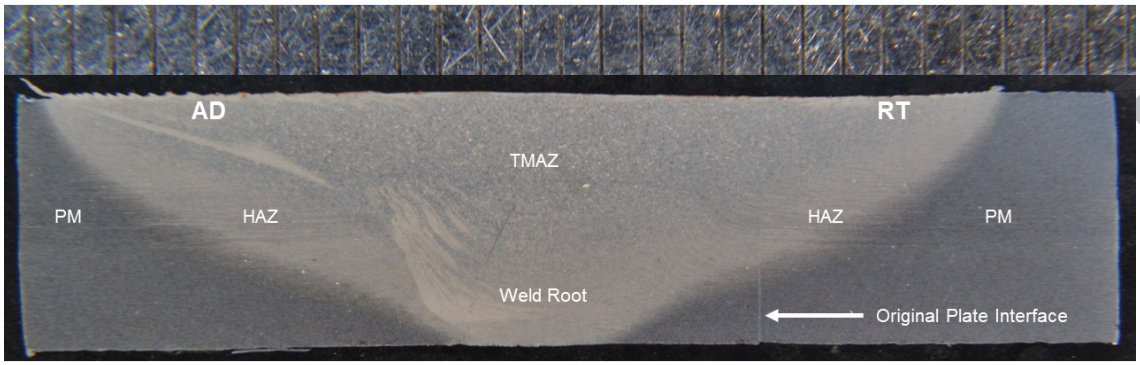
Figure 14 – Macro and micrograph of W03.11, centreline tool deviation of 4.3 mm towards the RT side of the weld. The original plate interface appeared on the advancing side of the weld.

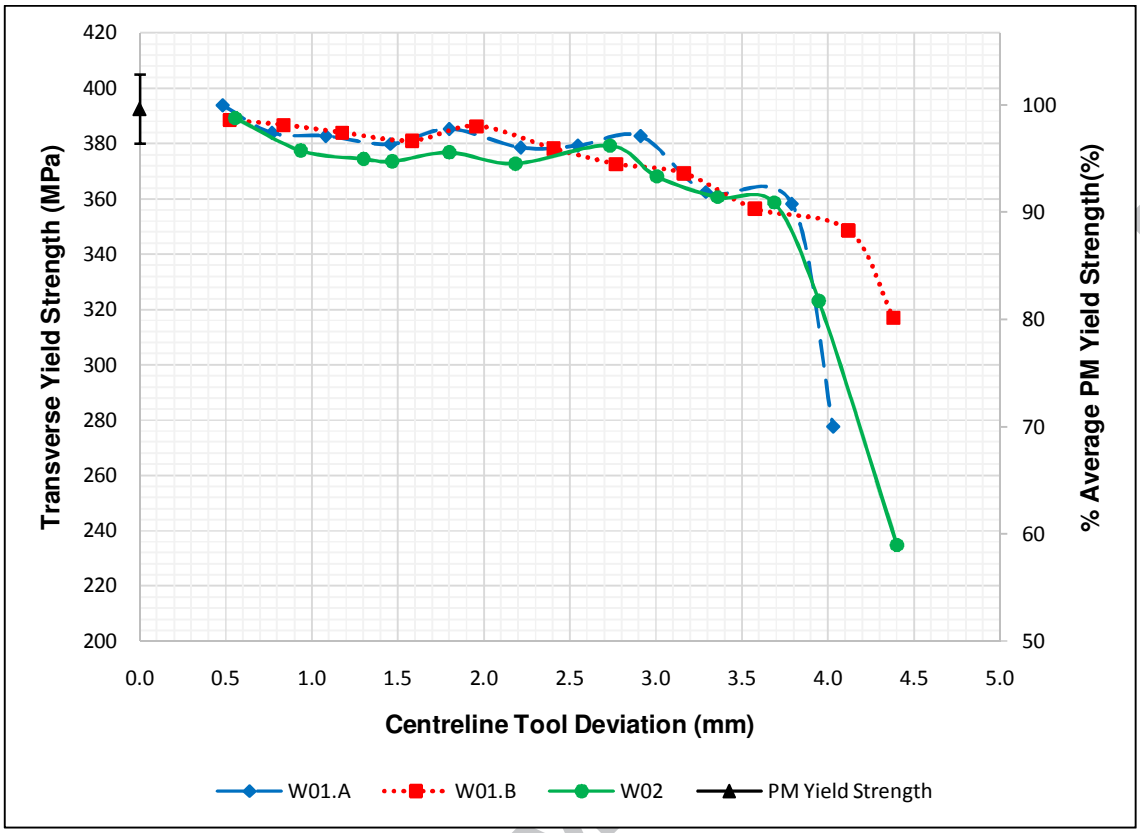


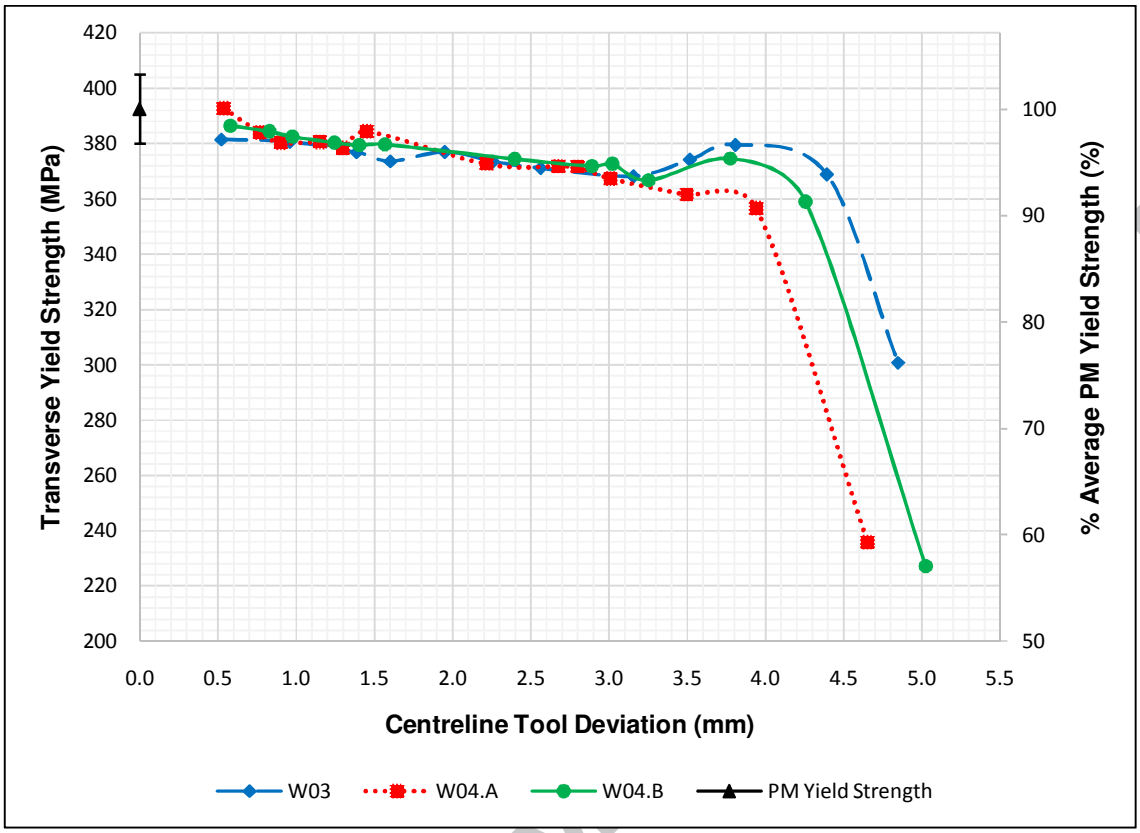


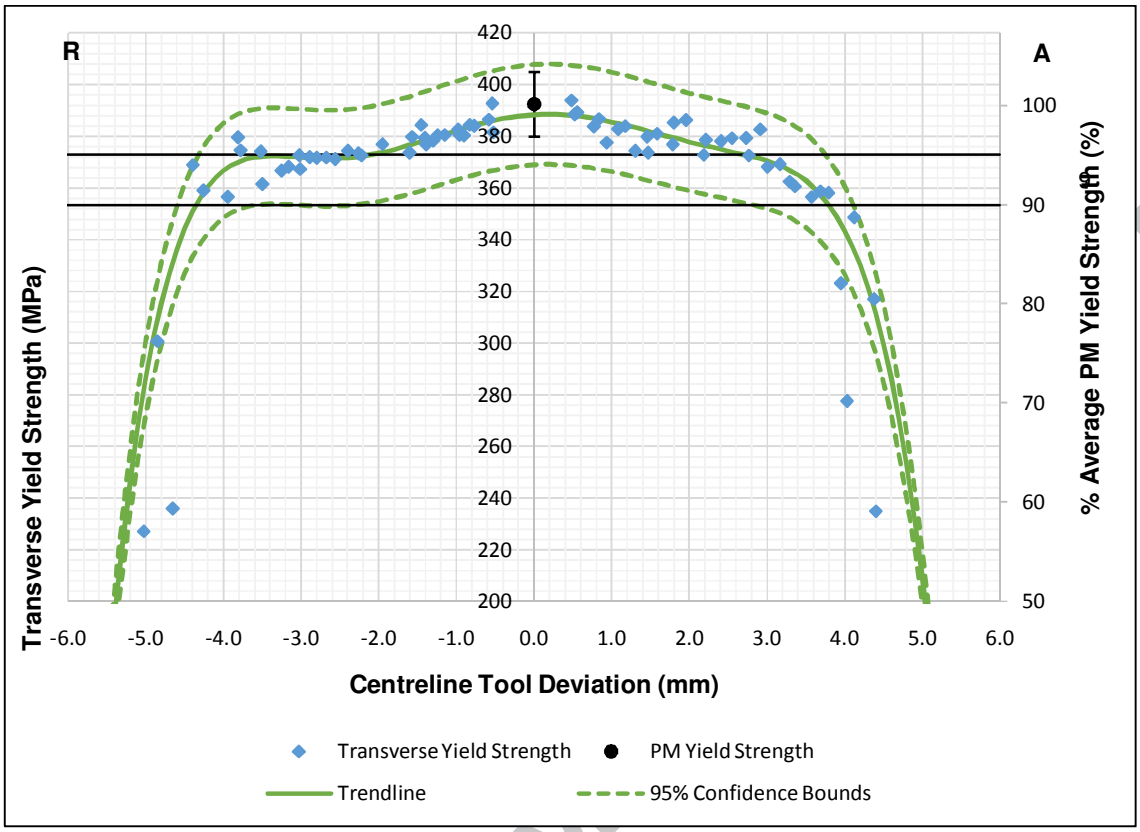


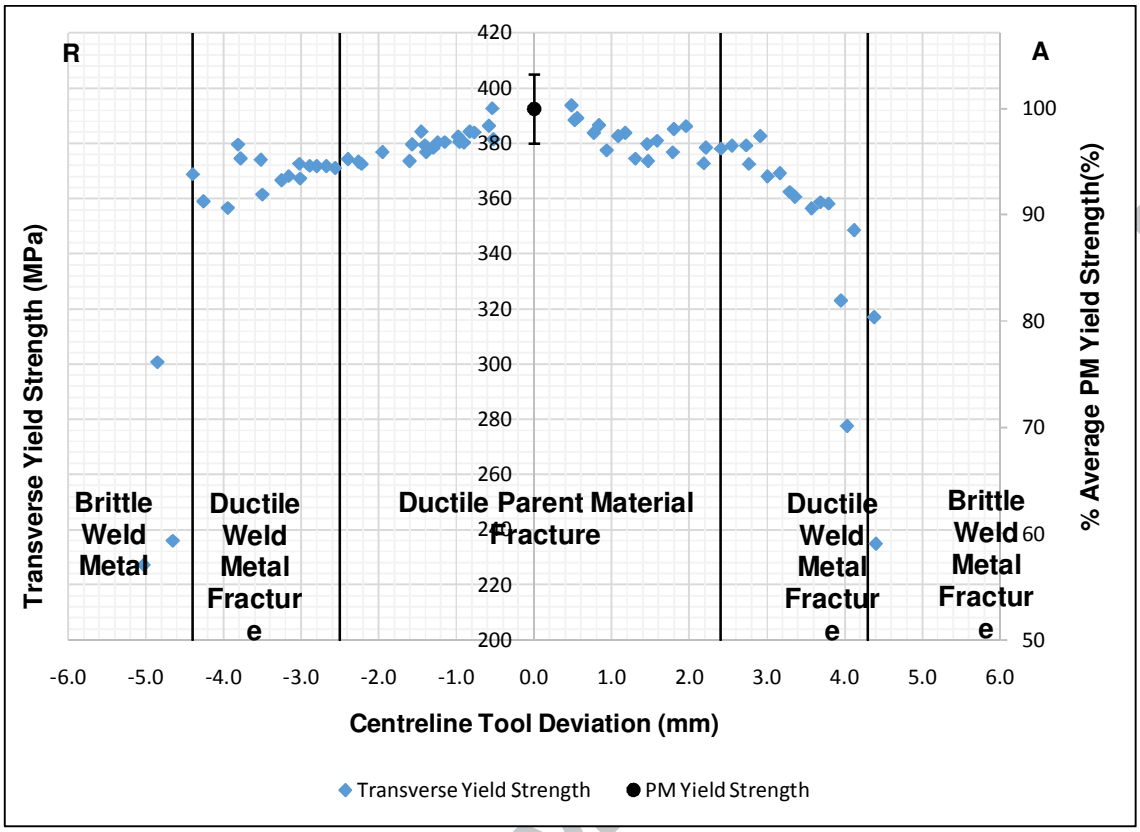


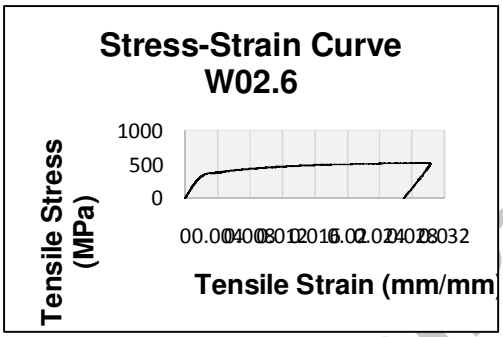


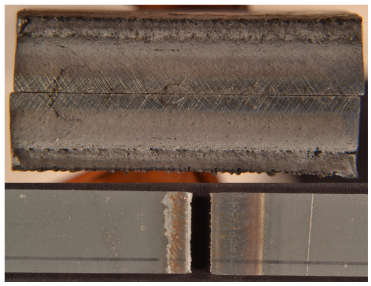


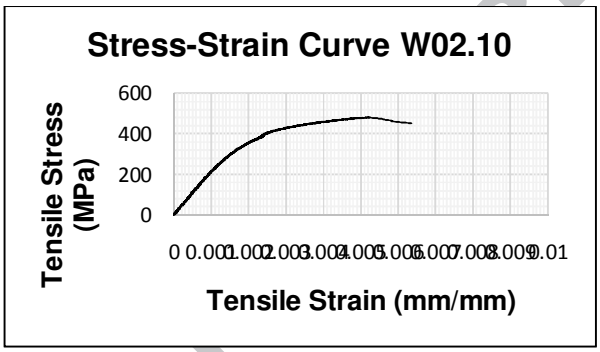


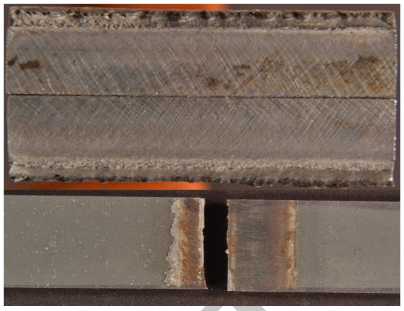


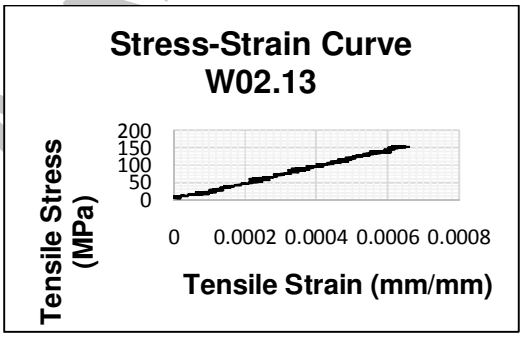


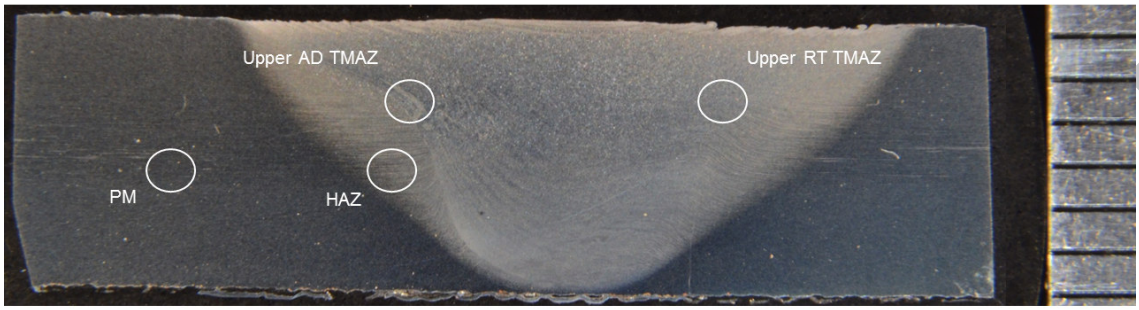


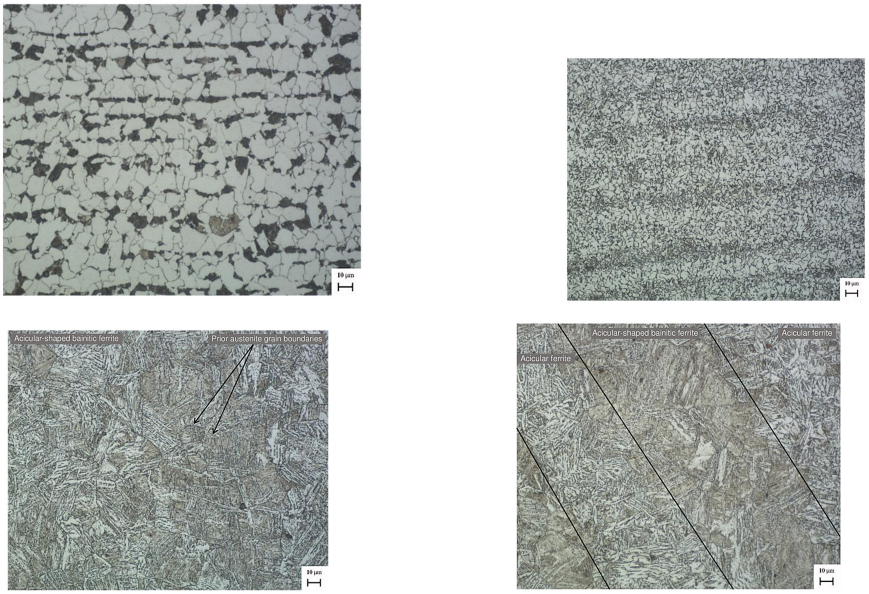


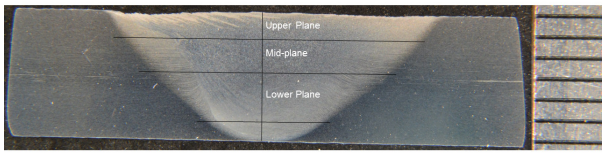


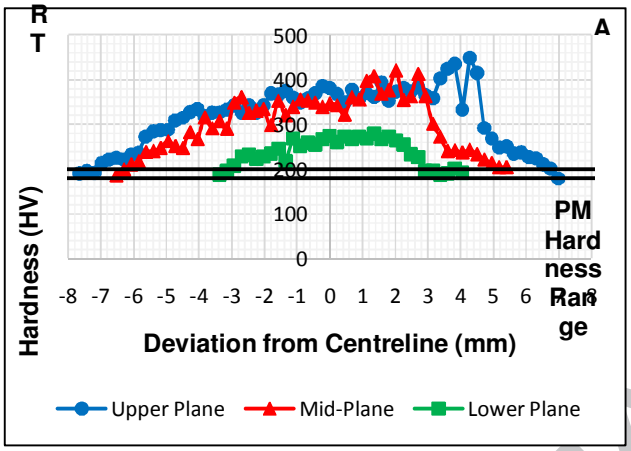


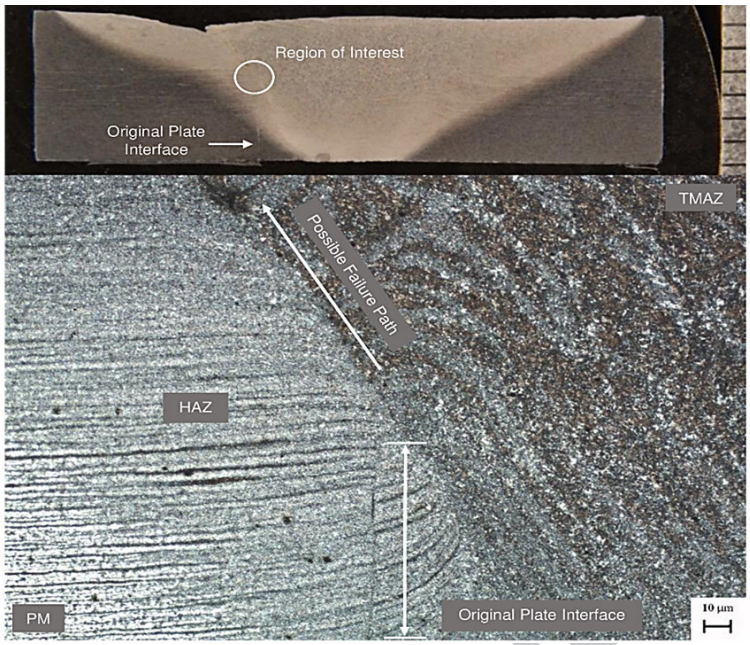












C	Si	Mn	P	S	Al	Nb	N
0.12	0.37	1.49	0.0014	0.004	0.02	0.02	0.003

% Average PM Yield Strength (%)	Centreline Tool Deviation (mm)			
	AD	*95% Confidence*	*RT*	*95% Confidence*
90	3.8	2.9 – 4.1	4.3	3.5 – 4.6
95	2.7	N/A	2.3	N/A

% Average PM Yield Strength	Advancing Side			Retreating Side		
	Tool Deviation	*Root Flaw Length*	*% Plate Thickness*	*Tool Deviation*	*Root Flaw Length*	*% Plate Thickness*
(%)	*(mm)*	*(mm)*	*(%)*	*(mm)*	*(mm)*	*(%)*
90	3.8	1.6	27	4.3	1.7	28
95	2.7	1.1	18	2.3	0.5	8

Highlights

1) FSW of DH36 was tolerant to a centreline defect induced by tool deviation.

2) High strength welds up to 2.5 mm centreline tool deviation with ductile PM fracture.

3) Critical tolerance to centreline tool deviation at 4mm with ductile weld metal fracture.

4) Brittle fracture above 4 mm deviation led to significant reduction in yield strength.