Chapter 2  Literature review

2.1  Introduction

Review of available published literature on any specific topic of research is the first valuable step to be taken to get an idea about the present status of that particular topic. To get familiarity with the established methods & procedures of conducting work as well as to avoid duplication of work & wastage of precious time, this stage of investigation must be carried out as comprehensively as possible. In this chapter, an effort has been made to present a review of the available & relevant information related to the different aspects of the investigation covered in the present study. The literature survey has been carried on the following topics:

a) Quenched & tempered low alloy steels
b) Welding of Q&T low alloy steels
c) Metallurgical and mechanical characterization studies
d) Wear studies

2.2  Quenched and tempered low alloy steels

Iron is the fourth most abundant element and makes up more than five percent of the earth’s crust and exists naturally in the form of ore (haematite and magnetite). Through smelting, the iron is extracted in the form of spongy, porous mass of relatively pure form intermixed with slag, which once removed forms wrought iron and when extracted through blast furnace produces cast iron. However, wrought iron has a small percentage of carbon (0.02–0.08%), just enough to make it hard without losing its malleability, and cast iron in contrast, has excessive amount of carbon (3–4.5%), which makes it hard, but brittle and non-malleable. In between these is steel (alloy of iron and carbon), with
0.2 to 1.5 percent carbon, making it harder than wrought iron, yet malleable and flexible, unlike cast iron.

The industrial revolution in Europe established new industries and put an enormous pressure on the steel requirements by industries like steam engines and boilers from 1769, steel ships from 1787, spinning and weaving machines in series from 1820, railways from 1829 in England and 1835 in Germany (Berns et al., 2008). During this period the steel was being produced using cementation process which was costly and inefficient in controlling the carbon level in iron. However, the invention of the Bessemer converter revolutionizes the steel production both qualitatively and quantitatively.

The changeover from wrought iron to mild steel marked the start of alloying. Wrought iron contained small amounts of Mn and Si, occasionally Cu or Ni as well, from the respective ore. However, intentional alloying was not possible owing to incomplete melting. Alloying was not used for cast iron because the individual elements were not available as a ferroalloy at that time. This changed rapidly after the introduction of mild steel as illustrated by the following examples: manganese steel by R. Hadfield in 1888, case hardenable nickel steel by Krupp in 1888, roller bearing steel by R. Striebeck in 1901, high speed tool steel by F.W. Taylor in 1906 and stainless steel by Krupp in 1912 (Berns et al., 2008).

This initial period of work resulted in empirical relationships relating the effects of additions of alloying elements to the strength and toughness of the resulting alloy. These alloy additions violated the definition for mild steel summarized in 1969 by Duckworth and Baird as an iron and carbon alloy without further deliberate alloying with the exception of manganese for oxygen control and sulphur stabilization. However, the early alloy additions were of insufficient quantity to qualify the resulting
steel as an ‘alloy steel’ as this category required derivation of properties from the primary alloying element added. The improvement in strength and toughness was correctly attributed to refinement of the ferrite grains, but according to Woodhead, as empirical relationships only. The discovery in 1951 by Hall that the lower yield point, $\sigma_{LYP}$, for very low carbon steel was proportional to the grain size, $d$, by the relation:

$$\sigma_{LYP} = \sigma \alpha d^{-1/2}$$  \hspace{1cm} (2.1)

where, $\sigma$ is the yield stress for a single crystal was supplemented in 1953 by a similar discovery by Petch. It was found that the fracture stress of mild steel could also be related to the grain size and formulated the equation:

$$\sigma_{LYP} = \sigma_0 + k^* d^{-1/2}$$  \hspace{1cm} (2.2)

where, $\sigma_0$ and $k^*$ are constants. Heslop, working with Peth, made another impressive contribution in 1958, where the ductile to brittle transition temperature ($T_c$), was related to the ferrite grain size via:

$$T_c = A - B \ln d^{-1/2}$$  \hspace{1cm} (2.3)

where, $A$ and $B$ are constants. These three discoveries formed the fundamental science that explained the empirical effects of improved strength and toughness as a function of refining ferrite grains from small alloy additions. Thus in the early 1960’s the stage was set for an impressive growth in metallurgical knowledge and a systematic correlation of alloy contribution to mechanical performance via metallurgical control (Weng, 2009; Weng et al., 2011).

A revolutionary phase in the development of high strength steels came in 1950’s, when a new concept of microalloying was introduced, which enabled the metallurgists to achieve the yield strength in the range of 300 MPa–350 MPa. Small addition of titanium, vanadium and/or niobium resulted in fine grained structure due to strong hindrances created by the precipitation of niobium carbonitrides as well as
vanadium carbonitrides to austenite grain growth. This strengthening through grain refinement results in high strength, good toughness an excellent weldability of steels.

Further increase in strength level is achieved by using thermo mechanical controlled processing (TMCP) technique of rolling in which the plates are hot rolled in a specific temperature interval, where no recrystallization takes place between the passes. This results in elongated and thin grains of austenite consisting of deformation bands and twin boundaries. After the final rolling pass, the austenite to ferrite phase transformation during cooling results in fine grained ferritic microstructure. The presence of microalloying elements along with sufficient content of carbon and nitrogen, results in carbides and carbonitrides precipitation on sub-grains, grain boundaries, twin boundaries and other defects. The precipitation during recrystallization stop temperature (950 °C–1025 °C in steels for heavy plate rolling) resulted in retardation or complete annihilation of austenite grain growth, leading to extra fine grains of austenite (Hansson, 2004).

When even higher levels of strength accompanying high hardness are required, a further increase in alloying content in combination with TMCP processing is not a suitable production route, as it would lead to a high alloying content which is detrimental from the weldability point of view. Thus the next step in processing is to quench and temper the actual plate after rolling. The quenching and tempering of plates results in a microstructure consisting of tempered lath martensite, which shows high yield strength ranging from 700 MPa–1200 MPa, in combination with good toughness and an excellent through hardness.

This led to the beginning of new era in 1990’s, with the development of new quenched and tempered low alloy abrasion resistant steels possessing excellent wear resistance and strength accompanied with excellent toughness and weldability.
2.3 Welding of Q&T low alloy steels

Welding is the main fabrication technique employed for joining steels in all areas of applications. However the ability of steels for being welded depends upon the weldability of the steel employed, type of welding process, joint design selection and the type of filler material used. Thus, this section discusses the previous attempts made on all such issues related to Q&T low alloy steels.

2.3.1 Previous studies on the weldability of Q&T low alloy steels

No material, however, can be considered to be satisfactory for use in structural applications unless it can be fabricated into components. One of the prime considerations as to the fabricability of a steel is its weldability, which has been defined by the International Institute of Welding (IIW. Document no. IIW/IIS-22-59) as: “A metallic material is considered to be weldable to a certain degree by a given process and for a given purpose, when a continuous metallic connection can be obtained by welding using a suitable procedure, so that the joints comply with the requirements specified, both in regard to their local properties and their influence on the construction of which they form a part” (Chandrasekharai, 1995).

Not only must the steel be capable of being welded, but the resulting weld joints must have adequate joint integrity (strength, ductility, toughness, hardness etc.). In addition to joint integrity considerations, the weld joint must be sound and free from porosity and cracks, and other types of flaws such as undercut, incomplete penetration, and rollover, which might result in stress concentrations (Mishler et al., 1959). Thus the design of welding procedures is dominated by the need to incorporate appropriate safeguards against cracking which has been classified into following four types:

1. Lamellar tearing—restricted to plate steels.
2. Hydrogen cracking in both HAZ and weld metal.
3. Cracking occurring at the later stages of solidification of the weld—solidification cracking of weld metal and liquation cracking usually restricted to the HAZ in ferritic steels.

4. Cracking during PWHT—restricted to a limited range of steels and weld metals and only when subjected to PWHT (Bailey, 1993).

Among these, the hydrogen cracking is the most dominating problem occurring in the high strength quenched and tempered steels. Hydrogen level/content, stress level, type of microstructure and temperature are considered to be the key factors responsible for hydrogen cracking. Thus in order to achieve the optimum results in the weld metal and heat affected zones of welds in Q&T steels, stringent requirements exist for controlling certain factors, such as carbon equivalent (CE), heat input, preheat/interpass temperature/post weld heat treatment, welding process, filler material and weld geometry (Bailey, 1993). All these factors are being explained under appropriate headings.

2.3.1.1 Carbon equivalent

The susceptibility of steel to hydrogen embrittlement i.e. parent steel, weld metal or HAZ depends on the inherent toughness of the steel. The tougher the material, the greater is the content of hydrogen needed to embrittle it sufficiently to cause it to crack. The likelihood of hydrogen cracking in weld and HAZ increases as their hardness (hardenability) increases. Since the hardenability of the steel is governed by its composition, and a useful way of describing hardenability is to assess the total contribution to it of all the elements that are present. This is done by an empirical formula which defines a carbon equivalent (CE) value and takes account of the important elements which are known to affect hardenability. The most common of these is termed as the IIW carbon equivalent (Bailey, 1993):
\[ CE_{IIW} = C + \frac{Mn}{6} + \frac{(Ni+Cu)}{15} + \frac{(Cr+Mo+V)}{5} \]  

(2.4)

The amounts of the elements listed are in weight percent. Values of CE_{IIW}, below 0.42 denote steel which is easy to weld without hydrogen cracking, whereas, steels with CE_{IIW} values above 0.5 are difficult to weld.

Several other carbon equivalent formulae have been proposed from time to time. Most of these have not been extensively used, because of their unfamiliarity, their excessive complexity, or because they were so close to the formula CE_{IIW}. However, one formula developed in Japan for steels of low carbon content, whose behavior with regard to hydrogen cracking is not well described by the IIW formula, is the \( P_{cm} \) (parameter crack measurement) formula (Yurioka, 2001):

\[ P_{cm} = C + \frac{Si}{30} + \frac{(Mn+Cu+Cr)}{20} + \frac{Ni}{60} + \frac{Mo}{15} + \frac{V}{10} + 5B \]  

(2.5)

Again the elements are described by weight percent; however, this relation is valid for carbon less than 0.22%. Compared with the IIW formula, \( P_{cm} \) gives an increased importance to carbon and adds the microalloying element boron. The value of the boron factor is large, because boron is a light element only added to steels in small quantities.

There are two more equivalents that have relevance to modern high strength steels and are:

\[ CE_{PLS} = C + \frac{Si}{25} + \frac{Mn}{20} + \frac{Cu}{16} + \frac{Ni}{60} + \frac{Cr}{20} + \frac{Mo}{40} + \frac{V}{15} \]  

(2.6)

\[ CE_{HSLA} = C + \frac{Mn}{16} + \frac{Ni}{50} + \frac{Cr}{23} + \frac{Mo}{7} + \frac{Nb}{5} + \frac{V}{9} \]  

(2.7)

These equations represent relationships for the pipeline steel (CE_{PLS}) and high strength low alloy steel (CE_{HSLA}) equivalents respectively (Yurioka, 2001). The primary difference between the CE_{IIW} and the more modern equivalents is the emphasis placed
on carbon. Because the CE_IW is derived from higher carbon steels, the effect of the alloying elements is more profound than small fluctuations in the carbon content.

2.3.1.2 Heat input

All fusion welding processes involve heat flow during welding to accomplish the desired joint. Depending upon the heating and cooling cycles involved, different types of microstructures are obtained in weld bead and the HAZ. This leads to varying mechanical properties of different zones of a weldment. Apart from the metallurgical effects of heat flow in welding there are other phenomena involved, including distortion, residual stresses, physical changes and chemical modifications. Thus, to achieve a weldment of desired specifications to perform satisfactorily in service, it is essential to know the effects of heat during welding. This can well be achieved by knowing the temperature distribution during welding so as to determine the cooling rates in different directions with respect to the weld metal (Parmar, 1997).

The heat input used to make a weld is directly influenced by the welding process and procedure adopted during welding. As per the relation shown in Equation (1.1), heat input is a function of welding current, arc voltage and travel speed. To increase the heat input, either the welding current should be increased or the travel speed should be reduced. The voltage has a minor effect, because it varies only slightly, when compared with the other factors. Further the higher heat input reduces the cooling rate (Cary, 2001).

In Q&T steels, the heat input must be used with caution, because too high a heat input will tend to soften the HAZ, and its strength level will be reduced. In relatively low hardenability steels, it is possible to produce an unhardened HAZ by increasing the heat input. In higher hardenability steels, the tendency toward cracking and the
maximum hardness will be reduced by a slower cooling rate. Thus limits are imposed to the amount of heat input that can be used during welding (C. Chen & Pollack, 1993).

During the earlier days of research on steels, low heat input was considered to be 2.0 kJ/mm but today low heat input corresponds to the values of 0.5 kJ/mm or lower due to the rise in strength level of steels. Prior to welding a particular Q&T steel, recommendations should be obtained from steel manufacturer concerning its weldability, hardenability, and preheat and heat input limitations (Nevasmaa et al., 1992).

(Madhusudhan Reddy & Mohandas, 1996) investigated the effect of heat input (0.6 kJ/mm, 1.2 kJ/mm and 1.8 kJ/mm) on the width of the soft zone and, in turn, its influence on hardness and ballistic performance of quenched and tempered steels, welded using austenitic stainless steel filler (18Cr-8Ni-6Mn) with SMAW process. The results revealed that the width of the softened zone (HAZ) is a function of the heat input i.e. greater the heat input, wider the extent of softening in the inter-critical and sub-critical regions, which resulted in inferior ballistic performance of the welded joint.

(Eroğlu et al., 1999) studied the effect of initial coarse grain size with varying heat inputs on microstructure and mechanical properties of weld metal and HAZ. It was observed that the initial coarse grain size had a great influence on the microstructure, hardness and toughness of HAZ of low carbon steel. Thus, taking into consideration the plate thickness, a higher heat input should be used with respect to the maximum toughness of the HAZ in the welding of coarse grained low carbon steels.

(Loureiro, 2002) carried out an experimental study to investigate the effect of heat input on the plastic deformation of undermatched welds of high strength quenched and tempered steel (RQT701). A multipass welding of 25 mm thick steel, with K type groove joint, was laid down using submerged arc welding process at two heat input
levels of 2 kJ/mm and 5 kJ/mm. Microstructural variations and real stress-strain curves of specimens sampling the WM, the WM in conjunction with the HAZ and the three zones altogether (WM, HAZ and BM) were analyzed. It was concluded from the investigation that the increase of heat input coarsened the microstructure and diminished the hardness in the WM and HAZ. A loss of hardness was also observed in the subcritical zone probably due to carbide precipitation. The increase of heat input increased the yield and tensile strength undermatching of the WM and also produced HAZ undermatching, which induced a concentration of plastic flow in the weakest zone and resulted in the loss of strength and ductility of the weld loaded in tension.

(Basu & Raman, 2002) experimented to obtain bead-in-groove weld under iso-heat input conditions by submerged arc welding using quenched and tempered HSLA steel. Depending on the welding current and travel speed combination used, significantly different dependencies on all the influencing parameters were observed even though the heat input was same. This was attributed to the differences in the weld bead morphologies. Different weld bead morphologies led to different weld cooling rates that affected the microstructure by itself, and also different microstructural features, such as austenite grain size, inclusion parameters, which in turn, further contributed to the final acicular ferrite content within the weld metal.

(Yajiang et al., 2003) studied the effect of weld heat input on microstructural and toughness variations of heat affected zone of HQ-130 super-high strength steel plates. Test plates of 12 mm thickness were welded using GMAW process at varied heat input levels of 0.92 kJ/mm, 1.31 kJ/mm, 1.86 kJ/mm and 2.64 kJ/mm. The results revealed that the impact toughness of HAZ reduced with the increase in heat input, and a quasi-cleavage fracture with river like pattern features was found to be the mode of failure in all the fractured surfaced, characterized using SEM and TEM.
(Liu et al., 2007) carried out double thermal experiments on copper bearing steel to investigate the effect of heat input on the impact toughness and embrittlement behaviour of welded joints in the intercritical region of CGHAZ. The results revealed that the increased heat input resulted in poor toughness and high embrittlement susceptibility due to the presence of pearlite on the interface of original austenite and coarse granular bainite. Thus, during multilayer welding, it was proposed to strictly control heat input. Single thermal cycle experiments showed that the copper-bearing steel has a narrow range of heat input and embrittlement can easily occur in the region of CGHAZ with higher heat-input. Granular bainite transformed from austenite led to embrittlement, and the softening started when t_{8/5} time was more than 7s. The dissolution of e-Cu and coarse lath bainite and more ferrite caused the softening of CGHAZ.

(Shome, 2007) studied the effect heat input on the prior-austenite grain size in the heat-affected zones of single-pass butt welded joints with K type groove in thick HSLA-100 steel plates. The plates were welded using GMAW and SAW welding processes at two heat inputs of 1 kJ/mm and 4 kJ/mm. The grain size was related with the temperature history of the location, which was obtained using an existing thermal model based on Gaussian heat source distribution.

The results revealed that for a heat input of 1 kJ/mm, the grain size in the CGHAZ was 80 µm corresponding to a peak temperature of 1450 °C, but falls sharply to 40 µm within a distance of 0.5 mm from the fusion line. At higher heat input of 4 kJ/mm, the grain size was 130 µm for the same peak temperature, and the coarse grain heat-affected zone was 1.1 mm wide. It was further observed, that the presence of more acicular ferrite in the CGHAZ of SAW welded plates was responsible for the lower hardness as compared to GMAW welded HSLA-100 plates.
(Keshav Prasad & Dwivedi, 2006) investigated the influence of the submerged arc welding (SAW) process parameters on the microstructure, hardness, and toughness of HSLA steel weld joints to analyze the results on the basis of heat input. The SAW process was used for the welding of 16 mm thick HSLA steel plates. The weld joints were prepared using comparatively high heat input (3.0 kJ/mm–6.3 kJ/mm) by varying welding current (500 A–700 A) and welding speed (200 mm/min–300 mm/min). Results showed that the increase in heat input coarsened the grain structure both in the weld metal and HAZ. The hardness was found to vary from the weld centre line to the base metal, and the peak hardness was found in the HAZ. The hardness of the weld metal was largely uniform. The hardness reduced with the increase in welding current and reduction in welding speed (increasing heat input), while the toughness showed mixed trend. The increase in welding current from 500 A to 600 A at a given welding speed (200 mm/min or 300 mm/min), increased toughness and further increase in welding current up to 700 A lowered the toughness.

(Suh et al., 2011) studied the effect of welding heat input on the fatigue life of a quenched boron steel. Boron steel under quenched conditions forming a lap joint was welded using ferrite-bainite steel in the heat input ranging from 0.29 kJ/mm to 0.67 kJ/mm. Boron, which can increase hardenability, affected the microstructure and hardness of the weld metal and HAZ. The hardness of the weld metal and HAZ increased with decreasing welding heat input, and the high hardness of the weld metal and boron steel HAZ prevented the initiation of cracks in the stress concentration area around the bead. The bead width increased with increasing heat input, and the results of finite element method (FEM) showed that the maximum stress in the notch of the weld joint decreased when the bead width was increased i.e. the fatigue life increased when the weld joint had wider bead width. Finally, while the fatigue life was affected by the
residual stress, the variation of the welding heat input used in this study had hardly any effect on the residual stress distribution.

2.3.1.3 Preheat / Interpass temperature / Post weld heat treatment

During welding, rapid heating and cooling take place which produces severe thermal cycles near weld centre line region. These thermal cycles causes non uniform heating and cooling in the material, thus generating harder heat affected zone, residual stresses and cold cracking susceptibility in the weld metal and base metal (Kasuya et al., 1995).

To get rid of these problems some heat treatment before welding (preheating), post weld heat treatment (PWHT) after welding and maintaining inter-pass temperature during welding are employed. Effective preheat and post heat are the primary means by which acceptable heat affected zone properties and minimum potential for hydrogen induced cracking are created (Bailey, 1993).

Preheating is the process of heating metal to some predetermined temperature before engaging in actual welding. The details and the modes may be different in various situations, but in general, the purpose is to influence the cooling behaviour after welding, so that the shrinkage stresses will be lower (relative to welding without preheating) and cooling rate will be milder (Scott, 1998). The minimum preheating temperature to be assured to avoid cracking depends on carbon equivalent expressing carbon and alloy content, condition of base metal prior to welding, thickness of base material, constraint level and the hydrogen level (Bailey, 1994).

The functions of a PWHT are to temper the martensite in the weld metal and HAZ, in order to reduce the hardness and increase the toughness, and to decrease residual stresses associated with welding (Okabayashi & Kume, 1988). The literature revealed that the recommendations for PWHT are usually dependent upon specific
alloys and filler metals involved. The necessity for PWHT depends on material and service requirements, plate thickness, joint design, welding parameters and restraint.

(Kasuya et al., 1995) while searching the methods for predicting maximum hardness of heat affected zone and selecting necessary preheat temperature for steel welding, concluded that the hard microstructure of the HAZ is responsible for the property deterioration of weld and cold cracking susceptibility.

(Scott, 1998) discussed the fundamentals of preheat and concluded that: (a) preheat can minimize cracking; (b) preheat must be used as per the guidelines suggested and recommended by the manufacturers and applicable standards; (c) finally, the interpass temperature should be checked to verify that the minimum preheat temperature has been maintained just prior to the initiation of arc for each pass.

(Okabayashi & Kume, 1988) while studying the preheating and post weld heating suitable for avoiding the heat-affected zone cracking in 9Cr-1Mo-Nb-V steel concluded that: (1) minimum preheating temperature to prevent the cracking is about 200 °C, which is lower than the values (over 300 °C) estimated from the steel chemistry; (2) The reheating temperature, however, becomes higher when the weld is made with an electrode of mild steel of 569 MPa tensile steel; (3) use of a martensitic weld metal lowers the residual stress and results in low susceptibility to cracking.

(Ghosh et al., 2004) while studying the influence of pre and post weld heating on weldability of modified 9Cr-1MoV-Nb steel plates under SMAW and GTAW welding processes concluded that the increase of preheating and PWHT coarsened the microstructures of weld and HAZ, and significantly influenced the properties of the welded joints.

(Francis et al., 2009) carried out an investigation to study the effect of weld preheat temperature and heat input on type IV failure (refers to the premature failure of
a welded joint due to an enhanced rate of creep void formation in the fine grained or inter-critically annealed heat affected zone). It was concluded that the effect of heat input on the tendency for type IV failure is small and there is a scope to improve resistance to type IV cracking in 9–12% Cr steels through the optimization of welding procedures.

(Olabi & Hashmi, 1993, 1995) studied the effect of post weld heat treatment on the mechanical properties of low carbon structural steel. In order to assess the effect of PWHT on the micro hardness, tensile strength and impact toughness, the tests were employed on the welded joints under as welded and as heat treated conditions. The results showed that the post-weld heat treatment improved the toughness by about 15%, without making any significant difference to the tensile strength and hardness. On the other hand, the residual stresses were reduced by about 70%.

(Ravi et al., 2005) investigated the influence of post weld heat treatment on fatigue life prediction of strength mismatched HSLA-80 steel welds. SMAW process was used to fabricate the welded joints using under matched, even matched and over matched filler electrodes. Fatigue life prediction tests revealed that the mismatch ratio had an inverse relationship with fatigue crack growth and crack initiation exponents. The PWHT did not alter the strength and impact toughness of weld metal greatly but slightly decreased the hardness and increased the percentage elongation i.e. fatigue performance of the joints increased irrespective of the weld metal strength mismatch.

(Frydman et al., 2008) investigated the influence of the multipass welding, welding preheat and welding post-heat treatments on the GMAW joints of Hardox 400 microalloyed steel. It was found that preheating the steel led to a beneficial action of crack free joint. The post-heat treatment strengthened the weld zone and improved the joint plasticity. This benefit from PWHT was found to be higher when the joint was
preheated before welding. Further, multipass welding was not found to have an advantage due to PWHT when compared to a single pass welding joint.

According to AWS, interpass temperature in a multipass weld is the temperature of the weld area between weld passes. When welding ferrous alloys, a controlled interpass temperature slows the cooling rate through an alloy’s critical temperature to prevent defects from happening during multipass welding (Kou, 2003). The interpass temperature for multipass welds is specified as either a minimum or a maximum temperature depending on the material being welded. This minimum temperature is used to prevent the weld from cooling too rapidly and causing the microstructure to transform from austenite to martensite, which could result in weld cracking because of rapid volume change and shrinkage (Cieslak & Michael, 1990). Weldability studies for steel alloys have been studied but interpass temperature was rarely studied separately as a welding parameter variable.

(Omar, 1998) carried an experimental study of dissimilar metal welds in carbon steel–austenitic stainless steel transition joints with electrode composition and preheat/interpass temperature as variables. This study found that interpass temperature and electrode composition did have an effect on hard zone formations in the weld joint. This research recommended an optimum preheat and interpass temperature along with electrode composition for the integrity of the hybrid welded joint.

2.3.2 Previous studies on the welding processes used for Q&T low alloy steels

In fusion welds the need to provide adequate heat for fusion must be balanced against the detrimental effects of heating as it develops different zones of various microstructures in the HAZ, whose property variations and crack susceptibility depends upon amount of heat supplied and cooling rates (Bailey, 1994). Thus the different
welding processes commonly used have been broadly divided into three categories based on the heat input.

a) Low heat input (MMA, GTAW)—1.0 kJ/mm to 1.8 kJ/mm

b) Intermediate heat input (SAW, GMAW, FCAW)—1.8 kJ/mm to 4.0 kJ/mm

c) High heat input (Electroslag, Electrogas etc.)—10 kJ/mm to 40 kJ/mm

While it is desirable to achieve the fastest possible deposition rate consistent with the properties required as per design, the welding process is generally dictated by several factors such as: base material, section thickness, type of joint, welding position, equipment and manpower availability, and shop/site environments (Chandrasekharaiah, 1995).

(Quintana & Johnson, 1999) investigated the effect of intermixed weld metal on mechanical properties of conventional carbon-manganese weld metals. In this case, two different shielded metal arc weld metals were combined with various self-shielded flux cored arc weld metals. The effects of dilution from the underlying self-shielded flux cored root layers on the mechanical properties of shielded metal arc weld metal were examined. The experimental investigations revealed that the effects on tensile results were limited to relatively minor changes in ductility. But significant reductions in Charpy V-notch impact energies were noted in all cases.

(Mohandas et al., 1999) investigated the effect of chemistry of the steel and the welding process on the softening of the heat affected zone. Three types of high strength low alloy steels with different chemical compositions were welded, using three welding processes namely; SMAW, GTAW and GMAW, using bead-on-plate technique.

It was observed that a steel with a high carbon-equivalent exhibited maximum softening. Steel with a low carbon equivalent with high martensitic start and bainitic transformation temperatures coupled with minimum critical cooling time for nil
martensite and full martensite exhibited the least softening in low heat input welding (SMAW), whilst, steel with longer critical cooling time for full martensite exhibited more resistance to softening in high heat input welding process (GMAW). In general, the extent and degree of softening was found maximum in GMAW, which is high heat input process. PWHT in the austenite region eliminated the softened zone. External cooling methods, such as copper backing and argon purging, were found to be useful in reducing the tendency for softening.

(Yayla et al., 2007) investigated the effect of SMAW, GMAW and SAW processes on the mechanical properties of HY-80 steel welded joints with both X and V groove designs. It was revealed from the investigations that welding processes had minimal effect on the tensile properties of the weldments, since all the welds fractured from the base metal. The Charpy V-notch impact test results had shown that, due to higher heat input, the SAW and the SMAW specimens gave better HAZ toughness than the GMAW process. Moreover, the hardness test results had shown that, the SMAW and SAW welding methods had given slightly higher hardness profile across the fusion zone and HAZ, than the GMAW method on the section 3 mm below the top surface of the weldments. It was found that for all the weldments, HAZ possessed the maximum microhardness (390 HV–430 HV), followed by weld metal and base metal. But the hardness gradient varied for different welding processes. The maximum hardness reached up to the 275 HV in the weld metal; which was well below than the HAZ hardness of 425 HV. In the roots of the weldments, the hardness distribution was lower than the upper surface of the weldments, which was mainly due to the tempering effect of the filler passes.

(G. Magudeeswaran et al., 2008b) investigated the effect of welding processes and consumables on tensile and impact properties of high strength quenched and
tempered steel joints. Two different consumables, namely, austenitic stainless steel and low hydrogen ferritic steel, were used to fabricate the joints by SMAW and flux cored arc welding (FCAW) processes. The joints fabricated by using low hydrogen ferritic steel consumables showed superior transverse tensile properties, whereas, joints fabricated by using austenitic stainless steel consumables exhibited better impact toughness, irrespective of the welding process used. The joints fabricated by using the SMAW process exhibited superior tensile and impact properties and less degree of coarse grained heat affected zone (CGHAZ) softening compared to their FCAW counterparts.

(G. Magudeeswaran et al., 2008a) studied the effect of welding processes and consumables on high cycle fatigue life of high strength quenched and tempered steel joints. Austenitic stainless steel and low hydrogen ferritic steel were used to fabricate the joints by SMAW and FCAW. The studies revealed that the joints fabricated using low hydrogen ferritic steel consumables showed superior fatigue performance than the joints fabricated using ASS consumables. Further, the joints fabricated by SMAW process endured higher number of cycles compared to FCAW counterparts i.e. SMAW joints showed 6% higher fatigue life than FCAW joints.

(Mittal & Sidhu, 2015) studied the influence of welding processes (SMAW, GTAW and their combination) on the mechanical and metallurgical properties of dissimilar T91/347H steel weldments welded using austenitic and nickel based filler electrodes. The experimental results revealed that the welded joint fabricated using GTAW welding process with nickel based filler electrode resulted in higher tensile strength and ductility as compared to the other welding combinations used during this study. The GTAW process was successfully used by the researchers (Dehmolaei et al., 2008) to fabricate dissimilar welds between HP heat resistant steel and Inconel 800
steel using the AISI 309 stainless steel and nickel based Inconel 82, 182 and 617 steel alloys.

### 2.3.3 Previous studies on the effects of joint designs

Previous research attempts made on the effect of joint design on the mechanical and metallurgical aspects of welded joints are discussed in the following paragraphs.

(Leijun Li et al., 2005) studied the effect of joint design on the mechanical properties of GTAW welded Al7075-T6 aluminum sheet. Using ER5356 filler metal, full-penetration welds were made on workpieces with various included joint angles. Testing of the mechanical properties of the joints was done in the as-welded, naturally aged and PWHT conditions. The results showed that by using crack-resistant filler and by selecting the proper joint design and PWHT, strong, dependable welds can be produced on thin AL7075 sheet material. An elasticity model of the weld joint was established to understand the mechanical behavior of the joints. An undermatched joint design was found to be capable of achieving a joint strength that matches the strength of the base alloy.

(Jalal, 2008) studied the relation between pre-straining before welding, joint design and microstructure in the welding of 7020 Aluminum alloy. In this investigation, the tensile characteristics and microstructure evolution of the welded plates was carried out which were welded at different pre-strains (5, 10, 20 and 30%) with varied joint designs (single and double V with angles equal to 70°, 80° and 90°). A significant improvement in the mechanical properties such as yield and tensile strength was observed for the single V butt welded joint with 90° groove angle welded with 20% pre-strain as compared to double V butt joint with same groove angle, and reduced with increasing pre strain to 30% due to phase transformation of the stable microstructure (MgZn2). It was found that, changing the joint design angles had little effect on the
mechanical properties improvement compared with joint design shape (single and double V).

(Sattari-Far & Farahani, 2009) investigated the influence of weld groove shape and pass number on the residual stresses measured by hole drilling method in butt welded pipes. The experimental results were used to develop the finite element model, which was used to study the hoop and axial residual stresses in pipe joints of 6 mm and 10 mm thickness of different groove shapes and pass numbers. From the study, it was found that the weld groove shape had no significant effect on residual stresses distribution on the surfaces of butt welded joints in thin pipes. But for thick pipes (10 mm), with X-groove shape, significant increase in the axial tensile stress on the inner surface of the pipes was observed compared with U-groove and V-groove shapes. The study also revealed that the weld pass number had no significant effect on residual stresses distribution in the outside surface of thin butt welded pipes, but the hoop residual stresses in the inner surface of thin pipes significantly decreased when pass number increased.

(Ghosh et al., 2010) studied the effect of pulse current on shrinkage stress and distortion in multipass gas metal arc welds of different groove sizes formed on 16 mm thick HSLA steel. It was observed that the variation in the joint geometries led to the variation in the amount of volume to be deposited, and the dilution level to which the joint would be subjected under the same heat input. This variation in the extent of base metal fusion significantly affected the temperature of the weld pool and consequently its solidification behaviour, influencing transverse shrinkage and deflection of weld joints. Narrow groove weld joint was found to give minimum distortion accompanied with reduced stresses in the joint.
(J. Chen et al., 2011) predicted the influence of groove angle on heat transfer and fluid flow for gas metal arc welding process. Different angles of V-groove were employed under the same welding parameters, and their influence on the weld pool behavior and weld bead geometry was calculated and analyzed, which was required for subsequent calculations of residual stresses and distortion of the workpiece.

It was found that the main flow pattern was more or less the same, although, the angles were different, but they changed the value of velocity, which led to the variation in the temperature distribution and shape of weld pool directly. The heat flow pattern developed numerically revealed that for small groove angles, a smaller process window is developed and larger V-groove angles facilitate the flow of molten pool to travel downward with heat energy at the front to increase the depth of the weld pool, but it evoked other problems such as overheating, much coarser grain structure and under-filling. Finally, it was concluded that, if the angle of groove can completely fill the groove with sufficient penetration, the smaller groove angle is the better choice.

(Balakrishnan et al., 2014) conducted a study to evaluate the effect of joint design on the ballistic performance of armour grade quenched and tempered steel welded joints. Equal double V and unequal double V joint configurations were fabricated using 4 mm thick tungsten carbide hardfaced middle layer, above and below which austenitic stainless steel layers were deposited on both sides of the hardfaced interlayer. Shielded metal arc welding process was used to deposit all the layers. The fabricated specimens were evaluated for their ballistic performance in terms of depth of penetration on weld metal. It was observed from the ballistic results that both the targets successfully stopped the bullet penetration at weld center line. However, the target made with unequal double V joint configuration offered maximum resistance to the bullet penetration at weld metal location without any bulge at the rear side. The higher
volume of austenitic stainless steel front layer and the presence of hardfaced interlayer after some depth of soft austenitic stainless steel front layer were found to be the primary reasons for the superior ballistic performance of unequal double V joint.

### 2.3.4 Previous studies on the effects of filler materials

The selection of welding consumables for Q&T steels is based on the tensile strength, composition, notch toughness and susceptibility to hydrogen induced cracking. The presence of hydrogen in the weldments is the main cause of hydrogen embrittlement, which can be controlled (within the limit of 0.4 wt %) by using low hydrogen electrodes while using SMAW process. The ability of austenitic stainless steel or nickel alloy consumables to dissolve hydrogen is also being used as an alternative to weld high strength steels when preheat levels necessary by other methods are unacceptably high (Bailey, 1993).

(Madhusudhan Reddy et al., 1995) studied the effect of austenitic stainless steel filler metals (309L and 18Cr-8Ni-6Mn) on the weldability of high strength low alloy steel (0.3C-3Ni-2.5Si-0.87Cr-0.52Mn-0.29Mo), welded using GTAW process. Welded joints were characterized on the basis of transverse tensile, hardness and impact toughness properties, and microstructural studies. It was inferred from the investigation that, the joint efficiencies of 309L and 18-8-6 austenitic stainless steel weld deposits were around 72% compared to that of the base metal (with respect to UTS). The microhardness values in the first pass of the weld bead were found to be greater than those of second pass in two-pass welding in all the welded joints. However, the toughness of the weld deposit was improved by a factor of nearly 1.5, compared to that of the base metal in case of 309L filler wire, while with 18-8-6 filler wire the toughness of the weld deposit was lower than that of the base metal.
(Govindaraj Magudeeswaran et al., 2009) studied the effect of welding consumables (viz. austenitic stainless steel (E307), low hydrogen ferritic steel (E11018-M), and high nickel steel (ENiCrFe3)) on fatigue performance of shielded metal arc welded quenched and tempered steel joints with ‘V’ joint configuration. The results revealed that, among all the joints, the ferritic steel joints exhibited superior fatigue performance and endured 14% more number of cycles than austenitic stainless steel joints, which however, endured 23% more number of cycles than nickel steel joints. It was found that, all tensile tested specimens fractured from the weld metal, and ferritic filler joint exhibited a joint efficiency of 66%, whereas, the joint efficiency of austenitic filler joint was found to be 50% and 48% for high nickel steel joint.

(Balakrishnan et al., 2011) investigated the effect of buttering and hardfacing on ballistic performance of shielded metal arc welded quenched and tempered steel joints with double V joint configuration. Before welding, the beveled edges of the double V groove were buttered with austenitic stainless steel (E307-16). Two joints namely; AHA and AHF were developed where a hardfacing alloy (EFeCr-A7) was sandwiched between E307-16 electrode at both root and cover passes, and E307-16 electrode at root pass and E11018-M at the cover pass respectively. The results revealed that the buttering layer improved the ballistic performance for both the joints due to the resultant microstructure and hardness distribution, and also the weld layers remained intact when the projectile was fired at interfaces and HAZ. The weld layer in both the joints offered least resistance against the projectile penetration as compared with other locations in the same targets.

(Bajic et al., 2011) investigated the effect of filler metal composition and heat input on the mechanical and metallurgical properties of shielded metal arc welded plates of microalloyed steel (class Nb/V) of different thicknesses (9.5 mm, 11.0 mm and
14.5 mm). The plates were welded using the basic electrodes E8018-C1 and were marked NM1 (alloyed with Ni and Mo) and N1 (alloyed with Ni) at two different heat input levels ($E_1=0.73 \text{ kJ/mm}$ and $E_2=1.85 \text{ kJ/mm}$) for each steel strip. It was inferred from the study that the formation of acicular ferrite was promoted with the addition of Ni content, while the addition of Ni and Mo had an effect on achieving the optimum ratio of structural components. Microstructural analysis of the HAZ of welded joints obtained with different levels of heat input showed that, the structure was finer with a lower level of heat input in the normalizing and over-heated zone. Finally, it was concluded that the best quality of welded joints was achieved with electrode NM1 and a heat input within the range of 1.37 kJ/mm and 1.65 kJ/mm.

(Balakrishnan et al., 2013) made an attempt to improve the ballistic performance of quenched and tempered steel welds by depositing a buttering layer of austenitic stainless steel (E307-16) on the double V beveled base metal, above which a multi layered structure was fabricated with three different hardfacing alloy (EFeCr-A7) interlayer thickness (4.0 mm, 5.5 mm and 7.0 mm) using shielded metal arc welding process. The study revealed that the buttering layer improved the ballistic performance of the armour steel welds due to the formation of preferable microstructure and required level of hardness at the weld metal region, which kept the weld layers effectively intact when the projectile was fired at fusion zone, interfaces and HAZ. Moreover, of all the three joints, the joint with 5.5 mm hard-facing interlayer thickness offered a maximum resistance to the bullet penetration with a depth of penetration of 14 mm, without any bulge at the rear side.

In general, the welding consumables chosen should deposit weld metal which can give joint efficiency of 100 percent. However, consumables depositing weld metal having lower strength than base metal are often adequate for welds that are subjected to
relatively low stresses. Thus the filler metals have been classified as undermatched, matched or overmatched. The filler metal is undermatched when the yield strength of the filler metal is below the yield strength of the base metal. Matched filler metals have the same yield strength as base metals, and overmatched filler metals have yield strength greater than the base metals. Preferably, the undermatched filler metals are adequate for welding Q&T steels (Duane & Miller, 1997).

Structural steels, whose yield strength lies between 235 MPa and 460 MPa, are usually welded with overmatched or matched filler material. The yield strength of structural steels is lower compared to high strength steels, and there are more possibilities when welding these steels. The flexibility has allowed for a greater variety of filler material research to be carried out with regards to structural steels.

(Satoh & Toyoda, 1975) carried out fundamental and joint performance studies to investigate the mechanical behaviour of the undermatched welded joint of HT-80 structural steels. The fundamental study was made theoretically and experimentally by using an idealized model joint including a soft interlayer. The joint strength was found to be increased toward the ultimate tensile strength of the base metal used, as the interlayer reduced in thickness. Experimentally, the joint performance of HT-80 steel plates with undermatching weld metal was studied by static tension, brittle fracture initiation test and fatigue test. A reasonable tensile strength level for the undermatching weld metal was found to be not less than 90% of the base metal strength. Burst tests of HT-80 steel pipe specimens including undermatching welded joints demonstrated that its fracture strength was as high as the tensile strength of the base metal.

(Satoh et al., 1979) conducted a study to identify the influence of undermatching on the preheating requirements for welding HT-80 penstock steel. In order to investigate this behaviour, two thicknesses (50 mm and 30 mm) plates of HT-80 steel
were welded using two commercially available electrodes namely; AWS E11016-G and E9016-G, and a newly developed electrode with a tensile strength of 637 MPa. The results showed that the undermatching E9016-G type electrode effectively lowered the preheating temperature by around 25 ºC, and the newly developed undermatched electrode lowered by 55 ºC as compared with the overmatching E11016-G type electrode.

(Umekuni & Masubuchi, 1997) carried out experimental and numerical investigations to study the effect of matching and undermatching welds on high strength steels (HT-80, HY-100 and HY-130) welded using GMAW process. The tensile test showed that the tensile strength of the undermatched weld increased due to restraint by surrounding matched welds and the base metal. Results of fatigue testing showed that both undermatched and matched welds exhibited a similar relationship between crack growth rates and the stress intensity factor. The results of restraint cracking tests indicated that the application of undermatched welds to high strength steels led to the reduction of minimum preheating temperatures, and thus preventing cold cracking of the weldment.

(Shi et al., 1998) in their investigation, studied the effect of weld strength undermatch on fracture toughness of the HAZ of 800 MPa HSLA steel, welded by SMAW process using different electrodes under same heat input conditions, and found that lower the weld strength mismatching, the higher the fracture toughness of the HAZ.

(Loureiro, 2002) investigated the effect of heat input variation (2.0 kJ/mm and 5.0 kJ/mm) on the yield strength mismatch of 25 mm thick high strength steel (RQT 710 grade) welded using SMAW process at the root pass, and SAW process at the fill passes. It was revealed from the experimentation that the weld metal yield strength undermatching levels induced a concentration of plastic flow in the weakest zone, and a
loss of strength and ductility of the weld when loaded in tension. Further, it was observed that the mismatching yield strength ratio (WM/BM) was 0.815 when heat input was 2.0 kJ/mm, and 0.765 when heat input was 5.0 kJ/mm.

(Pisarski & Dolby, 2003) carried out an investigation to study the significance of softened HAZ of 25 mm thick Q&T steel (RQT 501), welded by the SAW process at a heat input value of 2.4 kJ/mm. The target mismatch levels between weld metal and parent plate were kept at 0.75 and 1.25. The authors found that in assessing the toughness of softened HAZs, the test specimen must match the practical situation in terms of yield strength, mismatch between weld deposit and base metal. The research confirmed that in the worst case, fracture toughness of softened HAZs occurred when the HAZ undermatched in strength, in both the weld deposit and base metal. Higher toughness values were measured when either the weld metal or parent steel undermatched the HAZ in strength. The conclusions also elaborated that the tolerance to flaws in softened HAZs, critically depends on the fracture toughness of the HAZ region, where, tolerance reduces rapidly in a situation, where the cleavage was the dominant failure mechanism.

(Ravi et al., 2004) made an attempt to assess the influence of mismatch ratio (MMR) on the fatigue life of HSLA steel welds. Rolled plates of 12 mm thick HSLA-80 steel were welded using SMAW process using two types of low hydrogen ferritic electrodes having different yield strength to attain different mismatch ratios. Factorial experimentation technique and response surface approach were used to optimize the various factors influencing fatigue life of strength mismatched HSLA steels. The authors found that the overmatch weld metal offered enhanced resistance to crack initiation and crack propagation. Hence, the fatigue performance of the overmatched joint was found to be superior as compared to equal matched and under matched joints.
(Yong-Dong Li et al., 2008) carried out mechanical modelling for a non-homogeneous weldment with HAZs and fusion zones. The interface-perpendicular anti-plane fracture problems were analyzed for the HAZ and the weld metal, respectively using Fourier integral transform and Cauchy singular integral equation. Parametric studies in terms of stress intensity factor (SIF) resulted in three conclusions: (1) overmatching is more beneficial than undermatching to the reduction of the SIF of a HAZ crack, however, the latter is more effective than the former in reducing the SIF of a weld-metal crack; (2) the optimum value of mismatch factor is 1.0, and values too greater or too smaller than this should be avoided in engineering design; (3) when the mismatch factor is unequal to 1.0, the SIF could be reduced by increasing the absolute value of the non-homogeneity parameters of HAZs.

(Kocak, 2010) addressed the engineering significance of the relationship between different stages of the life of welded structures in terms of process, property and performance relationship. The issue of weld strength mismatch of high strength steels was extensively discussed and reviewed in this paper with a profound emphasis on the weldments property variations in terms of tensile strength, fracture toughness and fatigue strength. Apart from this, the application of the concept of weld strength mismatch in steel pipelines and aerospace structures had also been reviewed. It was further suggested that the heat input and $t_{85}$ time are considered to be important factors when undermatched filler metal is used in welding high strength steel. Both these factors are further dependent on plate thickness, preheating, current, voltage and welding speed which brings a desirable effect in the weld metal properties.

2.4 Metallurgical and mechanical characterization studies

Detailed examination of a welded joint reveals two distinctive regions, the fusion zone and the heat affected zone. The fusion zone is a region in which both the deposited
metal and base metal melted during welding can be found. Adjacent to this lies the heat affected zone in which the microstructure of the unmelted metal in the welded components undergoes significant changes due to the influence of heat. The microstructure which is an overall arrangement of grains, grain boundaries, and various phases is largely responsible for the properties of a metal. The microstructure of a weld metal and adjacent metal is greatly affected by the welding process and welding procedure, which influence the properties of the weld.

This part of the literature survey is concerned primarily with the metallurgical and mechanical aspects of the fusion zone and heat affected zone of the welded joints of high strength steels.

2.4.1 Microstructure of HAZ

Heat affected zone is the portion of the base metal lying next to the fusion line of the weld.

Figure 2.1: Different zones of steel weldment as represented on an Iron Carbon equilibrium diagram (Easterling, 1992)
This region undergoes considerable microstructural variations during welding as a result of thermal cycle variations in terms of sudden heating followed by rapid cooling. This divides the complete heat affected zone into different regions namely; coarse grained heat affected zone (CGHAZ), fine grained heat affected zone (FGHAZ), intercritical heat affected zone (ICHAZ) and subcritical heat affected zone (SCHAZ). Figure 2.1 shows how different zones of steel weldment can be indicated on iron-carbon equilibrium diagram.

(Nadkarni, 2005; Parmar, 1997) explained the different parts of heat affected zone in reference to the temperature encountered and development of various microstructural morphologies. In CGHAZ located adjacent to the fusion zone, the temperature lies between 1000 ºC and 1500 ºC. This results in larger grain growth with martensitic or bainitic microstructure at room temperature, thereby, leading to a region of hard microstructure which is highly susceptible to cold cracking. Adjacent to this lies the FGHAZ in which the temperature varies from 1000 ºC to 900 ºC, consisting of fine grains of bainitic or martensitic structure. In partially transformed or ICHAZ, the temperature lies between 900 ºC and 700 ºC. Finally, lays the SCHAZ in which the temperature lies below 700 ºC. This temperature does not affect the microstructure and grain size but leads to excessive tempering of the base metal, thereby, creating a softer microstructure in the form of tempered martensite.

(Lambert et al., 2000) studied the microstructure of the martensite-austenite constituent in HAZ of HSLA steel welds in relation to toughness properties. The material used in the research was HSLA steel, with yield strength of 433 MPa. The results indicated that the amount of M-A (Martensite-Austenite) constituents and coarseness of bainite are major metallurgical factors affecting the impact properties. The authors found that retained austenite and low carbon transformed martensite had
significantly different influences on cleavage fracture and impact properties of simulated HAZ microstructure, where, freshly transformed high carbon martensite is much more deleterious than retained austenite.

Metallographic investigations demonstrated that in the most brittle zones of intercritical CGHAZ, retained austenite was mostly located between bainitic packets, whereas, blocky martensite and mixed M-A constituents were located at prior austenite grain boundaries. In mixed M-A constituents, austenite was distributed at the periphery, while martensite was located at the centre. The presence of M-A constituent influenced the thermal stability of retained austenite, as they propagate before transformation. These observations constitute preliminary investigations of the transformation mechanism of retained austenite islands.

(Moon et al., 2000) carried out microhardness mapping analysis of high yield, quenched and tempered steel welds fabricated with ultra low carbon consumables using gas metal arc welding process. After fabricating the welded joints under low heat input value of 1.2 kJ/mm, the results were achieved through researching the microhardness and microstructural variations in the weld and HAZ areas. The results revealed that the heat affected zone of the base metal was found to be the hardest region in each weldment examined, regardless of filler metal type, base metal or heat input. The maximum hardness was observed about midway through the HAZ of each weldment studied, rather than adjacent to the fusion boundary.

(Juan et al., 2003) investigated the microstructural variations of the heat affected zone of quenched and tempered super high strength steel (HQ-130) welded joint, with the variation of heat input. The plates were welded using GMAW process under different heat input values varying between 0.92 kJ/mm and 2.64 kJ/mm. The test results inferred that heat affected zone of the weldments mainly consisted of lath
martensite (ML) with high dislocation density of around \((0.3\sim0.9) \times 10^{12}/\text{cm}^2\). However, it was realized that the variation of heat input varied the ML morphology to a significant extent, thereby influencing the impact toughness of the welded joints. Further, it was concluded that by controlling the weld heat input \((E<2 \text{ kJ/mm})\), the presence of carbides in the HAZ could be removed, and therefore the impact toughness in this zone could be assured.

(Pekalski, 2008) explored the structure and hardness changes in the welded joints of quenched and tempered Hardox 400 and Hardox 500 steels. The double V groove joints with 8 mm thickness were fabricated using submerged arc welding process under the recommendations of producer for the selection of welding materials and parametric combinations. The microhardness and microstructural mapping of the weldment revealed that the excessive tempering of base metal due to high heat input welding process resulted in extensively large HAZ width of 70 mm in Hardox 400 and 90 mm in Hardox 500 steels. Further, the heat treatment of the welded joints increased the weld material hardness by 70% for Hardox 400 and by 90% for Hardox 500 steels.

(Zeman, 2009) made experimental investigations to study the properties of 7 mm thick butt welded joints of ultrahigh strength Weldox 1100 steel, welded using laser welding, electron beam welding, plasma welding, activated TIG and metal active gas (MAG) welding processes. The linear energy of welding for laser, electron beam and MAG welding processes was \(\approx0.4 \text{ kJ/mm}\), and for other welding processes the value lies between 0.7 kJ/mm and 0.9 kJ/mm. It was observed, that in case of the joint made by the MAG method, the weld was characterized by its bainitic structure. In the HAZ, it was observed that a purely martensite structure or mixture of bainite and martensite structures was predominantly present. It was further found that the ultrahigh strength steels required the linear energy of welding to be precisely adjusted. If the linear energy
of welding was too low, there was excessive hardening of the HAZ, which increased the risk of cold cracking, whereas, if the linear energy of welding was too high, the strength properties reduced significantly.

2.4.2 Microstructure of fusion zone

Although ferrite and cementite are the two basic constituents of steel, but in weldments, depending on the composition, cooling rates and PWHT employed, they are found to exist in different microstructural phases with varied morphologies. The most basic form of microstructure that appears by direct solidification of the molten metal is the primary microstructure in the form of δ-ferrite and austenite. The solid phase transformation of austenite results in secondary microstructure consisting of the mixtures of allotriomorphic ferrite, Widmanstätten ferrite, acicular ferrite, polygonal ferrite and the different micro-phases (such as bainite, retained austenite, degenerate pearlite and martensite) (Bhadeshia & Svensson, 1993).

(Bhadeshia et al., 1983) reported that allotriomorphic ferrite or grain boundary ferrite, when present in thick layers at the austenite grain boundaries are detrimental to toughness as they offer little resistance to cleavage crack propagation. Thus, it becomes necessary to control the volume fraction of allotriomorphic ferrite in the welds. However, (Bhadeshia & Svensson, 1993) reported that some allotriomorphic ferrite should be retained in the weld microstructure in order to improve its high temperature mechanical properties. It was demonstrated, that the high temperature ductility and the creep rupture life of the welds deteriorated sharply in the absence of allotriomorphic ferrite. The associated intergranular fracture, with respect to the prior austenite grain boundaries, became intergranular when some allotriomorphic ferrite was introduced into the microstructure.
As described by (Parmar, 1997), ferrite side plate or Widmanstätten ferrite grows from the grain boundary ferrite into the original austenite grain as packets of parallel plates separated from each other by low angle grain boundaries. The presence of Widmanstätten ferrite has been considered as an undesirable constituent, and its presence leads to poor toughness in low alloy steel weldments. However, controlled experiments conducted by (Bodnar & Hansen, 1994) established that, when the microstructure is changed from one which is predominantly allotriomorphic ferrite, to one containing Widmanstätten ferrite, there is an improvement in both the toughness and strength, which would be expected due to the presence of large fractions of Widmanstätten ferrite associated with refined microstructures.

The increased use of high strength steels in applications such as pipelines, ships, oil platforms etc. had necessitated the development of weld deposits with much greater strength (>750 MPa). Such weld deposits are typically rather heavily alloyed. (Díaz et al., 1998) reported that the design of such high strength weld deposits must involve fine microstructures of predominantly acicular ferrite, bainite or martensite for such applications, where the fundamental strengths of allotriomorphic and Widmanstätten ferrite are insufficient.

(Zhang & Farrar, 1997) carried out a systematic study to investigate the influence of Mn and Ni on the microstructure and toughness of SMAW welded low alloy steel. The results showed, that increase in the content of manganese and nickel in the weld resulted in the high volume fraction of acicular ferrite at the expense of proeutectoid ferrite with small amount of martensite. It was further observed that optimum composition range (0.6–1.4% manganese and 1.0–3.7% nickel) resulted in balanced weld metal microstructure with better toughness. However, the additions
beyond this limit promoted the formation of martensite and other microstructural features, which were detrimental to weld metal toughness.

(Moon et al., 2000) studied the microhardness variations in quenched and tempered high yield strength steel welds fabricated with two ultra low carbon consumables using GMAW process at low heat input of 1.2 kJ/mm. Authors found that the fusion zone consisted predominantly of lath ferrite with varying amounts of untempered fine lath martensite, as well as small amounts of inter-lath retained austenite and oxide inclusions. No polygonal ferrite or solid-state precipitates such as carbides or carbonitrides were observed in the fusion zone. The local variations in microhardness correlate well with the local variations in the microstructure.

(Juan & Yajiang, 2003) made an attempt to study the microstructural development in the weldments of high strength quenched and tempered steels (HQ-130 and Q-J63), welded using GMAW process, under variable heat input conditions. The experimental results revealed that under low heat input condition of 0.96 kJ/mm, the acicular ferrite was found to be the dominating phase present in the grain with small content of pro-eutectoid ferrite present on the boundary of original austenite grains. However, the ferrite side plate was found to be the predominating phase at high heat input of 2.23 kJ/mm. Thus the authors concluded that in order to have the high toughness and crack resistant weld metals of HQ-130 and QJ-63 high strength steels, the weld heat input should be strictly controlled in the range 1 kJ/mm and 2 kJ/mm in order to control the content of pro-eutectoid ferrite within the limit of 25%.

(Babu, 2004) explained that the acicular ferrite is an intragranular nucleated form of bainite. Appearing to nucleate at non-metallic inclusions present in the weld pool, the phase is formed at intermediate temperatures, between that of allotriomorph ferrite and martensite formation. With its fine microstructure of disorganized plates, it
imparts good toughness and strength to weld metals, and is therefore a highly desirable phase. By presenting a torturous path to an advancing crack, more energy is absorbed per unit of propagation, thus improving toughness and overall strength.

(Keehan et al., 2006a) developed two high strength steel weldments by keeping the nickel content constant at 7 wt%, while varying the manganese level at 0.5 and 2 wt%. The experimental results revealed that in the high manganese weld metal significant amounts of coarse grained coalesced bainite formed, whereas, mainly upper bainite was seen with 0.5 wt% manganese weld metal. Reducing manganese increased the transformation temperature, promoting formation of upper bainite with a fine grain size, and dramatically reduced the amount and size of coalesced bainite. The resultant increased toughness was attributed to the finer grain size of bainite constituents and a more effectively tempered microstructure.

(Qinglei et al., 2011) investigated the influence of three filler wires namely; ER50-6, MK G60, MK G60-1 on the microstructure, tensile strength and impact toughness of the welded joints of Q550 quenched and tempered steel, welded using GMAW process without pre-heating and post weld heat treatments. The results revealed that the welded joint fabricated using MK G60-1 filler wire contained the maximum content of acicular ferrite which made the tensile strength and impact toughness of the joint to approach to that of base metal. However, the presence of crack initiating regions at soft proeutectoid ferrite phases in other joints resulted in dramatic loss of strength and toughness of welded joints. The properties variation due to variation in microstructural phases was further strengthened through fractographic studies. These studies revealed that the acicular ferrite microstructure region corresponded to relatively large dimples, while boundary ferrite microstructure corresponded to small dimples.
2.4.3 Multi-run welds

Apart from having primary and secondary microstructural products, a weld also constitutes a tertiary microstructure which is encountered only in multi-run welds due to progressive heating of the successive runs. Thus, the multi-run welds have complicated microstructural development depending upon the number and size of beads deposited. Some of the regions of original primary microstructure are reheated to temperatures, high enough to cause the reformation of austenite (re-austenitization), which during the cooling part of the thermal cycle may transform into a different microstructure denoted as reheated microstructure (Kou, 2003).

(Keehan et al., 2007) reported that the microstructure in reheated beads is often very difficult to interpret due to the complex morphology obtained after reheating. However, the as-deposited last bead can be used to understand the microstructure to some extent.

2.5 Wear studies

This section on wear studies discusses a brief introduction to wear followed by the previous attempts made in different aspects related to wear, namely; abrasion wear processes, wear in Q&T steels and measurement techniques, implementation of optimization techniques and wear of welded joints.

2.5.1 Introduction to wear

Wear is one of the chronic problems, from which every industrial component suffers and results in huge monetary losses due to failure of equipments, replacements of wear parts and down-time during repairs. In a more generalized form, wear is defined as the progressive removal of material from a surface due to mechanical movement with or without chemical processes. Among the various wear mechanisms, abrasion wear is the most important one due to its destructive character and its high occurrence frequency
(50% of total wear failures). In abrasive wear, detachment of material from surfaces in relative motion is caused by hard particles between the opposing surfaces or fixed on one of them. Thus its control and reduction not only depends on the appropriate selection of materials, but also on finding and analyzing the underlying mechanisms responsible for the abrasion wear (Eyre, 1978).

Because of its predominance, extensive research has been carried out in the area of abrasion wear, which resulted in the development of number of test methods to understand the working behaviour of abrasion phenomenon in different materials (Misra & Finnie, 1982). These test methods can be broadly divided into those where the abrading medium is loose as it passes over the test-piece (commonly termed as three-body abrasion), and those where the abrading medium is fixed in orientation as it passes over the test-piece (commonly termed as two-body abrasion). The most commonly employed test setup for three-body abrasion is dry sand rubber wheel (DSRW) test whose working procedure is encompassed in ASTM G65 standard. However, pin abrasion testing is the most commonly used method for simulating two-body abrasive behaviour, and ASTM G99 & ASTM G132 provides useful guide for carrying out these tests (I.M. Hutchings, 1992).

(Hawk et al., 1999) reviewed the different abrasion wear tests commonly used in the laboratories, and highlighted the various studies carried out for different alloy systems. The DSRW test has been employed to examine the abrasion behaviour of a very wide range of materials. In number of studies, the test is used simply to provide a quantitative ranking of the abrasion resistance of different materials. For example, the behaviour of a series of steels with a wide range of hardness has been tested, and whilst good correlation was found between wear rates and hardness, the operative mechanisms of wear were never examined. The Pin-on-disc abrasion tester has been used to study
the wear of tillage equipment materials where, depending on the load pressing a test specimen against a bonded abrasive paper, either low stress abrasion or high stress abrasion may be produced.

### 2.5.2 Abrasive wear processes

In the following paragraphs, a discussion on the literature related to the types of abrasive wear processes is made.

(Sarkar, 1980) described that wear can be classified in various ways and one of the usual classifications of wear is based on the fundamental mechanism that is operating. Wear can be divided into different modes such as adhesion, abrasion, erosion, surface fatigue and tribo-chemical reaction. Each wear mode can also be divided additionally into various wear mechanisms. (Davis, 2001) explained, that in studying the wear behaviour of materials, a specific mechanism of material removal may be dominant; however, commonly, several wear mechanisms operates at the same time.

(Sarkar, 1980) classified the abrasion wear process according to three factors: two or three body abrasion, low and high stress abrasion, open and closed abrasive wear.

#### 2.5.2.1 Two-body and three-body abrasion wear

Two-body abrasion wear involves the removal of material by abrasive particles which are held fixed (as in abrasive paper), while being moved across a surface. This process produces a grooving form of wear. Three-body abrasion involves loose particles which may rotate as well as slide as they contact the wearing surface. Compared to two-body abrasion, three-body abrasion is much more common and also much more complicated than two-body abrasion. Plastic indentation wear will be much more important in three-body abrasion than that in two-body abrasion. Furthermore, in three-body abrasion, the
movement patterns of abrasives are more complicated than in two-body abrasion, since the abrasives not only slide, but also roll. Thus, a relatively wide range of wear rates have been reported for three-body abrasion conditions, which depend not only on the material being tested, but also on the testing apparatus. In three-body abrasion of metals, cutting wear and plastic deformation wear coexist. As a consequence, two-body abrasion tests are said to produce wear rates one to three orders of magnitude higher than three-body abrasion under comparable loading condition (Gates, 1998).

2.5.2.2 Open and closed abrasive wear

Three-body abrasion is further subdivided into closed and open groups. The closed group covers the cases of fine abrasives between closely mating surfaces. Open three-body abrasion covers cases where there is a thick bed of abrasive, or the particles are so large that the two opposed surfaces are so far apart, that the mechanical properties of one have no influence on the other (Misra & Finnie, 1980, 1982).

2.5.2.3 High and low stress abrasive wear

Abrasive wear processes have also typically been grouped into two regimes: high or low stress. When abrasive particles are compressed between two solid surfaces, high-stress or grinding abrasion occurs. The high pressure produces dents and scratching on the surfaces and fractures and crushes the abrasive particles. Low-stress or scratching abrasion occurs when lightly loaded abrasive particles move across the wearing surface, generating cutting and ploughing on a microscopic scale, but with no damage to the abrasive particles (Moore, 1979).

2.5.3 Wear in Q&T steels & measurement techniques

This part of the section discusses the literature related to the problem of wear in quenched and tempered steels and the various measurement techniques employed for the measurement of the same.
(Jha et al., 2003) made an attempt to establish a correlation amongst the microstructural features and mechanical properties with three-body abrasion resistance of heat treated steels. The high strength low alloy steel was subjected to various heat treatment cycles (viz. annealed (T1), heated to 740 °C and water quenched (T2) and austenitized to 860 °C, and water quenched (T3)) for generating different combinations of microstructural features (viz. as received with martensitic structure, T1 with ferrite and pearlite, T2 with ferrite and martensite and T3 with coarse martensitic structure), and mechanical and wear properties. An analysis of the observations suggested that hardness had a direct correlation with mechanical properties, and optimum level of hardness led to the best wear performance of the steel. T2 conditioned steel offered best wear behaviour accompanied by plastic deformation and microcutting as wear mechanisms with optimum combination of strength and ductility. The samples in T3 and as received conditions revealing martensite experienced higher material removal rate predominantly through microcutting followed by T1 steel with highest material removal rate having softest microstructural matrix with ferrite and pearlite structure.

(Adamiak et al., 2010) compared the abrasion resistance of chromium cast iron wear resistant plates with typically used wear resistant plates made from Hardox 400 steel, and two different wear resistant materials cladded by welding. The low stress abrasive wear tests of these constructional materials were conducted on DSRW apparatus in accordance to ASTM G65 standard using quartz Ottawa sand. All the specimens of standard size (25 mm × 75 mm) were tested under same load of 130 N for 6000 revolutions at 200 rpm, when sand flow rate was set at 300 g/min–400 g/min. The results revealed that abrasion wear resistance of chromium cast iron plate was two times higher than wear resistance of layer made by welding technologies, and nine times higher than typical Hardox 400 steel plate.
(Jensen et al., 2010) investigated and compared the wear rates in abrasion-resistant high chromium white cast iron (21988/JN/HBW555XCr21), a heat-treated wear resistant steel (Hardox 400) and a plain carbon construction steel (S235) which are used in comminuting equipments. Abrasion tests were conducted on micro-wear tester, with quartz as abrasive medium under velocity (1 m/s–7 m/s) and pressure (70 kPa–1400 kPa). The developed wear maps revealed that chromium white cast iron in the low speed region, due to its horizontal topographical wear lines, marginally outperformed S235 and Hardox 400.

(Cheng et al., 1988) conducted an abrasive wear test on pin-on-disc apparatus, with SiC abrasive paper of 180 grit size under two different loads, on the specimens having various contents of tempered martensite, retained austenite and primary carbides of Cr$_7$C$_3$ and Cr$_{23}$C$_6$ in the microstructure of a D2 tool steel, in order to investigate their role in the wear characteristics. It was found from the research that under heavy load conditions, the primary carbides were having the dominating influence on the wear loss, but the role of retained austenite was significant only under the light load conditions.

(Pintaude et al., 2003) investigated the effect of abrasive particle size on the sliding friction coefficient of steels (AISI 1006 low-carbon steel and AISI 52100 bearing steel) using a spiral pin-on-disc apparatus. The 3 mm diameter pins were loaded against glass particle coated abrasive paper of size 80 mesh and 240 mesh, rotating with the tangential sliding speed of 0.08 m/s, with a normal force of 20 N and radial pin movement of 20 mm/min. Experimental results revealed that the friction coefficient of AISI 1006 steel was higher than AISI 52100 steel in both testing conditions. Further, it was found that for AISI 1006, which wore in a severe regime, the strain hardening intensity depended on the abrasive particle size, but for AISI 52100, which wore in a mild regime, the particle size had no effect.
(da Silva et al., 2006) studied the performance of cryogenically treated HSS tools using pin-on-disc test rig under different sliding abrasion wear testing conditions. It was observed that with the increase in abrasive grain size, the wear rate increased proportionally, since, the abrasive diameter changed from 15 µm (600 mesh) to 180 µm (80 mesh). Statistical analysis of the variance for the tests with 600 mesh abrasive paper for both cryogenically treated and untreated samples did not show significant difference in the average wear rates with 95% of reliability. However, for the 80 mesh, the average wear rate for the cryogenically treated sample was slightly smaller (3.3%) than the untreated sample.

(Rajasekaran et al., 2010) compared the abrasive wear resistance of thick tool steel coated with HVOF spraying and the standard high speed steel. A pin on disc wear test was carried out using different types and sizes of the abrasive papers under the same contact pressure of 1.3 MPa. The abrasive wear resistance of cold worked tool steel coated pins was found to be superior against soft and fine abrasive papers than the standard high speed steel. Besides, the performance of the coated pins against hard and coarser abrasive papers was found to be similar to standard high speed steel. The study showed the potential of HVOF spraying on the development of thick cold work tool steel coatings for wear resistance applications.

(Sahin & Kilicli, 2011) compared the abrasive wear behavior of SiCp/Al alloy composite (MMC), ductile iron (DI), partially austenitized and austempered ductile iron (PADI), and conventionally austenitized and austempered ductile iron (CADI) on a pin-on-disc configuration. The pins were abraded against the silicon carbide (SiC) abrasives and tested under different load conditions. The experimental results demonstrated that the wear resistance of MMCs was found to be better than those of DI, its alloy and heat treated samples, tested at the 70 µm size of the abrasives. The wear resistance increased
with wt% particles for MMCs and martensite volume fractions for PADI and CADIs respectively. Moreover, adhesion, chipping and abrasion were observed for the alloy matrix, but the abrasion became the most effective wear phenomenon for the CADI samples.

(Sarhan et al., 2011) studied the influence of silicon carbide abrasive particles of (20, 30, 40, 50 and 60) µm size on carburized digger tooth steels. Four types of steel, with different hardness, were tested at two constant linear sliding speeds under various loads of (10, 20, 30, 40 and 50) N. Tests were carried out for sliding time of (0.5, 1.0, 1.5, 2.0 and 2.5) min. Experimental results showed that there was consistent reduction in abrasive wear as the hardness of the materials was increased. It was found that wear increased with the increase of applied load, linear sliding speed and sliding time. Also, it was noticed that the wear increased with increase in abrasive particle size. Cutting and ploughing were found to be the dominant abrasive wear mechanisms in all types of steels.

(Kim et al., 2011) compared the sliding wear and three-body abrasive wear characteristics of plain carbon steel (0.19C-0.72Mn), heat treated under different conditions, to understand the mechanisms of both types of wear in the steel. Dry sliding wear tests were carried out at room temperature using a pin-on-disc wear tester against AISI 52100 bearing steel and the three-body abrasive wear tests were performed using a ball-cratering abrasive wear tester employing angular SiC abrasives. It was inferred from the investigation, that sub-surface strain-hardening and uniform-deformation were principal controlling factors for the sliding wear, and hardness was the main factor to control the abrasive wear of the steel with micro-ploughing and cutting as the main wear mechanisms under the given test condition.
2.5.4 Studies on wear behavior through optimization techniques

Wear is a complex problem that involves the simultaneous influence of large number of variables on the wear performance of the steels and their fabricated structures. In order to deal with such complexity, it becomes essential to identify, select and establish the operating ranges of the wear variables for obtaining optimized working conditions of steels under different wear environments. The following paragraphs discuss literature related to the issues stated above.

(Sahin, 2006) developed wear resistance model for three types of steels namely, AISI 1340, AISI 1020 and AISI 5150 in terms of abrasive grain size, applied load and sliding distance using the Taguchi method. Wear tests were carried out using a pin-on-disc type of apparatus under different conditions. It was observed that the type of materials was the major parameter among the controllable factors that influenced the weight loss of the steels. For AISI 1340 steel, the abrasive grain size exerted the maximum effect on the wear, followed by sliding distance and applied load. For AISI 1020 and AISI 5150 steels, however, the sliding distance was found to have a significant effect on the weight loss. The optimal combination of the testing parameters was determined and a good agreement within ±10% was observed between the predicted and actual wear resistance.

(Dharmalingam et al., 2010) optimized the dry sliding performances of the aluminum hybrid metal matrix composites using gray relational analysis. The L27 orthogonal array was employed for the experimental design to study the effect of different loads, sliding speeds and varying percentage of molybdenum disulfide on the specific wear rate and coefficient of friction using a pin-on-disc apparatus. The results indicated that specific wear rate of aluminum composites was mostly affected by molybdenum disulfide percentage (40.09%) followed by sliding velocity (30.49%) and
applied load (21.06%). In case of the coefficient of friction, applied load (63.43%) followed by molybdenum disulfide percentage (13.61%), and sliding velocity (5.38%) exerted a significant influence.

(El-Tayeb et al., 2010) investigated the cryogenic wear behaviour of titanium Ti54 alloy sliding against tungsten carbide at different speeds, loads and distances on pin-on-disc apparatus. Empirical models based on central composite rotatable design (CCRD) using response surface methodology (RSM) were developed to predict wear characteristics of Ti54 alloy as a function of sliding conditions at both room and cryogenic temperatures. It was found from the results that under cryogenic sliding conditions, wear volume was consistently lower than in dry sliding condition. Furthermore, load and speed influenced the wear with greater extent at longer distance. Delaminating wear was the dominated wear mechanism under dry and cryogenic conditions. Under cryogenic sliding, the main wear mechanism was abrasion; besides, some evidences of three-body abrasion due to entrapped wear particles. Cracks and fracture of brittle nature were also detected under cryogenic sliding due to alteration of titanium mechanical properties.

(Ravindran et al., 2012) studied the wear and sliding friction response of a hybrid aluminum metal matrix composite reinforced with hard ceramic (5 wt% of SiC) and soft solid lubricant (0, 5, and 10 wt% of graphite) fabricated by powder metallurgy. The influence of the percentages of reinforcement, load, sliding speed and sliding distance on both the wear and friction coefficient were studied using the pin-on-disc method with tests based on the factorial techniques. Analysis of variance showed that the most significant variables affecting the sliding wear of composites were the sliding distance followed by the sliding speed, applied load, graphite content and the interaction effect of the load with the sliding speed. However the applied load affected
the friction behaviour to the most followed by the sliding distance, sliding speed, graphite content and the interaction effect of applied load and the sliding distance. Further the researchers (Dinaharan & Murugan, 2012) used the response surface methodology with four factor five level central composite rotatable design to study the wear behaviour of AA6061/ZrB2 composite under varied dry sliding wear testing conditions. The wear rate of the composite was found to bear a proportional relationship with the sliding speed, sliding distance and the normal load.

2.5.5 Previous studies on the wear characterization of welded joints

Previous research attempts on the wear of welded joints are discussed in following paragraphs.

(Krishnan et al., 2006) studied the dry sliding wear behaviour of the different regimes (viz. Weld metal, different zones of HAZ) of a single pass carbon-manganese steel welded joint. Wear pins both of circular and rectangular cross sectional areas were loaded at a normal load of 196.1 N against the EN-31 steel disc rotating at different sliding velocities on the pin-on-disc apparatus. The wear behaviour of the various regimes of the weld-joint, specifically, that of the different sub-zones of the HAZ, were determined and analyzed using their microstructures, initial micro-hardness, work hardening during wear, and their inherent residual stresses. It was inferred from the experimental investigation, that due to the varying effect of work hardening and microstructural thermal stability, a significant difference exists between the wear behaviour of different regimes of a HAZ (possessing varied microstructure), when wear tests were carried out continuously from the weld metal to the base metal, and when these were done on isolated regions exhibiting only a specific subzone. Due to work hardening behaviour of each sub-zone, the hardness values of a subzone before and after the wear tests (under identical conditions) were found different.
(Garcia et al., 2008) studied the wear behaviour of the different zones of three welded joints, fabricated under different welding conditions from 9.2 mm thick pipeline steel (A355 Gr.P11). For the first welded joint (A), the plates were preheated to 100 °C, followed by the root pass laid down using GTAW process and cover passes by SMAW process, and finally PWHT at 700 °C. For the second welded joint (B), the welding procedure was kept same, however, the joint was not PWHT after welding. For the third joint (C), both preheated and PWHT was eliminated keeping the same welding procedure for joint fabrication. After joints fabrication, pins of diameter 4 mm were extracted from base metal, HAZ and fusion zone of all the welded joints. Those pins were subjected to wear test on pin-on-disc tribometer at 200 °C under the same wear test conditions of 10 N load, 200 rpm of rotational velocity and a track radius of 3 mm. The results revealed that the sample C showed maximum wear resistance owing to maximum hardness, followed by sample B and sample A with minimum wear resistance for all the zones of the weldment. Also, for each joint, HAZ was subjected to maximum wear rate due to minimum hardness owing to soft microstructural features present in this zone. This was followed by the base metal and fusion zone.

2.6 Problem formulation and objectives of research

In this section, gaps of the previous research are discussed and the main research gaps are identified, which is followed by objectives of the present work.

2.6.1 Research gaps identified

Extensive literature review was carried out covering various aspects related to the development and fabrication especially w.r.t. weldability, and serviceability of the quenched and tempered low alloy steels. Based upon the literature reviewed, it was found that ultra high strength with good toughness and excellent through hardness in these steels is the outcome of the combination of rolling processes and alloying
elements variations within these steels followed by quenching and tempering. These combinations of the properties make such steels extremely important for being extensively used in marine, military, mining, earthmoving and mineral processing industries. Despite of their widespread use in these areas, their extensively high hardness and strength leads to poor weldability, which creates a significant hindrance in realizing their engineering potential to their fullest. Some gainful experiences have been reported in the literature while researching with these steels which act as an important prior information for researching into various weldability related aspects of *Quenched and tempered low alloy abrasion resistant steels* which are a special category of these steels possessing sufficiently high hardness, ultra high strength and good toughness. These grades in particular are not standardized grades as the conventional ones and whatever information is available regarding their fabricability through welding, is mainly through literature which comprises of only the recommended procedures and that too indicative ones by their respective manufacturers (owing to obvious commercial reasons). So in view of the industrial importance of these particular grades of engineering alloys it was important to investigate/study their behaviour with an aim of gaining insights into their metallurgical and mechanical behavioral corresponding to certain specific changes that are brought into while designing and developing their welding procedures such that the data base related to their engineering usage could be broadened.

So in view of above mentioned important research gaps it was decided to select a specific grade of these alloys which in the present case was JFE-EH400 abrasion resistant steel (also broadly referred to as quenched and tempered low alloy abrasion resistant steel or low alloy martensitic steel) and investigate it in a systematic manner with special focus being on improving the mechanical and wear performance of their
welded joints, besides gaining understanding about the correlations existing between their microstructures and mechanical as well as wear behaviour.

So with the current status of the problem it was imperative to develop welding procedures which would help in enhancing the overall life, joint efficiency and wear performance of the welded joints fabricated from JFE-EH400 abrasion resistant steel. In view of the research gaps identified as mentioned above, the present research problem was formulated with the objectives as mentioned in the foregoing paragraphs.

2.6.2 Objectives of the research work

Primary focus of the work was to develop suitable welding procedures for Q&T low alloy abrasion resistant steels and evaluate the mechanical and wear performance of such welded joints under two-body and three-body abrasive wear conditions. The objectives of the present work are divided into four sub-objectives and are presented in this part of the section:

2.6.2.1 Establishing of welding procedures capable of giving sound weld quality and hence improved mechanical performance of quenched & tempered low alloy abrasion resistant steels

In order to accomplish the above stated objective, following secondary objectives were formulated:

a) To investigate the effect of joint design on the mechanical properties of Q&T welded joints.

b) To investigate the effect of varied weld metal composition induced via varied filler material composition on mechanical properties.
2.6.2.2 Investigation into improving wear performance of quenched & tempered low alloy abrasion resistant steel welds

In order to accomplish the above stated objective, following secondary objectives were formulated:

a) Optimization of the abrasion wear behaviour of the Q&T low alloy abrasion resistant steels under two-body and three-body abrasion wear conditions.

b) To study the wear behaviour of different welds under optimum wear testing conditions obtained in sub-objective 2(a).

2.7 Summary

The literature survey carried out in this chapter presented an exhaustive review of research primarily concentrated on studying the weldability related issues, mechanical and metallurgical characterization and tribological behaviour of quenched and tempered low alloy steels. Keeping in view the reviewed research work, the research gaps were identified and objectives of the proposed research were finalized. The next chapter describes the detailed experimental procedures adopted to accomplish the objectives formulated in this chapter.