Chapter 2

Literature Review

This chapter presents the past and current scenarios on the two extensively used Al alloys viz., Al-Mg and Al-Mg-Sc alloys. In addition, an overview of different welding processes such as FS, MIG, TIG and LB welding of Al-Mg alloys and of Al-Mg-Sc alloys is also presented.

2.1. CANDIDATE MATERIALS

2.1.1. ALUMINIUM-MAGNESIUM ALLOY 5083

Pure aluminium yield stress is 10 MPa only. The strength of Al is major constraint for their application as structural materials. However, adding various alloying elements such as Cu, Mg, Mn, and Zn, the strength of the alloy can be enhanced. Addition of Magnesium increases the tensile properties of the aluminium alloy system. The annealed state strength of AA5083 alloy is around 150 MPa, whereas, the strength of this alloy has been increased to above 250MPa by strain hardening techniques [Lloyd (1980)].

Amongst other alloying elements, Mg is found to improve the corrosion resistance in addition to yield stress. Mg forms substitution solid solution with Al as its atomic radius is 12 % larger than that of Al atom. The presence of Mg enhances strength as well as ductility. Minimum of 23 wt% ductility is observed up to the addition of 6 wt % of Mg and higher concentration above 6wt% is observed to result in difficulties in processing.

The yield stress of Al-Mg alloy increases with Mg content for an annealed AA5083 as shown by Fig.2.1. The curve depicts decrease in % elongation as Mg content increases.
Another findings show that the existence of other elements in solution (e.g. iron and manganese) can also provide a significant strengthening contribution, provided that it can be kept in solution. Alloying elements in solution with aluminium have different strengthening contributions as shown in Table 2.1.
Table 2.1. Effect of alloying elements on Yield Stress of Aluminium [Moataz (2007)]

<table>
<thead>
<tr>
<th>Alloying Element</th>
<th>Yield Stress, MPa/wt.%</th>
</tr>
</thead>
<tbody>
<tr>
<td>Mg</td>
<td>16 ~ 19</td>
</tr>
<tr>
<td>Mn</td>
<td>85</td>
</tr>
<tr>
<td>Fe</td>
<td>200</td>
</tr>
<tr>
<td>Si</td>
<td>20</td>
</tr>
<tr>
<td>Cu</td>
<td>70</td>
</tr>
<tr>
<td>Ti</td>
<td>45</td>
</tr>
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</table>

The strength of this AA5083-H321 alloy is attributed to the presence of various intermetallic compounds, such as, \( \text{Al}_3\text{Mg}_2 \), \( \text{Al}_6(\text{Fe},\text{Mn}) \), and \( \text{Mg}_2\text{Si} \) etc. The tensile properties and hardness of the base metal is significantly influenced by the presence of the intermetallic compounds. The specific weight and hardness of the intermetallic compounds are given in Table 2.2.

Table 2.2. Intermetallic Compounds in aluminium alloys and their Specific weight and Hardness [Sato et al(2001), Moataz M. Attallah (2007)]

<table>
<thead>
<tr>
<th>Intermetallic Compounds</th>
<th>Specific Weight, g/cm³</th>
<th>Hardness, Hv1</th>
</tr>
</thead>
<tbody>
<tr>
<td>( \text{Al}_3\text{Mg}_2 )</td>
<td>2.25</td>
<td>24 ~ 32</td>
</tr>
<tr>
<td>( \text{Al}_6(\text{Fe}, \text{Mn}) )</td>
<td>3.3</td>
<td>90 ~ 95</td>
</tr>
<tr>
<td>( \text{Mg}_2\text{Si} )</td>
<td>1.988</td>
<td>42</td>
</tr>
<tr>
<td>Al-Matrix</td>
<td>2.4</td>
<td>70 ~ 75</td>
</tr>
<tr>
<td>Al_{3}Fe</td>
<td>3.9</td>
<td></td>
</tr>
<tr>
<td>Al_{3}Ti</td>
<td>3.36</td>
<td></td>
</tr>
<tr>
<td>Al_{3}Sc</td>
<td>3.0</td>
<td></td>
</tr>
</tbody>
</table>

Lloyd and Court (2003) have investigated the effect of grain size on tensile properties of Al-Mg alloys. Grain size refinement is an important strengthening mechanism in Al-Mg (5000 series) alloys, which have a relatively large Hall-Petch slope compared with the other Al alloys. In addition, the high work hardening rate exhibited by Al-Mg alloys provides excellent formability.
AA5083 (Al-4.5 Mg-1.0 Mn) is a medium strength and non-heat treatable alloy that is generally known for its excellent corrosion resistance. However, it can become susceptible to stress corrosion cracking (SCC) when it is exposed to temperatures ranging from 50°C to 200°C for sufficiently long periods of time. This phenomenon is known as “Sensitization” and is widely associated with the precipitation of β-phase (Al₃Mg₂) on grain boundaries.

Lloyd (1980) have investigated the deformation of commercial Al-Mg alloys, viz., AA5052 (2.34 wt. % Mg), AA5154 (3.31 wt. % Mg) and AA5083 (4.46 wt. % Mg) over a wide range of strain rates and temperatures and the major factors influencing their deformation. The strain hardening strengthening increases the alloy strength to twice than that of the strength of the annealed state of this alloy.

Carroll et al. (2000) have investigated the effect of Zn addition on the grain boundary precipitation and corrosion of Al-5083. Stress corrosion cracking (SCC) concerns in aluminium alloys containing Mg levels greater than 3.5 % have been largely attributed to the formation of the β- phase (Al₃Mg₂) at grain boundaries. H31-type temper, which included a small amount of cold work (18 % reduction), followed by stabilization at 110°C for two hours reduces the grain boundary precipitation.

Jones et al. (2001) have investigated the effect of Mg on Stress Corrosion Cracking of Al-Mg alloy. As the solid solubility of Mg in Al is limited to about 2 wt. % at 200°C, AA5083 which contains 4.2 wt % of Mg rapidly cooled from processing temperatures above this temperature will result in a supersaturated solid solution of Mg.

The solid solution hardening is a result of an interaction between the mobile dislocations and the solute atoms [Oyvind Ryen et al. (2006)]. A number of mechanisms have been suggested. However, the most relevant mechanisms for substitutional alloying of aluminium are the elastic interactions due to (1) the size misfit, where the size of the solute atom differs from the size of the matrix atoms and creates a strain field around the atom, and (2) the modulus misfit, where the difference in binding force between the solute atoms and the matrix atoms results in a hard or soft “spot” in the matrix.

Christian and Murray (2006) have conducted an extensive analysis of various strengthening mechanisms contribution of strength in AA5083-H321 alloy. They have observed that solid solution strengthening contributes 50 % of the total strength, whereas
the other two strengthening mechanisms contributed around 25% of the total strength each. The presence of magnesium and manganese contributes the solid solution strengthening. The addition of Mn, Cu, Cr, and Ti contributes to the grain refinement effect and increases the grain size strengthening. The formation of the intermetallic compounds such as Al₆(Fe,Mn) and its distribution in the base metal contributes to the particle strengthening of the alloy.

Youssef et al. (2006) have investigated a nano-crystalline Al-Mg alloy with high strength and good ductility. The alloy possessed very high tensile strength. Grain refinement provided 90% of the total strength of the alloy. The solid solution strengthening contribution was only 10% of total strength.

Dharmendra Singh et al (2013) have conducted microstructural studies on AA 5083 alloy processed by cryorolling and afterwards annealing. They found that cryorolled alloys can be subjected to heat treatments such as annealing etc.

Thiyaneshwaran and Sureshkumar (2013) have conducted microstructural and mechanical property studies in Aluminium 5083 Alloy processed by equal Channel Angular Extrusion. It can be concluded that processing by ECAE increased the mechanical and wear properties of aluminium 5083 alloys.

2.1.2. ALUMINIUM-MAGNESIUM-SCANDIUM ALLOYS

The addition of scandium, first proposed by Willey in 1971, involves the formation of Al₃Sc precipitates during annealing as a result of decomposition of the Al-Sc solid solution. Apart from major alloying elements, rare earth elements added in micro-concentration helps to dramatically increase the strength of AA5083 alloy. The increase of strength results from both the precipitation of ordered Al₃Sc particles and the preservation of a fine grain structure. Indeed, high volume fraction and high dispersivity of these particles in the matrix promote a noticeable elevation of the recrystallisation temperature of wrought Al-Mg-Sc alloy semi products [Filatov et al. (2000)].

The thermal stability of Al₃Sc particles is very high (melting point of Al₃Sc: 1320°C) and the misfit in lattice parameter between Al and Al₃Sc is only 1.3% in binary Al-Sc alloys, which makes these particles stable against coarsening up to 350°C [Marquis and Seidman (2001)].
Roder et al (1996) have investigated the effect of scandium additions on an Al-5.2 % Mg alloy. An addition of 0.25% of Scandium reduces the grain size to half of its original size. The presence of Al₆Mn was also observed along with Al₃Sc. Addition of 0.26 % of scandium enhances the yield stress from 17 ksi to 42 ksi in AA5254 aluminium alloy, which means that Sc added alloy's yield stress has increased by 2.5 times than that of the ordinary alloy.

Aiura et al. (2000) have investigated the effect of scandium addition on the microstructure of AA5083 alloys. The dendrite arm spacing was reduced from 35 µm (base metal) to 20 µm in the scandium added alloy. Scandium presence reduced the dendrite arm spacing in the cast alloys. The size of the intermetallic compounds is also getting reduced. The reduction in dendrite arm spacing and fine precipitates caused increase in the values of hardness of the cast alloy. Sc added is mostly utilized in the Al₃Sc phase, not in the intermetallic compounds other than Al₃Sc. The growth of intermetallic compounds during solidification has been suppressed. The addition of Sc has increased the yield and tensile strength by 60 % and 30 %, respectively.

Davydov et al. (2000) have investigated the effect of zirconium and titanium presence in the scandium added aluminium alloys. Addition of zirconium and/or tungsten (about 0.1 %) has brought down the requirement of scandium addition from 0.55 % to 0.18 %. Such alloys were observed to be non-dendrite.

Decomposition of supersaturated scandium in aluminium, occurs during ingot annealing, give rise to dramatic strengthening. Micro hardness increases 2.5 times. A reduction in micro hardness in the case of prolonged cooling is caused by coagulation of the strengthening Al₃Sc particles. Scandium is one of the most effective modifiers of a cast grain structure in aluminium alloys. Continuously cast commercial scandium-bearing aluminium alloy ingots have, as a rule, a non-dendrite structure which has no internal structure.

The ability of scandium to refine grains is used in the case of fusion welding of aluminium alloys susceptibility of the alloys to crack formation reduces sharply, while mechanical properties of weld joints increase noticeably.

Titanium and vanadium, just as zirconium, are capable to be dissolved in the Al₃Sc phase to substitute for scandium forming intermetallics like Al₃(Sc₁₋ₓ,Tiₓ) and Al₃(Sc₁₋ₓ,Vₓ).
x, Vx). Ti and V, as Zirconium, reduce that maximum scandium content which results in the manifestation of the modifying effect. It is useful to limit Ti content within a range of 0.02-0.06 % keeping the cost of Titanium alloy.

Al3Sc dispersoid has a number of features (intermetallics crystalline structure, morphology and dispersivity of particles, particle distribution density) which cause it’s much more pronounced influence on a structure and properties of aluminium and a number of aluminium alloys in comparison with dispersoids of other transition metals.

Filatov et al (2000) have conducted and reported the effect of scandium additions on Al-Mg alloys with varying quantities of magnesium additions. The Russian alloy 01535 containing an average of 4.2% Mg had the properties as shown in Table 2.3.

| Table 2.3. Tensile properties of Russian alloy 01535 (Al-4.2% Mg-Sc) [Filatov et al (2000)] |
|-----------------------------------------------|-----------------|-----------------|
| Yield stress, N/mm² | Tensile strength, N/mm² | Elongation, % |
| 280               | 370              | 20              |

Yancy and Sanders (2000) have mentioned that in 5xxx alloys, strain hardening is the major strengthening mechanism. They also mentioned that the lattice parameter of Al, 0.4049 nm, and that of Al3Sc (Ll2), 0.4103 nm are similar resulting in little strain energy between the dispersoid and the matrix. The barrier to nucleation is small thus facilitating the decomposition of the supersaturated solid solution. The resultant dispersoids are very effective in retaining wrought microstructure approaching the solvus temperature.

The reason for such an enhancement of yield stress for the Sc containing alloy over the Sc-free alloy is attributed due to higher young's modulus value for the dispersoid Al3Sc compared to the modulus of Al matrix. On comparing the modulus values of both Al3Sc and Al (Al3Sc-164 GPa and Al-70.7GPa), the young's modulus of Al3Sc is higher by 230 % than that of Al [Marquis and Seidman (2001)].

Kendig and Miracle (2002) have investigated the effect of strengthening mechanisms in a specially prepared Al-6Mg-2Sc-1Zr alloy. The current work quantifies the active strengthening mechanisms at room temperature and explicitly considers solid solution strengthening, grain boundary strengthening, and Al3(Sc,Zr) precipitate strengthening. Existing strengthening models, together with data from microstructural characterization were used to determine the magnitude of individual contributions. Strengthening due to
the sub-micron grain size was the largest contribution to alloy strength, followed in decreasing order by precipitate strengthening and solid solution strengthening. The solid solution strengthening contribution is around 75 ~ 100 MPa, the grain boundary strengthening is around 280 ~ 395 MPa, and the particle strengthening is around 180 MPa. By increasing the secondary Al₃(Sc,Zr) volume fraction from 0.015 to 0.035 at the optimum particle size, an additional strength increase of 100 MPa is predicted.

In order to achieve higher strength in these alloys, it is advantageous to use the highest possible cooling rates in order to increase the supersaturation of Sc and Zr, and hence to increase the volume of fine Al₃(Sc,Zr) particles.

Lathabai and Lloyd (2002) have observed that addition of 0.17 wt% Sc to a cast Al-Mg alloy produced 55 % higher yield stress and 20 % higher tensile strength compared to Cast Al-Mg alloy. Addition of scandium improved the weld hot cracking resistance.

Hyde et al. (2002) have mentioned that small additions of the transition element Sc can produce Al-alloy castings and fusion welds with a more refined and uniform grain structure than can normally be achieved with traditional grain refiners. They have concluded that grain refinement was more effective in the Ti containing alloy.

Tolley et al. (2005) have quoted that Sc produces the largest increase in strength per atomic percent of solute. Addition of Zirconium together with Sc improves the effectiveness of Sc as an inhibitor of recrystallisation and increases the stability of the alloy during prolonged annealing at high temperatures. The addition of Zr to Al-Sc results in a large number density of smaller precipitates as shown in Table 2.4. Lowering the size of Al₃Sc particles leads simultaneously to the growth of the yield stress and to the increase of recrystallisation temperature.

<table>
<thead>
<tr>
<th>Precipitate</th>
<th>Size, nm</th>
<th>Volume Density, µm³</th>
</tr>
</thead>
<tbody>
<tr>
<td>Al₃Sc</td>
<td>74</td>
<td>33</td>
</tr>
<tr>
<td>Al₃(Sc₁₋ₓ,Zrₓ)</td>
<td>54</td>
<td>87</td>
</tr>
</tbody>
</table>

Kaiser et al. (2008) have tested the mechanical properties of Al-6Mg alloy doped with varying amounts of scandium ranging from 0.2 to 0.6 wt. %. They have concluded that the influence of addition of Scandium was much pronounced on yield stress than on the
tensile strength. Magnesium aluminide (Al<sub>3</sub>Mg<sub>2</sub>) does not enter the precipitate structure and hence the strengthening effect of Al<sub>3</sub>Sc is additive to the solid solution strengthening due to magnesium [Kaiser et al. (2008)].

Xu Wang et al. (2010) have conducted the microstructure and properties of Al alloys with high Mg content due to the alloying elements Sc, Zr, and Ti. They have observed that the refinement of grain size was more by the addition of Sc, Zr, and Ti.

Srinivasa Rao et al. (2010) have captured the Al<sub>3</sub>Sc precipitate, which was present in their experimental alloy and shown by TEM microstructure in Fig.2.2. The 20 ~ 40 nm sized precipitate Al<sub>3</sub>Sc and its chemical composition is given in Table 2.5.

Table 2.5. Chemical Composition of Al<sub>3</sub>Sc Precipitate [Srinivasa Rao et al. (2010)]

<table>
<thead>
<tr>
<th>Elements</th>
<th>Weight, %</th>
<th>Atomic, %</th>
</tr>
</thead>
<tbody>
<tr>
<td>Mg</td>
<td>0.70</td>
<td>0.80</td>
</tr>
<tr>
<td>Al</td>
<td>88.50</td>
<td>92.40</td>
</tr>
<tr>
<td>Sc</td>
<td>11.50</td>
<td>6.70</td>
</tr>
</tbody>
</table>

Fig.2.2. TEM image of Al<sub>3</sub>Sc precipitate in the aluminium matrix of AA2219 [Srinivasa Rao et al. (2010)]
2.2. MICROSTRUCTURE AND MECHANICAL PROPERTIES OF WELDMENTS

2.2.1. FRICTION STIR WELDING OF HEAT-TREATABLE ALUMINIUM ALLOYS

Madhusudhan Reddy et al (2009) studied the microstructure and mechanical property correlations in AA6061 aluminium alloy subjected to friction stir welding. The results indicate that a softened region occurred in the FS welding joints due to the dissolution of the strengthening precipitates/coarsening of the precipitates. The hardness profile across the weld was not consistent with grain size distribution. This was because the hardness profiles in the age hardenable aluminium alloy depend strongly on the precipitate distribution rather than on the grain size. The tensile strength of the joint was observed to be lower than that of the base metal.

Beathe Heinz and Birgit Skrotzi (2002) studied the micro-structural and mechanical properties of FS welded aluminium alloy 6013 in the T4 and the T6 tempered conditions, after the welding and after applying a post weld heat treatment (PWHT) to the T4 condition. The microstructural studies revealed that the elongated pancake like microstructure of the base metal was transformed into a dynamically re-crystallized microstructure of considerably smaller grain size in the weld nugget. Also, the strengthening precipitates that were present before welding in the T6 state were dissolved during welding in the nugget, while an over-aged state with a much larger precipitate size was established in the heat-affected zone (HAZ). The micro-hardness measurements and tensile tests showed that the HAZ is the weakest region of the weld and the welded sheet exhibited reduced strength and ductility as compared to the base metal.

An analysis of friction welds using the inverse problem approach in the friction stir welding of AA2517-T87 aluminium alloy was studied by Fonda and Lambrakos (2002). The study showed that the softening at the TMAZ/HAZ boundary was due to the coarsening and transformation of the strengthening precipitates during the welding process, and the fracture location corresponds to the region with least precipitate strengthening.

Lee et al (2003) reported that in AA6061 FS welded joints, the area of the stir zone reduced with a decrease in the tool traverse rate. Higher the tool rotating speed produced high temperature in the weld zone and slow cooling rate in the weld after welding. Under
the low rotational speed conditions less heat were produced due to lack of stirring. Also, under high rotational speed conditions, the excessive stirred materials moved to the upper surface, which resulted in voids defect in the weld zone. The grain size in the weld zone increased with increasing the rotational speed of the tool.

Tensile properties and fracture locations of friction stir welded joints of 2017 – T351 aluminium alloy were investigated by Liu et al (2004). As the rotational speed increased, the strained region widened, and the location of maximum strain finally moved to the advancing side from the original retreating side. This implies that the fracture location of the joint was affected by the rotational speed of the tool.

Ma et al (2006) investigated the FS welding to cast aluminium alloy A356 plates, to enhance the mechanical properties through micro-structural refinement and homogenization. The study disclosed that these micro-structural changes led to a significant improvement in both strength and ductility. The T6 treatment of the FS welding specimen had higher tensile strength and ductility than those of cast A356.

Elangovan and Balasubramanian (2008a) investigated the influences of the tool pin profile and welding speed on the formation of the friction stir processing zone in AA2219 aluminium alloy. In this investigation the effect of welding speed and tool pin profile on FSP zone formation in AA2219 aluminium alloy were analyzed. Five different tool pin profiles (straight cylindrical, tapered cylindrical, threaded cylindrical, triangular, and square) were used to fabricate the joints at three different welding speeds. The results indicated that the square pin profiled tool produce mechanically sound and metallurgical defect free welds compared to other tool pin profiles irrespective of welding speeds. The joints fabricated at a welding speed of 0.76mm/s showed superior tensile properties than the 0.37 mm/s and 1.25 mm/s, irrespective of tool pin profiles. Also, the joint fabricated using square pin profiled tool at a welding speed 0.76 mm/s exhibited maximum tensile strength, higher hardness and finer grains in the FSP region than other combinations used for this investigation.

Elangovan et al (2008b) studied the influences of the tool pin profile and axial force on the formation of FS processing zone in AA6061 aluminium alloy. The tensile strength of FS welded joints was lower at lower welding speed (0.25 mm/s). The tensile strength of the welded joint increases when the welding speed was increased from 0.25 mm/s, and
reaches a maximum value at 1.25 mm/s. Further, when the welding speed was increased above 1.25 mm/s, the tensile strength of the joint decreased.

Lakshminarayanan and Balasubramanian (2008) found that the tensile strength of FSW on RDE-40 aluminium alloys increases with an increase in the welding speed and decreases when welding speed increases after 45 mm/min. The percentage of contribution of FS welding process parameters was calculated. It was found that the traverse speed made 33% contribution to the tensile strength of welded joints and the optimum value of the process parameter of traverse speed is 45 mm/min.

Lakshminarayanan and Balasubramanian (2009) observed that at lower welding speed (less than 22 mm/min) the friction stir welded 7039 aluminium alloy joints were produced with a pin hole defect, due to the excessive heat input per unit length of the weld. The tunnel defect was found at the bottom during high rotational speed (75 mm/min) on the retreating side due to the inadequate flow of material caused by insufficient heat input. Response surface methodology (RSM) and artificial neural network (ANN) models were developed for predicting the tensile strength of the FS welded specimens. It is concluded that the rotational speed has more influence on tensile strength, followed by welding speed and tool axial force. The optimum parameters of the FS welded joints were 1460 rpm as rotational speed, 40 mm/min welding speed and the axial force was 6.5 kN.

The morphology of the precipitates in different regions of the FS welded sample is illustrated in AA6061 aluminium alloy recently. Some precipitates are highlighted with their average sizes being given. Size, distribution, and morphology of the precipitates before and after welding by the FS welding method in different regions were compared [Fahimpour et al (2013)]. Precipitates in the nugget zone have a nearly homogeneous distribution with relatively small size (2.5 µm). This is due to the violent stirring of this region during FS welding operation. Precipitates of TMAZ region are larger in size and less homogenous (in comparison with the nugget zone). Precipitates of HAZ region are even larger (in comparison with the nugget and TMAZ regions) with a much larger deviation.
2.2.2. FRICTION STIR WELDING OF NON-HEAT-TREATABLE ALUMINIUM-MAGNESIUM ALLOYS

AA5052 - O alloy (Al - 2.5 \% Mg):

Yong-Jai KWON, et al (2009) have conducted friction stir welding of automotive aluminium alloy 5052-O (annealed condition), whose magnesium content is around 2.5 \%. The experiments were conducted at tool rotation speed varying between 500 to 3000 rpm. Defect-free welds were obtained for welds at rotation speeds of 500 to 3000 rpm. The tensile strength of the FS welded joints at 500, 1000, and 2000 rpm were similar to that of the base metal. The average hardness was greater than the base metal by 33 \% at 500 rpm tool rotation speed. And also increase in tool rotation speed produced fine grains in the weld nugget.

Kumbhar and Bhanumurthy (2012) have investigated the effect of friction stir welding of AA5052 with AA6061 alloys.

Kumbhar et al (2011) have reported that in friction stir welding of AA5052 aluminium alloy the hardness values at the nugget-TMAZ interface was slightly greater on the advancing side than that of the base metal.

AA5086 - O alloy (Al - 4.2 \% Mg) and AA6061-T6:

Jamshidi Aval et al (2012(a) and (b)) have investigated the dissimilar friction stir welding of AA5086-O and AA6061-T6 aluminium alloys. The rotational speeds of 840 and 900 rpm and the linear speeds of 100 and 150 mm/min were found to be sound welds. The hardness at the Stir Zone of AA5086 side was higher than that of TMAZ and this may be attributed to small recrystallised grain size of SZ.

AA5456 - alloy (Al - 5.4 \% Mg):

Hua-Bin Chen et al. (2006) have studied about the external factors on the FS welding defects are abundant when the experiments of FS welding were conducted for 5456 automotive aluminium alloy. With the changes of tool tilt angle and material condition, defects can be generated. These defects can be conventional ones (lack of penetration or voids), or lazy S, which are unique to friction stir welding. However, the origin of the defects remains an area of uncertainty. In this paper an attempt has been made to investigate the formation of various defects. It was suggested that the distribution of
Al₂O₃ oxide particles in the local region did not affect the mechanical properties. In FS welding process, the oxide layer on the initial butt surface experiences intensively stirring, it would be smashed to oxide particle, so that the oxide particle distribution in the weld may be an initiation site for tension fracture. Channel defect is a very serious defection, which affects the tensile properties and elongation percentage greatly, and decreases the tensile strength and elongation percentage sharply.

**AA5083 - O alloy (Al - 4.2 % Mg):**

Dawes and Thomas (1996) have reported that AA5083-O (annealed condition) FS welded joint produced 100 % joint efficiency based on tensile strength. The joint efficiency based on yield stress is 98 %. The tensile testing samples failed outside the FS weld.

Svensson et al (2000) have investigated that FS weld in the 5083-O aluminium alloy fractured near the center of the weld. Hardness was approximately constant across the welded zone in AA5083. Aluminium alloys always contain some impurities, mainly iron and silicon. Owing to the low solid solution solubility of many elements, formation of intermetallic particles is very common. Iron, silicon, manganese, copper and chromium, for example, form intermetallic compounds. Thus, the possible intermetallics that can form are numerous. They are fairly coarse and easily resolved by optical microscopy. However, in manganese bearing alloys, a relatively fine dispersion of particles, of the Al-Mn-Si type and having a bcc crystal lattice, forms during homogenization. These dispersiods can have a profound effect as nucleation sites for finer precipitates, but in themselves are too coarse to have any significant effect on tensile properties. The non-hardenable alloys receive their strength from solid solution hardening and from deformation (i.e. dislocation strengthening).

Sato et al. (2001) have investigated the microstructural factors governing hardness in friction stir welds of the solid-solution-hardened Al alloys 1080 and 5083. The effect of grain boundary on the hardness was examined in an Al alloy 1080 which did not contain any second-phase particles. The weld of Al alloy 1080 had a slightly greater hardness in the stir zone than the base metal. The maximum hardness was located in the TMAZ. The stir zone consisted of recrystallised fine grains, while the TMAZ had a recovered grain
structure. The increase in hardness in the stir zone can be explained by the Hall-Petch relationship. On the other hand, the hardness profiles in the weld of Al alloy 5083 were roughly homogeneous. The present authors have examined the hardness profile associated with the microstructure in a FS weld of precipitation-hardened Al alloy 6063-T5. The hardness profile was found to be strongly affected by precipitate distribution rather than grain size in the weld. Such softening is caused by dissolution and growth of strengthening precipitates during the thermal cycle of the welding.

Generally, FS welding does not result in softening in the solid-solution-hardened Al alloys. Svensson et al. (2000) observed the microstructures of the stir zone and the base material in the weld of Al alloy 5083-O and then correlated the mechanical properties with the microstructure. The hardness profile was roughly uniform in the weld. The stir zone had fine equiaxed grains with a lower density of large particles (1 to 10 µm) and a higher density of small particles (0.1 to 1 µm). They concluded that the hardness profile mainly depends on dislocation density, because the main hardening mechanism for this alloy is strain hardening.

Recently, Sato et al. (2001) applied FS welding to an equal-channel angular (ECA) pressed Al alloy 1050. FS welding reproduced fine grains with a grain size of about 0.61 µm in the stir zone, which achieved suppression of the hardness reduction. These studies have suggested that the hardness is mainly affected by the grain size in FS welds of solid-solution-hardened Al alloys.

Consequently, the relationship between the microstructure and the hardness profile remains unclear in the welds of solid-solution-hardened Al alloys. Commercial solid-solution-hardened Al alloys often contain a high density of particles. A homogeneous distribution of many small particles may strengthen the Al alloy by Orowan mechanism. However, the dominant factor governing the hardness has not been clarified in the welds of solid-solution-hardened Al alloys.

A cross section perpendicular to the welding direction in the friction-stir welded Al alloy 5083-O contained an elliptical nugget shape in the stir zone. The average grain size of the base metal is about 18 µm, while the stir zone has an average grain size of about 4 µm.
Sato et al. (2002) have investigated the material Al alloy 5083-O subjected to FS welding. Rotation speed of the tool was varied in the range of 250 to 800 rpm, and the travelling speed was kept constant at about 5.9 mm/s. Rotation speeds more than 250 rpm resulted in defect-free welds.

The starting material has coarse grains with a low density of dislocations and a high density of fine particles. The fine particles can be presumed to be Al_{6}(Fe,Mn) phases by referring to the previous studies. The average grain size of the starting material was approximately 7.7 µm and its hardness value was in the range of 79 to 85 Hv.

The hardness value of the stir zone increases with decreasing the grain size. This result suggests that a Hall-Petch relationship between hardness (Hv) and grain size (d) is available through the following equation: \( Hv = H_0 + k_H d^{-1/2} \) where \( H_0 \) and \( k_H \) are appropriate constants.

A change in the slope of Hall-Petch equation may be caused by homogeneous distribution of fine particles in Al alloy 5083. Al alloy 5083 contains a high density of fine particles. These particles must be distributed throughout the FS processed material, because they have a higher solvus temperature than the melting temperature of the Al-matrix. The fine particles derive the strength of Al alloy by Orowan hardening. Orowan hardening depends on an average distance between particles, which was measured to be about 0.69 µm in the starting material. If many particles were homogeneously distributed in the grain interiors, they would work more effectively as obstacles to the dislocation movement than high angle boundaries, which diminishes the effect of Hv on the d.

Shigematsu et al (2003) have observed that at the 5083-5083 joining zone, the hardness was slightly higher than the original value because of grain refinement. The strength of the 5083-5083 annealed joint was 97 % of that of the original material.

Charit and Mishra (2004) have investigated the microstructure evaluation during FSP of AA5083 alloy. The parent material contained coarse primary intermetallic particles of Al_{6}(Fe,Mn), Al-Mn-Fe-Si, and Mg_{2}Si. However, the morphology of these particles was altered by FSP, that is, the size became finer, and the shape more equiaxed.

Fujii et al. (2006) have investigated the effect of tool shape on mechanical properties and microstructure of FS welded aluminium alloys. For 5083-O whose deformation resistance is relatively high. The weldability is significantly affected by the rotation
speed. For a low rotation speed (600rpm), the tool shape does not significantly affect the microstructures and mechanical properties.

Many researchers have investigated the relationship between the microstructure of stir zone and the mechanical property of FS-welded 5083 aluminium alloy in Annealed condition. They have observed that the microstructures of the stir zones consisted of fine equiaxed grains at various FS welded conditions in FS-welded 5083-O Al alloy. They have also found that the grain size of the stir zone decreased with the decrease in friction heat flow during FS welding [Hirata et al (2007)].

Leal and Loureiro (2008) investigated the effect of overlapping friction stir welding passes in the quality of the welds of aluminium alloys. In this investigation the overlapping passes were made in 3 mm thick plates of 5083-O and 6063-T6 alloys. Tunnel defects were detected in the first and second passes of welds in the 5083-O alloy but not in the 6063-T6 alloy. The first pass resulted in a small increase in hardness in the welds in the 5083-O alloy. However, in the 6063-T6 alloy, there was a substantial decrease in hardness and strength. Subsequent overlapping passes produced a modest increase in hardness and strength in both alloys as well the elimination of tunnel defects in welds in 5083-O alloy. The mechanical efficiency of the welds in alloy 5083-O is equal to one as opposed to welds in alloy 6063-T6 that show an efficiency around 0.47.

Min-Su HAN et al. (2009) have investigated the effect of FS welding tool rotation speed and traverse speed on the formation of defects on AA5083-annealed alloy. They have found out that the optimum tool rotation speed and traverse speed to get defect-free FS welds on AA5083 at annealed condition is 800 rpm and 124 mm/min.

Damjan Klobcar et al (2012) have investigated the FS welding of AA5083 in annealed condition for a 4 mm thick plate. The plan of experiments were prepared based on the abilities of a universal milling machine, where the tool-rotation speed varied from 200 rpm to 1250 rpm and the welding speed varied between 71 mm/min to 450 mm/min. The Vickers hardness HV1 (load equal to 9.807 N) was measured across the weld. They have concluded that the hardness was above that of the base metal for the samples which were heated above the temperature of recrystallization for 0.44 r/mm.
**AA5083-H111 alloys (Al – 4.2 % Mg):**

Chionopoulos et al (2008) have investigated the effect of tool pin on AA5083-H111 aluminium alloy with conical and screw type pin. Out of conical and screw type pin head used in their experiment, the conical pin geometry resulted in defect-free welds at specific welding parameters.

Palanivel et al. (2010) and Palanivel and Koshy Mathews (2011) have investigated the influence of five tool pin profiles on dissimilar welding of AA5083-H111 and AA6351 aluminium alloys. They have concluded that among five tool pin profiles used, square pin profile give better tensile strength and the stirred zone of the welded area has finer grains.

**AA5083-H131 alloys (Al – 4.2 % Mg):**

The presence of very fine Al₆Mn precipitates promotes/stimulates grain nucleation during the recrystallisation process resulting in an ultra-fine grain microstructure. [Grujicic et al.(2011)]. Qualitative Analysis of hardening Mechanisms within the FSW Joint: Solid-solution strengthening. This hardening mechanism is present in all four weld-zones and its contribution to the material hardness is expected to be fairly uniform across the entire weld region. Strain Hardening: When AA5083 is in a H131 temper condition, strain-hardening mechanism provides a contribution to the material hardness in the base metal zone which is larger than the contributions of the other two mechanisms. In the weld nugget region, material microstructure and properties are dominated by dynamic recrystallisation and, hence, the contribution of strain hardening to the overall material hardness in this region is minimal. Grain-Size Refinement: Since, to a first-order approximation, the average grain size does not change between the base metal zone, the HAZ, and the TMAZ, the contribution of this strengthening mechanism to the overall material strength is expected to be comparable in these three weld-zones. On the other hand, dynamic recrystallisation yields a very fine-grain structure within the nugget zone so that the overall contribution of the grain-refinement mechanism to the material hardness is expected to be largest in this weld zone.

**AA5251 and AA5083 alloys (Al – 1.2 % Mg and Al – 4.2 % Mg):**

Moataz et al. (2007) have investigated the effect of various strengthening mechanisms on hardness distribution along the cross-section of friction stir welded AA5251 (1.9 % Mg) aluminium alloy. They have concluded that the TMAZ/WN strength was found to be
primarily controlled by grain boundary strengthening and, in specific locations by dislocation-particle (Orowan) strengthening caused by the submicron Al$_6$(Fe,Mn) particles. The hardness was taken at 1 kg load.

**AA5086-H32 alloys (Al – 4.2 %Mg):**

Cam et al. (2008) have investigated the mechanical properties of friction stir butt-welded Al 5086 H32 aluminium alloy. The welding speed was taken as 1600 rpm. The traverse speed was varied between 175 mm/min to 225 mm/min. The best combination of strength and ductility performances, i.e., 75 % and 25 % was obtained from the joint produced with a traverse speed of 200 mm/min at the tool rotation speed of 1600 rpm.

**AA5083-H32 alloys (Al – 4.2 % Mg):**

Sangshik Kim et al. (2008) have investigated the fatigue crack propagation in FS welded AA5083-H32 aluminium alloy. The tool rotation speed and traverse speed was kept at 1600 rpm and 250 mm/min, respectively. The grains formed in the various locations of the FS weld are shown in Table 2.6. The tensile properties if AA5083-H32 alloy base metal and FS welds are shown in Table 2.7.

### Table 2.6. Grain size in various locations of FS weld of AA5083-H32 aluminium alloy [Sangshik Kim et al. (2008)]

<table>
<thead>
<tr>
<th>Location</th>
<th>Grain size, µm</th>
</tr>
</thead>
<tbody>
<tr>
<td>BM</td>
<td>19</td>
</tr>
<tr>
<td>TMAZ</td>
<td>16</td>
</tr>
<tr>
<td>WN</td>
<td>9.2</td>
</tr>
</tbody>
</table>

### Table 2.7. Tensile properties of FS welded joints of AA5083-H32 aluminium alloy [Sangshik Kim et al. (2008)]

<table>
<thead>
<tr>
<th>Process</th>
<th>Yield stress, MPa</th>
<th>Tensile strength, MPa</th>
<th>Elongation, %</th>
</tr>
</thead>
<tbody>
<tr>
<td>FS welding</td>
<td>150</td>
<td>301</td>
<td>17</td>
</tr>
<tr>
<td>AA5083-H32</td>
<td>137</td>
<td>306</td>
<td>22</td>
</tr>
</tbody>
</table>

**AA5083-H321 alloys (Al – 4.2 % Mg):**

Czechowski (2005) have conducted FS welding on AA5083-H321 alloy by double butt welding technique and obtained very good results as shown by Table 2.8. The FS welding produced 88 % efficiency based on yield stress and 93 % efficiency based on
tensile strength. The ratio between the percentage elongation of the weld and the base metal was around 53%.

<table>
<thead>
<tr>
<th></th>
<th>Yield stress, N/mm²</th>
<th>Tensile strength, N/mm²</th>
<th>Elongation, %</th>
</tr>
</thead>
<tbody>
<tr>
<td>FS Welded Joint</td>
<td>238.3</td>
<td>322.2</td>
<td>10.4</td>
</tr>
</tbody>
</table>

Table 2.8. Tensile properties of FS welded joints of AA5083-H321 alloy [Czechowski (2005)]

Higher traverse speeds were associated with low heat inputs, which results in faster cooling rates of AA5083 FS welded joint [Lomolino et al 2005]). This could significantly reduce the extent of metallurgical transformations taking place during welding, and hence, the local strength of the individual regions across the weld zone.

Steuwer et al (2006) studied the effect of the process parameters on residual stress of dissimilar friction stir welds in AA5083 – AA6082. The rotation speed is more useful process variable when attempting to optimize the properties of friction stir welds.

Christian and Murray (2006) have investigated the effect of friction stir processing on the MIG weld bead of AA5083-H321 alloy plate. They have concluded that FSP enhances the yield stress value by 110 Mpa over MIG welding. The increment in yield stress was due to combined increase in both grain boundary strengthening as well as particle strengthening of the FSP region. The solid solution strengthening contribution remained unaltered due to FSP.

The influence of the FS process parameters on the formability of AA5083 was studied by Hirata et al (2007). The friction stir welds were fabricated using a tool shoulder diameter of 12 mm, pin diameter of 4 mm and a tool inclination angle of 3°. The rotational speeds of the tool were varied from 500 rpm to 1000 rpm and the welding speed from 100 to 200 mm/min respectively. The yield strength of the specimens increased slightly with a decreasing rotational speed or welding speed. The tensile strength of the specimens was found to be the same for different friction stir welding conditions. The improvement in the ductility was observed during the changes in the combination of the rotational speed and welding speed.

Lombard et al (2008) studied the optimizing friction stir welding process parameters to minimize defects and maximize fatigue life in 5083 –H321 aluminium alloy. A flute with
threaded pin tool profile was used for this investigation. The rotational speed is the key parameter which governs the tool torque, temperature, and frictional power, and hence, the tensile strength of the FSW joint of AA5083-H32. The highest chance of defect free welds was between the rotational speeds 615 to 630 rpm. It has been observed that in these welds the rotation speed was also found to strongly influence the torque acting on the tool, the extent of material mixing in the stir zone, and the predicted peak temperatures in the TMAZ.

Sri Rangarajalu et al. (2008) have investigated the effect of pin profile on the mechanical properties and microstructure of AA5083-H321 aluminium alloy. Three different pin profiles were used in this study as shown in Fig.2.3. They are (1) Straight Cylindrical, (2) Taper Cylindrical, and (3) Threaded Cylindrical with Left handed threading. They have concluded that among the three tools used tapered pin profile resulted in defect free welds over a better range of parameters.

Fig.2.3. Different tool pin shapes used in FS welding of AA5083-H321 aluminium alloy plate[Sri Rangarajalu et al. (2008)]
The welding parameters were optimized for friction stir butt welding of AA5083 alloy in H321 condition using L9 Taguchi orthogonal design [Vijayan et al. (2010)]. It was found out that the optimum values of rotational speed to be 650 rpm for high tensile strength and the optimum welding speed and the axial force were 115 mm/min and 17 kN respectively.

Yazdipour et al. (2011) have investigated the microstructures and properties of MIG and FS welds in aluminium alloy 5083-H321. The results show that the extension of the heat affected zone is higher in the MIG welding than in FS welding. The weld microstructure of MIG specimens contains equiaxed dendrites as a result of solidification process during MIG welding while FS welded samples have wrought microstructures. The hardness distribution showed less loss of hardness in the FS weld compared to the MIG weld. The tensile properties of FS welding are shown in Table 2.9.

Table 2.9. Tensile properties of FS welded joints of AA5083-H321 alloy [Yazdipour et al. (2011)]

<table>
<thead>
<tr>
<th>Yield stress, N/mm²</th>
<th>Tensile strength, N/mm²</th>
<th>Elongation, %</th>
</tr>
</thead>
<tbody>
<tr>
<td>142</td>
<td>337</td>
<td>19.4</td>
</tr>
</tbody>
</table>

Koilraj et al. (2013) have conducted the dissimilar welding between AA5083 and AA2219 aluminium alloys.

Don-Hyun Choi et al. (2013) have conducted friction stir welding studies on AA5083 aluminium alloy plates. 4 mm thick plates were used in this study. The welding parameters used were tool rotational speed ranging between 700-1500rpm and welding speed of 100mm/min. They have observed that sound FSW joints were obtained without any defects under all welding conditions.

**AA5083-O and AA5083-H321 conditions (Al – 4.2 % Mg):**

Dickerson and Przydatek (2003) have analyzed the effect of flaws in FS welded joints of AA5083-O and AA5083-H321 aluminium alloys. Although the vast majority of FS welds will be free of flaws, it is not always possible to assume that they are. The properties of welds with flaws are needed to enhance confidence in the design and application of friction stir welded joints. They have concluded that root flaws in the 5083-O welds did not induce a significant loss in static or fatigue strength compared to
similar welds without root flaws. The larger root flaws in the 5083-H321 welds caused a reduction in the tensile strength and ductility when compared to welds without root flaws. The plate thickness is a key parameter to be controlled during FS welding to prevent or control root flaws.

2.2.3. METAL INERT GAS WELDING OF ALUMINIUM-MAGNESIUM ALLOYS

AA5456-H116 alloys (Al – 5.25 % Mg):

Moon and Metzbower (1983) have conducted MIG and LB welding of automotive aluminium alloy 5456 (Mg-5.25 wt. %, Mn-0.8 wt. % and Cr-0.1 %) on 12.7 mm plates. The tensile properties of MIG welded joint and the base metal is shown in Table 2.10. The weld joint efficiency based on yield stress was 62 %.

Table 2.10. Tensile properties of MIG welded joints of AA5456 alloy
[Month and Metzbower (1983)]

<table>
<thead>
<tr>
<th>Process</th>
<th>Yield Stress, MPa</th>
<th>Tensile strength, MPa</th>
<th>Elongation, %</th>
</tr>
</thead>
<tbody>
<tr>
<td>MIG welding</td>
<td>158</td>
<td>317</td>
<td>14.0</td>
</tr>
<tr>
<td>AA5456 (5.25 %Mg)</td>
<td>255</td>
<td>351</td>
<td>16.0</td>
</tr>
</tbody>
</table>

AA5083-H321 condition (Al – 4.2 % Mg):

Aluminium-Magnesium alloys offer weld yield strengths between 100 and 140 MPa during MIG welding [Christian and Murray (2006)], resulting in thicker structures and higher dead weight. Christian and Murray (2006) have conducted an extensive analysis of various strengthening mechanisms contribution to strength. The grain sizes at the base metal and MIG weld are shown in Table 2.11. The grain size in the MIG weld is larger than the grain size in the base metal. The tensile properties of the MIG welded joint and the base metal are shown in Table 2.12. The weld joint efficiency based on yield stress was 51 %.

Table 2.11. Grain sizes in MIG welded joints of AA5083-H321 alloy
[Christian and Murray (2006)]

<table>
<thead>
<tr>
<th>Location</th>
<th>Grain size, µm</th>
</tr>
</thead>
<tbody>
<tr>
<td>BM</td>
<td>32</td>
</tr>
<tr>
<td>Weld</td>
<td>65</td>
</tr>
</tbody>
</table>
Table 2.12. Tensile properties of MIG welded joints of AA5083-H321 alloy
[Christian and Murray (2006)]

<table>
<thead>
<tr>
<th>Process</th>
<th>Yield Stress, MPa</th>
<th>Tensile strength, MPa</th>
<th>Elongation, %</th>
</tr>
</thead>
<tbody>
<tr>
<td>MIG welding</td>
<td>117</td>
<td>259</td>
<td>10.8</td>
</tr>
<tr>
<td>AA5083-H321</td>
<td>228</td>
<td>317</td>
<td>16.0</td>
</tr>
</tbody>
</table>

Tabun and Kaluc (2007) have found out that 70 µm pores have been formed in the double-pass MIG welded joints of aluminium alloy 5086-H32. Prokic-Cvetkovic et al (2010) and Sandra Kastelec-Macura et al (2008) have observed similar pore formation during MIG welding of Al-4.5Mg-Mn alloy in the heat affected zone.

Yazdipour et al. (2011) have investigated the microstructures and properties of metal inert gas and friction stir welds in aluminium alloy 5083-H321. The results show that the extension of the heat affected zone is higher in the MIG welding than in FS welding. The weld microstructure of MIG specimens contains equiaxed dendrites as a result of solidification process during MIG welding while FS welded samples have wrought microstructures. The hardness distribution showed more loss of hardness in the MIG weld compared to the FS weld. The tensile properties of the MIG welded joint and the base metal are shown in Table 2.13. The MIG weld yield stress is 134 MPa.

Table 2.13. Tensile properties of MIG welded joints of AA5083-H321 alloy
[Yazdipour et al. (2011)]

<table>
<thead>
<tr>
<th>Process</th>
<th>Yield Stress, MPa</th>
<th>Tensile strength, MPa</th>
<th>Elongation, %</th>
</tr>
</thead>
<tbody>
<tr>
<td>MIG welding</td>
<td>134</td>
<td>311</td>
<td>14.0</td>
</tr>
</tbody>
</table>

In general, it is an accepted fact that finer grains could improve both hardness and strength of a metal according to the Hall-Petch relation. Therefore, higher strengths observed for FS welds could simply be related to their finer grains produced in the stir zone. It should be noted that the decrease in strength in both FS welded and MIG welded samples is inevitable. That is because, as mentioned before, 5083-H321 is a non-heat treatable alloy; its hardening mechanism is therefore by work hardening or refining its structure. Therefore, heating the base metal due to FS or MIG welding will modify the structure of the work-hardened base metal and it will decrease the strength. Moreover, it
should be noted that increasing the heat input to the weld region will increase the extension of the heat affected zone, and will further decrease the strength.

The AA5083 aluminium alloy used in the construction of high speed trains (HSTs) in China has been taken for analysis by Yao Liu et al (2012). Both MIG and TIG welds were applied over the material. The welding was done with the help of suitable TIG weld wires. They have concluded that MIG welded joints contained micro porosities of average sizes of 77 µm.

2.2.4. TUNGSTEN INERT GAS WELDING OF ALUMINIUM-MAGNESIUM ALLOYS

Pronounced loss of alloying elements and the resulting changes in the weld composition during TIG welding have been well documented. Previous emission spectroscopic investigations indicated that during TIG welding of steels, iron and manganese have evaporated. The magnesium loss phenomenon was demonstrated by Dudas and Collins (1966). Loss of volatile alloying elements, such as magnesium, due to selective vaporization is a common occurrence in the LB welding of aluminium alloys [Moon and Metzbower (1983)]. 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.

Prachya and Anucha (2012) and Arun Narayanan et al.(2013) have investigated the effect of shielding gas on porosity formation of AlMg4.5Mn weld metals.

The AA5083 aluminium alloy used in the construction of high speed trains (HSTs) in China has been taken for analysis by Yao Liu et al (2012). They have found out that TIG welded joints were pore-free. They have recommended that for HSTs TIG welded joints compared to MIG welded joints.

In general, the severity of a number of weld defects can be reduced if the solidification structure is refined. Certain novel welding techniques like dc pulsed current welding and magnetic arc oscillation have been employed to improve hot cracking resistance and mechanical properties in the late seventies.

AC pulsed current (PC) welding resulted in substantial grain refinement and increase in yield strength of aluminium –lithium alloy 1441 weld metal. Significantly large amount of work was conducted on the issue of changing the fusion zone solidification
structure from epitaxial columnar grains to fine equiaxed grains.

From the literature [Tseng and Savage (1971) and Kumar and Sunderesan (2006), Kumar et al. (2007) and Kumar et al. (2008)], the most important process parameters which are having greater influence on the weld bead geometry and fusion zone grain refinement of pulsed welding process have been identified. They are: Peak current, and Pulse frequency [Kumar and Sunderesan (2009) and Balasubramanian et al. (2008a)] for Al-Mg alloys.

Apart from the mechanical considerations of joint design, the welding process, filler material, heat input, number of weld passes, etc., will influence the microstructure of the weld at the joint and in turn will influence the extent of heat affected zone and residual stresses that will build up in the base metal [Balasubramanian et al. (2008b)]. Weld fusion zones typically exhibit coarse columnar grains because of the prevailing thermal conditions during weld metal solidification. This often results inferior weld mechanical properties and poor resistance to hot cracking. While it is thus highly desirable to control solidification structure in welds, such control is often very difficult because of the high temperatures and higher thermal gradients in welds in relation to castings and the epitaxial nature of the growth process.

Senthil Kumar et al. (2007) have observed solidification cracking occurs when the thermal stresses that build up during freezing exceed the strength of the solidifying weld metal. Methods that are commonly used to reduce the tendency for solidification cracking include: altering weld metal composition, through the addition of a filler wire, close process control, and controlling the grain structure within the fusion zone. It is widely accepted that by changing the weld’s grain structure, from coarse columnar to fine equiaxed, better cohesion strength can be obtained, and the remaining eutectic liquid present during the final stages of solidification can be fed more easily and heal any cracks that may form may be healed [Kou and Le (1986) and Norman et al. (2003)].

In order to manufacture high quality welded joints, it is essential to keep the heat input as low as possible [Meran et al. (2004)]. This situation helps the formation of a small heat affected zone (HAZ) in the weld and helps to protect the welding zone against the surrounding medium. For this purpose, TIG welding is preferred in various welding applications, since its energy concentration is lower in the welding pool.
Among other things, the reliability of welded structures depends upon the presence of fabrication defects and flaws developed during service as a result of degradation processes such as pitting corrosion and stress corrosion [Madhusudhan Reddy et al. (2001)].

During welding, the weldment is locally heated by the arc; thus, the temperature distributions in the weldment are not uniform. Welding temperatures also vary, along with welding times. Typically, the weld metal and base metal near the fusion zone immediately adjoining it are at a temperature substantially above that of the unaffected base metal. During the welding cycle, non-uniform thermal strains are induced in both the weld metal and adjacent base metal. The thermal strains produced during heating are accompanied by plastic upsetting. The non-uniform thermal stresses resulting from these strains combine to produce internal forces that cause welding distortion. It was found that the welding distortions can affect the fabrication, precision (shape and dimensional tolerance), and function (reliability and stability) of the finished structures.

Fine equiaxed grain structures in welds have better resistance to hot cracking. The development of grain refinement procedures is therefore an important topic in current welding research. The grain structure of normal TIG welds depends on both welding conditions and alloy contents [Pearce and Kerr (1981)]. Recently, it has been shown that transition to more refined structures in TIG welds of aluminium alloys are caused mainly by Ti and/or Zr providing heterogeneous nuclei. This is consistence with the well-known effects of Ti and Zr on grain refinement in Al-base castings.

For the weld metal zone, generally, the columnar-to-equiaxed grain transition (CET) may reduce solidification cracking and brittle fracture and improve mechanical properties of the welded joints. This is because equiaxed grains accommodate strains more uniformly or permit easier transport of liquid between grains. However, unlike in castings, the natural alloy system the morphology of the solidification structure is controlled by the solidification parameters--- the solidification growth rate R and the thermal gradient in the liquid GL. That is to say, the ratio of the two parameters GL/R changes from a maximum value at the fusion boundary to a minimum along the center of the weld. These changing solidification conditions result in a weld solidification structure
changing from planar at the weld boundary to columnar dendrite and then to equiaxed dendrite grain along the weld center.

In a recent study, Kou and Le (1985) have discovered a new effective mechanism for reducing solidification cracking in the autogenously GTA welds (i.e., gas-tungsten arc welds made without filler metals) of 2014 aluminium alloy. The effectiveness of this mechanism, i.e., alternating grain orientation, was most pronounced in welds made with transverse arc oscillation of low frequency and high amplitude, and solidification cracking was reduced significantly in these welds. This mechanism was effective due to the fact that solidification cracking is Intergranular and that columnar grains which reverse their orientation at regular intervals force the crack to change its direction periodically, thus making crack propagation difficult.

In general, approaching from the base metal toward fusion boundary coarsening of aluminium grains was noticed while aluminium grains in the weld metal were finer than that of the base metal, HAZ and fusion boundary [Rajesh Mandi et al. (2008)]. Second phase particles and other intermetallic compounds were found in comparatively lesser amount as probably they get dissolved in aluminium matrix in a region close to the fusion boundary. This dissolution is termed as reversion in welding metallurgy.

Precipitates and second phase particles act as barrier to the movement of dislocation which in turn increases hardness and strength.

Yao Liu et al. (2012) have investigated the mechanical properties and microstructure of aluminium 5083 weldment by TIG and MIG welding. Weldment produced by both is mechanically softer than the base metal. It is revealed that AA5083 weldment processed by TIG is mechanically more reliable than those by MIG welding.

2.2.5. Nd:YAG LASER BEAM WELDING OF ALUMINIUM-MAGNESIUM ALLOYS

Zhao et al. (1999) have investigated the problems in automotive aluminium alloys laser welding. Porosity is a common problem in LB welded aluminium alloys. The detrimental effect of porosity on mechanical properties of aluminium welds has been documented in the literature. The effects of weld porosity on the tensile and bend test performances of AA5086-H116 alloy welded with 5356 electrodes were studied and they reported that weld porosity is detrimental to the static tensile properties and bend ductility
of the welds. The elongation can be reduced by 50% from its highest level as the porosity level is increased to 4 VPP (volume percent porosity). The yield strength is reduced only slightly by porosity levels up to about 4 VPP. The tensile strength is unaffected by a small amount of porosity, but drops below 241 MPa when the porosity is higher than 3.6 VPP. It is also noted that when porosity level is higher than 2.5 VPP, the reduction in tensile strength is more severe; this may exceed the effect caused by reduction in cross-sectional area due to porosity.

Pastor et al. (1999) have investigated the influence of various welding parameters on porosity and under fill formation and magnesium loss during continuous wave Nd: YAG laser beam welding of thin plates of aluminium-magnesium alloys 5182 (4.44 wt.%) and 5754 (2.82 wt.%). The experimental results showed that the instability of the keyhole was the dominant cause of macro-porosity formation during laser welding of thin plates of aluminium alloys 5182 and 5754.

The applicability of CO₂ and Nd: YAG laser beam welding of 5754-O aluminium alloy for automotive applications is reviewed by Cao et al. (2003 a and b). They have also reported an exhaustive review of LB welding and their parameters influence on the microstructure and mechanical properties. In this review, the research and progress in laser welding of wrought Al alloys have been critically discussed from different perspectives. The primary objective of this review is to understand the influence of welding processes on joint quality and to build up the science base of laser welding for the reliable production of Al alloy joints. Two main types of industrial lasers, carbon dioxide and Neodymium-doped yttrium aluminium garnet are applied. But special attention is paid to Nd: YAG laser welding of 5000 and 6000 series alloys in the keyhole mode.

Punkari et al. (2003) have investigated the effects of magnesium content on dual beam Nd: YAG laser welding of Al-Mg alloys. Bead-on-plate welds on 1.6 mm thick AA1100, AA5754 (3.2 wt-% Mg) and AA5182 (4.6 wt-% Mg) were done. Hardness profile in the 5754 and 5182 welds showed a gradual increase in hardness from the base metal values through the heat affected zone to a peak in hardness in the weld metal adjacent to the fusion boundary. It was suggested that this increase in hardness may be the result of the
presence of $\text{Al}_3\text{Mg}_2$ or metastable $\text{Al}_3\text{Mg}_2$ precipitates in this region of the weld and HAZ.

Ion (2000) has reviewed the laser beam weldability of wrought aluminium alloys by CO$_2$ and Nd: YAG welding. The necessity for a minimum gas flow rate to prevent plasma shielding has been demonstrated. Since the ionization potential of aluminium (6 eV) is lower than that of steel (7.8 eV), plasma formation is more likely to occur, and appropriate measures must be taken to deal with the problem. In theory, helium is the most suitable gas for plasma suppression because its high ionization potential (24.5 eV) limits plasma formation, its high thermal conductivity maintains the transparency of the plasma, and its low molecular weight aids recombination of electrons and ions in the plasma. However, its low density dictates that high flow rates are required, which may produce a cooling effect and disrupt the weld pool.

Haboudou et al. (2003) have investigated the reduction of porosity content during Nd: YAG laser welding of A356 and AA5083 aluminium alloys. Porosity formation is greatly influenced in aluminium alloys by the low vaporization point element (Mg, Zn) content, or by process instability such as key-hole closures that tend to entrap occluded gases during welding. Another important contribution comes from the hydrogen content, because of a very high solubility in molten aluminium that favors micro porosity generation. It was concluded that surface preparation as well as dual beam welding are adequate methods for reducing porosity formation tendency in laser assemblies. At last, the metallurgical modifications (microstructure, micro hardness, compositional change) induced by laser welding will be analyzed and discussed.

AA5083 is a wrought aluminium alloy delivered in 4 mm-thick plates in an O-annealed state. Like most of the 5000 alloys, 5083 is primarily strengthened by a solid solution of Mg in the aluminium matrix. Thus, there is a quasi-linear dependence between the tensile properties (the yield stress, $\sigma_Y$) and the magnesium content (Refer Fig.2.1). In our case, after annealing treatment, grain sizes are nearly 50-100 µm, and the base metal hardness is about 85 Hv. The two alloys have a low solidification cracking susceptibility, mainly due to rather narrow temperature range between liquidus and solidus (50 ± 5K) on Al-Mg and Al-Si diagrams.
In both cases, the laser beam was focused at the very surface of the plates, even if many previous studies have evidenced the beneficial effects of a negative defocusing on the penetration depths, and of a positive defocusing on the porosity content.

Welding speeds of 3 and 5 m/min were considered in single beam mode. Cross-sectional analysis of 3 m/min weld beads indicates a 3.5 mm penetration depth, and a 3 mm width, whereas 5 m/min laser beads reveal a 2.5-3 mm penetration depth and a 2.5 mm width, all along the 100 mm length.

Morphology and localization of pores: Two kinds of porosity are generated during YAG laser welding: the first one are micro porosity (50-200 µm), which are mostly ascribed to hydrogen solubility in aluminium, and even more in Al-Mg alloys. The second kind may be called macro cavities. Their shape is less circular than hydrogen occluded pores, and their sizes range between 300 µm and 600 µm. These pores are not attributed to hydrogen rejection because the high cooling rate during laser welding could not allow pores to reach such an important size. They are mostly due to key-hole closure or shrinking and process instabilities, especially on Al-Mg alloys. Mg oxides are revealed on the porosity walls, which confirm that Mg bubbles have been entrapped in the molten pool, due to selective boiling of Mg above 1090°C. These cavities are much less circular than micro porosity and have certainly been initiated at the key-hole root, which is a common occurrence during laser welding in key-hole mode. They are localized mostly at the edges and at the root of the bead, due to convections movements of liquid aluminium going up from the bottom of the melt zone and following the liquid-surface interface towards the surface.

Porosity contents are lower at 5 m/min than at 3 m/min (nearly 3-4% pore ratio). This may be ascribed to the formation of a deeper and more instable keyhole in low welding speed conditions, driving to a larger number of occluded pores.

The major modification evidenced in 5083 weld beads is a nearly + 10 % hardening of weld zones, which is due to a very fine dendrite structure, especially in the equiaxed zone, despite a 0.5% Mg loss, that tends to decrease mechanical properties.

Joseph (2010) have investigated the maximum absorptivity of AA5083 aluminium alloy using 3.5 kW CO₂ laser beam welding machine. A 4 mm thick 5083 aluminium alloy plate was used in his study. He found out that a maximum of 12 % of the laser
energy was absorbed. The remaining part of the laser energy would have been re-absorbed into the plasma region and others lost as heat to coolest parts of the base metal, which is normally the boundary region. The poor absorption of laser energy by aluminium is found to be due to the high reflectivity of aluminium alloys.

2.2.6. CO₂ LASER BEAM WELDING OF ALUMINIUM-MAGNESIUM ALLOYS

Although several studies have reported many advantages in using laser welding, there still exists the fundamental problem that the welding reliability of aluminium alloys is lower than that of other industrial metals, such as steels, due to its higher reflectivity, higher thermal conductivity and lower viscosity. In fact, the thermal conductivity of the aluminium alloys is about one order of magnitude higher than that of steels. Thus, the thermal conductivity of low carbon steel is ~14 W m⁻¹ K⁻¹, being 143 W m⁻¹ K⁻¹ for the aluminium alloy AA5083 [Sanchez-Amaya et al. (2009)].

In general, the absorption efficiency for CO₂ laser with 10.6 µm wavelengths by most metallic materials was quite low, for example, the absorption portion by Al was only 2 % at room temperature and less than 4 % at 1000°C. However, once the input laser power density reaches a critical value (~ 10⁴ W mm⁻² for metals), plasma will be induced and the work piece absorption rate rises significantly [Fritsch (1987)]. At even higher power input, a lower absorption will again result due to the strong plasma shielding. Aluminium alloys are characterized by their low ionization energy, resulting in a high tendency to form a plasma plume above the work piece. This effect will be beneficial to the initial stage, but detrimental to later stages (once a stable key-hole is established) due to the plasma screening effect [Lee et al (1996)].

Owing to the high fluidity and high vapor pressure of Al alloys, the key-hole during welding is less stable compared with those in steels, resulting in the tendency to form pores, splashes, fluctuation of weld depth, and unevenness of the weld top-surface. Generally, gas pores were found to be present in most cases.

High-power density at the work piece is crucial to achieve keyhole welding and control the formation of welds. Compared with steel, aluminium alloys have higher initial reflectivity to laser beams and greater thermal conductivity, thus greater power densities are needed. If the power density is too low, the coupling of laser energy to work piece and penetration may be lost [Riches and Ion (2000)]. Power density that is too high may
cause spatter, undercut, under fill, and drop-out [Dawes (1992)]. In practice, it is generally advisable that power density should not exceed $10^7$ W/cm$^2$ to avoid heavy ejection of molten material. The laser power should be set according to the material and its thickness. Thicker material will require greater beam intensity at a set welding speed because an increase in laser power yields a proportional increase in penetration depth [Cao et al (2003a)]. For a set material thickness, higher laser power will increase welding speed. Higher power has also been reported to enlarge the operating window [Steen (1998)]. However, more laser-induced plasma or plume can be emanated in deep penetration welding with the increase of laser power.

At the same output power, smaller spot-size means higher power density but the welds may become narrower than necessary or even not fully fused. Larger weld seams are usually less than one-quarter the width of a TIG weld for the same material thickness. A small spot size may also lead to more loss of elements by vaporization, causing undercut and under fill defects resulting from too high a power density [Dawes (1992)].

The two fundamental modes of laser welding are: a) conduction welding and b) keyhole or penetration welding. The basic difference between these two modes is that the surface of the weld pool remains unbroken during conduction welding and opens up to allow the laser beam to enter the melt pool during keyhole welding. Conduction welding offers less perturbation to the system because laser irradiation does not penetrate into the material being welded. As a result, conduction welds are less susceptible to gas entrapment (porosity) during welding. With keyhole welding, intermittent closure of the keyhole can result in porosity [Duley, 1999]. Since maximum penetration depth is desired, keyhole welding is the method used for this study. Usually, since aluminium alloys have high reflectivity and thermal diffusivity, it is difficult to generate a keyhole in laser penetration welding. This must be overcome by sufficient laser power and proper focusing, in order to achieve the required power density. Welding in keyhole mode resulted in a much larger weld pool and less pronounced compositional change compared to conduction-mode.

The mechanical properties of LB welded joints are mainly depends on defects and metallurgical changes during welding. When the joints are defect-free variations in alloy composition is the reason for inferior tensile properties.
Moon and Metzbower (1983) have conducted LB welding of automotive aluminium alloy 5456 (Mg-5.25 wt. %, Mn-0.8 wt. % and Cr-0.1 %) on 12.7 mm plates. The redistribution of solutes and precipitates in the fusion zone was investigated and correlated with the enhanced toughness of the laser beam weldment. Metallographic examinations indicated both substantial purification and refined grain structures in the fusion zone. The tensile properties of LB welded joint and the base metal are shown in Table 2.14. The weld joint efficiency based on yield stress was 77 %.

<table>
<thead>
<tr>
<th>Process</th>
<th>Yield stress, MPa</th>
<th>Tensile strength, MPa</th>
<th>Elongation, %</th>
</tr>
</thead>
<tbody>
<tr>
<td>LB welding</td>
<td>193</td>
<td>289</td>
<td>9.2</td>
</tr>
<tr>
<td>AA5456-H116</td>
<td>255</td>
<td>351</td>
<td>16.0</td>
</tr>
</tbody>
</table>

In addition to the fine structure, the fusion zone became remarkably cleaner compared to the base metal in terms of inclusions. Only 5 % of the insoluble precipitates are retained in the fusion area after laser beam welding.

It is relatively well established that porosity diminishes the tensile strength of a metal. It has been reported that the effects of porosity on Al-5000 series and A-6000 series alloy welds and concluded that weld porosity was detrimental to the tensile strength as well as the ductility of the weld.

The phenomenon of evaporation within the keyhole during laser beam welding of aluminium alloys A5052 and A5083 was investigated by Xijing et al. (1997) under different welding conditions. The character of the molten pool was compared and analyzed. It was found that the evaporation of the main alloying element for these alloys, magnesium, greatly influenced the reaction force induced between the metal vapour and thermal plasma, which in turn affected the degree of penetration. The results of these experiments also confirmed that increasing shielding gas flow rate, within a limit, and a slight increase in the entrance angle of the laser beam improved meltability and increased penetration depth. Surface preparation was also observed to improve beam absorption and increase penetration.

The loss of magnesium due to laser beam welding of Alloys A5052 and A5083 was 0.48 % and 0.74 %, respectively. The results indicates that the level of alloying content
with the base metal has a direct effect on the amount of evaporation within the weld, that is, the loss of Mg was related to the Mg content within the alloys. Alloy A5083, which contained a greater amount of Mg, displayed greater evaporative losses.

Venkat et al (1997) have mentioned that power densities of at least $2 \times 10^6$ W/cm$^2$ are required for keyhole mode welding of Al-Mg and Al-Mg-Si alloys. That work also reported that threshold power densities of various aluminium alloys depend on the alloy content. The presence of increased amounts of magnesium or zinc tends to reduce the minimum power density required for keyhole mode welding. The hardness profile across the fusion zone of a 5754-O laser weld made at 3 kW has shown a slight increase across the HAZ and fusion zone relative to the base metal. The average knoop hardness number for the base metal is 73, while that for the fusion zone is 79.

Ion (2000) have reviewed laser beam welding of wrought aluminium alloys applicable to various industries. He has also mentioned in his review that laser beam welding of 5083-O (annealed conditioned) has attracted much attention, because of its widespread use in welded constructions. Sections of 6 mm in thickness have been welded autogenously at 6.5 m/min using a power of 10 kW. The weld tensile strength values equal to the base material value of 270 MPa have been obtained.

Ancona et al (2007) have performed welding trails on AA5083 alloy by using a 2.5 kW CO$_2$ laser source. Ancona et al (2007) have conducted the weld quality studies on AA5083 (annealed condition) aluminium alloy with thickness 3 mm. The welding parameters selected were ranging between 2 kW-2.5 kW and welding speeds between 3.6 m/min – 7.2 m/min. They have concluded that for the maximum power (2.5 kW) and welding speed the joint efficiency achieved was slightly more than 80 %. Formation of porosity and penetration of the weld played an important role in the enhancement of tensile strength of the AA5083 butt-welds.

This is probably due to the fact that laser welding of aluminium alloys is far from being a reliable technology, owing to the low coupling efficiency of the beam energy to the material because of its reflectivity at the laser wavelength, to the loss of volatile alloying elements in the joints and to the significant incidence of the porosity and hot cracking [Sibilano et al. (2006)]. Furthermore, since aluminium reacts very easily with active gases, like oxygen, composing ambient air, the role of the shielding gas in
aluminium alloys laser welding is very important because it prevents the formation of oxides that can affect the metallurgical properties of the weld. 2 mm thick AA5083 aluminium alloy plates were used in this experiment.

The use of aluminium alloys is continuously growing in several industrial sectors due to their low weight, good mechanical strength, and high corrosion resistance properties. The welding of these alloys is normally carried-out by using TIG and MIG processes but the results are not good enough for thin sheets because of the high thermal gradient inducing high distortion on final components and low characteristics of the welded joints. The laser welding represents a valid alternative to the above welding processes thanks to the low thermal fluxes, normally concentrated in the welding zone, high penetration and high processing speed. However, the achievement of high quality welded joints is difficult if the process is not accurately designed and controlled. Several factors influencing the final product quality related to the power beam, travelling speed, alloy characteristics, laser-matter interactions, etc.

Kuo and Lin (2006) have investigated the effects of shielding gas and power waveform on Nd: YAG laser welding of AA5754-O aluminium alloy. A part of the plasma cloud is ejected from the keyhole as a result of ionization of the metal vapor by the laser beam. A high metal vapor pressure is required to maintain the keyhole, and the presence of plasma is thought to be one of the characteristics of keyhole mode welding. Therefore, when the keyhole is formed, more laser beam energy is absorbed due to multiple reflections within the keyhole. Consequently, the penetration depth increases, e.g. during the tp stage of the PW mode and over the entire cycle of the CW mode. If the plasma cloud is not generated, this implies that the power density is insufficiently high to cause the metal vapor to form a keyhole. The results indicate that the average hardness of the weld metal is higher than that of the base metal. Similar results were also found by Venkat et al (1997), and Ramasamy and Albright (2000) in the Nd: YAG laser welding of AA5754 aluminium alloy. Hence, the higher hardness of the weld bead appears to be a result of the microstructure refining effect prompted by the higher cooling rate of laser welding. Mg loss in weld- 0.42 wt % (base metal-Mg-2.9 wt %). Magnesium loss reduces the hardening quality of the solid solution, with the result that both the tensile strength and the ductility decrease.
El-Batahgy and Kutsuna (2009) have investigated the LB welding of AA5052, AA5083 and AA 6061 aluminium alloys. Two mm thick plates were used in their study and were in H12 condition. The laser incident power used was ranging between 3 kW to 5 kW. The welding speed varied between 3 m/min to 6 m/min. They have used a backing strip below the plates to be welded. It is obvious that using a backing strip resulted in remarkable improvement in weld bead shape, where it became convex, and the recommended and acceptable shape concerning hot cracking. Another important finding in their research was the absence of weld metal solidification cracking. 0.2 % proof stress and tensile strength of both 5052 and 5083 welded joints were reduced to 87 %, while their elongation was reduced to about 75 % of those of the base metal where all the samples failed in the weld metal.

The metallographic observation and the micro hardness values of the processed samples have not shown evidence of heat affected zone between the base metal and the weld beads. In general terms, it can be observed that for all the systems, the Hv values taken in the weld beads are higher than those measured in the base metal. It can be seen that the difference between these zones are ~ 10 Hv, most bead data belonging to the interval between 85 and 100 Hv, while most metal base data range between 75 and 90 Hv.

In the recent literature, it has been observed that the weld beads of 5xxx alloys can have lower magnesium content than the metal base. Obviously, the magnesium composition of the weld depends on the extent of the magnesium evaporation, which is closely related to the processing conditions. The magnesium depletion of weld beads generated under different laser keyhole welding conditions has been measured in literature for different Al-Mg alloys. Thus, weld beads of AA5754 samples ranged between 2.32 and 2.55 wt. %, depending on the welding conditions, the base metal being 2.9 wt. % [Kuo and Lin (2006)]. This means that the magnesium loss during the laser welding under keyhole regime ranged between 12 and 20 %. In AA6061 samples, a magnesium loss of ~ 20 % has been also observed in the weld fusion zone, with respect to the base metal, while in AA5083 samples, a loss of magnesium between 13 and 22 % in the middle of the keyhole is reported [Sibillano et al. (2006)].
Joseph (2010) have investigated the maximum absorptivity of AA5083 aluminium alloy using 3.5 kW CO₂ laser beam welding machine. A 4 mm thick 5083 aluminium alloy plate was used in his study. He found out that a maximum of 12 % of the laser energy was absorbed. The remaining part of the laser energy would have been re-absorbed into the plasma region and others lost as heat to coolest parts of the base metal, which is normally the boundary region. The poor absorption of laser energy by aluminium is found to be due to the high reflectivity of aluminium alloys.

Jun Shen et al (2013) have studied the effect of welding speed on the microstructures and mechanical properties of laser welded AZ61 magnesium alloy joints, welded using 3.0 kW continuous wave CO₂ laser beam without a filler wire. The β-phase Al₃Mg₂ precipitates were found to be reduced, when the welding speed was increased. It has been concluded that the grain refine is the main reason for the increase in hardness, but not the amount of β-phase Al₃Mg₂ precipitates.

2.2.7. METAL INERT GAS WELDING OF WROUGHT ALUMINIUM-MAGNESIUM-SCANDIUM ALLOY

Jiang Feng et al (2005) have conducted the MIG welding of Al-Mg-(Sc, Zr) alloy plates with Al-Mg and Al-Mg-Sc filler wires. They have concluded that the yield stress of the plate welded with Al-Mg-Sc filler wire was 100 MPa greater than that welded joint welded with Al-Mg filler wire.

2.2.8. FRICTION STIR WELDING OF WROUGHT ALUMINIUM-MAGNESIUM-SCANDIUM ALLOY

Blanka Lenkowitzki et al (2000) have done extensive work on FSW of C557 alloy (wrought Al-Mg-Sc alloy) and reported the advantage of FSW joints. The fuselage share of the structural weight of a commercial aircraft is extremely high. A reduction in weight together with improving resistance to corrosion decisively influence the direct operating costs of fixed-route services for airlines and hence the cost effectiveness. An additional advantage is yielded by reducing the manufacturing costs of a welded structure. These requirements led Dasa Airbus to develop a welded, integral fuselage construction. However, the prerequisite for this is the availability of the appropriate weldable materials, such as the Aluminim-Magnesium-Scandium (Al-Mg-Sc) alloys, with the necessary technological properties and efficient welding methods. The extremely favorable
properties of the Al-Mg-Sc alloys were exploited in the former Soviet Union for use in welded structures (welded housings, tanks) in spaceflight. This led to a reduction in weight together with an increase in efficiency.

Lapasset et al (2003) have investigated the tensile properties of Wrought Al-Mg-Sc alloy plates. As the global joint failed outside the weld, they have prepared component part joints and found out the strength of the FS weld.

Koteswara Rao et al (2005) have investigated the effect of scandium additions to 2219 aluminium alloy weld metal. They have concluded that Sc levels above 0.37 % produced better tensile properties compared to Sc levels of 0.16 %.

The microstructure and mechanical properties of fusion zones of medium strength Al-Zn-Mg alloy (RDE-40) welds were investigated by Dev et al. (2007). They have conducted the welding trials with the help of different fillers containing various amount of scandium as well as commercial AA5556 filler. They have concluded that addition of scandium above the eutectic level (0.55 wt. %) produced significant grain refinement in the fusion zone. Improvement in tensile properties and reduction of solidification cracking were obtained with Sc contained fillers. An increase of 50 MPa was observed in the yield stress values of the welded joint due to scandium addition in the filler rod.

The microstructure and mechanical properties of FSW and A-TIG for wrought Al-4.5Mg-0.26Sc heat-treatable alloy was investigated by Munoz et al (2008). Emphasis is laid on hardening precipitates in the weld areas. The effect of a post-weld heat treatment on both microstructure and mechanical properties is further examined. The results suggest that hardening precipitates are comparatively more affected by the TIG than by the FS welding process. This results in a substantial reduction in mechanical properties of TIG welds that can be partially recovered through a post-weld heat treatment.

Munoz et al. (2008) have conducted the welding experiments on 4 mm thick rolled plates of Al-4.5Mg-0.26Sc in H116 condition. The base metal contained coherent Al$_3$Sc particles of size 5 to 15 nm. The incoherent Al$_3$Sc inclusions were also found with a bigger size of 30 nm. The base metal hardness was 130 Hv. The decrease of hardness was observed as 27 Hv and 60 Hv respectively in the FS and TIG welded joints. After annealing, the hardness values of TIG welded joint has approached that of the base metal. Both YS and TS are affected by welding operations but more significantly for TIG
process. The decrease of YS between FS joint and BM is about 20 %, whereas it is almost 50 % between TIG and BM. But post-weld treatment increased the value of YS. The tensile properties of A-TIG, FS welded joints and the wrought base metal is tabulated in Table 2.15.

Table 2.15. Tensile properties of A-TIG, FS welded joints and wrought Al-4.2Mg-0.3Sc alloy [Munoz et al. (2008)]

<table>
<thead>
<tr>
<th>Process</th>
<th>Yield Stress, MPa</th>
<th>Tensile strength, MPa</th>
</tr>
</thead>
<tbody>
<tr>
<td>A-TIG welding</td>
<td>145</td>
<td>250</td>
</tr>
<tr>
<td>FS welding</td>
<td>235</td>
<td>300</td>
</tr>
<tr>
<td>Wrought Al-4.2Mg-0.3Sc alloy</td>
<td>290</td>
<td>360</td>
</tr>
</tbody>
</table>

The important softening occurred in the TIG welded zone is due to a loss strain hardening associated with fusion and above all to the absence of Al₃Sc precipitates because of their dissolution during the melting phase. This dissolution was expected because of the high temperatures attained during TIG welding.

The long columnar grains present in the fusion zone have to be replaced into equiaxed grains and the size of the grains has to be reduced. This is achieved by creating and maintaining free nuclei in the molten weld pool by addition of nucleants to the pool. This is achieved by the addition of Sc into the weld metal. Scandium is one of the most effective modifiers of a cast grain structure in aluminium alloys.

Juan Zhao et al (2010) have conducted the comparative investigation of TIG and FS welding of wrought Al-Mg-Sc alloy plates. They have conducted their experiment on (Russian alloy 1570 equivalent) an alloy containing 6 % - Mg and 0.4 % - Sc + Zr with a filler wire of similar composition. The tensile properties of TIG and FS welded joints and the wrought base metal are tabulated in Table 2.16.

Table 2.16. Tensile properties of TIG, FS welded joints and wrought Al-6Mg-Sc+Zr alloy [Juan Zhao et al (2010)]

<table>
<thead>
<tr>
<th>Process</th>
<th>Yield Stress, MPa</th>
<th>Tensile strength, MPa</th>
</tr>
</thead>
<tbody>
<tr>
<td>TIG welding</td>
<td>217</td>
<td>332</td>
</tr>
<tr>
<td>FS welding</td>
<td>286</td>
<td>394</td>
</tr>
<tr>
<td>Wrought Al-6Mg-Sc+Zr alloy</td>
<td>320</td>
<td>430</td>
</tr>
</tbody>
</table>
This indicates that the yield and tensile strength of FS welded joint are 19 – 31 %
greater than those of TIG welded joints. The minimum micro hardness was observed on
the advancing side and the minimum was higher by 30 Hv in the TIG welded joint weld
metal. The reason for less strength properties in TIG welded joints are non-retention of
cold-worked microstructure in the weld and dissolution of Sc, Zr elements in the Al
supersaturated solid solution, with only a few of Al₃(Sc,Zr) particles precipitated in the
welded joint, therefore, the precipitation strengthening of Al₃(Sc,Zr) particles is weak in
the TIG welded joints.

HE Zhen-bo et al (2011) have conducted the comparative investigation of TIG and FS
welding on wrought Al-5.8Mg-Sc alloy. They have used truncated cone type tool pin for
FS welding. The relationships between microstructures and mechanical properties of the
welded joints were investigated. The yield strength of Al-Mg-Mn-Sc-Zr alloy increases
by 60 % [YIN Zhi-min et al. (2000)] compared with conventional Al-Mg-Mn alloy.

The change of any kind of these strengthening mechanisms all can lead to the change
of strength and hardness of the welded joint. During FS welding, the temperatures of
WNZ, TMAZ and HAZ are 500-530°C, 480-500°C, and 450-490°C, respectively. The
tensile properties of TIG and FS welded joints and the wrought base metal are tabulated
in Table.2.17.

<table>
<thead>
<tr>
<th>Process</th>
<th>Yield Stress, MPa</th>
<th>Tensile strength, MPa</th>
<th>Elongation, %</th>
</tr>
</thead>
<tbody>
<tr>
<td>TIG welding</td>
<td>246</td>
<td>365</td>
<td>10.2</td>
</tr>
<tr>
<td>FS welding</td>
<td>257</td>
<td>391</td>
<td>10.9</td>
</tr>
<tr>
<td>Wrought Al-5.8Mg-Sc alloy</td>
<td>322</td>
<td>423</td>
<td>20.7</td>
</tr>
</tbody>
</table>

2.3. GAPS IN THE EXISTING LITERATURE

FS welding of AA5083 in annealed condition produced 100 % joint efficiency based
on yield stress, whereas FS welding of AA5083 in strain-hardened condition produced
joint efficiency less than 100 %. It has been noted down from literature that FS welding
joint efficiency for AA5083 in H321 condition was around 88 %.
The Al-Mg alloy, i.e., AA5083-H321 alloy when processed by autogenously MIG welding has produced yield stress values ranging between 117 to 158 MPa [Christian and Murray (2006), and Yazdipour et al. (2011)]. The weld joint efficiency based on yield stress was below 60 %.

In the case of laser welding of aluminium alloy, a maximum incident power of 2.5 kW was utilized by Ancona et al. (2007) for laser welding of aluminium alloy AA5083 used in marine applications. They have concluded that lower power produced lesser joint efficiency due to formation of micro porosity and other defects, whereas higher power (2.5 kW) laser produced better weld joint efficiency of around 80 %.

Much research work in India was done by adding Scandium in welding filler rod and by that process an enhancement of mechanical properties of welded joint was observed in the case TIG welding.

The latest solid state welding technique, viz., FS welding was performed on the cast Al-Mg-Sc plates and the characterization studies were conducted. The microstructure and mechanical properties of FS welded cast Al-Mg-Sc alloy plates are compared with FS welded wrought aluminium alloy AA5083-H321 plates.

2.4. PROBLEM FORMULATION

From the literature review, it is well understood that very less work has been carried out on welding of AA 5083–H321 aluminium alloy. Moreover, the joint efficiency of conventional fusion welded AA5083- H32 aluminium alloy joint was found to be only 45 to 60 %. Application of a solid state welding process namely friction stir welding can be beneficial in improving the joint efficiency of AA 5083 welded joints. Hence, this present investigation is aimed at to compare the fusion welding processes namely Tungsten Inert Gas Welding and Laser Beam Welding with a solid state process namely friction stir welding on the properties of AA 5083–H321. Further, the effect of Sc addition on the properties of Al-Mg alloys has not been studied systematically. Hence, the effect of scandium addition on the properties of friction stir welded cast Al-Mg-Sc alloy is also investigated.