Effect of Hardfacing Consumables on Ballistic Performance of Q&T Steel Joints

M. BALAKRISHNAN a,*, V. BALASUBRAMANIAN b, G. MADHUSUDHAN REDDY c

Abstract

This study was carried out to evaluate the effect of hardfacing consumables on ballistic performance of armour grade quenched and tempered (Q&T) steel welded joints. To evaluate the effect of hardfacing consumables, joints were fabricated using 4 mm thick tungsten carbide (WC)/chromium carbide (CrC) hardfaced middle layer; above and below which austenitic stainless steel (SS) layers were deposited on both sides of the hardfaced interlayer. Shielded metal arc welding (SMAW) process were used to deposit all (hardfaced layer and SS layers) layers. The fabricated joints were evaluated for its ballistic performance, and the results were compared with respect to depth of penetration (DOP) on weld metal and heat-affected zone (HAZ) locations. From the ballistic test results, it was observed that both the joints successfully stopped the bullet penetration at weld center line. Of the two joints, the joint made with CrC hardfaced interlayer (CAHA) offered better ballistic resistance at weld metal. This is because its hardness is higher due to the presence of primary carbides of needle shape, polyhedral shape and eutectic matrix containing a mixture of $\gamma + M_7C_3$ carbides in the CrC hardfaced interlayer. The scattering hardness level in the WC interlayer, the matrix decomposition resulted lower hardness and the co-existence of $\delta$ ferrite in the interface between hardfacing and SS root/SS cap could be attributed to the inferior ballistic resistance of the joint made with WC hardfaced interlayer (WAHA joint).

Keywords: Ballistic performance; Armour steel; Hardfacing; Microstructure; Hardness

1. Introduction

The possibility of improving the penetration resistance of a target by layering it with materials having different properties was known in the late 1800 s. At that time, the ballistic performance of armour plate was first improved by hardening its surface. It has been shown that a hard surface layer to resist impact indentation, backed-up by a tough and ductile inner layer to absorb the kinetic energy of the projectile is an efficient combination to resist the projectile impact [1−3]. The hardfacing alloy selection is guided primarily by the wear and cost considerations. However, other manufacturing and environmental factors must also be considered, such as the parent metal, deposition process, and impact, corrosion, oxidation, and thermal requirements.

Among the hardfacing alloys, the high chromium-containing hardfacing alloys have been used most extensively because of their excellent hardness, corrosion resistance, and wear resistance as well as inexpensiveness [4]. These properties are obtained from the large volume fraction of hard chromium carbides [5]. The work on these alloys has
focused on the property enhancement, the microstructural modification, and the high temperature application [6]. Ceramics are very strong materials, especially in compression. They are well suited for armour applications when subjected to high pressures during impact and penetration. It is generally agreed that ceramic materials exhibit strength after they are failed and that this strength is pressure dependent [7]. Historically, tungsten-base carbides were used exclusively for the hardfacing applications.

Tungsten carbide (WC) has a number of valuable properties, which make them the most promising material for use in various new fields of technology [8]. Tungsten carbide is a high-density ceramic with mechanical properties that make it attractive for applications related to high-velocity impacts. To utilize WC’s high hardness and improve its toughness, it is coupled with a metallic binder. The majority of them utilize cobalt (Co) as the binder, but nickel (Ni) and chromium (Cr) are also used. From the above discussions, it is clear that both W based and Fe based carbide consumables are used for hardfacing application. Experiments related to projectiles impact resistance on Q&T steel weld metal at high velocities have produced beneficial results in recent years. The majority of these researches focused on sandwich joints of shielded metal arc (SMA) hardfaced interlayers [9–14] due to low cost, high thickness hardfacing and high dilution, to some extent, it is expected to be beneficial for the ballistic impact of SMA hardfacing process [15]. From the available literatures, it is understood that there is no published information on the comparative evaluation of ballistic performance of hardfacing consumables. Hence, in this investigation, one consumable from tungsten — based hardfacing alloy group (WC) and another consumable from iron — base hardfacing alloy group (CrC) were selected for depositing hardfaced middle layer to study the effect of hardfacing consumables on ballistic performance enhancement.

2. Materials and methods

The parent metal (PM) used in this study is 18 mm thick high strength low alloy Q&T steel closely confirming to AISI 4340 specification. The heat treatment adopted for the steel was to austenitise at a temperature of 900 °C followed by oil quenching. Subsequently, the steel was subjected to tempering at 250 °C. The chemical compositions of PM and filler metals used in this investigation are presented in Table 1. SMAW process was selected as it is generally employed in welding of combat vehicle construction. Austenitic stainless steel (SS) electrode was selected because it inhibits the delayed cracking tendency of the Q&T steel weldments. In this study, two different hardfacing consumables, namely tungsten carbide (WC) and chromium carbide (CrC) consumables, were used to deposit 4 mm thick hardfaced interlayer. The root and capping front layers were deposited using SS filler. The welding parameters used to fabricate the joints are presented in Table 2. Necessary care was taken to avoid joint distortion and to obtain defect free welds. The preheating and interpass temperatures were maintained at 150 °C during the welding of all the different layers to avoid both cold and hot cracking tendency.

Table 2 Welding parameters used for fabricating the joints.

<table>
<thead>
<tr>
<th>Parameters</th>
<th>Unit</th>
<th>SS buttering</th>
<th>Hardfacing (SMAW)</th>
<th>SS WC</th>
<th>CrC</th>
</tr>
</thead>
<tbody>
<tr>
<td>Filler wire diameter</td>
<td>mm</td>
<td>3.15</td>
<td>4</td>
<td>4</td>
<td>4</td>
</tr>
<tr>
<td>Preheat temperature (°C)</td>
<td></td>
<td>150</td>
<td>150 150 150 150</td>
<td>150 150 150 150</td>
<td></td>
</tr>
<tr>
<td>Interpass temperature (°C)</td>
<td></td>
<td>150</td>
<td>150 150 150 150</td>
<td>150 150 150 150</td>
<td></td>
</tr>
<tr>
<td>Welding current (A)</td>
<td></td>
<td>110</td>
<td>153 153 153 153</td>
<td>153 153 153 153</td>
<td></td>
</tr>
<tr>
<td>Arc voltage (V)</td>
<td></td>
<td>21</td>
<td>22.3 22.3 22.3</td>
<td>22.3 22.3 22.3</td>
<td></td>
</tr>
<tr>
<td>Welding speed (mm/min)</td>
<td></td>
<td>217</td>
<td>207 207 207 207</td>
<td>215 215 215 215</td>
<td></td>
</tr>
<tr>
<td>Heat input (kJ/mm)</td>
<td></td>
<td>0.6391</td>
<td>1.05 0.97 0.97</td>
<td>0.95 0.95 0.95</td>
<td></td>
</tr>
</tbody>
</table>

Fig. 1 shows the unequal double Vee butt joint configurations and schematic illustration of welding sequence for fabrication of the joints. Two joints were prepared. The joint with buttering layer on the beveled edge, unequal Vee joint design, SS root, WC hardfaced interlayer and SS capping font layer was labelled as WAHA (Fig. 1(a)). Similarly, the other joints with SS buttering layer, unequal Vee joint design, SS root, SMA – CrC hardfaced interlayer and SS capping font layer was labelled as CAHA (Fig. 1(b)). Here, the presence of CrC hardfaced interlayer is the only difference. In both the joints, on the unequal double Vee joint configuration, SMA — hardfacing of 4 mm thickness was sandwiched in between the root and capping weld. Table 2 shows the parameters used to fabricate the joints. The sequential welding procedure was applied to avoid distortion, as illustrated in Fig. 1(a)–(b). The fabricated joints were evaluated for their ballistic performance and the results were compared in terms of depth of penetration on weld metal and HAZ locations. Fig. 1(d) represents the weld coupon design and target plate dimensions for the fabrication of target.

Table 1 Chemical composition of parent metal and filler metals used to fabricate the joints.

<table>
<thead>
<tr>
<th>Element</th>
<th>C</th>
<th>Mn</th>
<th>Si</th>
<th>S</th>
<th>P</th>
<th>Cr</th>
<th>Ni</th>
<th>Mo</th>
<th>V</th>
<th>W</th>
<th>Fe</th>
</tr>
</thead>
<tbody>
<tr>
<td>Parent metal (AISI 4340)</td>
<td>0.35</td>
<td>0.54</td>
<td>—</td>
<td>—</td>
<td>—</td>
<td>1.25</td>
<td>1.75</td>
<td>0.52</td>
<td>—</td>
<td>—</td>
<td>Bal</td>
</tr>
<tr>
<td>SS electrode</td>
<td>0.08</td>
<td>3.3</td>
<td>0.90</td>
<td>0.015</td>
<td>0.04</td>
<td>20.30</td>
<td>8.50</td>
<td>1.5</td>
<td>—</td>
<td>—</td>
<td>Bal</td>
</tr>
<tr>
<td>AWS E 307-16 Hardfacing alloy (CrC)</td>
<td>4.0</td>
<td>1.0</td>
<td>1.50</td>
<td>—</td>
<td>—</td>
<td>30.00</td>
<td>—</td>
<td>2.0</td>
<td>0.50</td>
<td>—</td>
<td>Bal</td>
</tr>
<tr>
<td>AWS E FeCr-A7 Hardfacing alloy (WC)</td>
<td>2.4</td>
<td>0.18</td>
<td>0.18</td>
<td>0.04</td>
<td>—</td>
<td>0.30</td>
<td>0.36</td>
<td>—</td>
<td>56.4</td>
<td>Bal</td>
<td></td>
</tr>
<tr>
<td>AWS E WC1 30/40 Hardfacing alloy (CrC)</td>
<td>—</td>
<td>—</td>
<td>—</td>
<td>—</td>
<td>—</td>
<td>—</td>
<td>—</td>
<td>—</td>
<td>—</td>
<td>—</td>
<td>—</td>
</tr>
</tbody>
</table>
Both the fabricated joints were tested as per the military standard (JIS.0108.01) in a ballistic testing tunnel located at Defence Metallurgical Research Laboratory (DMRL), Hyderabad in standardized testing conditions. The fabricated joints were tested with 7.62 mm armour piercing incendiaries (API) projectiles. The ballistic testing procedures were dealt with elsewhere [9–14]. Few numbers of the preliminary experiments were performed, and adjustments were made to obtain the required impact velocity of the projectile onto the joint. The velocity of projectile was measured to be $820 \pm 10$ m/sec. The schematic illustration of experimental setup used for ballistic testing is displayed in Fig. 2. For each fabricated joint, at least one shot was fired at weld metal.

3. Experimental results

3.1. Ballistic test results

Fig. 3 represents the photographs of ballistically tested WAHA joint. Fig. 3(a) is the front view and Fig. 3(b) is the

![Fig. 2. Schematic diagram of experimental setup used for ballistic testing.](image)

![Fig. 3. Photographs of WAHA joint after ballistic testing.](image)
rear view of the joint after ballistic test. All the projectiles fired at various locations of the WAHA joint were successfully stopped.

Fig. 4(a) is the front view and Fig. 4(b) is the rear view of CAHA joint after the ballistic testing. From the ballistic test results, it is found that both the combination (WAHA and CAHA) successfully stopped the projectiles and the summary of ballistic test results is presented in Table 3.

3.2. Macrostructure and microstructure

The microstructures of various locations of the multi layered WAHA joint are shown in Fig. 5. The undiluted SS capping layer microstructure (Fig. 5(a)) clearly depicts the presence of grain boundary $\delta$ ferrite in a dendritic austenitic matrix. The weld interface regions of WC hardfacing and SS capping (Fig. 5(b)) reveals that the epitaxial growth of $\delta$ ferrite in austenitic matrix from the hard phase eutectic WC in the hardfaced interlayer.

In this micrograph, the crossing over of delta ferrite grain from the SS capping to hardfaced interlayer is also visible. The undiluted WC hardfaced region consisting of large amount of very fine needles of WC in austenitic matrix is witnessed in Fig. 5(c). Fig. 5(d) shows the macrographs of WAHA joint before ballistic testing. From this macrograph, it is clear that WAHA joint has good joint integrity between layers. In addition, the absence of any macro level defects like crack at the interfaces of hardfaced layer, slag inclusions etc., was the indication of good fashioning of the joint. The weld IF region of WC hardfacing and SS root (Fig. 5(e)) has similar structure in the interface between SS capping and hardfacing (Fig. 5(b)). Both the interfaces (Fig. 5(b) and (e)) reveal the presence of very narrow unmixed region in the interface. This white unmixed zone is obviously perpendicular to the normal grain growth direction in both sides of the interface. SS root (Fig. 5(f)) comprised of the vermicular $\delta$ ferrite in the austenitic matrix. The HAZ microstructure (Fig. 5(g)) clearly depicts the presence of bainite and martensitic features.

The microstructures of various locations of the multi layered CAHA joint are shown in Fig. 6. The undiluted SS capping layer microstructure (Fig. 6(a)) clearly depicts the presence of grain boundary $\delta$ ferrite in a dendritic austenitic matrix. The weld interface region of CrC hardfaced layer and SS capping (Fig. 6(b)) reveals the fine anchoring between these layers. The microstructure of the undiluted CrC hardfaced interlayer (Fig. 6(c)) consisting of very complex mixture of hexagonal $M_7C_3$ carbide and metastable austenite contains a high chromium concentration [5]. Fig. 6(d) shows the macrographs of CAHA joint before ballistic testing.

From this macrograph, it is clear that CAHA joint also has good joint integrity. In addition, the absence of any macro level defects, such as crack at the interfaces of the hardfaced layer, slag inclusions etc., was the indication of good fashioning of the joint. While in the interfaces, between SS capping front layer/CrC hardfaced interlayer (Fig. 6(b)) and CrC hardfaced interlayer/SS root layer (Fig. 6(e)), the unmixed region is totally absent. Good anchoring of SS cap to hardfacing and hardfacing to SS root showed a clear bonding of the layers in the microstructure. SS root (Fig. 6(f)) is comprised of the vermicular $\delta$ ferrite in austenitic matrix. The HAZ microstructure (Fig. 6(g)) clearly depicts the presence of bainite and martensitic features.

3.3. Hardness

The hardness measurement was carried out in two different directions to evaluate the hardness disparity along and across the weld cross section. More than fifteen readings

<table>
<thead>
<tr>
<th>Joint type</th>
<th>Shot No (as labelled in photographs)</th>
<th>Velocity</th>
<th>Location</th>
<th>Result category</th>
<th>Depth of penetration (DOP) in mm</th>
<th>Illustration</th>
</tr>
</thead>
<tbody>
<tr>
<td>WAHA</td>
<td>1 (Fig. 4(a))</td>
<td>825</td>
<td>Weld</td>
<td>$S_2^{ab}$</td>
<td>16</td>
<td>Stopped</td>
</tr>
<tr>
<td></td>
<td>2 (Fig. 4(a))</td>
<td>812</td>
<td>FL</td>
<td>$S_2^{b}$</td>
<td>17</td>
<td>Stopped</td>
</tr>
<tr>
<td></td>
<td>3 (Fig. 4(a))</td>
<td>829</td>
<td>HAZ</td>
<td>$S_1^{a}$</td>
<td>16</td>
<td>Stopped</td>
</tr>
<tr>
<td></td>
<td>4 (Fig. 4(a))</td>
<td>823</td>
<td>Weld</td>
<td>$S_1^{a}$</td>
<td>15</td>
<td>Stopped</td>
</tr>
<tr>
<td></td>
<td>5 (Fig. 4(a))</td>
<td>822</td>
<td>BM</td>
<td>$S_1^{a}$</td>
<td>14.0</td>
<td>Stopped</td>
</tr>
<tr>
<td>CAHA</td>
<td>1 (Fig. 5(a))</td>
<td>821.90</td>
<td>Weld</td>
<td>$S_1^{a}$</td>
<td>14</td>
<td>Stopped</td>
</tr>
<tr>
<td></td>
<td>2 (Fig. 5(a))</td>
<td>828.09</td>
<td>Weld</td>
<td>$S_2^{b}$</td>
<td>15</td>
<td>Stopped</td>
</tr>
<tr>
<td></td>
<td>3 (Fig. 5(a))</td>
<td>823.86</td>
<td>HAZ</td>
<td>$S_2^{b}$</td>
<td>15</td>
<td>Stopped</td>
</tr>
<tr>
<td></td>
<td>4 (Fig. 5(a))</td>
<td>839.46</td>
<td>BM</td>
<td>$S_1^{a}$</td>
<td>14</td>
<td>Stopped</td>
</tr>
</tbody>
</table>

$^a$ Stopped without any bulge at the rear side of the target.

$^b$ Stopped with bulge at the rear side of the target.
were taken at close proximity and the values are presented in Figs. 7 and 8.

Fig. 7(a) represents the hardness profile along the weld center line and depicts the existence of the high hardness hardfaced layer between two low hardness soft zones (i.e. SS root and SS capping front layer). By comparing the hardness graphs shown in Fig. 7(a)–(b), it is observed that there is a considerable difference in hardness of the hardfaced layer. The highest hardness in WC layer was recorded to be 722 HV050. This is 11% higher hardness as compared to CrC hardfaced interlayer, since the peak hardness of CrC hardfaced layer was measured to be 650 HV050. In addition, Fig. 8 represents the hardness profile across the weld center line for both (WAHA and CAHA) joints. From the hardness profile across the weld center line on SS capping and SS root on both the joints are more or less similar. But, the hardness level of SS root is little lower (225 HV050) than SS capping front layered (235 HV050) joint (Fig. 7(a) and (b)). From Fig. 8(C)–(d), it is clear that a low hardness zone is predominant between the parent metal and the weld metal in all four hardness profile and it is mainly due to the heat affected zone softening reported elsewhere [15–20].

The width of the soft zone (<400 HV) is measured to be 2–4 mm in hardness profile across both the layers. Softening can be due to the heating of the base plate in the inter-critical and subcritical regions, resulting in microstructures other than a fully martensitic microstructure. This indicates that the soft zone does not influence the ballistic property, since the projectiles fired in this investigation are 7.62 mm armour piercing incendiary projectiles. The hardness profile across WC hard-facing (WAHA) showed a scattered hardness value (735, 713, 783, 762, 714, 767, 690, 655, 647 and 656 HV050 in the WC hardfaced region. While in CAHA joint, the hardness values are not much scattered (653, 651, 660, 645, 653, 634, 645, 634 and 647 HV050) in the CrC hardfaced region.

4. Discussion

From the previous paragraphs, it is well understood that, with the presence of buttering layer, WC/CrC hardfaced middle layer between soft SS root and capping front layers has enhanced the ballistic resistance of armour grade Q&T steel welds. The impact resistance is primarily taken care by the presence of very hard hardfaced interlayer. But, the impact resistance property of hardfacing is achieved from the matrix toughness and the hardness of the reinforcing particles which
directly depends on the volume fraction, size and distribution of the hard phases [21]. The cemented carbides increase the bulk hardness of the composite materials, which prevent most of the projectiles from penetrating into the substrate [22]. Even if a projectile managed to penetrate into material, it will collide with the hard-phase particles leads to the termination of the further penetration of bullet. During these processes, the projectile gets worn off or smashed, and finally, loses its function of penetration. As a result, the depth of projectile penetration is greatly limited. From these considerations, an introduction of WC carbides and CrC is expected to be beneficial to improve the impact resistance of hardfaced deposits.

Welded hardfacing deposits are, in effect, the mini-castings characterized by variable composition (segregation) and solidification kinetics that influence the deposit microstructure. It is not surprising, therefore, that the properties and quality of welded hardfacing deposits should depend on welding process and technique, as well as on the alloy selection [4]. The projectile penetration resistance of various armour materials is dependent on the best combination of hardness and toughness. Both are required to avoid the cracking tendency, to avail the better ballistic resistance and to battle the consequent disintegration of the material [23,24].
ballistic tests have shown that the joint made with CrC hard-faced interlayer offers better ballistic performance than the joint made with WC hardfaced interlayer (WAHA) with respect to the front layer damage and DOP. The primary reason for the superior ballistic resistance of the CrC hardfaced interlayered joint is discussed in the following sections.

4.1. Role of hardfaced interlayer microstructure on ballistic performance

The microstructures and ballistic properties of hardfaced deposits vary depending on solidification kinetics and dilution. Solidification kinetics tends to be somewhat slower in

![Fig. 8. Hardness profile across weld center line of WAHA and CAHA joints.](image)

(a) WC hardfaced region
(b) WC hardfaced region (High magnification)
(c) CrC hardfaced region
(d) CrC hardfaced region (High magnification)

![Fig. 9. Undiluted hardface microstructures of WC and CrC layer alone.](image)
conventional weld hardfacing processes. This slow solidification rate produces widely different microstructures and widely different properties regardless of dilution [4]. The undiluted WC hardfaced region consisting of large amount of very fine needles of WC in austenitic matrix is witnessed in Fig. 9(a). In the high magnification microstructure (Fig. 9(b)), it was observed that a significant contact had occurred between the carbides and a number of micro-voids were observed in the center of the carbides. This could be attributed to the beginning of solidification; a great positive temperature gradient exists in the liquid side of solid/liquid interface, so the crystals grow in a planar manner on solidification to form the interface. After solidification, the matrix is composed of thick primary crystal and polytropic eutectic. The blocks are probably the undissolved or partly dissolved WC particles that have relatively high melting point. In the rapid solidification process, these WC particles, either undissolved or partly dissolved, play a role as nucleating sites and prevent the grains from growing, which results in lamellar plate like structure in the WC deposited hardfaced layer.

In CrC hardfaced interlayer, the microstructure (Fig. 9(c)) shows the existence of complex morphology. It primarily exhibits two microconstituents: namely, primary carbides of needle (A) and polyhedral shape (B) and the eutectic matrix containing mixture of γ + M7C3 carbides (C) [25]. The needle and cuboid shape primary carbide particles (bright area) are present in the matrix of eutectic (dark zone). The eutectic is composed of F.C.C. austenite and eutectic M7C3 carbide in these layers. It was reported that typical composition of M7C3 depends on the local composition of alloy and cooling rate [26,27].

Due to this complex structure, the microstructure can able to absorb more impact energy and the presence of γ in the matrix enhances the energy absorption capacity of the hardfaced interlayer. At higher magnification (Fig. 9(d)), the presence of eutectic mixture is clearly visible in the matrix with the very fine hexagonal precipitates (B) of M7C3. The interface microstructure of CrC hardfacing and SS capping (Fig. 6(e))/SS root layer (Fig. 6(f)) suggested that the solidification in the alloys usually begins with the formation of primary M7C3 carbides, the residual liquid eventually decomposed into a mixture of austenite and more M7C3 carbides by a ternary eutectic reaction [28]. But, the undiluted CrC hardfaced interlayer consists of microstructure of large primary M7C3 carbides, which are all found to be surrounded by precipitate-free zones (Fig. 9(c)–(d)). These zones are believed to arise due to the depletion of chromium in the liquid close to the primary carbides [29]. Due to this microstructure, the energy absorption capacity of the interface and undiluted hardfaced layer is increased and finally the projectiles were stopped with a lesser DOP (Table 3).

On the other hand, the undiluted WC hardfaced interlayer and the interface of WC hardfaced layer with SS layer (Fig. 6(c) and (e)) are highly diluted by mixing with the SS layer, so that its microstructure is found to be largely of primary austenite dendrites. This is the reason for more austenitic phase and thus resulting in poor ballistic performance by allowing the projectile to a higher depth of penetration (Table 3). Some of the austenite dendrites close to the SS layer (where dilution is expected to be at a maximum) were found to decompose into δ ferrite, which is a good sign of bonding between layers. Since the adhesive damage is started within the dendrite phase of the hardfacing alloy, the impact resistance of WC hardfacing alloy is strongly believed to be dependent on the δ-ferrite content of the dendrites [30]. The fracture-toughness values of the hardfacing alloys are defined by a crack-bridging model, which involves the plastic stretching and necking of dendrites in the wake of a spreading crack tip. The δ-ferrite content of the dendrites has a strong effect on the amount of plastic stretching and the degree of crack-bridging toughening, which controls the fracture toughness of hardfacing alloy. The lower δ-ferrite content outcomes with greater amount of crack-bridging toughening and higher fracture toughness values. The WC hardfaced layer contains the highest fraction of δ ferrite in the dendrites, which reduces the level of dendrite stretching and amount of crack-bridging toughening to result in its fracture toughness value being much lower than that of CrC hardfaced interlayered joints.

From the microstructural observation, it is clear that Cr in presence of Ni is more effective in producing finer microstructure. Nickel does not produce any second phase particle. Ni is found mostly in the form of solid solution in ferrite [31—33]. So Ni increased the strength of steel by solid solution strengthening. Besides that Ni also lowers the transformation temperature [31—33], so the lower transformation temperature produced smaller ferrite grains. Chromium in the form of chromium carbide precipitates increased the strength by means of precipitation strengthening. Secondary chromium carbides pin the grain boundaries and inhibit the grain growth, resulting in grain refinement. The presence of second phase particles also makes the dislocation movement more difficult. Second phase particles like chromium carbide in the matrix increases the energy required for elastic/plastic deformation, hence creates higher strength in the alloy [34]. Nickel in solution and chromium as chromium carbide precipitates increases the yield strength of the hardfaced layer, but the effectiveness of chromium carbide precipitates in the increment of yield strength is found to be more than that of nickel. In the presence of nickel, the contribution of chromium carbide to increase yield strength is more.

Besides the above discussion, the combined effect of interface microstructure, SS capping front layer and SS root bottom layer microstructure play a key role towards the better ballistic performance of this CAHA joint. The observed results show that the fully austenitic weld metal using SS electrode having a microstructure of δ-ferrite in a plain austenitic matrix is the beneficial microstructure for moderate strength welds with good crack resistance. This layer would allow the projectile with reduction of initial velocity towards hardfaced interlayer. Then the hardfaced layer had completely stopped the projectile penetrating into it due to its higher hardness and tough SS root layer.

### 4.2. Role of hardfaced interlayer hardness on superior ballistic performance

From the microhardness survey carried out in the joint with WC hardfaced interlayer, the hardness profile across the weld...
center line (Fig. 8(c)) indicated the scattered results in the middle hardness values. This could be attributed to the presence of irregular of WC in an austenitic matrix and the recovery, recrystallization and grain growth which occurred slowly in this layer leading to the reduction the dislocation density with increased grain size and relieving the internal stresses within the grain. These transformations led to carbide dissolution and accounted for the reduction in hardness of the lower layer as observed in Fig. 8(c). So, the better ballistic performance of CAHA joint can be related to the combined effect of the soft SS front layer (232 HV050) and the higher hardness (650 HV050)) CrC hardfaced interlayer and lower hardness (215 HV050)) SS root layer. The reason for this hardness variation is primarily dominated by the above discussed microstructural constituents. The presence of this hardness level in this SS capping front layer, hardfaced interlayer, SS root layer and interface zone helps the joint against disintegration at the time of projectile attack.

5. Conclusions

From this investigation, the effect of hardfacing consumables was evaluated and the following important conclusions are derived:

1. Of the two joints, the joint made with CrC hardfaced interlayer (CAHA) offered better ballistic resistance at the weld metal. The CAHA joint successfully stopped the bullet with 13.5 mm DOP only. This is 14% lesser DOP as compared to WAHA joint; which offered a DOP of 15 mm.

2. The hardness of CAHA joint is higher due to the presence of the primary carbides of needle shape, polyhedral shape and eutectic matrix containing a mixture of \( \gamma + M_7C_3 \) carbides in the CrC hardfaced interlayer.

3. CrC hardfaced layer is highly toughened due to the existence of soft SS layer in all the four sides (two as buttering layer and another two sides as root and capping layer). The combined effect of joint design, presence of CrC hardfaced layer, SS buttering layer, SS root and capping layer is the primary reason for the successful projectile stop with lesser DOP of the CAHA joint.

4. The scattering hardness level in the WC interlayer, the matrix decomposition resulting in lowering the hardness and the co-existence of \( \delta \) ferrite in the interface between hardfacing and SS root/SS cap could be attributed to the relatively lower ballistic resistance of this joint by allowing to a DOP of 15−16 mm.

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