Chapter- 4
Results and Discussions

This chapter comprises of the results and discussions of the various welding processes on aluminium alloy AA5083 in H321 condition. The analysis part is divided into four sections. Section- I deal with the structure-property determination of the weldment made on AA5083-H321 aluminium alloy made by Friction Stir welding technique. The results of chemical composition, metallographic examination of the FS weldment as well as the hardness, and tensile strength of this weldment are reported and discussed with respect to the existing literature. Section- II deals with the structure-property correlation of the weldment of AA5083-H321 aluminium alloy fabricated by Tungsten Inert Gas (TIG) welding process. Moreover, the results of macro and microstructural analysis, vaporization of volatile elements, as well as the hardness; and tensile strength of the TIG weldment have been reported and discussed in details. Section- III deals with the metallurgical and mechanical behavior of AA5083-H321 weldment fabricated by Laser Beam welding technique. Section- IV details on the structure-property relationship on Cast Al-Mg-Sc alloy weldment made by Friction Stir welding technique. Section-V deals with the comparative evaluation of TIG, LB and FS welding of Wrought AA5083-H321 aluminium alloy and FS welding of Wrought AA5083-H321 and Cast Al-Mg-Sc alloys.

4.1. FRICTION STIR WELDING OF AA5083-H321 ALUMINIUM ALLOY

FS welding is performed on AA5083-H321 aluminium alloy plate of 5 mm thickness and the details of the weld parameters are provided in section.3.1.2.

4.1.1. BEAD APPEARANCE OF FRICTION-STIR WELDED AA5083-H321

The photograph of FS weldment of AA5083-H321 aluminium alloy is shown in Fig.4.1. It is very clear that the formation of flash is higher on the retreating side and this is attributed to higher axial load. When the FS welding tool is pressed on the plates to be welded, the formation of flash occurs. The welding tool rotation direction and weld plate traverse direction coincides on the advancing side and this made the flash to be formed on the advancing side.
4.1.2. COMPOSITIONAL ANALYSIS OF FRICTION STIR WELDED AA5083-H321

The compositional analysis has been carried out on this weldment on base metal as well as weld nugget zone and the same is presented in Table 4.1. In addition, the concentration of Mg alone is focused in this research since it mainly influences the metallurgical and mechanical properties of weldment as reported by Cao et al (2003 a & b). The difference in Mg contents in the base metal as well as in FS welding is given in Table 4.1. It is clearly evident from the analysis, that there is an increase (9.7 %) in Mg content in the weld zone, when compared to the base metal. From the chemical analysis of the FS welded joint, the 9.7 % increase in Mg in the FS weld is attributed to the presence of Al₃Mg₂ secondary precipitates. According to Al-Mg equilibrium phase diagram, 6 % Mg present in Al - 3.5 % Mg alloys in liquid at elevated temperature act as the source for the formation of intermetallics. This results corroborates well with the reported on Friction stir spot welding of 5083[Senkara and Zhang (2000)]

Table 4.1. Compositional analysis of Mg in the FS welded joint of AA5083-H321 alloy

<table>
<thead>
<tr>
<th></th>
<th>Mg %</th>
<th>Gain of Mg, %</th>
</tr>
</thead>
<tbody>
<tr>
<td>Base Metal</td>
<td>4.214</td>
<td>-----</td>
</tr>
<tr>
<td>FS Welding</td>
<td>4.622</td>
<td>+ 9.7</td>
</tr>
</tbody>
</table>
4.1.3. OPTICAL MACROSTRUCTURE OF FRICTION STIR WELDED AA5083-H321

The macrostructure of the cross-section of the FS weld is shown by Fig.4.2. It is observed that the defect free FS weld nugget is relatively finer than the TMAZ and HAZ. The demarcation line between the weld nugget and TMAZ on the advancing side is shown clearly in the Fig.4.2.

There are several characteristic areas in the weld zone viz., distinct layer of mixed materials in the central part of the weld called weld nugget, TMAZ, HAZ and unaffected material. The stirred zone, i.e., the weld nugget or dynamically recrystallized zone, approximately coincides with the shape of the tool and contains fine equiaxed grains. On both sides of the stir zone are thermo-mechanically affected zone, which contains highly deformed grains from the stirring action. The macrograph (Fig.4.2) shows the lack of symmetry along the centerline of the weld. The advancing side (the edge of the weld where direction of tool rotation is the same as the travel) of the weld nugget is typically sharp and readily discernable; whereas the retreating side (the direction of tool rotation is opposite to the travel) of the weld nugget is much more diffuse. This is due to the torsion (due to rotating motion of the tool) and the circumventing (due to the translation motion of the tool) velocity fields having opposite directions on the advancing side, whereas these velocities have the same direction on the retreating side. The fine recrystallized zone at the weld nugget is due to heavy plastic deformation followed by dynamic recrystallisation due to thermo-mechanical processing.

![Fig.4.2. Optical Macrostructure of FS Weld of AA5083-H321 showing the presence of various regions of the weld joint](image-url)
4.1.4. OPTICAL MICROSTRUCTURE OF FRICTION STIR WELDED AA5083-H321

The cross-sectional image of AA5083-H321 weldment is presented in Fig.4.3 to 4.6. It shows three distinct regions like stirred zone, thermo-mechanically affected zone and heat affected zone. The stirred region has considerable grain refinement with the precipitates equally distributed owing to re-precipitation which is in accordance with the previous studies [Moataz M. Attallah (2007)].

The microstructures at the junction of TMAZ and Weld nugget at the top and middle of the FS weld are shown in Fig.4.3 and 4.4. The weld nugget at the top of the FS weld is shown by Fig.4.5, whereas the weld nugget at bottom is shown in Fig.4.6.

The microstructure of FS weld nugget at the top side of the weld is finer than the microstructure of weld nugget at bottom. During FS welding, most of the heat is produced by friction between the shoulder and the surface of the sheet. This generates an asymmetry between the top and the bottom of the plate.

It is generally agreed that the refined grain size of the nugget zone is a result of recrystallisation. Hence, the retention of most of the mechanical properties has been observed in this experimental study. As per Hall-Petch rule, the grain refinement increases both strength and hardness in the case of FS welded joint.

Fig.4.3. Microstructure of the top of the weld nugget taken at advancing side showing the clear demarcation between TMAZ and weld nugget
Fig. 4.4. Microstructure at the middle of the weld nugget

Fig. 4.5. Microstructure of weld nugget at top showing fine grains

Fig. 4.6. Microstructure of weld nugget at bottom showing the presence of coarse grains
4.1.5. MICROSTRUCTURE OF FRICTION STIR WELDED AA5083-H321

The SEM image of the weld zone of the FS welded joint is shown in Fig.4.7. It can be identified that the weld nugget of the FS welded joint is having very fine microstructure and the density of the various precipitates found in the weld nugget is also quite high compared to the SEM image of the base metal AA5083-H321 as shown in Fig.3.2(b). The stirring and heating during the welding are possible reasons for the decrease in size of the large particles, the coarsening of the small particles, and the increase in the volume fraction.

However, the present study did not detect any particles except $\text{Al}_6(\text{Fe},\text{Mn})$ and $\text{Al}_3\text{Mg}_2$. The solvus temperature of $\text{Al}_6(\text{Fe},\text{Mn})$ is calculated to be about 908K (635°C) using the thermo chemical database for light metal alloys. The solvus temperature of $\text{Al}_6(\text{Fe},\text{Mn})$ is higher than the solidus temperature of Al alloy 5083 (about 848K (575°C)). Considering that FS welding is solid-state bonding, the $\text{Al}_6(\text{Fe},\text{Mn})$ is stirred with the softened matrix during the welding, which probably leads to fragments or aggregations of the $\text{Al}_6(\text{Fe},\text{Mn})$ particles [Sato et al. (2001)]. FS welding heats up the stir zone to 773K (500°C), which may result in growth of the pre-existing $\text{Al}_6(\text{Fe},\text{Mn})$ precipitates or precipitation of new $\text{Al}_6(\text{Fe},\text{Mn})$ particles during the thermal cycle. Dissolution of $\text{Al}_3\text{Mg}_2$ precipitate in weld nugget was absorbed.

4.1.6. HARDNESS OF FRICTION STIR WELDED AA5083-H321

The Vickers hardness distribution on the cross-section perpendicular to the tool traverse direction of the FS welded specimen is shown in Fig.4.8. The Stir zone and the
TMAZ exhibited a higher hardness than the remaining zones. In addition, there was a significant difference in the hardness distribution between the RS and the AS. That is, the hardness within the SZ was higher on the AS than on the RS. The maximum hardness region was also located on the AS. These results suggest that the hardness distribution within the SZ was not symmetric with respect to the tool rotation axis. Moreover, in the transition zone between the unaffected zone and the SZ, the hardness variation was more notable on the AS than on the RS. This result corresponds well to the microstructural difference. It is noteworthy that, the average hardness of the SZ increased markedly, reaching a level about 28 % greater than that of the base metal. The maximum hardness value observed on the advancing side is around 100 Hv1. The average value of hardness outside the FS welded zone is similar to that of the base metal.

The higher hardness could be due to evolution of the refined grain with micro-precipitates. The higher hardness in the stirred zone could be due to random distribution of precipitates shown in SEM morphology (Fig 4.7). The variation in hardness values indicates that the FS weld nugget is relatively harder than the AA5083-H321 base metal. The variation in the size and shape of the precipitates and recrystallisation of fine grains may be the reason for such increase in hardness in the FS welded joints of AA5083-H321 aluminium alloy.

![Hardness distribution in AA5083-H321 alloy FS weld](image)

**Fig.4.8.** Distribution of hardness in AA5083-H321 alloy FS weld
4.1.7. TENSILE PROPERTIES OF FRICTION STIR WELDED AA5083-H321

The tensile properties of FS welded joints at Welding speeds of 115, 135 and 158 mm/min were carried out and the results of which are shown in Table 4.2. The best tensile properties were observed at 158 mm/min welding speed.

<table>
<thead>
<tr>
<th>Welding Speed, mm/min</th>
<th>Yield Stress, MPa</th>
<th>Tensile Strength, MPa</th>
<th>Elongation, %</th>
</tr>
</thead>
<tbody>
<tr>
<td>115</td>
<td>231</td>
<td>273</td>
<td>17.3</td>
</tr>
<tr>
<td>135</td>
<td>230</td>
<td>276</td>
<td>18.7</td>
</tr>
<tr>
<td>158</td>
<td>240</td>
<td>285</td>
<td>20.8</td>
</tr>
</tbody>
</table>

The tensile properties of the FS welding are evaluated based on the application of axial load 17 kN. In addition an axial load of 9 kN was also applied. The tensile test results are listed out in Table 4.3. As FS welding at 9 kN axial load produces superior tensile strength compared to 17 kN axial load, and further characterization studies are carried out at 9 kN FS welded samples. The tensile properties of the FS welded AA5083-H321 alloy are better than the results published in literature. The yield stress value of the FS weld is 245 MPa, which is 93% of that of the base metal, and the best value for this particular alloy (previous best was 88% by Czechowski (2005)).

<table>
<thead>
<tr>
<th>Axial Load, kN</th>
<th>Yield Stress, MPa</th>
<th>Tensile Strength, MPa</th>
<th>Elongation, %</th>
</tr>
</thead>
<tbody>
<tr>
<td>9</td>
<td>245</td>
<td>278</td>
<td>18.2</td>
</tr>
<tr>
<td>17</td>
<td>245</td>
<td>266</td>
<td>22.7</td>
</tr>
</tbody>
</table>

Fig.4.9.Failure location of tensile specimens of FS welded joints of AA5083-H321 alloy at 9 kN
4.1.8. FRACTOGRAPHIC STUDIES OF FRICTION STIR WELDED AA5083-H321

From the fractographic studies, it was evident that all the three weld samples fractured at weld nugget. The fracture surface of FS weld at 9 kN axial load is shown in Fig.4.10 and the failure is ductile.

The fracture at the weld surface makes an angle of 45° with the base plate. This implies that the weld failed in ductile manner. The fractography of FS weld consists of dimples, which indicates that the tensile specimens failed in a ductile manner under the action of tensile loading. Since fine dimples are characteristic feature of ductile failure, FS welded joints have shown higher ductility (18.2 %).

![Fracture surface of FS welded joint of AA5083-H321 alloy exhibits dimples indicating the ductile failure mode of this weld joint](image)

4.1.9. DISCUSSION ON FRICTION STIR WELDING OF AA5083-H321 ALLOY

The relative merits and demerits of FS welding techniques with respect to the AA5083-H321 alloy are discussed here.

**Effect of microstructure on Hardness:** There are three valid reasons for such behavior. They are (1) refinement of weld microstructure due to dissolution of low density particles, such as, Al₃Mg₂ and Mg₂Si, (2) increase in density of higher hardness intermetallic compounds in the FS weld nugget compared to the base metal due to fragmentation of Al₆(Fe,Mn) particles and nucleation of new Al₆(Fe,Mn) particles in the FS weld nugget, and (3) retention of wrought alloy structure in the FS weld nugget.

In fact, Sato et al (2001) suggests the hardness profile is mainly affected by the distribution of the precipitates in the weld. High hardness spots have been previously
observed in 1xxx, and 5xxx-series welds [Salem (2003)]. Shigematsu et al (2003) have reported that at the 5083-5083 metal joining process, the hardness at the weld nugget was higher than that of the base metal. They have ascertained that due to grain refinement the weld hardness was more than that of the base metal. Cantin et al (2005) have reported that in AA5083-O alloy, the finer grain size in the nugget zone may be partly responsible for increasing the hardness in the weld nugget region. Yong-Jai KWON et al (2009), YAN Yong et al (2010) and Kumbher et al (2011) have found out that the hardness at the weld nugget of FS welded 5052 alloy was more than the AA5052 alloy base metal hardness. Jamshidi Aval et al (2012) have reported similar increase in hardness in the FS weld for AA5086 alloy. Damjan Klobcar et al (2012) have reported similar increase in the FS weld of AA5083-O alloy. Rajesh Kumar Gupta et al. (2011) have investigated the hardness distribution of AA7475 aluminium alloy and concluded that the hardness at stir zone is more than that of the base metal.

The contradicting behavior of increased hardness values and decreased tensile properties of FS welded joint compared to the AA5083-H321 base metal was similar to earlier reports of FS welding in AA5083-O [Shigematsu et al (2003)] and in aluminium-lithium alloy 01420 [Shitong Wei et al. (2007)]. They have concluded that the hardness values within the SZ were higher than those of the BM and the maximum tensile strength of the joints is equivalent to 86 % that of the base metal.

**Tensile properties:** In FS welding, the work piece does not reach the melting point and the mechanical properties of the welded zone are much higher compared to the traditional techniques, in fact, the undesirable microstructure resulting from melting and resolidification, characterized by low mechanical properties, is absent in FS welds leading to improved mechanical properties such as ductility and strength. The FS weld nugget of AA5083-H321 contained recrystallised grains. In addition to the microstructural refinement in the case of FS welded joint, absence of porosity and shrinkage void losses also enhances the tensile properties. The microstructure of the FS welded joint weld nugget compared to the microstructure of the base metal was refined. Hence, the retention of most of the mechanical properties has been observed in this experimental study. As per Hall-Petch rule, the grain refinement increases both strength and hardness in the case of FS welded joint.
The FS welding is done at 0.8 times the melting temperature of aluminium alloys; the strain-hardening effect is also retained in the weld. Hence, the tensile properties reduction compared to the base metal is minimal. If suitable welding parameters were selected in future, the FS weld strength can reach that of the base metal.

The presence of very fine Al₆(Fe,Mn) precipitates promotes/stimulates nucleation during the recrystallisation process resulting in an ultra-fine grain microstructure. Such microstructure enables plastic deformation by grain-boundary sliding and provides super-plastic behavior to the material when deformed under low deformation rate/high temperature conditions. Since, the material residing in the nugget of FS welded region is normally subjected to very high levels of plastic deformation and tends to recrystallize dynamically, one would expect formation of a very fine-grain microstructure in this region.

**Strengthening mechanisms:** The AA5083-H321 alloy is strengthened by solid-solution, grain-size, precipitation and strain-hardening strengthening mechanisms.

The grains in the weld nugget were formed due to recrystallisation and their sizes will be less than that of the grains in the base metal. Hence, an increase in grain-size (Hall-Petch effect) strengthening is possible. The reduction in size and increase in volume density due to fragmentation and nucleation of new intermetallic particles will have a tendency to increase the precipitation strengthening (Orowan strengthening) mechanism of the FS weld nugget. Thus, it is possible that the local variations in Al₆(Fe,Mn) dispersoids could explain the reason for the inconsistent micro hardness distributions in the FS welds.

4.1.10. CONCLUSIONS

- FS weld nugget contained 9 % higher concentration of magnesium than the wrought base metal.
- The macro and microstructure of FS weld revealed that the weld nugget grains were formed due to dynamic recrystallisation resulting in finer structure.
- Microstructural analysis performed using SEM reveals that fragmentation of the intermetallic compound Al₆(Fe,Mn) and formation of new compounds. Dissolution of the β-phase Al₃Mg₂ precipitates is also observed.
• The hardness distribution was asymmetric and hardness was low on the retreating side. The hardness in the TMAZ of both sides and the weld nugget was higher than the base metal hardness of AA5083-H321 alloy plate. There is 28 % increase in hardness in the case of FS welding on the advancing side (100 Hv1) compared to the AA5083-H321 alloy (77Hv1) hardness value.

• Among the FS welded joints produced with axial loads of 9 kN and 17 kN, FS weld at 9 kN produced same yield stress and 5 % increase in tensile strength.

• Transverse tensile test results at 9 kN axial load showed a minimum decrease in strength compared to the base metal. The joint efficiency in terms of yield stress and tensile strength is found to be about 93 % (245MPa) and 95 % (278 MPa).

• Transverse tensile test results at 9 kN axial load showed a decrease in ductility. The joint efficiency in terms of elongation was at 68 %.

• An increase in grain boundary strengthening due to recrystallized grains formed in the weld nugget is observed. Particle strengthening due to increase in density and decrease in size of intermetallic compounds Al₆(Fe,Mn) are responsible for near base metal tensile strength and higher hardness than the base metal in the FS welded joint.
4.2. TUNGSTEN INERT GAS WELDING OF AA5083-H321 ALUMINIUM ALLOY

TIG welding is performed on AA5083-H321 aluminium alloy plate of 5 mm thickness. The welding is done autogenously with parameters as shown in section 3.1.3.

4.2.1. BEAD APPEARANCE OF TUNGSTEN INERT GAS WELDED AA5083-H321

The weld bead contained under fill defect as shown in Fig.4.11. As the present welding experiment was carried out autogenously (without using filler wires), the shape of weld bead obtained is having the concavity. The usage of filler wires is not employed in the present study, in order to compare this process with FS welding, which does not use filler wires.

4.2.2. COMPOSITIONAL ANALYSIS OF TUNGSTEN INERT GAS WELDED AA5083-H321

During TIG welding process, the low boiling point elements such as Mg, Zn, Mn, Cu and Fe were subjected to evaporation. Magnesium, manganese and zinc can be easily vaporized during TIG welding, due to their high vapor pressure and low boiling point. The Boiling and Melting temperatures of elements present in the AA5083 alloy are given in Table 4.4.
Table 4.4. Boiling and Melting Temperatures of elements in AA5083 alloy

<table>
<thead>
<tr>
<th>Elements</th>
<th>Al</th>
<th>Mg</th>
<th>Mn</th>
<th>Fe</th>
<th>Si</th>
<th>Cr</th>
<th>Cu</th>
<th>Zn</th>
<th>Ti</th>
</tr>
</thead>
<tbody>
<tr>
<td>Boiling Temp, °C</td>
<td>2467</td>
<td>1090</td>
<td>1962</td>
<td>3027</td>
<td>2355</td>
<td>2672</td>
<td>2590</td>
<td>907</td>
<td>3257</td>
</tr>
<tr>
<td>Melting Temp, °C</td>
<td>660</td>
<td>649</td>
<td>1244</td>
<td>1535</td>
<td>1407</td>
<td>1857</td>
<td>1083</td>
<td>419</td>
<td>1667</td>
</tr>
</tbody>
</table>

The results of compositional analysis of the TIG welded samples are tabulated in Table 4.5. The amount of losses due to evaporation are listed in Table 4.6.

Table 4.5. Compositional analysis of TIG weld

<table>
<thead>
<tr>
<th>Elements</th>
<th>Mg</th>
<th>Mn</th>
<th>Fe</th>
<th>Si</th>
<th>Cr</th>
<th>Cu</th>
<th>Zn</th>
<th>Ti</th>
<th>Al</th>
</tr>
</thead>
<tbody>
<tr>
<td>% Wt.</td>
<td>2.549</td>
<td>0.662</td>
<td>0.308</td>
<td>0.198</td>
<td>0.009</td>
<td>0.034</td>
<td>0.096</td>
<td>0.01</td>
<td>96.085</td>
</tr>
</tbody>
</table>

Table 4.6. Percentage loss of volatile elements in TIG weld

<table>
<thead>
<tr>
<th>Elements</th>
<th>Magnesium</th>
<th>Manganese</th>
<th>Iron</th>
<th>Zinc</th>
</tr>
</thead>
<tbody>
<tr>
<td>% Wt.</td>
<td>39.5</td>
<td>15.9</td>
<td>24.3</td>
<td>20</td>
</tr>
</tbody>
</table>

Therefore, during welding at temperatures of above 2000 °C, Mg, Zn and Mn would have already reached the boiling points shown by Table 4.3, while the other alloying elements, Si, Cu, and Cr have not yet even melted. The percentage evaporation of Mg, Mn, Fe, and Zn were around 39.5 %, 16 %, 24.3 % and 20 % respectively. Previous emission spectroscopic investigations indicated that during TIG welding of steels, iron and manganese have evaporated. Similar results of vaporization of Manganese up to 50 % in the weld compared to the base metal composition were reported by Khan and Debroy (1984) in LBW of AISI 202 stainless steel.

The 39.5 % magnesium evaporation has reduced the yield stress values in the welded joint, as the percentage of magnesium is directly proportionate to the yield stress of the material as shown by Fig.2.1. The yield stress of AA5083 annealed alloy is around 150 MPa. The yield stress of AA5052 annealed alloy, which contains around 2.5 % Mg, is around 90 MPa. Thus reduction of 1.7 % of Mg in the annealed state loses around 60 MPa of yield stress as shown in Fig. 2.1.

The reduction of Mn also affects the solid solution strengthening of the alloy. The evaporation of Fe combined with Mn is responsible for formation of minimum
intermetallic compounds in the welded joints. The reduction of Zn increases the tendency for stress corrosion cracking in the welded joints.

4.2.3. OPTICAL MACROSTRUCTURE OF TUNGSTEN INERT GAS WELDED AA5083-H321

The optical macrostructure of the TIG welded joint is shown in Fig.4.12. The columnar grains formed in the HAZ of TIG welded joint are clearly visible in the macrostructure. The columnar grains grow approximately normal to the welding direction when welded at higher speeds in the case of LB welded joints [Cao et al. (2003a)]. In contrast, welds produced at lower travel speeds (in the case of MIG or TIG welding) have their columnar grains curved away from the normal to the welding direction and align themselves with the welding direction. The orientation of the grains was reported to have important effects on the ductility of the welds [Ramasamy and Albright (2001)]. The columnar grain structure is very susceptible to hot tearing [Guitterez et al (1996) and Gaumann and Kurz (1998)]. Compared to anisotropic columnar grains, equiaxed grains are isotropic [Cao et al. (2003b)]. Equiaxed microstructures increase weld strength and reduce the susceptibility to solidification cracking because of better resistance to crack formation and propagation [Gaumann and Kurz (1998) and Jones et al (1998)].

![Columnar grains at Fusion line](image)

Fig.4.12. Optical Macrostructure of TIG welded joint of AA5083-H321 alloy
4.2.4. OPTICAL MICROSTRUCTURE OF TUNGSTEN INERT GAS WELDED AA5083-H321

The microstructure of the TIG weld cross-section is shown in figures 4.13 to 4.18. Fig. 4.13 shows the HAZ of TIG weld, while Fig. 4.14 depicts the shrinkage voids present in the weld zone. The presence of irregular voids and micro pores in the top portion and the middle portion of the weld zones are clearly revealed on the Fig. 4.15 and 4.16. The microstructure of the TIG weld at top and bottom are shown in Fig. 4.17 and 4.18.

During solidification of aluminium, there is a 6% reduction in the volume of the melt. This causes the shrinkage cavity as shown by Fig. 4.14 and 4.15 and other casting voids are formed in the TIG welded joint is shown by figures 4.16 and 4.17. Equiaxed grains are present in the TIG weld. Absence of micro pores and casting voids are noticed in the bottom portion of the TIG weld. The formation of casting voids and micro pores of size 80 micrometers are more on the top side of the weld. Fig.4.17 shows a cross-section at which absence of micro pores is noticed at the middle of the welded joint. The weld microstructure at the bottom of the weld portion is shown in Fig.4.18.

Due to the low density of aluminium, the buoyancy is too small for the bubbles to quickly move upward and escape from the weld pool during the solidification after welding [Gao et al. (2009)]. It can be seen from TIG welded joint in Fig.4.16 that pores are formed at the upper part of the weld. Besides the buoyancy factor, the relatively long escaping distance resulted from the thick weld reinforcement also increases the possibility of the formation of pores at the top part of the weld [DONG hong-gang et al. (2012)].

On the other hand, the AC pulses in the welding process contribute to the rapid solidification of the metal. Therefore, the grains of the weld are coarsened to equiaxed grains of larger size compared to the base metal [Yong Dongxia et al. (2012)]. The sizes of the coarse grains in the MIG weld of AA5083-H321 alloy is around 65 µm, whereas the base metal grains are at 32 µm [Christian and Murray (2006)]
Fig. 4.13. Microstructure of Heat affected zone of TIG welded joint of AA5083-H321 alloy

Fig. 4.14. Microstructure of TIG weld showing shrinkage cavity

Fig. 4.15. Microstructure of TIG weld at the top showing the presence of micro pores and voids in AA5083-H321 alloy
Fig. 4.16. Microstructure of weld at the middle showing voids only

Fig. 4.17. Microstructure of TIG weld showing equiaxed grains

Fig. 4.18. Microstructure of TIG weld at the bottom
4.2.5. MICROSTRUCTURE OF TUNGSTEN INERT GAS WELDED AA5083-H321

The SEM image of TIG welded joint is shown in Fig.4.19. The distribution of precipitates is less in TIG weld as shown in Fig.4.19. Whereas, the distribution of precipitates is very dense in base metal, which is shown by Fig.3.8 (b).

The distribution of precipitates is not uniform in the base metal as well as in the TIG weld shown by SEM images. The density of intermetallic compound was observed to be higher and in addition coarsening of intermetallic compound is noticed in Fig. 4.19 compared to Fig.3.8(b).

![Fig.4.19. SEM image of TIG welded joint of AA5083-H321 alloy](image)

4.2.6. HARDNESS OF TUNGSTEN INERT GAS WELDED AA5083-H321

Hardness measurement was taken in the transverse direction, i.e. parallel to the base plate surface. The hardness distribution for the AA5083-H321 alloy and its TIG weld is shown in Fig.4.20. It is observed that the hardness curve is symmetrical with respect to the weld centerline. The main reason for this fact is that the melt flow field on the both sides of weld center is uniform.

The hardness values of TIG welded joints have been found to be less than that of the base metal as shown by Fig.4.20. The lowest value of hardness at the TIG was is 66 Hv1. The hardness is reduced in TIG welded joint fusion zone of AA5083 weld zone by 14 % due to higher heat input.
4.2.7. TENSILE PROPERTIES OF TUNGSTEN INERT GAS WELDED AA5083-H321

The tensile properties such as yield stress, tensile strength and percentage elongation of AA5083-H321 aluminium alloy subjected to TIG welding are presented in Table 4.7. The yield stress and tensile strength of TIG welded joints are 176 MPa and 266 MPa, respectively. The strength coefficient of welded joint (known as the ratio between the yield stress of a welded joint and that of base metal) is 67 %. However, the elongation of TIG welded joints is 13.5 %. The reduction in tensile properties of TIG welded joints are as expected, owing to the presence of weld defects found in the TIG weld. During TIG welding of Al alloy, the effect of strain hardening is lost and weld exhibits a cast structure and thus leading to yield stress values of annealed state (160 MPa).

<table>
<thead>
<tr>
<th>Table 4.7. Tensile Properties of TIG weld</th>
</tr>
</thead>
<tbody>
<tr>
<td>Yield Stress, MPa</td>
</tr>
<tr>
<td>176</td>
</tr>
</tbody>
</table>

4.2.8. FRACTOGRAPHIC STUDIES OF TUNGSTEN INERT GAS WELDED AA5083-H321

TIG welding of AA5083-H321 alloy resulted in cast structure and the formation of coarse dendrites are shown in Fig. 4.21. Fractograph shows a mixed mode failure with dimples along with cleavage facets. This microstructure (with dendrites) is seen to affect the properties of the welded joints.
The formation of coarse dendrites in the TIG welded joint had an impact on the tensile properties as well as hardness of the welded joint. In TIG welding the coarse dendrite arm spacing (DAS) may be more due to slow heating and cooling nature of the process.

![Coarse Dendrites and Cleavage Facets](image)

**Fig.4.21.** Fracture surface of TIG welded joint of AA5083-H321 alloy exhibiting the presence of coarse dendrites and cleavage facets

### 4.2.9. DISCUSSIONS ON TUNGSTEN INERT GAS WELDING OF AA5083-H321 ALLOY

The relative merits and demerits of TIG welding techniques with respect to the AA5083-H321 alloy are discussed here.

**Effect of Mg vaporization on Yield Stress:** The concentration of Mg (wt.%) in the fusion zone is 2.55 % (Table 4.4), which is lower than that in the base metal (4.214 %) (Table 3.1) due to the evaporation of Mg during the welding process. Since Mg is the important strengthening element in Al-Mg alloys, the concentration reduction of Mg should be responsible for the yield stress decrease of TIG welded joint.

**Effect of Mn and Fe vaporization on Yield Stress:** Moreover the Al₆(Fe,Mn) particles distribute heterogeneously in the base metal, as shown in Fig.3.8(b), which leads to significant Orowan strengthening. The dissolution of Mn and Fe will minimize the formation of the intermetallic compound Al₆(Fe,Mn) in the weld zone after welding as shown in Fig.4.19. The reduction in this particle presence will affect the yield stress as well as hardness of the welded joint.
Effect of defect on yield stress: The formation of micropores, shrinkage cavities and voids also affected the properties by reducing the effective area of cross-section of the weld. All this defects acted as stress raisers and the failure of the welded joint has occurred at very low levels of yield stresses. All the defects mentioned above caused drastic loss of yield stress of the welded joint.

Effect of microstructure on elongation: The formation of columnar grains at the intersection of base metal and the weld reduces the ductility of the welded joints. The anisotropic nature of the columnar grains was responsible for 50% reduction in elongation percentage.

Effect of intermetallics on hardness: TIG welding process is a high heat input process. During slow heating and cooling of the process, the particles might have been dissolved in the welded joint. The TIG welded joint contained less number of the intermetallic compounds and their sizes were large compared to the sizes of the intermetallic compounds at the base metal. Reduction of the volume density of intermetallic compounds affects the hardness of the welded joint.

Effect of coarse dendrites on hardness: In case of TIG welding, very high arc temperature increases the peak temperature of the molten weld pool causing a slow cooling rate. This slow cooling rate, in turn, causes relatively wider dendrite spacing in the fusion zone [Ratnesh and Pravin (2010), and Akio Hirose et al (1997)]. These microstructures generally offer lower resistance to indentation and this may be one of the reasons for lower hardness.

Strengthening mechanisms in TIG welding: Grain coarsening due to TIG welding and formation of coarse dendrites caused more drop in yield stress due to Hall-Petch (Grain Boundary Strengthening) effect.

Particle strengthening (APB/Orowan) is reduced much in TIG welding due to dissolution of intermetallics. Effect due to strain hardening is totally lost in TIG welding.

In addition, the work hardening disappears in the as-cast fusion zone. The strength of TIG joints is lower than that of base metal. The improvement gained in yield stress due to cold-working of the aluminium alloy for the H-321 condition (Roughly around 261-150 = 111 MPa) is totally lost and the cast microstructure of the TIG welded joint weld zone attains yield stress levels equal to AA5083 annealed state (150MPa).
• For TIG welding process, it has been observed that 39.5 % of magnesium, 15.9 % of manganese, 24.3 % of iron and 20 % of zinc were subjected to evaporation.

• The macrostructure of TIG weld revealed the formation of columnar grains in the fusion line. Columnar grains having anisotropic properties have led to reduction in elongation.

• SEM studies revealed that dissolution of the intermetallic compound Al₆(Fe,Mn), Al₃Mg₂ and Mg₂Si in the TIG weld zone. The TIG weld contained relatively less number of the intermetallics. The size of the intermetallic was larger than that found in AA5083-H321 alloy.

• A maximum hardness reduction of 14 % was observed in the TIG welded joints compared to AA5083-H321 alloy. The distribution of hardness was similar to the weld center.

• The tensile properties of the TIG welded joints were inferior to that of the base metal. The joint efficiency based on yield stress and tensile strength are around 65 % (176Mpa) and 91 % (266MPa), respectively. The joint efficiency in terms of elongation was 50 % of that of AA5083-H321 alloy.

• Formation of coarse dendrites was observed in the fractograph of tensile testing specimens.

• The reduction in the volatile elements normally affects the strengthening of the welded joints. The reduction in magnesium and manganese reduced the solid solution strengthening and reduction in manganese and iron reduced the particle strengthening.

• The grain coarsening of TIG weld due to the high intensity of heat applied causes reduction in grain boundary strengthening.

• The presence of micro porosity, shrinkage voids, reduction in volatile elements content in the fusion zone were the main reasons for 35% reduction in yield stress.

• Vaporization of volatile elements, grain coarsening in the TIG weld, formation of coarse dendrites and dissolution of intermetallic compounds in the TIG weld are responsible for 14 % decrease in hardness in the welded joints.
4.3. LASER BEAM WELDING OF AA5083-H321 ALUMINIUM ALLOY

4.3.1. BEAD APPEARANCE OF LASER BEAM WELDED AA5083-H321

The bead appearance of LB weld at 3 kW is shown in Fig.4.22. The top surface of the weld bead shown in Fig.4.22 had “ropelike” irregular chevron geometry with some under fill. The bottom weld bead quality of LB welds produced in this alloy is typically very poor with numerous “spike like” projections of ejected and solidified weld metal that were up to 350 µm length.

4.3.2. COMPOSITIONAL ANALYSIS OF LASER BEAM WELDED AA5083-H321

The high-power density used for LB welding may cause selective vaporization of some alloying elements which has low fusion temperatures such as Magnesium, Manganese and Zinc in aluminium alloys because of their higher equilibrium vapor pressure than aluminium. The chemical composition of the LB weld is found out using spectrometer. The chemical compositions of the base metal and the three weld beads were listed out in Table 4.8 along with their percentage loss of the volatile elements in the weld bead in Table 4.9.

LB welding resulted in loss of certain alloying elements. The Magnesium, Manganese, Iron and Zinc present in the base metal were subjected to evaporation during fusion welding processes of aluminium alloys. It has been observed that the increase in incident power of the LB welding has increased the vaporization of volatile elements in the weld bead as shown in Table 4.8. The higher incident power LB welding produced maximum loss of volatile elements as shown in Table.4.9
Table 4.8. Composition analysis of LB welds

<table>
<thead>
<tr>
<th>Wt.%</th>
<th>Mg</th>
<th>Mn</th>
<th>Fe</th>
<th>Si</th>
<th>Zn</th>
</tr>
</thead>
<tbody>
<tr>
<td>Base metal</td>
<td>4.214</td>
<td>0.787</td>
<td>0.407</td>
<td>0.16</td>
<td>0.12</td>
</tr>
<tr>
<td>LB weld at 3.00 kW</td>
<td>4.190</td>
<td>0.650</td>
<td>0.373</td>
<td>0.164</td>
<td>0.01</td>
</tr>
<tr>
<td>LB weld at 3.25 kW</td>
<td>3.891</td>
<td>0.392</td>
<td>0.287</td>
<td>0.164</td>
<td>0.01</td>
</tr>
<tr>
<td>LB weld at 3.50 kW</td>
<td>3.678</td>
<td>0.401</td>
<td>0.320</td>
<td>0.164</td>
<td>0.01</td>
</tr>
</tbody>
</table>

Table 4.9. Percentage Loss/Gain of volatile elements in LB welds

<table>
<thead>
<tr>
<th>Processes</th>
<th>Magnesium Loss, %</th>
<th>Manganese Loss, %</th>
<th>Iron Loss, %</th>
<th>Silicon Gain, %</th>
<th>Zinc Loss, %</th>
</tr>
</thead>
<tbody>
<tr>
<td>LB weld at 3.00 kW</td>
<td>-1</td>
<td>-17.4</td>
<td>-8.35</td>
<td>+2.5</td>
<td>-91.7</td>
</tr>
<tr>
<td>LB weld at 3.25 kW</td>
<td>-7.7</td>
<td>-50.2</td>
<td>-29.48</td>
<td>+2.5</td>
<td>-91.7</td>
</tr>
<tr>
<td>LB weld at 3.50 kW</td>
<td>-13</td>
<td>-49</td>
<td>-21.48</td>
<td>+2.5</td>
<td>-91.7</td>
</tr>
</tbody>
</table>

4.3.3. OPTICAL MACROSTRUCTURE OF LASER BEAM WELDED AA5083-H321

The analysis of weld cross-sections allowed studying the influence of the main process parameters on the selected response variables. The Macrograph of the three LB welded joints at incident powers of 3 kW, 3.25 kW and 3.5 kW are shown in figures 4.23 to 4.25. The macrostructures of all the three LB welded joints contained weld defects. The top of the weld shown in Fig.4.22 is free of under fill defect; however, a small undercut is formed. The top of the weld shown in Fig.4.23 is having under fill defect of depth 500 µm. The top of the weld shown in Fig.4.24 is having maximum under fill defect of depth 750 µm. Increase in under fill depth has been observed, when there is increase in incident power of laser beam. The presence of under fill defect reduces the effective area of cross-section of the welded portion by reducing the thickness at the weld to 4 mm from 5 mm. Premature failure of LB welded joint is favored.
Fig. 4.23. Macrostructure of LB welded joint of AA5083-H321 alloy at 3 kW

Fig. 4.24. Macrostructure of LB welded joint of AA5083-H321 alloy at 3.25 kW

Fig. 4.25. Macrostructure of LB welded joint of AA5083-H321 alloy at 3.5 kW
4.3.4. OPTICAL MICROSTRUCTURE OF LASER BEAM WELDED AA5083-H321 AT 3 kW

The optical microstructure of LB weld at 3 kW incident power is shown in Fig.4.26. The enlarged view of the columnar grains is shown in Fig.4.27. The weld microstructure of the LB welded joint is shown in Fig.4.28. The intersection of the LB weld and the base metal at 3 kW is shown in Fig.4.26. The HAZ of the LB welded joint contains columnar grains adjacent to the base metal as shown in Fig.4.27. As discussed earlier in TIG welding of AA5083-H321 alloy in section 4.2.9, the presence of columnar grains has direct impact in the properties of the welded joint. The presence of fine dendrites of the Al-Matrix is noticed in this Fig. 4.28. In the current study, the LB welds have a finer microstructure than the base metal and TIG weld as shown in Fig. 4.28.
4.3.5. OPTICAL MICROSTRUCTURE OF LASER BEAM WELDED AA5083-H321 AT 3.5 kW

The optical microstructure of LB weld at 3.5 kW incident power is shown in Fig.4.29. The resolidified microstructure of LB weld contains a macro pore and fine dendrite as shown in Fig.4.30. The formation of fine dendrites is shown separately in Fig.4.31. The weld microstructure of the LB welded joint is shown in Fig.4.32.

The intersection of the LB weld and the base metal at 3.50 kW is shown in Fig.4.29. The HAZ of the LB welded joint is shown in Fig.4.29 consisting of columnar grains. The presence of columnar grains have a definite role in the property evaluation. The presence of macropore and fine dendrite are shown in Fig.4.30. The formation of spherical macropore of size 0.2 mm has been observed at the lower half of the welded joint. Presence of micropores were reduced and formation of the macropore initiated at the 3.5 kW incident power LB weld. The presence of the macropore introduces stress concentration at the cross-section and responsible for strength reduction in the welded joint. The fine dendrite is shown in Fig.4.31. The weld microstructure at 3.5 kW LB welded joint weld is shown in Fig.4.32.
Fig. 4.29. Microstructure of Fusion line of LB weld at 3.5 kW

Fig. 4.30. Microstructure of LB weld showing Macro porosity and Fine dendrite

Fig. 4.31. Microstructure of LB weld showing Fine dendrite
4.3.6. MICROSTRUCTURE OF LASER BEAM WELDED AA5083-H321 AT 3 kW

The SEM image of the Base Metal and LB weld is shown in Fig.4.33 and 4.34. In order to find out the presence of the various precipitates and dispersoids EDS have been taken and shown in Figures 4.35 and 4.36. The presence of $\text{Al}_6(\text{Fe,Mn})$ dispersoid in the weld was observed and shown in Fig.4.35, whose chemical composition is given in Table.4.10. The irregular shaped intermetallic compounds sizes are varying between 5 to 10 µm.

The presence of intermetallic compounds $\text{Al}_6(\text{Fe,Mn})$ in the weld region of base metal welded using 3 kW laser were clearly releaved in SEM images (Fig. 4.34), while the presence of dendrites are observed in the weld zone.
Fig. 4.34. SEM image of LB welded joint of AA5083-H321 alloy at 3 kW

Fig. 4.35. EDS of Al₆(Fe,Mn)

Fig. 4.36. EDS image of Al₆(Fe,Mn)
Table 4.10. Chemical Composition of Al₆(Fe,Mn)

<table>
<thead>
<tr>
<th>Element</th>
<th>Al</th>
<th>Fe</th>
<th>Mn</th>
<th>Cr</th>
</tr>
</thead>
<tbody>
<tr>
<td>Wt.%</td>
<td>72.94</td>
<td>17.01</td>
<td>9.69</td>
<td>0.03</td>
</tr>
</tbody>
</table>

4.3.7. MICROSTRUCTURE OF LASER BEAM WELDED AA5083-H321 AT 3.5 kW

The SEM image of LB weld is shown in Fig. 4.37. In order to find out the presence of the various precipitates and dispersoids in the LB welded joint at 3.5 kW incident power, EDS have been taken and shown in figures 4.38 to 4.40. Two different particles were captured by SEM and are shown in Fig. 4.38. The presence of both Al₃Mg₂ and Al₆(Fe,Mn) are not observed in the EDS analysis as shown in figures 4.39 and 4.40 as well as Tables 4.10 and 4.11.

The SEM image of LB weld at 3.5 kW is shown in Fig. 4.37. The dendrites shown in this figure are coarser than the dendrites in 3 kW LB weld (shown by Fig. 4.34). Absence of the intermetallic compounds Al₆(Fe,Mn) were observed in this SEM microstructure shown in Fig. 4.38, 4.39, and 4.40 (with the help of the Table 4.11 and 4.12).
Fig. 4.38. SEM image of LB welded joint of AA5083-H321 alloy at 3.5 kW showing the presence of uniform distribution of particles.

Fig. 4.39. EDS image of particle 1.

Fig. 4.40. EDS image of particle 2.
Table 4.11. Chemical Composition of Particle 1

<table>
<thead>
<tr>
<th>Element</th>
<th>Al</th>
<th>Mg</th>
<th>Mn</th>
<th>Cu</th>
<th>O</th>
</tr>
</thead>
<tbody>
<tr>
<td>Wt.%</td>
<td>96.53</td>
<td>1.22</td>
<td>0.62</td>
<td>0.64</td>
<td>0.99</td>
</tr>
</tbody>
</table>

Table 4.12. Chemical Composition of Particle 2

<table>
<thead>
<tr>
<th>Element</th>
<th>Al</th>
<th>Mg</th>
<th>Mn</th>
<th>Fe</th>
<th>Cu</th>
<th>Ti</th>
</tr>
</thead>
<tbody>
<tr>
<td>Wt.%</td>
<td>95.43</td>
<td>3.05</td>
<td>0.66</td>
<td>0.45</td>
<td>0.36</td>
<td>0.05</td>
</tr>
</tbody>
</table>

4.3.8. HARDNESS OF LASER BEAM WELDED AA5083-H321

The hardness values measured across the LB welds at 3.00 kW, 3.25 kW and 3.50 kW incident powers are shown in Fig.4.41. The average Vickers hardness of the base metal and Butt welds for the three power modes considered in this study is shown in Fig.4.41. The hardness values of LB welded AA5083-H321 samples at 3 kW and 3.25 kW power range were nearly 9% and 4% greater than that of the base metal whose average was 77 Hv1. The hardness values for the remaining set of experiments conducted at 3.5 kW were less than that of the base metal value of 77 Hv1 by 4%. The hardness values decreases as the incident power increases and similar observations were made by by Ancona et al (2007), and Cao et al (2003a). The effects of heat input on the hardness of the FZ and the HAZ are shown in Fig.4.41. It can be seen that the lower the heat input was, the higher was the hardness of the FZ and the HAZ. In general, the increase of the hardness can be attributed to the grain refinement and the strengthening effect of the brittle and hard phase. During a welding process, grain coarsening occurred in both the HAZ and the FZ with an increase of heat input. Hence, a higher hardness value was achieved with lower heat input. In addition, it can be seen that with a further increase of heat input, the hardness of the FZ and HAZ changed slightly. This is because with a relatively high heat input, more granular phase was formed in the HAZ and in FZ, which partially offsets the effect of grain coarsening on the decrease of the micro hardness of HAZ and FZ of the welded joint.

Higher Dendrite Arm Spacing has a tendency to reduce the properties of the welded joint. When laser incident power increases, the dendrite arm spacing is increased. The increase in dendrite arm spacing decreases the hardness values of the fusion zone. Such microstructures offer lower resistance to indentation and this may be one of the reasons for lower hardness values at higher power. Venkat et al (1997) have reported hardness
profiles of laser welded Al-Mg alloys showing an increase in hardness in the weld metal due to a smaller grain size owing to the rapid cooling rates associated with the LB welding process.

Yamaguchi et al (2001) found out an overall hardness increase in the weld metal of AA5052 aluminium alloy laser welded joints. The small dendrite arm spacing coupled with micro-segregation of Mg in the dendrites was reported as reasons for increase in hardness at laser weld.

Fig. 4.41. Distribution of hardness in AA5083-H321 alloy LB weld

4.3.9. TENSILE PROPERTIES OF LASER BEAM WELDED AA5083-H321

The average values of the yield stress and the tensile strength of the LB welded joints at the three incident power ratings are tabulated in Table 4.13. The best tensile properties of the AA5083-H321 alloy were obtained for 3.5 kW incident power LB welding. The weld yield stress and weld tensile strength for 3.5 kW power range was obtained around 226 MPa and 248 MPa respectively. The percentage of elongation for the LB welded joints is within 12 to 13 %, which is nearly equal to 50 % of the percentage elongation of the base metal value. The tensile testing of the welded specimens revealed joint efficiencies (i.e., ratio between the yield stress or tensile strength in the weld and in the base metal) of around 85 % for 3.5 kW power.

When a high power laser is used, the heat input per unit weld length will increase, which results in more molten metal and a larger weld width [Kamel Abderrazak et al
Obviously, increasing the heat input leads to a significant increase in dendrite arm spacing and, consequently, a decrease in solidification rate [Akin Odabasi et al (2010)]. Increase in LB incident power increased the tensile properties of the LB welded joints due to the formation of fine grains and these results are well corroborated with studies performed by Kim and Park (2011).

<table>
<thead>
<tr>
<th>Incident Power, kW</th>
<th>Yield Stress, MPa</th>
<th>Tensile Strength, MPa</th>
<th>Elongation, %</th>
</tr>
</thead>
<tbody>
<tr>
<td>3.00</td>
<td>219</td>
<td>243</td>
<td>12</td>
</tr>
<tr>
<td>3.25</td>
<td>222</td>
<td>246</td>
<td>12</td>
</tr>
<tr>
<td>3.50</td>
<td>226</td>
<td>248</td>
<td>13</td>
</tr>
</tbody>
</table>

Fig. 4.42. Failure location of tensile specimens of AA5083-H321 alloy LB weld at 3 kW

Fig. 4.43. Failure location of tensile specimens of AA5083-H321 alloy LB weld at 3.25 kW
4.3.10. FRACTOGRAPHIC STUDIES OF LASER BEAM WELDED AA5083-H321

The fractograph of the LB weld shown in Fig.4.45 was taken by a scanning electron microscope. The fractograph of LB weld contains 20 µm micro pores.

4.3.11. DISCUSSIONS ON LASER BEAM WELDING OF AA5083-H321 ALLOY

The relative merits and demerits of LB welding techniques with respect to the AA5083-H321 alloy are discussed here.

Effect of Mg vaporization on Yield Stress: It has been observed that the vaporization of volatile elements were proportional to the LB incident power applied. When the LB incident power increases, the absorption of heat by the aluminium plate also increases, this in turn increases the vaporization of volatile elements in LB welds. The reduction of
magnesium plays an important role in reducing the yield stress of the welded joint as shown by Fig.2.1. Approximate reduction in annealed condition of the alloy is around 10 ~ 15 MPa. The base metal, which has been used in this study, is in strain-hardened condition and the reduction of yield stress might be more than 15 MPa.

**Effect of Mn and Fe vaporization on Yield Stress:** The manganese evaporation from the LB weld affects the solid solution strengthening of the welded joint. The grain refining effect caused by the addition of Mn is reduced in the respective welds. The vaporization of the alloying elements such as, Mn and Fe has caused reduction of the intermetallic compounds $\text{Al}_6(\text{Fe,Mn})$ in the LB welds. The dissolution of the compounds affects the precipitate strengthening of the LB welded joints. Hence, reduction in yield stress is evident.

**Effect of Mn and Fe vaporization on Hardness:** The hardness values of LB welded joints at 3 kW and 3.25 kW are higher than the hardness values of the base metal because of the presence of the intermetallic compounds. The reduction in the chemical composition of the vaporizing element viz., Mn and Fe has normally caused dissolution of $\text{Al}_6(\text{Fe, Mn})$ particles in the higher incident power laser beam welding. The hardness values of LB welded joint at 3.5 kW are less than that of the base metal because of the absence of the higher hardness intermetallic compounds. This observation of absence of the various particles in the weldzone and HAZ was correlating with earlier observation done by Moon and Metzbower (1983). Moon and Metzbower have found that only 5 % of the precipitates were remaining in the LB weld zone compared to the base metal.

**Effect of defect on yield stress and tensile strength:** The porous cross-section at 3 kW LB weld reduced the tensile properties. Reduction of porosity in the subsequent LB welds at 3.25 kW and 3.5 kW increases the tensile properties.

The reduced cross-sectional area of the LB weld act as stress raiser and decrease in tensile properties has been observed. The fine grains formed due to LB welding generally increase the tensile properties. Due to combined effect of fine grains and stress concentration at the cross-section, a minimum decrease of 15 % was observed in tensile properties at 3.5 kW incident power LB weld.

**Effect of microstructure on mechanical properties:** Generally, in order to maintain the mechanical properties during welding of aluminium alloys, the heat input and time of
exposure to very high temperatures must be minimized. In comparison with arc welding processes, laser welding offers the benefits of low-heat input and extremely rapid cooling rate, all of which will help to minimize the metallurgical problems in the fusion zone. For example, high cooling rate will tend to slow down the development of blisters because of the short time in which the diffusion of hydrogen can take place. In addition, the low heat input will tend to keep a very narrow HAZ then, retaining some to the strength of the materials.

**Strengthening mechanisms in LB welding:** In LB welded AA5083-H321 alloy decrease in magnesium and manganese were observed. So there is a possibility of slight decrease in solid solution strengthening mechanism. The grains in the weld are fine dendrites and their sizes are also smaller than that of the grains in the base metal [Yang Dongxia et al. (2012)]. Hence, an increase in grain-size (Hall-Petch effect) strengthening is possible.

Hence, reduction in the tensile properties when compared to the base metal is maximum. The reduction in solid-solution strengthening and particle strengthening mechanisms is compensated by the increase in grain-size strengthening mechanism; the LB welds produced better tensile properties of 85 % of that of the base metal. If suitable welding parameters were selected in future, the LB weld strength can reach that of the base metal.

4.2.12. CONCLUSIONS

- From the compositional analysis of the LB weld bead, it has been observed that magnesium, manganese, iron and zinc were subjected to evaporation. Reduction of volatile elements has increased when there was an increase in the Laser beam power applied.
- The depth of under fill and drop through increased when the incident power has increased. Increase in evaporation of volatile elements was responsible for increase in under fill depth.
- Anisotropic behavior of the columnar grains formed at the fusion line was responsible for low ductility (50 %) of the LB welded joints of AA5083-H321 alloy.
• LB weld produced very fine microstructure at 3.5 kW compared to LB welding at 3 kW and 3.25 kW incident powers. Rapid heating and cooling of LB welding produced finer dendrites in LB welded joints.
• Maximum amount of micro pores of size less than 20 µm were observed in the weld cross-section at 3.00 kW incident power, whereas reduction of micro pores were observed during higher incident power LB welding.
• From SEM microstructural studies, dissolution of the intermetallic compounds Al₆(Fe,Mn) was also observed in LB welding at higher incident power LB welding.
• Presence of the intermetallic compound (Al₆(Fe,Mn)) in the low incident power LB welds had caused higher hardness above base metal, whereas the dissolution of the compounds had reduced the hardness below that of the base metal.
• The hardness values were greater than that of the base metal for LB welds at low incident powers (3 kW and 3.25 kW). The hardness values of LB weld at 3.5 kW were less than that of the base metal by 4 %. Increase in laser power had reduced the hardness values at the welded joints.
• The best tensile properties of LB weld were obtained at 3.5 kW incident power rating. The joint strength based on yield stress was around 85 % (226MPa), whereas it was 80 % at 2.5 kW incident power rating.
• Ductility of the LB welded joints was at 50 % of AA508-H321 alloy plate.
• Fractography of tensile testing specimens revealed that 20 µm size micro pores in the LB weld.
• Vaporization of important volatile elements has negative effect on mechanical properties of the welded joints. Reduction of magnesium and manganese reduces the solid solution strengthening of the LB welded joint.
• Vaporization of manganese and iron from the LB weld had reduced the precipitation of the intermetallic compound Al₆(Fe,Mn), which in turn reduces both tensile properties and hardness of the LB welded joint.
4.4. FRICTION STIR WELDING OF CAST ALUMINUM-MAGNESIUM-SCANDIUM ALLOY

4.4.1. COMPOSITIONAL ANALYSIS OF FRICTION STIR WELDED CAST Al-Mg-Sc ALLOY

The chemical compositional results of base metal and FS welded Cast Al-Mg-Sc alloy plate is listed out in Table 4.14. On the analysis of chemical composition of the FS weld, it has been observed that there is a slight increase (12.4%) in magnesium content in the weld zone compared to the base metal.

The comparison of the chemical composition of the FS weld with that of the base metal is shown in Table 3.1, and a gain of 12.5% magnesium in the fusion zone of the FS welded samples. The other elements were subjected to minor variations with that of the base metal.

<table>
<thead>
<tr>
<th>Element</th>
<th>Mg</th>
<th>Mn</th>
<th>Fe</th>
<th>Sc</th>
<th>Ti</th>
</tr>
</thead>
<tbody>
<tr>
<td>Wt. %</td>
<td>4.45</td>
<td>0.645</td>
<td>0.366</td>
<td>0.287</td>
<td>0.005</td>
</tr>
</tbody>
</table>

4.4.2. OPTICAL MACROSTRUCTURE OF FRICTION STIR WELDED CAST Al-Mg-Sc ALLOY

The macrostructure of FS welded joints is shown in Fig.4.46. The different FS weld zones are marked on the figure. The FS weld nugget is relatively finer than the TMAZ and HAZ. Two different kinds of defect are shown in this figure. One is void formation and the other is tunnel defect.
4.4.3. OPTICAL MICROSTRUCTURE OF FRICTION STIR WELDED CAST Al-Mg-Sc ALLOY

The microstructure shown by Fig.4.47 and Fig 4.48 are taken from the advancing and retreating side of TMAZ of the FS weld. The microstructure of the FS welded joint of the cast Al-Mg-Sc alloy near the shoulder of the FS welding tool is shown in Fig.4.49. The microstructure taken near the centre of the plate is shown in Fig.4.50. The microstructure of the TMAZ reveals the presence of finer structure compared to the cast base metal. Fig.4.51 shows the microstructure of the FS welded joint very near the back plate or for away from the shoulder. The weld intersection at the bottom of the weld tool is shown in Fig.4.52. The microstructure of the weld nugget at the top and at the bottom of the welded joint is shown in Fig.4.53 and 4.54 respectively.

During FS welding process the temperature attained in the weld nugget is around 550°C [Hirata et al. (2007)]. The Al₃Sc precipitate will harden any structure in which it is present. So the resistance offered by the Al₃Sc precipitate in the cast alloy made lesser penetration in the FS welded joint as shown by Fig.4.49. The partial penetration of FS weld tool implies that more axial load may be required for full penetration.

The weld nugget contains fine, fragmented precipitates and the adjoining TMAZ contains slightly coarser microstructure compared to the weld nugget. Clear boundaries of the two zones are evident by a demarcation of the zone on the advancing Side.

There is some difference in the grains observed between the two sides. HE Zhen-bo et al. (2011) explained the reasons for such variations in the microstructures. The relative rate between rotating pin and base metal reached a peak in advancing side and it was the lowest in the retreating side. The strain degree and rate were greater in the advancing side, and the distortion of fibrous microstructure in the advancing side is more severe than that in the retreating side which brought the asymmetry of micro hardness distribution as well as microstructures.

The grains are distributed in a homogeneous pattern. The grains of FS welded joints are very fine compared to the grains of the cast base metal as shown in Fig.3.8 (a). The dynamically recrystallised grains are finer and fragmented from the original grains formed during casting. The intermetallic compounds Al₆(Fe,Mn) have converted into globular precipitates in the FS weld nugget. Compared to the microstructure shown in
Fig. 4.57, the microstructure of the weld nugget at the top shown by Fig. 4.56 is very fine and fragmented into smaller size due to forging and extrusion caused by the FS welding tool. Similar observations of fine microstructure at the weld nugget top portion were reported earlier [Hirata et al. (2007)].
Fig. 4.49. Microstructure of TMAZ and nugget zone at the top

Fig. 4.50. Microstructure of the TMAZ and nugget zone at the middle

Fig. 4.51. Microstructure of TMAZ and nugget zone at the bottom
Fig. 4.52. Microstructure of Intersection at weld bottom

Fig. 4.53. Microstructure of Weld Nugget at the top

Fig. 4.54. Microstructure of Weld Nugget at the bottom
4.4.4. MICROSTRUCTURE OF FRICTION STIR WELDED CAST Al-Mg-Sc ALLOY

The SEM image of FS welded joint of Cast Al-Mg-Sc alloy is shown in Fig 4.55. It can be identified that the weld nugget of the FS welded joint is having very fine microstructure. The SEM image of the same shown in Fig.4.58 is very fine and fragmented into smaller size due to forging and extrusion caused by the FS welding tool. Most of the precipitates were subjected to fragmentation due to stirring the softened metal formed in the FS welding process, which enhances the tensile properties of the FS welded joint of cast Al-Mg-Sc alloy. This observation is very well correlated with the earlier research [Liu et al (2004)] for FS welded joints of cast aluminium alloy AC4A.

4.4.5. HARDNESS OF FRICTION STIR WELDED CAST Al-Mg-Sc ALLOY

The Vickers micro hardness variation in along the cross-section perpendicular to the tool traverse direction of the FS welded specimen produced at a tool rotation speed of 650 rpm is shown in Fig.4.56. The microstructural variations along the weld zone have led to difference in hardness at different zones. The average hardness value for the base metal was around 96 Hv1. The hardness values of FS welding on Cast Al-Mg-Sc alloy plate are not symmetrical with respect to the weld nugget. It is clearly identified that the hardness values on the advancing side (84 Hv1) is greater than the hardness values on the retreating side (81 Hv1) from Fig.4.48. This drop in hardness in the case of friction stir welded plates on the retreating side is only 2-3 % compared to the advancing side. The maximum drop of hardness was 14 % less than that of the base metal.
4.4.6. TENSILE PROPERTIES OF FRICTION STIR WELDED CAST Al-Mg-Sc ALLOY

The tensile properties of the Cast Al-Mg-Sc alloy FS weld performed using various loads is shown in Table 4.15. FS welding produced at 23 kN axial load produced superior tensile strength compared to 9 kN and 17 kN axial load. Increase in axial force increased heat input to the weld nugget which in turn increased the tensile properties of the welded joint of the cast alloy. Similar results were produced by Elangovan et al(2009). As the best properties were obtained for 23 kN axial load, FS welding and further characterization studies were carried out at 23 kN FS welded samples.

On comparing the tensile properties of friction stir welded samples of cast Al-Mg-Sc alloy plates with the cast base metal properties. It is observed that there is an enhancement of 14 % in the values of yield stress and 17.3 % in the values of tensile strength. The broken samples of the friction stir welded cast Al-Mg-Sc alloys at 23 kN after the tensile testing is shown in Fig.4.57. The tensile failure occurred in the base metal irrespective of the tunnel defect present in it. None of the sample has failed in the weld nugget of the friction stir welds. This indicates that the weld is stronger than the cast base metal.
Table 4.15. Tensile properties of FS welded joints at three different loads

<table>
<thead>
<tr>
<th>Axial Load, kN</th>
<th>Yield Stress, MPa</th>
<th>Tensile Strength, MPa</th>
<th>Elongation, %</th>
</tr>
</thead>
<tbody>
<tr>
<td>9</td>
<td>207</td>
<td>230</td>
<td>6.8</td>
</tr>
<tr>
<td>17</td>
<td>212</td>
<td>242</td>
<td>5.6</td>
</tr>
<tr>
<td>23</td>
<td>231</td>
<td>251</td>
<td>7.2</td>
</tr>
<tr>
<td>Base Metal</td>
<td>203</td>
<td>215</td>
<td>7.0</td>
</tr>
</tbody>
</table>

The tensile failures of the FS weld occurred outside the weld as shown in Fig. 4.57. In order to further evaluate the tensile properties of the weld, the gauge length of the tensile specimens was reduced to 4 mm as represented by Ren et al (2007) and as shown in Fig. 3.11.

The yield stress, the tensile strength and the percentage of elongation of the weld were measured to be 271 MPa, 316 MPa, and 22.7 % respectively. This indicates that the tensile properties of the weld are much higher than those of the base metal and the heat affected zone, which are 33.5 % and 47 % higher than the base metal values. The ductility of the FS welded joint was higher by 217 %. The failed samples of component part joints are shown in Fig. 4.58. The tensile properties of component part joints were tabulated in Table 4.16.
4.4.7. FRACTOGRAPHIC STUDIES OF FRICTION STIR WELDED CAST Al-Mg-Sc ALLOY

For FS welded joints the fracture location of the global joint was observed at the base metal, but the failure location in the case of component part joints was in weld region. The fractography of the FS welded joint is shown in Fig.4.59.
4.4.8. DISCUSSIONS ON FRICTION STIR WELDING OF CAST ALUMINIUM-MAGNESIUM-SCANDIUM ALLOY

The variation in hardness values indicates that the FS weld nugget is relatively softer than the cast base metal. This contradicting behavior of reduced hardness values of Friction Stir welded joint compared to the cast base metal was earlier reported in FSP of cast aluminium 2285 alloys [Karthikeyan et al. (2009)]. The variation in the size and shape of the precipitates and dispersiods may be the reason for such reduction in hardness in the FS welded joints of cast Al-Mg-Sc alloys.

**Effect of microstructure on properties:** During friction stir welding of cast Al-Mg-Sc alloys, as the casting defects are totally eliminated from the weld nugget, the FS welds have shown an improvement in the tensile properties compared to the cast base metal. FS welding resulted in enhancement of tensile properties due to fine and fragmented dynamically recrystallised grains. This study agrees with a previous study reporting an improvement in tensile strengths and ductility in aluminium alloys on FS processing [Nakata et al. (2006)].

**Effect of intermetallics on hardness:** It can be observed from Fig.4.48, that a hardness degradation region composed of a weld nugget, two TMAZ and two HAZs have occurred in the joints. From the plotted micro hardness it was inferred that the retreating side of the weld (nearby the tool pin) had lower hardness compared to the advancing side. Also the advancing side had high hardness due to the strain developed in the plate by the welding tool. In fact, the hardness profile is mainly affected by the distribution of the precipitates in the weld [Sato et al. (2001)].

**Effect of intermetallics on tensile properties:** The size and volume density of the intermetallic compounds in the FS weld nugget were found to be varying compared to that in the base metal. All these precipitates could refine grains, inhibit movement of dislocations, stabilize substructures, impede growth of sub-boundaries to high-angle grain boundaries, and inhibit nucleation and growth of recrystallisation process. The reduction in size and increase in volume density due to fragmentation and nucleation of new intermetallic particles, viz., $\text{Al}_6(\text{Fe,Mn})$ and $\text{Al}_3\text{Sc}$ will have a tendency to increase the precipitation strengthening (Orowan strengthening) mechanism of the FS weld nugget.
**Strengthening mechanisms:** The grains in the weld nugget were formed due to recrystallisation and their sizes will be less than that of the grains in the base metal. Hence, an increase in grain-size (Hall-Petch effect) strengthening is possible. Hence, the tensile properties of FS welded joints are greater than that of the cast base metal.

4.4.9. CONCLUSIONS

- Increase in the presence of magnesium was observed in the FS weld nugget.
- Due to elimination of casting defects and grain refinement in the FS weld microstructure, better tensile properties were obtained.
- Reduction of hardness was observed in the FS weld nugget. The unsymmetrical distribution of hardness was due to accumulation of intermetallic compounds on advancing side of the FS weld. The minimum hardness (81Hv1) at the retreating side was 14 % less than the cast base metal Al-Mg-Sc hardness (96 Hv1).
- FS welding of Cast Al-Mg-Sc alloy at 23 kN axial load has produced best tensile properties compared to the cast Al-Mg-Sc alloy plate FS welds at 9 kN and 17 kN axial loads.
- The FS welded cast Al-Mg-Sc yield stress, tensile strength and percentage elongation were 271 MPa, 316 MPa and 22.7 % respectively.
- The yield stress and tensile strength of the Cast Al-Mg-Sc alloy FS weld are much higher than those of the Cast Al-Mg-Sc alloy plate by 33.5 % and 47 % respectively.
- The percentage elongation of the Cast Al-Mg-Sc alloy FS weld was higher by 217 % compared to the Cast Al-Mg-Sc alloy plate.
- Increase in grain boundary strengthening and particle strengthening were the main cause of increase in tensile properties. The solid solution strengthening remained the same in the cast base metal as well as the FS welded cast Al-Mg-Sc alloy plate.
4.5. COMPARATIVE DISCUSSIONS

4.5.1. AA5083-H321 ALLOY

The Marine Aluminum alloy AA5083-H321 was subjected to TIG, LB and FS welding. The Comparative performance of the three processes on AA5083-H321 alloy is shown in Fig. 4.60, 4.61.

![Graph showing tensile properties of TIG, LB and FS welded joints of AA5083-H321 alloy]

**Fig. 4.60.** Tensile properties of TIG, LB and FS welded joints of AA5083-H321 alloy

![Graph showing hardness values of TIG, LB and FS welded joints of AA5083-H321 alloy]

**Fig. 4.61.** Hardness values of TIG, LB and FS welded joints of AA5083-H321 alloy
4.5.2. COMPARATIVE DISCUSSION OF FS WELDED JOINTS OF AA5083-H321 ALLOY AND CAST Al-Mg-Sc ALLOY

The Marine Aluminum alloy AA5083-H321 and cast Al-Mg-Sc alloy were subjected to FS welding. The FS welding process on AA5083-H321 alloy and cast Al-Mg-Sc alloy in comparison with their respective base metals are shown in Fig. 4.62, 4.63 and 4.64.

![Tensile properties chart](image1)

Fig.4.62. Tensile properties of FS welded joints of AA5083-H321 alloy and Cast Al-Mg-Sc alloy in comparison with respective base metals.

![Ductile property chart](image2)

Fig.4.63. Ductile property of FS welded joints of AA5083-H321 alloy and Cast Al-Mg-Sc alloy in comparison with respective base metals.
Fig. 4.64. Hardness values of FS welded joints of AA5083-H321 alloy and Cast Al-Mg-Sc alloy in comparison with respective base metals.
4.6. CONTRIBUTIONS TO THE LITERATURE

1. FS welding of AA5083-H321 alloy can be safely carried at 650 rpm tool rotation speed, 158 mm/min welding speed and at 9 kN axial load to produce weld joint efficiency based on yield stress above 90%.

2. CO₂ LB welding of AA5083-H321 alloy can be safely performed at 3.5 kW power and 3500 mm/min welding speed to produce a weld joint efficiency based on yield stress above 85%.

3. FS welding of Cast Al-Mg-Sc alloy plate can be done at 650 rpm tool rotation speed, 158 mm/min and at 23 kN axial load to produce a weld yield stress above 270 MPa.