Studying the strengthening mechanisms and mechanical properties of dissimilar laser-welded butt joints of medium-Mn stainless steel and automotive high-strength carbon steel

Atef Hamada\textsuperscript{a,∗}, Sumit Ghosh\textsuperscript{b}, Mohammed Ali\textsuperscript{b,c}, Matias Jaskari\textsuperscript{a}, Antti Järvenpää\textsuperscript{a}

\textsuperscript{a} Kerttu Saalasti Institute, Future Manufacturing Technologies (FMT), University of Oulu, Nivala, FI, 85500, Finland
\textsuperscript{b} Materials and Mechanical Engineering Centre for Advanced Steels Research, University of Oulu, Oulu, FI, 90014, Finland
\textsuperscript{c} Steel Technology Department, Central Metallurgical Research and Development Institute, Helwan, 11421, Cairo, Egypt

\smallskip

\textbf{ARTICLE INFO}

Keywords:
Medium-Mn stainless steel
High-strength carbon steel
Laser welding
Post-weld heat treatment
Tempered martensite

\textbf{ABSTRACT}

In this study, a post-weld heat treatment (PWHT) at a high temperature of 700 °C for 10 min was proposed to significantly improve the mechanical performance of dissimilar laser-welded butt joints between medium-manganese stainless steel (MMn-SS) and high-strength carbon steels (HS-CS) as base metals (BMs). Electron backscatter diffraction and transmission electron microscopy were employed to characterise the microstructures and study the deformation mechanisms in the paired metals of the dissimilar joints during PWHT and tensile straining. Micro-indentation hardness tests were performed to analyse the mechanical responses in the weld zones after PWHT. The results revealed that PWHT at 700 °C significantly affected the microstructure and mechanical strength of the BMs and their weldments. A fully tempered martensitic microstructure with a recrystallised grain region was formed in the fusion zones (FZs) of the weldments, which were processed at low and high specific point energies (SPEs) of 19, 25, and 30 J. BM HS-CS was significantly strengthened owing to the interphase precipitation hardening of vanadium carbide precipitates during the proposed PWHT. The strength-ductility balance of BM MMn-SS was improved by combining the martensitic transformation and mechanical twinning mechanisms. Although, the tensile strength of BM HS-CS increased from 500 to 630 MPa and that of the metal in the FZ decreased from 1100 to 800 MPa owing to PWHT. Hence, HS-CS is a weak metal for dissimilar joints, and joint failure was promoted in it.

1. Introduction

Dissimilar metal welding between steel plates of various grades, such as carbon, martensitic, and stainless steel, has recently received extensive attention, as this process is essential for improving the cost-effectiveness in several industries such as thermal and nuclear power plants [1–4]. In the automotive industry, different joints between carbon and austenitic stainless steels are required to assemble a modern car body, that is, a body in white (BIW). Dissimilar welded joints of low-carbon and austenitic stainless steels are also employed in the chemical and shipbuilding industries owing to their complementary properties [5]. During the welding of dissimilar steels, mixing and alloying through melting are the main processes responsible for the formation of a unique structure in the weld. Thus, welded metals exhibit significantly different performances than the base metals (BMs) [6]. In this context, Liu et al. [7] and Cao et al. [8] studied the effect of laser welding parameters on the microstructural evolution, elemental distribution, and mechanical properties of dissimilar laser-welded butt joints between carbon steel-graded AH36 steel and 304 stainless steel, and between austenitic stainless steel (AISI 316L) and carbon steel-graded EH36 steel, respectively. In both studies, it was observed that a fully martensitic structure was formed in the fusion zone (FZ) of the dissimilar butt joints. Hietala et al. [9] examined the strength and microstructure of dissimilar laser-welded joints of martensitic steel, abrasion-resistant steel AR600, and cold-hardened austenitic stainless steel. They reported that the microstructure of the FZ was composed of a two-phase structure, with predominantly martensite and a small fraction of austenite.

Medium-Mn stainless steel (MMn-SS) with a low Ni content is used in structural engineering applications as a cost-effective candidate material...
to replace the widely used Cr–Ni stainless steel, AISI 304 SS [10]. However, the dissimilar metal welding of MMn-SS remains complex and needs further investigation. To date, there are only a few studies regarding dissimilar laser-welded joints between medium-Mn, low-Ni stainless steel and other metals. For instance, Vashishtha et al. [11] studied the microstructural and mechanical properties of dissimilar welded butt joints between MMn-SS grades AISI 201 and AISI 304 welded using the GTAW technique and found that the weld structure was mainly ferritic-austenitic. Ibrahim et al. [12] investigated the microstructure and tensile strength of dissimilar joints of low-Ni austenitic MMn-SS and Ni–Cr austenitic stainless steel processed at two different energy inputs. They found a duplex structure between the austenite matrix and delta ferrite in the weld. Chuainpan and Srijaroenpramong [13] studied the gas tungsten arc (GTA) welding of dissimilar welded joints between austenitic stainless steels AISI 201 and AISI 304. They determined the optimal welding parameters for the weld metal with corrosion properties comparable to those of BM AISI 304.

In a subsequent study, the same authors [14] employed GTA welding to weld dissimilar butt joints between MMn-SS grade 204Cu and 304 SS with different filler metals. Subsequently, they analysed the resulting microstructure in the FZ and its mechanical properties, as well as the pitting corrosion resistance. Mondel et al. [15] performed friction stir welding to successfully fabricate dissimilar joints between low-Ni austenitic stainless steel and ferritic stainless steel with active dynamic recrystallisation in the latter, resulting in a very fine microstructure.

Medium-Mn stainless steel with low Ni, as an economical engineering material, is freshly developed for use in the automotive industry. The high-strength carbon steel is frequently applied in the manufacturing of different structural applications, such as pipelines, and power plant industries as well as the automotive industry. Thus, the metal joining between medium-Mn stainless steel and automotive high-strength carbon steel is welding significant issue and is highly required to achieve cost-effective the renewed vehicle designs.

In our recent study [16], dissimilar butt joints between MMn-SS and high-strength carbon steel (HS-CS) were processed by laser welding at three different energy inputs. The tensile mechanical characteristics of the joints, such as strength and formability, were studied using tensile strength and Erichsen cupping tests, respectively.

It is well established that a post-weld heat treatment (PWHT) is applied to weldments to improve their mechanical performance, such as in terms of the tensile strength and toughness [17–22], overcome the heterogeneity in microstructures across the weld structure [23], and relieve the weld-induced residual stresses arising from rapid cooling, phase transformation in the FZ, and different thermal expansion coefficients of the BMs and weldments [24–27].

Moreover, PWHT significantly affects the microstructural characteristics, hardness, tensile and mechanical properties, and fatigue strength, as well as the absorption energy of weldments. Prabakaran and Kannan [28] investigated the effect of PWHT on the microstructural and tensile mechanical properties of dissimilar laser-welded joints of austenitic stainless steel (AISI 316) and low-carbon steel (AISI 1018). They found that the tensile strength and elongation of the joints improved considerably after PWHT at 960 °C. Khan et al. [29] studied the effect of the PWHT temperature on the tensile strength and bending properties of dissimilar joints between AISI 304 stainless steel and AISI 1020 carbon steel. They found an improvement in the mechanical properties of the welded joints at a lower PWHT temperature. Liu et al. [30] investigated the effects of PWHT on the microstructure of dissimilar joints of ASTM A508 carbon steel and 309L stainless steel subjected to 20 h of PWHT at 607 °C. They observed the formation of a weak carbon-depleted zone, thereby the hardness was remarkably decreased. In contrast, the study by Rozumek et al. [31] showed that heat treatment had a negative effect on the fatigue behaviour of welded joints of hot-rolled high-strength low-alloy C steel (Docol S355– a brand of steel developed by SSAB Europe Oy (Raah, Finland) for automotive industry).

Despite the importance of dissimilar laser-welded joints of low-Ni MMn-SS and carbon steel for several applications, e.g., for the automotive industry, few studies have been conducted on the subject. In our previous study [16] related to dissimilar butt joints of Docol S355 and (Type 201) MMn-SS welded by a laser at different energy inputs, we investigated the microstructural evolution, mechanical strength, and stretch-forming properties of the as-welded joints. The aim of the current study is to investigate the effect of PWHT on the microstructure and tensile strength of low-Ni, medium-Mn austenitic stainless steel (Type AISI 201) and Docol S355 and to compare the results with those of their conventional as-welded counterparts.

The deformation mechanisms in the structures of the BMs and their weldments during tensile straining were examined by comprehensive microstructural characterisation using electron backscattering diffraction (EBSD) and transmission electron microscopy (TEM).

2. Experimental procedures

2.1. Materials

The BMs used in this study were Type AISI 201 and Docol S355 supplied by Outokumpu Stainless Oy (Tornio, Finland) and SSAB Europe Oy (Raah, Finland), respectively, in the form of sheets with a thickness of 2 mm. The chemical composition of the experimental BMs is listed in Table 1.

2.2. Welding experimental setup

Laser welding was conducted between MMn-SS and HS-CS plates with dimensions of 100 mm × 100 mm. Fig. 1 illustrates the laser welding process. Autogenous welding experiments were carried out by employing a Yb: YAG diode-pumped disk laser machine (Trumpf HDL-4002, Ditzingen, Germany). In this study, different laser welding parameters, such as laser power, welding speed, and focal position level correlated with the laser beam diameter, were applied to generate different specific point energy (SPE) inputs, as shown in Table 2.

The specific point energy SPE, delivered energy to the weld specimen, is calculated based on the applied welding parameters and according to the following equation [32]:

\[
SPE = \frac{P \times D}{V}
\]

Where P is the laser power in watt, D is the laser beam diameter in mm and V is the welding speed in mm/s.

After welding, the dissimilar joints were subjected to PWHT at 700 °C for 10 min in a muffle furnace.

A high-temperature tempering process (heat treatment at 700 °C) for a short time of 10 min was applied to the weld specimens to significantly reduce the dislocation density in the tempered martensite and induce both ultrafine grain sizes in tempered lath martensite, through recrystallisation, and very fine nanoscale carbides.

2.3. Mechanical tests on the dissimilar joints

The tensile properties of the experimental BMs and the dissimilar weld joints subjected to PWHT were measured using a Zwick Z 100 tensile machine (Zwick Roell GmbH, Ulm, Germany) at a strain rate of 10 \(^{-3}\) s \(^{-1}\). An external extensometer was employed to measure the tensile properties accurately.

Micro-indentation hardness measurements of the weldments were conducted using an instrumented micro-indentation tester (CSM Instruments, United States) equipped with a Vickers indenter. The micro-indentation hardness tests were performed by ramping the load up to a maximum force of 2 N at loading and unloading rates of 66.66 mN/s with a pausing time of 15 s. Based on the loading-unloading cycles, the force-penetration depth (F-D) curves were obtained up to the maximum
indentation force (2 N).

2.4. Microstructural characterisation

The fine features of the phase structure in the FZ and heat-affected zones (HAZs) were studied using an EBSD detector with a field-emission gun scanning electron microscope (Carl Zeiss Ultra Plus, Germany). The EBSD mapping was obtained using an accelerating voltage of 15 kV. The TEM images were obtained using the JEOL 2200FS EFTEM/STEM instrument. Thin lamellae were extracted from the BM and welded joints for the TEM analysis using the focused ion beam (FIB) technique. The TEM instrument was operated at 200 kV, enabling high-resolution imaging at a high magnification to analyse the evolution and co-existence of different phases and precipitates. Accordingly, the selected area electron diffraction (SAED) patterns of the corresponding phases and precipitates were obtained.

For the microstructural examination using the EBSD technique, the MMn-SS/HS-CS welds were sectioned perpendicular to the welding direction. The weld samples, transversely sectioned, were undergone metallographic preparation according to the standard preparation technique, i.e., mechanically polished using SiC papers, then down to 1 μm by using a diamond suspension, and finally chemically polished using a 0.05 μm colloidal suspension of silica. The microstructure maps acquired by EBSD were processed using HKL-CHANNEL 5 software (Oxford Instruments, UK) to obtain detailed statistical information on the structural characteristics. For TEM imaging, the sectioned foils by FIB were electropolished for thinning using a Struers TenuPol-5 twinjet electropolisher in a solution of 10 vol pct perchloric acid and applying a voltage of 30 V.

3. Results

3.1. Characteristics of the BMs

The initial microstructures of the experimental BMs and dissimilar metals were significantly different in terms of the phase structure and grain size, whereas that of MMn-SS displayed a fully austenitic structure with a grain size of 36 μm, and that of HS-CS displayed a ferritic matrix with a small amount of pearlite. The microstructures of the BMs used in this study are described in detail elsewhere [16]. The engineering tensile flow curves of MMn-SS and HS-CS at room temperature before and after PHT at 700 °C for 10 min are shown in Fig. 2. The mechanical properties, such as the yield strength (YS), ultimate tensile strength (UTS), and elongation (El), are listed in Table 3. It can be observed that PHT has a significant effect on the flow curves of the BMs.

A striking feature of HS-CS after PHT was the disappearance of the Lüders bands after a distinct yielding. Moreover, the UTS increased from 500 to 630 MPa after PHT. In contrast to those of HS-CS, the mechanical properties of MMn-SS deteriorated: the YS decreased from 700 to 600 MPa after PHT, while the UTS slightly decreased from 900 to 880 MPa. This was attributed to the deformation modes, i.e., twinning-induced plasticity (TWIP) and transformation induced plasticity (TRIP), during the tensile straining of MMn-SS (see Section 4 for more details).

3.2. Microstructural characteristics of the welded steels after PWHT

The microstructural features and characteristics of the welded metals at different SPEs were discussed in detail in our previous study [16]. A typical lath martensite structure with a high density of low-angle grain boundaries (LAGBs) of ~57 pct is observed in the FZ structure, as shown by EBSD results in Figs. 3 and 4. In this study, the effect of PWHT on the microstructure and related tensile properties at 700 °C was investigated.

The EBSD maps of the microstructure along the weld cross-section of the metals after PWHT at an SPE of 19 J are shown in Fig. 3. The structural morphology of the grains in the FZ, highlighted by the yellow lines, is significantly different from that of the BMs, as shown in the

![Fig. 1. Schematic of the laser welding process between MMn-SS and HS-CS.](image)

![Fig. 2. Quasi-static tensile properties of the BMs, MMn-SS (blue), and HS-CS (black) at room temperature under post-heat treatment (PHT) at 700 °C for 10 min. (For interpretation of the references to colour in this figure legend, the reader is referred to the Web version of this article.)](image)

![Table 1. Chemical composition of the BMs for laser welding (in wt. %).](table)

<table>
<thead>
<tr>
<th>Steel Code</th>
<th>C</th>
<th>Si</th>
<th>Mn</th>
<th>P</th>
<th>S</th>
<th>Cr</th>
<th>Ni</th>
<th>N</th>
<th>V</th>
<th>Nb</th>
</tr>
</thead>
<tbody>
<tr>
<td>AISI 201</td>
<td>0.027</td>
<td>0.4</td>
<td>6.8</td>
<td>0.031</td>
<td>0.001</td>
<td>17.2</td>
<td>4.5</td>
<td>0.203</td>
<td>–</td>
<td></td>
</tr>
<tr>
<td>Docol S355</td>
<td>0.1</td>
<td>0.03</td>
<td>1.58</td>
<td>0.025</td>
<td>0.01</td>
<td>0.03</td>
<td>0.34</td>
<td>0.004</td>
<td>0.2</td>
<td>0.02</td>
</tr>
</tbody>
</table>

![Table 2. Laser welding parameters of dissimilar joints of HS-CS and MMn-SS.](table)

<table>
<thead>
<tr>
<th>Laser power, kW</th>
<th>Welding speed, mm/s</th>
<th>Focal position level, mm</th>
<th>Laser beam diameter, mm</th>
<th>Specific point energy, J</th>
</tr>
</thead>
<tbody>
<tr>
<td>2.4</td>
<td>60</td>
<td>–5</td>
<td>0.62</td>
<td>25</td>
</tr>
<tr>
<td>2.4</td>
<td>80</td>
<td>–5</td>
<td>0.62</td>
<td>19</td>
</tr>
<tr>
<td>3</td>
<td>80</td>
<td>–7</td>
<td>0.80</td>
<td>30</td>
</tr>
</tbody>
</table>
image quality (IQ) map in Fig. 3 (a). MMn-SS exhibits austenitic coarse grains, while HS-CS exhibits ferritic fine grains. However, the packet and block morphologies of the martensite structure are evident in the microstructure of the FZ. The corresponding grain-boundary misorientation map is shown in Fig. 3 (b). There is a significant increase in the density of the LAGBs (shown in green) in the tempered structure of the FZ.

It has been reported that the lath boundaries of the fresh and tempered martensite (TM) are LAGBs (<15°), while other boundaries of the martensitic structure, for example, prior austenite grains (PAGs) as well as block and packet boundaries, are high-angle grain boundaries (>15°) [33,34].

It is well known that when a highly dislocated fresh martensite is subjected to PWHT at a high temperature, the dislocation and LAGB densities decrease because of annihilation and recovery [35]. The corresponding phase map is used to examine thermal transformations of the martensitic structure of the FZ subjected to tempering (Fig. 3(c)). Both the martensite phase in the FZ and the ferritic matrix of HS-CS were indexed by EBSD mapping as an α-bcc crystal. The FZ and BM HS-CS are shown in red. No austenite formation is observed in the FZ. However, EBSD could not index precipitates or carbides that evolved during tempering. This will be examined by secondary electron (SE) imaging.

The magnified microstructures of the FZ and HAZs of BM MMn-SS after PWHT are shown in Fig. 4. As shown in the IQ map in Fig. 4(a), the grain structure of the FZ after PWHT is heterogeneous, as some regions exhibit more of the recrystallised grains (highlighted by yellow dashed circles). This is attributed to tempering at a high temperature of 700 °C. Consequently, recrystallisation of the martensitic lath structure is achieved. Recent reports have indicated that the recrystallisation of fresh martensite occurs in the tempering temperature range of 600–700 °C [36]. Moreover, some residual stresses generated during welding are relieved. Misorientation boundary maps of the martensite structure are shown in Fig. 4(b). The density of the misorientations along the recrystallised grains is lower than for those in the other regions of the lath structure. The corresponding phase map in Fig. 4(c) indicates the constituent phase in the FZ, that is, TM without retained austenite.

The microstructure of HC-CS on the other side of the FZ is shown in Fig. 5. In a similar manner as before, as highlighted by the yellow circles,

<table>
<thead>
<tr>
<th>Steel</th>
<th>Mechanical properties</th>
<th>Before PHT</th>
<th>After PHT, at 700 °C</th>
<th>Before PHT</th>
<th>After PHT</th>
</tr>
</thead>
<tbody>
<tr>
<td>HS-CS (Docol S355)</td>
<td>YS, MPa</td>
<td>420</td>
<td>440</td>
<td>700</td>
<td>600</td>
</tr>
<tr>
<td>MMn-SS (Type 201)</td>
<td>UTS, MPa</td>
<td>500</td>
<td>630</td>
<td>900</td>
<td>880</td>
</tr>
<tr>
<td></td>
<td>EL, %</td>
<td>32</td>
<td>18</td>
<td>60</td>
<td>72</td>
</tr>
</tbody>
</table>

Fig. 3. EBSD maps across the weld section of the dissimilar laser-welded butt joint at an SPE of 19 J after PWHT: (a) IQ map, (b) grain-boundary misorientation map, and (c) phase maps.
Recrystallised grain regions emerge from the lath martensite structure (Fig. 5(a)). A prominent feature of the PAGs is the appearance of an elongated columnar morphology touching the fusion boundary of HS-CS. However, this morphology is not observed in the FZ next to MMn-SS (Fig. 4). It is reasonable to assume that the columnar morphology of the PAGs, which appears in the FZ near HS-CS, might be caused by the high thermal conductivity of carbon steel (compared to stainless steel), which results in rapid cooling, thereby leading to an increase in the unidirectional heat flux towards HS-CS.

The inverse pole figure (IPF) colour map in Fig. 5(b) shows the orientation of the grain structures on the FZ and HAZ of HS-CS. It is well known that a single PAG is composed of several packets that are aggregated into blocks with the same orientation plane $\{111\}_\gamma$ [37]. Here, every PAG exhibits packets of different orientations. Furthermore, the recrystallised martensitic grains have different orientations than the parent packets.

The PAGs and effective grain size of the tempered martensitic structure in the FZ were analysed by processing the EBSD maps using MATLAB software. Based on the analysis and observations of the processed EBSD maps in Fig. 6, the SPE input has a significant effect on the average size of the PAGs. As the SPE increases from 19 to 30 J, the PAG size increases from 32 to 43 $\mu$m. The formation of coarse grains larger than 100 $\mu$m is stimulated at a high SPE (30 J), as shown in Fig. 6(h) by the yellow-coloured grains. The corresponding histogram of the PAG distributions in Fig. 6(i) verifies the presence of the large grains ($>100$ $\mu$m). In agreement, Hietala et al. [38] found that the higher the energy inputs, the higher the PAG size during the welding of the martensitic steel AR600, as evidenced by the significant increase in the PAG size.
from 34 to 67 μm at energy inputs of 60 and 320 J/mm, respectively.

The analytical results for the effective grain sizes of the martensitic structure in different FZs at different SPEs are shown in Fig. 7. The lath structure morphology of the martensitic structure gradually increases with increasing SPE input (from 19 to 25 J), as shown in the IQ maps in Fig. 7(a, d, and g). Small and fine grains are visible at a lower SPE of 19 J (Fig. 7(a)), with an effective grain size at 90% of the cumulative grain size distribution ($D_{90\%}$) of 4.8 μm, as shown in the corresponding histogram of the grain size distribution in Fig. 7(c). Upon increasing the SPE input to 25 J, $D_{90\%}$ significantly increases to 10.8 μm. This can be attributed to two factors: first, the PWHT treatment at 700 °C enhances recrystallisation of the martensite phase [36], and second, the small PAGs that form at a low energy input drive the formation of small packets and blocks in the martensitic structure [39]. Consequently, the density of the packet/block boundary nuclei (i.e., pre-existing nucleation sites) used for recrystallisation increases. Thus, more of the fine small-grained regions are formed at an SPE of 19 J. However, at a high SPE of 30 J, $D_{90\%}$ decreases to 6.3 μm. This can presumably be attributed to the promotion of partial auto-tempered martensite at high energy input. This agrees with the study reported by Hietala et al. [38].

3.3. Investigation of the microstructural evolution using SE imaging

SE imaging at different magnifications reveals the typical morphology of the TM (Fig. 8). The TM features show narrow raised bands, i.e., ridge-like regions composed of clusters of fine martensite laths (highlighted regions in Fig. 8(a)). Babu et al. [40] reported similar observations of ridge-like regions for auto-TM in low-C steel. At a higher magnification of the highlighted regions shown in Fig. 8(b), the orientations of the ridges vary in the different regions. Interestingly, carbides are absent within the ridge-like regions, as observed in Fig. 8(a) and (b). At a high SPE of 30 J, a broken lath-like morphology is predominant in the lath structure (Fig. 8(c)). SE imaging using an in-lens detector was employed to detect the carbides. Fig. 8(d) clearly shows nanoscale particles within the tempered structure. These particles are discussed in more detail in Section 4.

It is now well established that the precipitation of carbide particles during the tempering of martensite is often initiated at three different types of boundaries in the TM: the PAG boundaries and block boundaries, lath boundaries (inter-lath carbide), and within the tempered structure (intra-lath carbide) [41–44]. Note in particular the aligned blocks in one packet with the same crystallographic orientation in the
TM, marked by the red arrows in Fig. 8(e). The size of the block is approximately 1 μm. Zhuo et al. [45] reported that the size of martensitic laths increased during tempering. They found that the average lath width increased from 387 nm for quenched martensite to 489 nm for TM.

### 3.4. Tensile properties of dissimilar metal joints

The engineering stress-strain curves of the dissimilar weld joints before and after PWHT are shown in Fig. 9. The flow curves of the joints are significantly affected by PWHT. Generally, the tensile properties of dissimilar joints subject to welding and PWHT are similar to those of BM HS-CS, since the behaviour and shape of the tensile flow curves for both joints are identical to the corresponding flow curves of BM HS-CS, as shown previously in Fig. 2. The flow curves of the as-welded joints display a discontinuous yielding behaviour with the Lüders effect, which disappears in the joints after PWHT. This can be attributed to the tensile deformation being transferred to the weak metal in the dissimilar joint, which is BM HS-CS. Meanwhile, the structures of BM MMn-SS before (metastable γ-fcc) and after (TM) welding show high deformation resistance and plastic deformation capacity during the tensile tests, as will be discussed in detail in Section 4. As a result, the BM HS-CS side is exposed to a higher stress than its YS, leading to fracture of the dissimilar joints with a necking phenomenon on the HS-CS side.

In the present study, the Vickers hardness (HV) of the two dissimilar joints welded at SPEs of 19 and 30 J subjected to PWHT was evaluated using a Vickers diamond indenter equipped with an instrumented micro-indentation tester (CSM Instruments). The load–penetration depth (P-h) curves extracted from the indentation testing of the different zones in the joints after PWHT are displayed in Fig. 10. The FZ (red colour) shows very low penetration depths of 2.1 and 2.3 μm at SPEs of 19 and 30 J, respectively. The corresponding Vickers hardness values are significantly high, 538 and 500 HV, respectively. This is attributed to the TM structure of the FZ. In addition, BM MMn-SS and its HAZ exhibits lower penetration depths (i.e., higher indentation resistance) than BM HS-CS and its HAZ.

### 4. Discussion

In this section, the deformation and strengthening mechanisms of the
BMs and their weldments are interpreted based on the detailed observations of the microstructural features using EBSD and TEM. The BMs (MMn-SS and HS-CS) of the dissimilar joints exhibit significant variations in their mechanical properties after the heat treatment, as shown previously in Fig. 2. This is attributed to the various operative deformation mechanisms of each metal. It is interesting to observe that MMn-SS reveals two deformation mechanisms: martensitic transformation (TRIP mechanism) with a fraction of approximately 15% and mechanical twinning (TWIP mechanism) with a fraction of approximately 18% (Fig. 11(a)). It is well established that both mechanisms enhance the strain hardening and ductility of steel [46–48]. Hence, MMn-SS displays a superior elongation of approximately 70% and a high tensile strength of 900 MPa. At a high magnification, \(\alpha'\)-martensite is mainly formed in the banding pattern, as shown in Fig. 11(b). According to several studies, \(\alpha'\)-martensite is generated within highly localised shear strain bands [49–51]. The orientation map in Fig. 11(c) reveals the occurrence of an active deformation mechanism related to the grain orientation. For instance, the newly generated TRIP \(\alpha'\)-martensite is induced as a rod-shaped morphology in grain \(A\) with a 111 orientation, which is the favoured orientation for martensitic transformation. This is in agreement with the shear model of deformation-induced martensitic transformation in metastable austenitic steels with a low stacking fault energy (SFE) reported by Lee et al. [47]. However, deformation twins with \(\Sigma3\) twin boundaries (yellow colour) are located in grain \(B\) with a 101 orientation. Grain \(C\) with a 001 orientation shows neither the TWIP nor TRIP mechanism. Grain \(C\) might have a higher Schmid factor for twinning and martensitic transformations. Reports from previously published studies confirm that deformation twinning and \(\alpha'\)-martensitic transformation have mainly occurred in grains with relatively small Schmid factors for the Shockley partial dislocations to promote intrinsic and extrinsic stacking faults [46,52]. Accordingly, the feasibility of martensitic transformation and twinning in 111- and 101-oriented grains with low Schmid factors depends on the formation of the local strain concentration required for the nucleation of the martensitic

Fig. 8. SE images of TM: (a) ridge-like regions at an SPE of 25 J, (b) high-magnification image of the highlighted regions shown in (a), (c) broken lath-like morphology of TM at a higher SPE of 30 J, (d) image of TM using an in-lens detector, and (e) magnified view of the martensite laths.

Fig. 9. Effect of PWHT on the tensile flow curves of dissimilar laser-welded butt joints processed with different SPEs.
transformation and twinning. A prominent feature is observed in the misorientation map shown in Fig. 11 (d). The deformation-induced martensite shows a higher local misorientation angle in grain A compared to the local misorientation upon twinning in grain B and a perfect slip in grain C. Kamali et al. [53] reported that \( \alpha' \)-martensite increases in the fraction of LAGBs during the bending deformation of low-Ni-Cr-Mn stainless steel.

The tensile-induced microstructural features of BM HS-CS were observed using EBSD, and the results are shown in Fig. 12. Because HS-CS is deformed via a dislocation-mediated plasticity mechanism, the IQ map in Fig. 12(a) displays a unified grain structure, \( \alpha \)-bcc, without inducing another deformation mechanism such as TWIP or TRIP. The
kernel average misorientation (KAM) map of the deformed grain structure is shown in Fig. 12(b). A higher local misorientation angle in the grain boundaries than in the grain interior is observed in the KAM map, indicating a strain heterogeneity distribution. This is attributed to the slip activity of BCC ferrite (α-bcc) via the slip and cross-slip of dislocations on different glide planes, which, in turn, causes dislocation annihilation and relieves the local strain concentration [54]. However, the strengthening mechanism of HS-CS after PWHT was not detectable using EBSD imaging. Hence, TEM was performed to examine the microstructural evolution after PWHT.

Fig. 13 shows TEM images of the microstructure of BM HS-CS after PWHT at 700 °C for 10 min. In summary, very fine precipitates are formed on the grain boundaries and inside the grains. The various types of precipitates responsible for the enhancement of the tensile strength of BM HS-CS to approximately 630 MPa were determined using TEM (Figs. 2 and 9). A polygonal ferrite (α-bcc) grain structure is observed in the bright-field image (Fig. 13(a)). A typical arrangement of interphase precipitates aligned in a certain direction within the polygonal ferrite is observed in Fig. 13(b) at higher magnification. Energy-dispersive X-ray spectroscopy (EDS) was conducted to quantitatively measure the precipitates, and a clear vanadium peak is observed in Fig. 13(c). Furthermore, the interplanar spacing, which was measured using the obtained SAED pattern of the [111] zone axis, confirmed the formation of vanadium carbide (VC; Fig. 13(d)). In addition, heat treatment at 700 °C facilitated the formation of interphase VC precipitates, which enhanced the strengthening of HS-CS.

The strength of dissimilar joints is closely related to the initial failure of the weakest metal. The joints presented in this study are composed of three different metals: MMn-SS, HS-CS, and the metal in the weld zone. Tensile strength tests of the joints revealed that the failure is located on the HS-CS side (Fig. 9). Hence, we explored the effect of PWHT on the mechanical strength of the metal in the weld zone by measuring the tensile strength of this zone only. Tensile flow curves obtained from the weld zone with and without tempering are shown in Fig. 14. As expected, softening is enhanced during tempering, mainly because of the reduction in dislocation densities within the lath crystals of TM. The strengths of the as-welded and tempered weld zones are 1080 MPa and 800 MPa, respectively. However, these values are higher than those of HS-CS (630 MPa) tempered at the same temperature. Interestingly, the ductility of the tempered joints increases to twice its original value.

![Fig. 13. TEM images of BM HS-CS tempered at 700 °C for 10 min: (a) bright-field image of the microstructure at a low magnification, (b) magnified view of the highlighted region shown in (a), (c) EDS profiles of the highlighted zone shown in (b), and (d) SAED pattern of the precipitates highlighted in (b).](image-url)
Hence, it is expected that the high ductility of TM in the tested dissimilar joints will result in a superior strength-toughness balance.

TEM analysis of the FZ subjected to PWHT revealed a low dislocation density within the fine, equiaxed recrystallised grains of TM, suggesting the recrystallisation of martensite and recovery of the dislocation substructure for the heat-treated FZ (Fig. 15(a)). The magnified view of a selected location in 15 (a) is shown in Fig. 15 (b). Fine precipitates are located on the grain boundaries in Fig. 15(b). The EDS profiles of the two precipitates show peaks of chromium and carbon (Fig. 15(c and d)). This indicates that chromium carbide precipitates are formed in the FZ. The electron probe microanalysis (EPMA) analysis and distribution of the alloying elements in the FZ were reported in our previous work [16]. We found that the chromium content in the FZ increases with increasing SPE as a result of the contribution and dilution of MMn-SS in the weld pool. Based on the comprehensive TEM observations, it is reasonable to assume that the proposed PWHT promotes the recovery, recrystallisation, and formation of chromium carbide precipitates in the tempered structure of the FZ.

5. Conclusions

The effect of PWHT on the microstructure and tensile strength of dissimilar laser-welded metal joints between MMn-SS and HS-CS,
processed at different SPEs, was investigated using EBSD and TEM. The following conclusions can be drawn based on the experimental observations from this study:

1. A fully tempered martensitic structure was formed in the FZs of the dissimilar metal joints subjected to PWHT. Moreover, EBSD and TEM observations revealed the existence of recrystallised grain regions in the TM zones.

2. PWHT enhanced the tensile strength of BM HS-CS from 500 to 630 MPa and improved the ductility strength of MM-SS by activating the TRIP and TWIP mechanisms. Accordingly, the tensile strengths and flow behaviours of the dissimilar joints were improved by the strengthening of BM HS-CS.

3. The fine microstructural features of the TM, such as PAGs and the fresh martensite, were affected by the SPE. The effective grain size at 90% of the cumulative grain size distribution, D90 eff, of the TM increased from 4.83 to 10.80 μm with increasing energy input from 19 to 25 J, respectively. However, at a high energy input of 30 J, D90 eff decreased to 6.32 μm.

4. PWHT at 700 °C led to a significant increase in the ductility of the TM and loss in the UTS compared to the fresh martensite in the as-welded specimens. The reason for that was a partial recovery of martensite, which resulted in a reduction in the dislocation density. This presumably improved the toughness of the dissimilar joints subjected to PWHT.

5. The hardening mechanism of the precipitates in the BM due to the formation of carbide precipitates of VC, after PWHT on low-alloy C steel at 700 °C, was revealed by the TEM images. Furthermore, the TEM images revealed the existence of equiaxed recrystallised grains of TM and chromium carbides in the FZ after tempering at 700 °C.

Acknowledgements

The authors express their gratitude to the Interreg Nord Program and the Regional Council of Lapland for funding this research through the InTeMP project, No. NYS 20202486. Also, “Jane and Aatos Erkko Foundation” is appreciated for partly funding this study. Outokumpu Stainless Oy (Tornio, Finland) and SSAB Europe Oy (Raashe, Finland), are gratefully acknowledged for providing the experimental materials.

References


