MECHANICAL AND MICROSTRUCTURAL INVESTIGATION
OF WELD BASED RAPID PROTOTYPING

A thesis submitted in partial fulfilment of the requirements of the
degree of Doctor of Philosophy in Mechanical Engineering.

by

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‘In the name of Allah, the Beneficent, the Merciful. ‘

_Praise be to Allah, the Lord of the Worlds,_

_The Beneficent, the Merciful._

_Master of the Day of Judgment,_

_Thee (alone) we worship; Thee (alone) we ask for help._

_Show us the straight path,_

_The path of those whom Thou hast favoured; Not the (path) of those who earn Thine anger nor of those who go astray._

(Surah Al-Fatihah, Al-Qur’aan)


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Abstract

At present the commercially available rapid prototyping (RP) machines can mainly produce parts that can be used either as models for visualization or for rapid tooling. The emphasis of the ongoing research in this field is to produce parts that can physically imitate and work like a component produced by a conventional manufacturing technique. Thus the idea is to produce “form-fit-functional” parts rather than prototypes for visualization. Parts made by metals are of specific interest and welding based RP has good prospects in this regard; with the specific possibility to produce fully dense metallic parts and tools. However, the big draw back of using welding as the deposition process is the large heat input to the substrate or to the previously deposited layers, thus causing high temperature gradients and resulting in deformations, warpage, residual stresses, delamination and poor surface quality. In addition the layer by layer additive manufacturing nature results in non-homogeneous structures, porosity and anisotropic material properties. Parts thus produced are of near net shape and out of tolerance. In order to predict and minimize these problems, knowledge of thermal gradients and temperature history during manufacture is important. Moreover, to overcome the problem of surface quality and out of tolerance parts a hybrid welding/CNC milling based RP system can be a good option. These problems associated with the use of welding as RP tool needs to be minimized by the proper investigation of the different deposition parameters and process conditions e.g. intermediate machining, deposition patterns, heat sink size, interpass cooling time, preheating and constant control temperatures on the material properties and mechanical behaviors of the finally produced parts.

This dissertation presents an analysis based on a numerical and experimental approach for the effects of different deposition and process parameters on welding based rapid prototyping process. The entire work is divided into two main parts. The first part is an experimental comparison of microstructure and material properties of the simple GMAW based layered manufacturing (LM) with the hybrid
welding/milling based LM process. Material properties were investigated both on a macro and microscopic level. The microstructure for the two deposition procedures were studied and compared. The hardness test results for the two procedures were investigated and the results were studied in the light of the respective microstructures. Tensile test samples were developed and testing was performed to investigate the directional properties in the deposited materials. Reaustenitised and un-reaustenitised regions were found in the entire body of deposition without machining (DWM) while these are confined to the top layer of deposition with intermediate machining (DWIM) changing alternatively across the weld direction with intervals equal to the inter-bead spacing. The central layers of the DWIM deposit comprise only of reaustenitised region varying sequentially in grain size in both longitudinal and perpendicular direction. This sequential variation is in accordance with the inter-bead spacing in the across direction, and with the layer thickness in the perpendicular direction. The hardness results are in good agreement with the variation of the microstructure both for DWIM and DWM. The hardness values are higher at the top and interface layer while it is comparatively less in the central layers of DWIM samples. However, in DWM samples the hardness values are relatively higher in the top layer only. The correlation for hardness values as related to the tensile strength also holds within normal expectations. The tensile test results show no variation in the yield strengths of samples produced longitudinal and perpendicular to the deposition direction; however there is a slight difference in elongation. Moreover a sharp yield point was observed in the DWIM samples in contrast to the DWM samples.

The second part presents a finite element (FE) based 3D analysis to study the thermal and structural effects of different deposition parameters and deposition patterns in welding based LM. A commercial finite element software ANSYS is coupled with a user programmed subroutine to implement the welding parameters like Goldak double ellipsoidal heat source, material addition, temperature dependent material properties. The effects of interpass cooling duration were studied and it was found that an intermediate value of interpass time is suitable for a nominal level of deformations and stresses. A similar finding was made from the studies about different weld bead starting temperatures. The studies regarding different boundary conditions revealed that the deformations are least for adiabatic case while isothermal case produced the maximum deformations. Simulations carried out with various deposition sequences
revealed that the thermal and structural effects, on the work piece, are different for different deposition patterns. The sequence starting from outside and ending at the center is identified as the one which produces minimum warpage. The results presented are for deposition by gas metal arc welding but can be applied to other deposition process employing moving heat source. The parametric results suggest that in order to minimize the harmful effect of residual stresses, proper combination of deposition parameters is essential. Proper selection of deposition patterns, substrate thermal insulation, and nominal interpass cooling / control temperature can reduce the part warpage due to residual stresses.
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In recent years the competitiveness of manufactured products in the world market has intensified tremendously. In addition mass customization, accelerated product obsolescence and continued demands for cost savings are forcing companies to look for new ways to improve their business processes. Fast and cost effective development of high quality products with a swift response for market demands regarding rapid changes in design and functionality is the requirement of today’s global economies. It is extremely important for new products to reach the market as early as possible and that too before the competitors. To achieve this, the processes involved in the design, test, manufacture and market of the products require to be squeezed, both in terms of time and material resources. This therefore requires new tools and approaches to tackle these issues, and many of these tools and approaches have already evolved or are being evolved. These are mainly technologically-driven, usually involving computers. The rapid development and advancement in such technologies over the last few decades has made it possible to achieve such results.

In product development, the pressure of time is a major factor that drives the direction of development and success of new methodologies and technologies for enhancing its performance [1]. To meet these challenges of decreased time to market for mass products and decreased delivery times for low production parts a variety of measures can be taken. In response to these challenges, the industry and academia have worked on a number of technologies that help to develop new products and broaden the number of product alternatives. Examples of these technologies include feature-based design, design for
manufacturability analysis, simulation, computational prototyping, and virtual and physical prototyping.

Rapid Prototyping (RP) is an early design exploration and validation process, in which executable models of a system are created that reflect some subset of properties of interest [2]. It is a process that automatically creates a physical prototype from a three-dimensional computer aided design (CAD) model, in a short period of time [3,4]. It is an indispensable tool for shortening product design and development time cycles [5,6]. Furthermore, it is a powerful communication tool that bridges design, marketing, process planning and manufacturing, and can facilitate the implementation of concurrent engineering [7,8]. Most designers are of the idea that “getting physical fast” is critical in exploring novel design concepts. The sooner designers experiment with new products, the faster they gain inspiration for further design changes. The evaluation of a design is required so as to quickly uncover design errors or undesirable properties and this helps refine the requirements. The main idea is to quickly create prototypes to reveal any shortcomings in the product and with less effort than required for full system implementation and production. If prototype models are available for testing and visualization the time needed to design a product can considerably decrease.

Since RP refers to a class of technologies that can automatically construct physical models from Computer-Aided Design (CAD) data. Therefore CAD/CAM systems are required that can quickly produce physical objects directly from CAD models. The main idea is to decompose the 3-D computer models into thin cross-sectional layers, followed by physically forming the layers and stacking them up “layer by layer”. The basic RP process chain is presented in figure 1.1.

The idea of creating 3D objects in a layered fashion is almost as old as human civilization. The earliest constructions such as the Egyptian pyramids were built block by block, layer by layer. Stacking up of individually shaped material layers also has a long tradition in a number of manufacturing applications such as tape casting and shape melting [9]. In the present form RP has its beginning in the mid-1980s with the debut of 3D Systems Inc.’s (Valencia, CA) SLA-1 [8]. In its early days of development, the automotive and aerospace industries were the leading users of RP applications [10]. This
Figure 1.1: Typical RP process chain

- CAD 3-D Modelling
  - STL file conversion

- RP System Computer
  - Slicing

- RP Fabrication
  - Layer-by-layer buildup

- Post Processing
  - Cleaning, hardening
  - Finishing

- 3D Physical Part
however, is no longer the case as the use of RP has spread into many other industries. In the frame of research a few space programs have investigated the possibility to use RP and rapid manufacturing techniques during space instrument development [11]. At present there are about 28 manufacturers worldwide offering a total of more than 56 different RP systems to meet the diverse demands of end-users [12]. This shows how fast the RP technology is developing.

1.1 The RP process: basic methodology

Fundamentally, RP is a fabrication process. Therefore, the three basic fabrication processes i.e. additive, subtractive and formative, could be applicable to RP [5].

- In the additive and/or incremental processes, thin layers with distinct shape are stacked one upon other to build the models. The shaping method however, of each layer varies for different processes. Most of the commercial RP systems belong to this category. Such processes are also called layered manufacturing (LM) or solid freeform fabrication (SFF). LM implies that the artifacts are built in layers which greatly simplifies the processes and enables their automation. Dutta et al. [13] has given a comprehensive review of LM processes and related process planning issues. An important feature in LM is the raw material, which is either one-dimensional (e.g. liquid and particles) or two-dimensional (e.g. paper sheet) stocks. This is in contrast with the three-dimensional raw material stocks in case of subtractive RP processes. Stereolithography apparatus (SLA), selective laser sintering (SLS), three dimensional printing, fused deposition modeling (FDM), contour crafting (CC) etc. are few examples of LM.

- In the subtractive or material removal (MR) processes models are build by cutting of excessive material from the raw material stocks. There are not as many subtractive prototyping processes as that of additive processes. A good example of the subtractive process is the sculpturing robot (SR) system at the Delft University of Technology [14]. A commercially available system is DeskProto [15], which is a three-dimensional computer aided manufacture (CAM) software package for RP and manufacturing. Another commercially available system is the Roland’s SRP process that allows designers to create precise, smooth prototypes from a wide
choice of non-proprietary, inexpensive plastics [16]. The major feature of pure subtractive RP processes is that a model is made from a single stock, thus fully compact parts of the same material as per actually required for end use is possible. Hence the idea of form fit and functional parts is possible. The added advantages are the accuracy of the part dimensions and the surface quality that can be achieved by the subtractive machining approach. However the MR processes are limited in geometric complexity as compared to the LM processes. The cutting methods used include computer numerical control (CNC) milling, water-jet cutting, laser cutting and the Millit [17] package.

- In formative or deforming processes, a part is shaped by the deforming ability of materials. At present there is no commercial forming-based RP system in the market [18].

The pros and cons of LM and MR processes are quite obvious. The constraint of the geometric complexity of models is relaxed to a significant extent due to the layer by layer buildup in LM. Many features (such as deep cavities and freeform internal surfaces), which are difficult or not possible to fabricate by conventional MR processes, can be easily built by LM. Thus additive processes have great potential for hands-off free form production of very complicated parts. However the geometric accuracy and surface quality of the part built in LM is far less than that in MR. One of the main error sources arises due to the staircase effect: a model made by LM is a zeroth-order approximation of the design model. The current LM processes are also restricted due to the limited build envelope and choices of raw materials. To obtain the benefits of both the LM and MR processes, there have been increasing efforts in the recent years for the integration of the two. This integration leads to a hybrid RP system in which the staircase phenomena can be eliminated and a better surface quality can be acquired without losing the manufacturability for complex features. The common ground of the hybrid RP systems is to build models in additive ways, while shaping each layer in subtractive ways. Thus complex geometries can be produced due to layered buildup and accuracy and surface quality achieved due intermittent machining. Typically such systems include shape deposition manufacturing (SDM) [19], Computer-aided manufacture (CAM) of
laminated engineering materials (LEMs) [20], laminated object manufacturing or cured at temperatures which do not (LOM) [21], hybrid plasma deposition and milling (HPDM) [22,23], Paper lamination technology (PLT), Solid ground curing (SGC), Solidscape and the solvent welding freeform fabrication technique (SWIFT) [24].

1.1.1 Sacrificial Support Material

In most additive SFF processes three dimensional shapes of arbitrarily complex geometries are created by incremental material deposition of 2 1/2 dimensional, cross-sectional layers embedded in complementary shaped, sacrificial support material (as shown in figure 1.2 for the case of SDM). This support material is required during buildup to accommodate the geometries with overhanging features. In addition these are useful for delicate features such as islands, internal cavities and thin-walled sections. Since pieces of a part can only be located by being built in contact with previously deposited material or the support platform, special procedures are required to build layers with islands or overhanging features. In these cases, special support structures are built upward from the platform to support these objects. The previous layer, embedded in the support material, provides the necessary flat surface for the overhangs, thus allowing manufacturing without any change in strategy. This approach facilitates the automatic planning and execution of fabrication by eliminating the need for part specific tooling or fixtureing. The support structures are designed to be only as large or as strong as necessary for support since they are removed once the part is completed.

In some SFF processes the support material is the same as part material which remains unused / unprocessed during the buildup (e.g. as in SLS, LOM etc.). However, in other processes the two materials are different (e.g. as in SDM, FDM etc.). In the second category as each part is encased in a sacrificial support material the process has to deal with more than one material even for homogeneous parts. The basic requirement is the compatibility of part and support materials, subject to the constraints of the deposition and removal processes. Part and support materials must be physically and chemically compatible. These materials must be deposited or cured at temperatures which do not damage previously shaped features. Also the part and support materials must have minimal intersolubility, and neither material may inhibit curing processes of the other
Figure 1.2: Schematic of SDM
There must be a suitable degree of adhesion between the two materials without the formation of any defects, e.g. internal voids.

1.2 Rapid Tooling and Rapid Manufacturing

In addition to prototypes, RP techniques can also be used to make tooling (referred to as *rapid tooling*) and even production-quality parts (*rapid manufacturing*). Rapid tooling (RT) technology, which embodies the creation of prototype or production tooling based on RP parts, complements RP when large quantities of similar parts containing complex features, made economically utilizing materials close to or identical to end production materials and with normal production processes are required [26]. This is a common feature in concurrent engineering environments where different cross-functional design teams simultaneously require prototype parts or when prototypes produced using end production materials and processes require to be tested for performance. As such, RT not only allows the prototyping of products but also the production processes [27].

RP is a technology for quickly fabricating physical models, functional prototypes and small batches of parts directly from computer aided design (CAD) data. RT generally concerns the production of moulds and tooling inserts using RP. Since the beginning of RP, parts fabricated by RP systems have been employed as patterns for investment casting (IC) to cut tooling costs and lead times [28]. This use of RP for IC is one of its more popular tooling related applications [29]. However, the economic benefits from such RP patterns are limited to small production quantities due to high RP material costs [30,31]. The current research focus has shifted from RP pattern fabrication to the development of RT for producing IC patterns. For large number of castings, RT can be economically and effectively used to produce wax patterns ranging from tens to millions. The traditional foundries are the main beneficiaries since they make wax patterns. However, for the non-wax RP patterns, changes are required in the IC process for successful runs [32]. Manufacturers are increasingly looking towards RT, not only as an alternative to RP, but especially for short production runs which do not justify the investment required for conventional hard tooling [33].

In the industry several RT technologies are now commonly available. Some of these technologies produce the tool directly from the RP process. However, the majority of RT
technologies use the model created by the RP process in a secondary process to produce the tool.

A variety of tooling can currently be produced using different RP technologies. The tooling is basically classified as hard or soft tooling and also as direct or indirect tooling [34]. For short manufacturing runs the tooling used is often known as soft tooling. As the name suggests these tools are often made from softer materials such as silicon rubber, epoxy resins, low-melting-point alloys, or aluminium, which are easier to work with than tooling steels. On the other hand the tooling for longer manufacturing runs is known as hard tooling and is usually made of hard tooling steels. When the tool or the die is created directly by the RP process it is known as direct tooling. In injection moulding, for example, the RP process is used to produce the core and cavity, together with the gating and ejection system. On the other hand in the indirect tooling, the RP technology is only used to create the master. This master is used to make a mould out of a material such as silicone rubber, epoxy resin, soft metal, or ceramic.

RP parts produced till to date are mostly used for prototyping or tooling purposes; however, in the coming years majority of these may be produced as end products. In this context the term ‘rapid manufacturing’ (RM) is introduced in which RP technologies are used as processes for the production of end products. In this everyday changing world there is an increased and ongoing demand for new and improved products. This demand derives the way for newer versions of the product with lesser product development time and therefore smaller production volumes. The application of RP technologies for production rapid manufacturing may therefore grow considerably as the production volumes decrease. The RM operation can greatly improve the competitive position of companies adopting it. It is a natural extension of RP. The key enabling technologies of rapid manufacturing are RP and RT [35]. RM will never completely replace the conventional manufacturing techniques, especially in large production runs where mass-production is more economical. However, due to no tooling requirements for short production runs, RM is much cheaper [36]. It also allows geometric freedoms such as variable wall thickness. Furthermore, manufacturing without tooling also permits distributed manufacture so that parts may be made in or near the location where they are
required. An example of this may be seen by NASA’s adoption of fused deposition modeling (FDM) to make spare parts on the international space station [37].

Most systems developed so far mainly have a single function and have one material on one stage. A new concept of functional prototype development (FPD) is newly proposed that provides the necessary prototype functions such as mechanical, optical, chemical and electrical properties in order to meet the broad requirements of the industry [38].

1.3 Rapid Prototyping Pros and Cons

1.3.1 Advantages of RP

RP implies that complex shapes are as easy to build as simple shapes, since the planning and manufacturing processes are automated. RP machines are actually "three dimensional printers" that allow designers to quickly create tangible prototypes of their designs, rather than just two-dimensional pictures. The main benefits of RP are:

- RP allows Designers to make products faster and less expensively.
- Rapidly Prototyped parts show great time, cost and material savings.
- Quick product testing is possible.
- Expeditious design improvements.
- Fast error elimination from design.
- Increased product sales
- The speed of system development is increased.
- Rapid manufacturing is possible
- Automated, toolless, patternless RP systems can directly produce functional parts in small production quantities.
- The ability to experiment with physical objects of any complexity in a relatively short period of time.
- Optimize part design to meet customer requirements, with little restrictions by manufacturing.
• Minimize time consuming discussions and evaluations of manufacturing possibilities.

• Minimize time and cost for design, manufacturing and verification of tooling. Since RP encourages creation of detailed design at the beginning of a product’s development. This can lead to a more usable product in a shorter period of time. In fact, development teams that prototype, require 45% less effort to produce systems than development teams that did not prototype [39].

• Reduce the labor content of manufacturing, since part specific setting up and programming are eliminated, machining/casting labor is reduced, and inspection and assembly are also consequently reduced, however more manual finishing is required for the RP parts.

• Reduction in material cost waste disposal cost, material transportation cost, inventory cost.

• Can avoid design misinterpretations.

• Can quickly modify design.

• Allows early customer involvement. RP parts make excellent visual aids for communicating ideas with co-workers or customers. There is better communication between the user and designer as the requirements and expectations are expressed in the beginning itself.

• Opportunity to separate design from manufacturing.

• Research has shown that the total costs for new products can be reduced by as much as 30 to 60% [40]. Therefore consumer can buy at lower prices, since the manufacturers savings will ultimately be passed on.

1.3.2 Disadvantages of RP

Rapid prototyping also poses some special challenges that are usually worth overcoming. Some of the disadvantages of rapid prototyping are described below.

• Some people are of the opinion that rapid prototyping is not an effective model of instructional design because it does not replicate the real thing. It could so happen
that many important steps of instructional design are forfeited for a faster, cheaper model.

- Many problems may be overlooked that results in endless rectification and revision.
- Rushing in to develop a prototype may exclude other design ideas.
- Design features may be limited by the scope of the prototyping tool.
- The user may have very high expectations about the prototype’s performance and the designer is unable to deliver these.
- The system could be left unfinished due to various reasons or the system may be implemented before it is completely ready.
- The producer may produce an inadequate system that is unable to meet the overall demands of the organization.
- Too much involvement of the user might hamper the optimization of the program.

1.4 A Survey of Current RP-Technologies

Detailed descriptions of RP techniques can be found in literature [41,42]. Another comprehensive review of LM processes and related process planning issues was given by Dutta et al. [13]. As already discussed above, in a broad description, RP can be classified as either involving material addition or material removal, with a few cases involving both LM and MR. The LM processes as per stated by Kruth [43] can be classified by the state of the material before the manufacturing of a part. The prototype material can be liquid polymer, molten material, powder, solid sheet or electroset fluids. Based on these classifications the different RP techniques are discussed below:

1.4.1 Material Addition based RP-Approaches

1.4.1.1 Processes involving a liquid

1.4.1.1.1 Stereolithography (SL)

Patented in 1986, stereolithography started the revolution in rapid prototyping. This process is based on a photosensitive liquid resin which when exposed to ultraviolet (UV)
light solidifies and forms a polymer. SL systems consist of a build platform (substrate) which is mounted in a vat of resin and a UV helium–cadmium or argon ion laser [44]. The information obtained from the three-dimensional solid CAD model is used by the laser system to image the first layer of the part on the resin surface. Once the layer is hatched after scanning the contour, the platform is lowered and a new layer of resin is applied. The next layer is then scanned. After the completion of the part, it is removed from the vat, the excess resin drained and supports are broken off. The ‘green’ part is then placed in a UV oven for post-curing. The SL approach is depicted in figure 1.3.

Stereolithography Apparatus (SLA) machines have been made since 1988 by 3D Systems of Valencia, CA [45, 46]. Since it was the first technique, SL is regarded as a benchmark by which other technologies are judged. In the beginning SL prototypes were fairly brittle and prone to curing-induced warpage and distortion, but recent improvements have largely corrected these problems. To broaden the application area of SL, research and technology development efforts are being directed towards process optimization [47, 48].

1.4.1.1.2 Solid ground curing (SGC)

This system again utilizes photopolymerizing resins and UV light. This is a complicated process developed by the Cubital Corporation in which thermoset part material is imaged using an ultraviolet light to form layers of a part [49, 50]. To build a layer, a thin coating of ultraviolet-curable photopolymer is spread over the bottom of the build chamber. An electrostatically-charged roller is used to apply toner to a mask. This will shield all the undesired areas of photopolymer. The mask is then placed over the resin surface and the entire layer is illuminated with a powerful UV lamp to cure the desired portion of this cross-section of the part. The uncured photopolymer is then removed and hot wax is filled in areas where cured part material is not present. A cooled platen is then used to solidify the wax, which acts as a support material for undercut features. Finally, the unwanted wax and photopolymer on the top of the layer is planed by a milling cutter to a desired height. This cycle is repeated for each layer to ultimately build the entire part. The part is taken out from the chamber at the end of this process, and the support material is removed manually by heating. The company plans to develop a reverse-cycle process,
Figure 1.3: Stereolithography

Figure 1.4: Solid ground curing
in which photopolymers would be used as support materials while other polymer, metallic, or ceramic materials are used as part materials [51].

A distinct feature of this technique is the capability to produce numerous parts in a timely fashion in a single batch. Because the processing time for each layer is independent of the part size or geometry, multiple parts packed into a single batch (e.g., the four replicates next to each other in figure 1.4) can be fabricated in the same time as is required to build a single part. This reduces the average cost and time required to build parts. This machine was a very complicated rapid prototyping machine, with a reputation of requiring a lot of maintenance [52]. However this machine is not available commercially since 1999.

1.4.1.1.3 Rapid micro product development (RMPD)

The RMPD process is a mask-based technology very similar to that of photolithography as used in microelectronics manufacture [53]. Laser polymerization of a liquid photoresin is done in a layer-by-layer fashion employing masks produced by CAD data. This technique has the capability to build micro level components. Parts with a minimum layer thickness of 1 µm and X–Y resolution of 10 µm are possible. This process can also be used to create complex micro systems in which electronics, optical and mechanical components are integrated.

1.4.1.1.4 Liquid thermal polymerization (LTP)

This process is similar to SL except that instead of a photosensitive resin a thermosetting resin and in place of UV light an infrared laser is used to create voxels (three-dimensional pixels). In this case the heat dissipation corresponds with the size of the voxels, and this may also cause unwanted distortion and shrinkage in the part [41]. The system is still in research stages.

1.4.1.1.5 Holographic interference solidification (HIS)

In place of layer by layer solidification, a holographic image is projected into the resin, causing an entire surface to solidify. The imaging data are still obtained from the CAD model, although not in the form of slices [41]. There are no commercial systems available yet.
1.4.1.6  *Objet Quadra process.*

This technique employs 1536 nozzles for spreading layers of photosensitive resin while building a part. After spreading the resin each layer is cured by using two UV lights. The intensity of the lights and the exposure are controlled in such a way that the parts produced do not require post-curing. A second material is also deposited for support overhanging areas and undercuts that can be easily separated later on from the model [54].

1.4.1.2  *Solidification of an Electroset fluid*

1.4.1.2.1  *Electrosetting (ES)*

A conductive material such as aluminium is used as electrodes on to which the layer cross-section is printed. After printing of all the layers, they are stacked onto each other, immersed in a bath of electrosetting fluid and energized. As a result the fluid in between the electrodes solidifies to form the part. The composite is then removed and drained. The unwanted aluminium is trimmed from the part. This technology is advantageous due to the control available over certain material properties. This is done by adjusting the voltage and current applied to the aluminium electrodes which may result in the control of part density, compressibility, hardness and adhesion. Silicon rubber, polyester, polyurethane or epoxy parts can be made by this technique. The hardware for such a system may be inexpensively bought off the shelf [55].

1.4.1.3  *Processes involving solidification of molten material*

1.4.1.3.1  *Ballistic particle manufacture (BPM)*

Ballistic Particle Manufacturing is another inkjet RP process, commercialized by BPM Technology, Inc. [56, 57] (figure 1.5). In this process an inkjet nozzle mounted on a five-axis motion system is used to spray thermoplastic part material at an object under construction. The stream separates into droplets that hit the substrate and immediately cold-weld to form the part [58]. The five-axis characteristic of the process reduces the stairsteps which are otherwise created in 2-1/2D LM processes. The material used in this process is suitable for investment casting patterns or for the production of soft tooling to make functional plastic parts [59]. Designed to break away easily after part completion,
Figure 1.5: Ballistic Particle Manufacturing BPM, Inc. (U.S.)

Figure 1.6: Fused Deposition Modeling Stratasys, Inc. (U.S.)
the support structures were made from the same material. Commercial systems based on this process were available until 1998, which has since gone out of business.

1.4.1.3.2 Fused deposition modeling (FDM)

In this technique, filaments of heated thermoplastic are extruded from a tip that moves in the x-y plane [60] (figure 1.6). The extrusion head deposits very thin beads of material in a controlled fashion onto the build platform (like a baker decorating a cake) to form a layer. The material is heated just above its melting point so that it solidifies immediately after extrusion and cold-welds to the previous layers [61]. The platform is kept at a lower temperature, so that the thermoplastic quickly becomes firm. The deposited layers can be as fine as 0.005 inch (0.127 mm) thick. After deposition of a layer the platform lowers and the extrusion head deposits the next layer on the previous. Supports are built alongside the deposition where required. These are fastened to the part either with a second weaker material or with a perforated junction.

FDM is a direct digital manufacturing process patented by Stratasys, Inc. Stratasys is renowned as a technical innovator. In 2007, Stratasys supplied 44% of all additive fabrication systems installed worldwide, making it the unit market leader for the sixth consecutive year [62]. The FDM process creates functional prototypes, tooling and manufactured goods from engineering thermoplastics, such as ABS, sulfones and polycarbonate, as well as medical versions of these plastics. Unlike some additive fabrication processes, FDM requires no special facilities or ventilation and involves no harmful chemicals and byproducts.

1.4.1.3.3 Three-dimensional welding (3DW)

In this system an arc-welding robot is used to deposit material on a substrate. Welding based systems are discussed in detail in the next chapter.

1.4.1.3.4 Shape deposition manufacturing (SDM)

This technique is still in its development stages. Molten metal is sprayed on to a substrate in a near-net shape and then unwanted material is removed via numerically controlled (NC) operations [63, 64]. The support material is added in the same way; however this is done either before or after the deposition of the prototype material. This is done
depending on the fact whether the layer contains undercut features or otherwise (Figure 1.2). If the features of the layer are complex, it may be required to add the support material both before and after the prototype material. To remove residual stresses, each layer is shot-peened. The process is then repeated layer after layer. The prototype is shifted from one station to another using a robotized pallet system. Typically stainless steel parts have been produced by this technique using copper as support material. Copper can be removed afterwards by immersion in nitric acid. The prototypes produced by this technique have a structure similar to cast or welded parts and the accuracy of NC milled components.

1.4.1.3.5 Selective Laser Melting (SLM)

Selective laser melting is a generative fabrication process for the manufacture of metallic components (Figure 1.7). The parts are built layer by layer with the metal powder being melted locally by an intensive laser beam that traces the layer geometry. The layerwise design may be compared in principle with that of stereolithography. On a building platform, a thin layer of metal powder is applied, which is locally melted by the laser beam energy. The building platform is lowered by a defined layer height, coated with the metal powder, and melted again. The process is repeated until the component is completed.

The process builds up to 100% dense metal parts direct from design data using customary metal powder. Very fine details are achievable such as thin vertical walls of less than 100µm thickness. Metals like stainless Steel, Titanium, Tool Steel, Cobalt Chrome and Inconel can all be processed. Unlike SLS, the SLM process is hard to control, due to the large thermal stresses and the presence of a liquid pool around the laser spot. Recently Elsen et.al. made an effort to describe the SLM process by making the parameters dimensionless which helped in comparing the results of different research groups [65]. SLM technology is commercially available from Staffordshire based MCP Tooling Technologies [66].

1.4.1.3.6 Electron Beam Melting (EBM)

Electron Beam Melting (EBM), patented by Arcam AB (Mölndal, Sweden) [67] and distributed in the U.S. by Stratasys, Inc. (Eden Prairie, MN) offers another method for
Figure 1.7: Selective Laser Melting (SLM).

Figure 1.8: Electron Beam Melting (EBM).
rapid manufacturing and prototyping metal components (Figure 1.8). It can produce fully-dense metal components with properties similar to or better than that of wrought materials [68]. In this process, powder metal is spread over a vertically adjustable surface. A computer guided electron beam then traces the cross section of the modeled part, first heating then melting and forming the first layer. The surface is then lowered and the process repeated for each successive layer, forming the three-dimensional object modeled. The process takes place in a vacuum to prevent loss of energy that would be caused by the fast-moving electrons colliding with air or gas molecules. Added advantages of vacuum are clean environment (resulting in excellent material characteristics) and higher energy efficiency (lower operating cost and faster build times—five to 10 times greater than laser technology.

1.4.1.3.7 Multi jet modeling (MJM)

In this process a technique similar to inkjet or phase-change printing, applied in three dimensions, is used to build models [69]. A ‘print’ head consisting of a linear array of multiple jets is used to build models in successive layers. Each of the jet individually deposits a specially developed thermopolymer material only where necessary (Figure 1.9). Similar to a line printer, the MJM head shuttles back and forth along the X axis. If the MJM head is narrow than the part being produced, the platform is repositioned (Y axis) and the layer buildup carries on. After completion of a layer, the platform moves apart from the head (Z axis) and the next layer commences. Finally after complete buildup, the support structures are removed to finish the model.

1.4.1.3.8 Precision droplet-based net-form manufacturing (PDM)

This is a droplet-based net-forming RP technique [70]. The capillary instability phenomenon of liquid jets for producing uniform liquid metal droplets is utilized in this process. The microstructure of the deposit can be tailored by controlling the thermal state and mass flux of the droplets. There is no commercial system based on this process.
Figure 1.9: (a) Multi-Jet Modeling 3D Systems, Inc. (U.S.) (b) Multi jet modeling head

Figure 1.10: Selective laser sintering
1.4.1.4 Processes involving discrete particles

1.4.1.4.1 Fusing particles by laser

1.4.1.4.1.1 Selective laser sintering (SLS)

Selective laser sintering (SLS) is a process developed at the University of Texas. It is an indirect process for manufacturing metal parts and molds. It was initially available commercially from DTM Corporation but was later bought over by 3D Systems in 2001. In SLS a fine powder is heated with a CO\textsubscript{2} laser which causes sintering of the powder particles and as a result they are mutually bound. The heating is done to a level so that the surface tension of the particles is overcome and they fuse together. In order to minimize thermal distortion and facilitate fusion to the previous layer the entire bed is heated to a temperature just below the melting point of the material and this is done prior to the commencement of sintering [71]. The laser is modulated in such a way that only the grains which are directly exposed to the beam are affected. After the laser beam has scanned a layer on the powder bed, the bed is lowered. The powder-feed cartridge is then raised and the counter-rotating roller evenly spreads the next layer of powder over the build area. The sintered material forms the part while the un-sintered powder remains in tact as support structure which is cleaned away and recycled after the completion of the part (Figure 1.10). The laser sintering technology (LST), is another process based on the same physical principles. Presently such dual laser systems are available for the processing of thermoplastics and sand. To widen the range of applications of SLS and LST a significant amount of development efforts have been directed towards process optimization [44,71–74].

To receive fully dense ceramic parts a French company ‘Phenix’ has developed a system that completely melts the powder. The so called PM 250 is designed as a high temperature laser sintering device with a fibre laser. According to the company all ceramic powders (and even metallic powders) can be processed. Ceramic powders must be sintered in a subsequent oven process.

1.4.1.4.1.2 Laser engineering net shaping (LENS)

This process involves simultaneous feeding of powder, through a nozzle, and fusing it with a laser on the part bed (Figure 1.11) [75]. The nozzle can be on a side of the bed or
Figure 1.11: LENS process.

Figure 1.12: Three Dimensional Printing MIT, and Soligen, Inc. (U.S.)

Figure 1.13: Multiphase Jet Solidification Fraunhofer Institutes IFAM and IPA (Germany)
coaxial with the laser beam. If it is on a side, the solidified sections cause shadowing areas for the sections to be built. To prevent this, a constant orientation to the part creation direction must be maintained. If the powder feeder is coaxial, inaccuracies may occur in the geometry of the part and the layer thickness if the beam and the powder feeder move out of alignment. Some other systems based on the same principle have also been developed e.g. direct metal deposition (DMD) [76] (commercially available by POM and Trumpf) and AeroMet laser additive manufacturing [77]. Another technique uses wire based alloy in place of metal powder [78]. In a latest attempt laser is used in conjunction with plasma, thus a new direct metal prototyping technology is developed by using hybrid plasma-laser deposition manufacturing (PLDM) for obtaining near net shape full density and good surface intractable high-temperature alloy prototype [79].

1.4.1.4.1.3 Gas phase deposition (GPD)

In gas phase deposition process, a laser is used to produce a solid by decomposing the molecules of a reactive gas [44]. The resulting solid adheres to a substrate to form the part. Three slightly different methods to construct a part have been investigated. The first method is known as the selective area laser deposition (SALD). In this technique only the solid component of the decomposed gas is used to form a part. It is possible by the SALD technique to produce parts made of carbon, silicon, carbides and silicon nitrides. In the second method called the laser-assisted chemical vapour deposition (LCVD), a thin covering of powder is spread in each layer and the spaces between the powder grains are filled in by the decomposed solids. The third method is called the selective laser reactive sintering (SLRS). In this technique the laser initiates a reaction between the gas and the spread layer of powder. As a result a solid part of silicon carbide or silicon nitride is formed. As yet there are no GPD based systems available commercially.

1.4.1.4.2 Joining particles by a binder

1.4.1.4.2.1 3-D Ink-Jet Printing

This technique refers to an entire range of machines that employ ink-jet methodology. The first equipment of its type (Three-Dimensional Printing or 3DP) [80] was developed at MIT and licensed to Soligen Corporation, Extrude Hone, and others (figure 1.12). In 3DP a layer of powder is spread onto a substrate and the cross-sectional pattern of the
layer is scanned by spraying a binder material into the powder using ink jets. Another layer of powder is then added onto the preceding one and the process is repeated. The unbounded powder remains in its position and acts as a support material. Upon completion the unbounded powder is removed and the “green” part is sintered. In best cases the porosity of the “green” and debindered parts lays between 20 and 30% [81]. Typical powders used are ceramic powders such as aluminum oxide, silica, zirconia and silicon carbide. Experiments with stainless steel resulted in a low density of the part (around 78%) and significant shrinkage during the sintering process [82]. Ceramic cores and shells produced by 3DP are successfully being used for investment casting. Soligen, Inc. has commercialized the process for ceramic core and mold applications under the trade name Direct Shell Production Casting. The Z Corp. 3D printing is the fastest additive technology commercially available on the market. In place of single-jet technology it uses inkjet print heads with a resolution of 600 dpi (dots per inch), that focuses on a drop-on-demand approach [83].

1.4.1.4.2.2 *Multiphase Jet Solidification (MJS)*

It is another extrusion-based RP technique (Figure. 1.13). MJS was jointly developed by the Fraunhofer Institutes for Applied Materials Research (IFAM, Bremen) and Manufacturing Engineering and Automation (IPA, Stuttgart). The later is working on the development of software and the former on material aspects. In this process metal or ceramic slurries are extruded using the metal injection molding technique. The slurry is a mixture of wax and metal (low-melting alloys) or ceramic powder mixed in a ratio of about 50:50. It is kept in a heated container and a computer-controlled screw-activated plunger pumps it through an attached nozzle. Parts are manufactured layer by layer and the “green parts” are debinded and sintered to reach final density [84]. The use of ceramic materials for the MJS process is under research at Rutgers University, New Jersey, USA.

1.4.1.4.2.3 *Direct photo shaping (DPS)*

In this process polymerizable compositions are selectively photocured layer-by-layer utilizing a digital micromirror device (DMD™) array [85] as a mask. More than 500,000 microscopic mirrors are integrated in the DMD array that can electronically tilt to reflect
visible light on to the photocurable slurry [86]. DPS based system is commercialized by EPFL in Switzerland.

1.4.1.4.2.4 SPATIAL FORMING (SF)

This technology was developed for prototyping specialized, high precision, metallic, medical equipment within a small build envelop of 2 x 2 x 300 mm. It was conceived and demonstrated as a method of manufacturing parts for cardiac catheter systems [87]. The process combines several technologies to produce solid metallic microstructures from fine powder. Cross section data from computer solid models are used for patterning of a chrome mask which images a lithographic printing plate like those used in the publishing industry. An offset printing press is then used to print “negative” material (the space around the parts) with ceramic pigmented organic ink (averaging 0.5 µm thick) on a ceramic substrate in multiple registered layers. Each layer is successively cured with UV light. After a period of approximately 30 layers, the positive space (non-image voids) left by the printing, (corresponding to the part cross section) is filled using another ‘ink’ which contains metal particles. This material is also UV cured and the surface milled flat. The entire process is repeated until the desired thickness is reached. The semi-finished part is then debinderized to remove organic ink components, and sintered in a furnace with nitrogen atmosphere to join the metal particles. The ceramic negative material either crumbles or removed in an ultrasonic bath. This also results in the separation of the finished part from the substrate.

When designing a part, shrinkage of up to 20 % in all directions caused by the sintering process, needs to be taken into account. Further research includes optimizing the binder removal process and automating the addition of the positive material and the later milling [88]. Since this process generates structures from polymers and particulates, numerous other materials should be adaptable to it, including many metals, ceramics, piezoceramics, plastics, and combinations thereof. So far, no commercial system is available and only extruded parts with a constant cross section can be produced. In theory, however, completely arbitrary geometries should be feasible.
1.4.1.5  Joining solid sheet

1.4.1.5.1  Laminated object manufacturing (LOM)

Laminated Object Manufacturing [21,89] is a process based on the principle of lamination. In this technique shapes are build with layers of paper or plastic (Figure 1.14). The binding together of the laminates is brought about by means of a thermally activated adhesive. A heated roller is used to glue the laminate to the previous layer [90]. The outline of the part cross-section for each layer is then cut using a laser. The unwanted material of each layer is scribed by the laser into a cross-hatch pattern of small squares. As the process is repeated layer after layer, the cross-hatches build up into tiles of support structure. After the part completion the support structures are removed and the tiled cross-hatched structure facilitates this removal. Since only the layer contours are scanned, large parts can be built by LOM in relatively lesser durations. However, the removal of the excess material from LOM models can still be a lengthy and tedious task. A method that speeds up and simplifies this process has been reported by Karunakaran et al. [91]. This technique was developed and commercialized by Helisys Corporation (U.S.), though it ceased operation in 2000. However the company's products are still sold and serviced by a successor organization, Cubic Technologies.

1.4.1.5.2  Paper lamination technology (PLT) or the solid center (SC)

This technique employs the same basic building approach as that used for the LOM machine. However, the Kira Solid Center (SC) machine [92] (Figure 1.15) is operated in a significantly different manner. A conventional laser printer is used to feed standard printing paper to the SC machine. An adhesive-based toner is used by the printer to print the outline of the cross-section. In addition a cross-hatched bonding pattern is also printed on each piece of unwanted paper. A hot plate then laminates the paper to the previous layers. A carbide knife, mounted on a swivel base, then cuts the cross-sectional outline. Additionally, to facilitate the removal of the support material, segments of “parting-plane” sections are also cut.

1.4.1.5.3  Solid foil polymerization (SFP)

In this technique semi-polymerized foils, soluble in monomer resin, are used to build up a part. The foil solidifies and bonds to the previously deposited layer on exposure to UV
Figure 1.14: Laminated Object Manufacturing Helisys, Corp. (U.S.)

Figure 1.15: Solid Center Kira Corp. (Japan)
light. Once the cross-section has been exposed, a new foil can be applied. The unexposed areas of foil acts as support structure and are later on easily removed by dissolving in the resin [41, 93]. There are no commercial systems available as yet.

1.4.2 Machining-Based RP Approaches

In this section, the major developments in machining (MR) based RP approaches are briefly reviewed.

1.4.2.1 Sculpturing robot (SR) system

This technique, at the Delft University, employs a six-degree of- freedom industrial robot in conjunction with a three-axis milling to build CAD models [14]. The raw material stock is mounted on a turntable. A milling device is mounted on the end effector of the robot. This device constitute of a milling tool and a tool holder. A six-grid voxel data structure is used to generate the tool path. An interference-free milling tool path is generated by applying the Minkowski operations [94]. An interactive simulation of the milling process is also implemented [95]. The SR is a subtractive RP system without the involvement of any layer. It machines a part from a single piece of stock.

1.4.2.2 Computer-aided manufacture (CAM) of laminated engineering materials (LEMs)

The CAM of LEMs (CAM-LEM Inc., OH) is a commercialized hybrid RP process [20]. It is capable of fabricating metal or ceramic tooling using pre-formed stainless steel or ceramic sheet material. The sheet material is made from steel or ceramic powders, held together using a binder. Part slices are cut from the stock sheets using a high-power laser with five degrees of freedom. These slices are then assembled together with the help of a selective-area gripper and registration system (pins and holes) in a correct relative position. To obtain a monolithic structure the laminated model is fired after assembly. The CAM of LEMs parts have a relatively low accuracy. The error is mainly caused by two reasons: one is the post-assembly process; and the other is unpredictable shrinkage due to the firing process. The typical shrinkage rate is as high as 12-18 per cent. The potential source of error can be the registration system which destroys the continuity of
material properties within the final part. The build envelope of the available CAM of LEMs machines is also limited to a cubic space of side 150 mm.

1.4.2.3 Solvent welding freeform fabrication technique (SWIFT)

This process is based on solvent welding and CNC contour machining [24, 95]. As is the case with most RP processes, SWIFT builds parts one layer at a time. A laser printer applies a thin film and prints a high-density polyethylene (HDPE) image for each layer. The HDPE image acts as a solvent mask and specifies areas of the downward surface for welding to the previous layer. Acetone solvent is then applied to the bottom side of the sheet and stacked onto the previous layers. A pressing force is then applied which welds the current sheet to the previous. The current sheet is then machined to the required shape with a 3-axis CNC machine. The part is completed by repeating the above cycle for all the layers. SWIFT has cost, accuracy and speed advantages over most commercially available RP processes. Although, a large geometric error exists due to the uniform stock layer thickness which is limited by the feeding system in SWIFT and the zeroth-order or first-order approximation to the design surface.

1.4.2.4 Millit package

Millit [17] is a commercial software package which generates a numerical control (NC) tool path from STL or 3DS (three-dimensional slice) files. With the help of a build orientation, it decomposes a model into layers free of undercuts, and these layers are called ‘components’. Dual-sided milling can be employed to mill each component from the base material. Post-assembly operations are then facilitated by adding borings to the components’ mating surfaces. All the components are laid out on the raw material sheets with user-defined thickness. The leftover sheet material is called a ‘frame’. The frames are connected to the components via bridges, which also act as fixtures during the milling process. To ensure the accuracy in dual-sided milling, positioning borings are added to frames. The components are removed from frames, after milling, by breaking the bridges. These components are then assembled by inserting fitting pins into the borings. To complete the final physical model, material-specific bonding is needed.

The Millit process is far from an automatic manufacturing technology. A number of operations have to be done manually, i.e. the design of borings and bridges, the turnover
of frames during dual-sided milling, and the extraction and post-assembly of components. In addition, the registration of frames and components and the post-assembly operation are potential error sources.

1.4.2.5 Thick layered object manufacturing (TLOM)

The TLOM project was carried out in order to achieve $G_1$ (tangential) or $G_2$ (curvature) surface continuity in building large-volume models [96]. The main feature of TLOM is the cutter, which is a deformable blade and is known as flexible blade tool. A pair of electromechanical rollers controls the blade. The blade profile can be adjusted instantaneously to fit the local shape of the model features. The layered manufacturing process is called hot blade cutting. First of all the model is sliced into various thick layers. The blade profile and tool path is then calculated for each layer and the layer is shaped accordingly. After all layers are made, they are manually assembled into a physical model and finishing touches if needed are performed. The complexity of models that can be built by this technique is limited due to the clumsy tool profile controller assembly. The final part geometry may have errors due to the introduction of post-assembly.

1.5 Conclusions

The ability to rapidly produce the designs of parts and objects with complex 3D geometry is a basic necessity in today’s market place. RP has emerged over the last 20 years based on the principle of creating 3D geometries using computer aided design (CAD), directly and quickly with little human interaction. It is a technology used in facilitating concurrent engineering, thus it has started to change the way companies design and build products. The main goal is to reduce product development, manufacturing costs and lead times, thereby increasing competitiveness. In the last two decades impressive steps towards this goal have been made. However the field of RP is still developing, with much effort being expended on improving the speed, accuracy and reliability of RP systems and widening the range of materials for prototype construction. Another area for improvement is the costing, as most RP systems are currently too expensive to be affordable by any but the larger firms. However, RP technology is available to most companies via the services
provided by different bureaus. The future is likely to see more user-owned RP machines as their costs are reduced.

To date most RP parts are used for prototyping or tooling purposes; however, in future the focus may be to produce parts as end-use products. In this context the term ‘rapid manufacturing’ (RM) is used in which RP technologies are employed as processes for the production of end-use products. The introduction of non-polymeric materials, including metals, ceramics, and composites, represents a much anticipated development. These materials would allow RP users to produce functional parts. Today’s plastic prototypes work well for visualization and fit tests, but they are often too weak for function testing. More rugged materials would yield prototypes that could be subjected to actual service conditions. In addition, metal and composite materials greatly expand the range of products that can be made by RM. In this context the focus of the current work is on metals; specifically with respect to 3D welding approach, with main emphasis on the material properties and deformations due to the high temperatures involved in welding. A more detailed discussion on the welding based systems is presented in the forth coming chapter.
Chapter 2

Use of Welding as a Rapid Prototyping Tool.

Rapid prototyping (RP) is one of the fastest growing automated manufacturing technologies with the capacity to go directly from CAD models to finished components. In layered manufacturing (i.e. the methodology of RP), three dimensional parts of arbitrarily complex geometries are build by sequential deposition of material layers. One of the main advantages is the ability to build complex parts in a very short time with very little human intervention. The present and future emphasis is to produce “form-fit-functional” parts rather than prototypes for visualization thus ultimately leading towards the concept of rapid manufacturing. Manufacturing of metallic parts with good accuracy and mechanical properties is one of the main aims of solid freeform fabrication (SFF). Different deposition process are being developed and studied all around the world. Welding as a deposition process has shown promise for RP of metallic parts. It has been recognized that the principles of the welding process could be used in the development of a cost effective method for layered deposition manufacturing of fully dense metallic parts and tools. This chapter presents a detailed review, concentrating on the current research and developments in the use of welding systems as a RP tool. A report related to this chapter has been published in [98].

2.1 Introduction

Today’s global markets require rapid product development and manufacture of new designs. Different tools, for visualization purpose, can play a major role in the development of a final product, from an initial conceived idea. The fast development in the field of computer graphics provides good tool for the three-dimensional representation of the part to be manufactured. In the last 10 to 15 years lots of research
and progress has taken place in the field of rapid prototyping [99-103]. The basic idea of RP is to build an artifact through the repeated layer additions by converting complex 3D geometry into simpler 2.5D representation, which does not require any part-specific fixtures or tools [104]. The typical advantages of RP are less turn around time, ease to manufacture complex geometries, and ease of automatic planning, specifically for smaller lot size [104,105].

At present, a number of RP techniques are in the developmental stage, but a few are available commercially. These current techniques including stereolithography, laminated object manufacturing, fused deposition manufacturing, 3D printing etc. can only produce prototypes made from wax, plastic, nylon, and polycarbonate materials. The properties of these materials are far from those required in the final products, especially for metallic parts. These products therefore can be used only as models, giving a feel of the actual part, or at the most may be used to check the fit up in certain assemblies. The present and future efforts are primarily focused to manufacture fully functional components rather than prototypes for visualization [106]. But there are many problems related with the production of structurally strong and dimensionally accurate metallic parts. For this purpose research is going on in the fields of materials [107,108], tolerance [109], software and system design [106,109-112].

2.2 Metal based RP

The main methods for metal based RP include sintering [98], laser deposition [101], and brazing/soldering. Each of these techniques has their shortcomings. Some might cause inclusion of unwanted materials or require post-processing to achieve fully dense parts [100], others cause accumulation of residual stresses thus causing warpage, delamination or poor surface quality. Although among the different RP techniques, laser deposition systems are more flexible but are expensive, bulky and energy inefficient.

Another process for metal part RP is the droplet based manufacturing. This process includes different techniques for metal deposition such as thermal spraying, microcasting and welding [111,113-117]. The first technique produces parts having porosity, and comparatively lesser mechanical strength. Drawbacks associated with the microcasting are relatively expensive setup, stair-step surface texture which leads to poor precision and
dimensional accuracy, as well as warpage and delamination caused by the buildup of residual stresses. An improved manufacturing technique based on microcasting is called Shape deposition manufacturing [113,114,63]. In this process molten droplets are produced in a non-transferred mode characterized with low substrate temperature. This process requires a relatively expensive setup and still does not have the capability to control the uniformity of the metal droplet size and its detachment rate.

Welding on the other hand has better control of size, flux, velocity, and trajectory of the droplet. However, because of low resolution, parts produced are generally ‘near net shape’ (Figure 2.1). A hybrid RP process, based on welding and milling shows good promise to produce fully dense, metallurgically bonded, dimensionally accurate and cost effective metallic parts and tools [106,109].

2.3 Welding as a RP tool

The use of welding for creating free standing shapes was established in Germany under the process name Shape Welding in the 1960’s. This led to companies such as Krupp, Thyssen, and Shulzer developing welding techniques for the fabrication of large components of simple geometry, such as pressure vessels which could weigh up to 500 tons. Other work in this area has been undertaken by Babcock and Wilcox under the process name Shape Melting; that was used mainly for building large components made of austenitic materials [118]. Also, work by Rolls-Royce has centered on investigating 3D welding as a means of reducing the waste levels of expensive high performance alloys which can occur in conventional processing. They have successfully produced various aircraft parts of nickel based and titanium based alloys. Research work on 3D welding has been in progress at the University of Nottingham, UK [115-117], University of Minho, Portugal, University of Wollongong, Australia [119-121], and Southern Methodist University, Dallas, TX [122-125]. Two more research groups, one from Korea [126] and another consisting of researchers from Indian Institute of Technology Bombay and Fraunhofer Institute of Production Technology and Automation [127] presented their conceptual ideas of combining a welding operation with milling.

While extensive documentation and experience is available on selection and optimization of welding parameters for the production of welds with exceptional joint quality, this
Figure 2.1: Tube-shaped “near net shape” RP part made by GMAW [109].

Figure 2.2: Parts produced by VPGTAW based RP (a) cone (b) pipe reducer [129]
information cannot be successfully applied for the rapid prototyping process since the criteria for "good weld" and “good prototyped layer” differ significantly. For example, the requirements for build-up height, penetration depth into the previous layer, and the ratio of these two variables are very different for welding, and rapid prototyping by welding. The welding process suitable for 3D welding needs to meet several requirements. The process should provide low heat input into the work piece in order not to destroy the underlying geometry, and in order to generate the lowest possible level of residual stresses. Furthermore, shallow depth of penetration; just enough to provide sound metallurgical bonding between layers, and high material buildup are desirable features.

The welding based RP machine consists of a mechanical system and a welding system. The mechanical system includes a 3D positioner and a welding torch holding mechanism. The mechanical system is computer controlled to provide motion to the welding torch with the desired path. Feedback control is established by means of thermocouples, high frame rate imaging or laser based control systems. This feedback system monitors the levels of preheat, part temperature, pulse rate, droplet size and droplet detachment. A grit blasting nozzle may also be installed to minimizes the oxidation of the part with a suction pump and vacuum nozzle to remove excess water vapour and grit [118,128].

Among the available welding processes two seem to have potential for use in the RP process i.e. gas metal-arc welding (GMAW) and gas tungsten-arc welding (GTAW). These processes remelt the substrate resulting in excellent metallurgical bonds and shielding the weld puddle guarantees oxide free material. Excessive heat from the arc in combination with slow cooling rates creates relatively large heat affected zones with disrupted microstructure. Plasma-arc welding (PAW) is a technique derived from GTAW. In a process called “Shape Melting” [128] PAW was used to incrementally build engineering parts. The shapes created with this process have been limited in size, complexity and geometry.

It is important to note that most of the research work focuses on the RP of parts made of steel. However, aluminium alloys also have a wide spread history of applications in the industry. A new deposition process for directly building cylindrical parts of the 5356
aluminium alloy was developed at the Southern Methodist University Dallas, TX, with the variable polarity gas tungsten arc welding (VPGTAW) [129]. This process can be used to produce parts of variable thickness, but cannot yet be used to produce asymmetric samples (Figure. 2.2).

NASA’s Langley Research Center, Houston, TX. and Johnson Space Center, Hampton, VA. have developed a solid freeform fabrication system, called the electron beam freeform fabrication (EBF³), utilizing an electron beam energy source and wire feed stock in a vacuum environment [130,131]. This rapid metal deposition process works efficiently with a variety of weldable alloys. Thus far, this technique has been demonstrated on aluminium and titanium alloys of interest for aerospace structural applications, however, it can be expanded to nickel and ferrous based alloys (Figure. 2.3).

2.4 Hybrid RP process based on welding and milling

In spite of all the improvements and control of the welding process, parts produced by welding are generally of ‘near net shape’. RP process based on 3D welding alone does not provide satisfactory dimensional accuracy and surface quality. Because of complete melting, the accuracy as well as the surface quality of the parts is generally lower than that of machined parts. To overcome this difficulty, a combination of welding as an additive process with a subtractive technique such as milling is an appropriate solution [132]. A computer controlled hybrid RP system integrates the controlled welding process, which provides the controlled heat and mass transfer and precision control of bead penetration with a CNC end and face milling operation. This system offers a way of building metallic parts in layered fashion with full density, high mechanical and metallurgical properties, high dimensional accuracy and good surface quality with complex geometrical features and sharp edges [106] (figure 2.4, 2.5). Song [126] proposed to combine welding and 5-axis CNC milling for direct prototyping of metallic parts while Karunakaran [127] has proposed to combine welding with 2-1/2 axis milling, where complex shapes of the layers will be obtained by using angle cutters. The brazing process is proposed to deposit the masking material at the edges of each layer in order to allow the formation of overhangs. The technical and economic viability of the hybrid process has been proved through a real life case study by Karunakaran et.al. [133].
Figure 2.3: 2219 Al shapes built using EBF3 [131]

Figure 2.4: Sample made by hybrid RP based on GMAW and CNC milling [106]
Figure 2.5: A sample with complex network of conformal channels made by hybrid RP based on GMAW and CNC milling [106]
and cost of tool making and the material cost was found substantially lower than that of CNC machining and other RP methods.

2.5 Areas under Research

At present no welding based RP equipment is commercially available. The most of the work going on in this regard is in the development / research stage. The main research work available in the literature is focused on, process planning, thermal control, control of metal transfer, and simulation and modeling. The ongoing research in these areas will be discussed here.

2.5.1 Software development and process planning

Most RP processes begin with a design created on a CAD surface or solid modeler. An output file is generated (e.g. an STL file) which approximates the part surface. The geometric model is then transformed into slices by a computer program, to generate cross-sectional layers for constructing the part layer by layer. However the welding process operates quite differently. For example the welding process has ignition and the weld pass is wider. If the same interface, used for conventional RP processes, is used between the geometric modeling data and deposition system in a weld based RP system, the quality and accuracy of the product is not acceptable. Therefore a separate interfacing process is required for weld based RP. This is usually done by using computer numerically controlled (CNC) programs generated directly from the CAD files instead of employing slice data. The interface for welding deposition systems are usually based on the IGES format, which is compatible with most solid modeling environments [109] (figure 2.6). However based on the methods of data processing for STL model, aimed at one multi-parameter, coupling strongly complicated welding process, one procedure “Shaping-Repair-Slicing-Compensation-Planning-Machining” was proposed by Ling-na et.al. [134].

In RP systems, there are different imaging strategy (deposition patterns) approaches used to define the image of a single layer. One approach is to use a raster, where the image is created as a series of contiguous or overlapping straight line segments, as shown in Figure. 2.7(a),(b). Another is to draw at least the outline of the image, as illustrated in Figure. 2.7(c) and then fill the interior with the raster pattern (Figure. 2.7(d)). The
Figure 2.6: Flow chart for a welding based RP system [109]

Figure 2.7: Imaging strategies (a) raster fill for laser (b) raster fill for welding (c) vector fill for outline (d) imaging strategy for welding RP [109].
fundamental trade-off here is between speed and precision. As the width of the weld pass is much larger than that of other RP methods, it is very important to follow the second strategy [109]. Otherwise, the surface error will be significant. The imaging strategy used to deposit a layer of material has a significant effect on the deflection of the manufactured part. When depositing in a raster path, material should not be deposited parallel to the longest part dimension. Because the curvature (deformation) is greatest parallel to the deposition direction, depositing parallel to the longest part dimension would result in greater warping deflections and loss of tolerance [135,136]. A third strategy, the spiral pattern, is for rectangular or circular sections, where the image is scanned either from inside to outside or from outside to inside as shown in figure 2.8. Among all these strategies the spiral pattern scanned from outside to inside produces low and uniform deflections [137]. Fessler et al. [101] reported that continuous bidirectional raster deposition of stainless steel and Invar produced significantly more deformations as compared to a tower technique (depositing in alternate lines then filling in between). Similar technique also termed as the spiral overlay welding was employed by Spencer et al. [117] at the University of Nottingham, UK, in which multiple beads were used to enable the production of parts wider than normally possible from a single bead.

The raster fill method for welding is quite different because each weld pass has ignition and ending process, which do not exist in other RP systems. Extra procedures are necessary to maintain the accuracy and quality of the beginning and ending portions of the pass. Due to the heat sink, the penetration is lower at the start. Thus the thickness is higher than in the normal portion of the weld. At the end of the weld pass, because of the flowing of the melted metal, the slope shape is created gradually. These uneven portions of the parts influence succeeding layers (Figure 2.9a). The control method therefore, for the start and end portions of the weld pass are managed by applying different deposition parameters at these positions. In the start portion of the weld pass, the current and travel speed decreases from higher values to the normal ones. In the end portion of the pass, the current and travel speed reduces gradually (figure 2.9b).

2.5.2 Temperature control
High heat inputs during welding effects the surface finish and thus the part quality. Surface finish can be improved using simple temperature control to prevent the
Figure 2.8: Spiral pattern (a) scanned from inside to outside (b) scanned from outside to inside.

Figure 2.9: (a) Sample showing accumulated error at start and end. No. start and end control is used (b) Sample deposited with control of the start and end portions. [109].
deposition of the next layer of weld if part temperature was too high. Excess residual heat not only affects surface finish but also the bead height, due to delayed solidification. This results in large quantities of porosity, poor surface finish and increased material flow. The reduction in weld bead height causes the programmed step height to exceed the actual bead height. The resulting increase in arc to electrode contact tip length can have a significant influence on the quality of weld.

Simple temperature control techniques can help improve surface finish. Spencer et.al [117] has reported the use of an infra-red remote sensor, to monitor the starting temperature of a weld pass. Improvements to the surface finish of three-dimensional welded parts have been achieved, but this was at the expense of reduced process efficiency. Thus although, it is critical to part quality, temperature control significantly slow the three-dimensional welding process. Klingbeil et.al [132,136] has reported that mechanically constraining the substrate, as well as preheating and insulating it, can give substantial payoffs in limiting residual stress-induced warping. The implementation of cooling methods and techniques require extensive study to find the effects these may have upon the properties of the parts produced.

2.5.3 Controlling the metal transfer process

Weld parameters selected for deposition are important in the control of high temperatures. The importance of understanding and controlling the metal transfer process is the key issue in controlling the quality of the resultant weld and/or generated layer of metal for rapid prototyping. Modulation of welding current is used extensively in arc welding for control of droplet detachment and transfer. Existing methods for metal transfer control in welding have two major flaws: uncertain detachment instant and inconsistent droplet size. These methods, specifically for pulsed arc welding, rely on the one-drop-per-pulse (ODPP) approach by properly selecting the duration of the peak current. To guarantee the detachment, the peak current has to be larger than the transition current. However, in addition to detaching the droplet, such a high current causes superheating of the droplet, results in a very high impact speed of the droplet, deforms the underlying geometry, and yields irregular droplet volume. The detachment control
approach requires elimination of uncertainty in the detachment instant associated with the conventional method, as well as the capability to control the size of the metal droplet.

Many sensing systems have been developed in this regard. e.g. sensors monitoring current or voltage signal, acoustic sensing, video imaging and laser based structured-light machine vision. Out of these the last two systems have been reported in literature for use with GMAW based RP [106].

In the video imaging technique real time image processing is used to regulate a closed loop control of welding process. A high frame rate vision system is used to monitor the metal transfer process. Information obtained by the real time image processing is used as feedback in the closed loop control (figure 2.10). This technique ensures the ODPP transfer mode and constant arc length, adjusts automatically pulse rate and size, controls droplet size, and senses droplet detachment. The welding process remains stable even for very low current, which makes this technique suitable for 3D welding. Control of up to 10 droplets per second has been reported successful by Kovacevic and Kmecko [106].

One way to make metal deposition more accurate is to keep the arc length short so that a detached droplet cannot be diverted out of the weld pool. Laser beam based control system placed across the arc manages the arc to any desired length (Figure 2.11). An arc can be just long enough to avoid short circuiting. Besides, this technique ensures the ODPP transfer mode, adjusts automatically pulse rate and size, controls droplet size, and senses droplet detachment. The control device consists of a laser diode and photo diode positioned in such a way that the laser beam passes through the arc (Figure 2.11) [106].

With these control systems droplet transfer rate can be adjusted automatically based on wire feed speed. It is especially useful in 3D welding situations where for a short period of time the quantity of metal added to the weld pool has to be reduced (e.g. corner) or increased (e.g. root filling). The use of these control techniques allow welding with low heat input, reduced spatter and shallow weld penetration.

Another technique, i.e. mechanically assisted droplet transfer, separates the control of arc energy and arc force in a way to improve the welding quality and to obtain the projected metal transfer mode. In this technique the welding system is composed of an oscillating wire feeder which exerts an additional force on the metal transfer process. It was reported
Figure 2.10: Experimental setup video imaging [106]

Figure 2.11: Experimental setup laser based control [106]
that an appropriate electrode vibration produces an additional mechanical force to shake droplets off the electrode, and the minimum current for welding of thin sheets can be reduced by 10-20 percent compared with pulsed arc welding [138]. The additional mechanical force makes the droplet size smaller and produces a high droplet transfer rate. The oscillating frequency plays a key role in the droplet transfer process, thus resulting in high surface quality with a uniform spatter-free appearance, which is of significant importance for weld based RP [139]. The use of this dabber technique with GTAW had made precision edge build up possible with wall thickness of as minimum as 1 mm [140].

2.5.4 Simulation and modeling

The main aim of research in modeling of RP process is to simulate the whole process so as to predict different process parameters based on minimization of certain process constraints like deformation, delamination, warping and residual stresses. Moreover control of penetration depth and high temperature gradients are also important to produce parts with uniform properties and microstructure. The modeling of welding as RP process is very much similar to the multi pass welding simulation with few differences like deposition on substrate, in contrast to conventional welding where material is deposited in groove. Since the complete welding simulation required for RP is computationally very expensive therefore the major effort has been on modeling the process based on assumptions such as the bead is deposited at once or even the layer is deposited at once. Other simplicities are the simulation confined to a droplet level deposition, limiting the model to a 2-dimensional analysis or considering it to be an axi-symmetric case [132-137, 141-146]. The simplified approaches can be useful for understanding the physics behind the deposition process and can give some qualitative estimations regarding deformations, stresses and effects of boundary conditions etc. but the actual deposition process is complicated due to the presence of a moving heat source, material addition in a 3D space and thermal cycling. Various analyses have shown that the deposition patterns, bolting conditions and base thermal conditions, affect the buildup of residual stress and deformations. For parts with insulated and constrained base, the deformation is reported to be the minimum [145]. Conflicting results about the effects of deposition sequences on warping were reported [132-137,143]. A complete 3D model of deposition process has
been developed to predict the deformations and residual stresses caused by the deposition [146].

**2.6 Need and Scope of Present Work**

Since the introduction of rapid prototyping (RP) in 1986, several techniques have been developed and successfully commercialized in the market. However, with a few exceptions, most commercial systems currently use resins or waxes as the raw materials. As a result, due to limited mechanical strength for functional testing, this is regarded as an obstacle towards broader application of RP techniques. The present and future emphasis is to produce “form-fit-functional” parts rather than prototypes for visualization thus ultimately leading towards the concept of rapid manufacturing. To overcome this problem, direct metal deposition methods are being investigated worldwide for RP and even for rapid tooling applications. The commercially available metal based Solid Freeform Fabrication (SFF) systems mainly rely on lasers or electron beam technology to melt the metal for layer buildup. These technologies are comparatively very expensive and have relatively slower buildup rate. Welding as a deposition process has shown promise for RP of metallic parts. Welding has been in existence before current developments in RP started, and is a much more mature technology. It has been recognized that the principles of the welding process could be used in the development of a cost effective method for layered deposition manufacturing of fully dense metallic parts and tools. However, welding is one of those processes where high heat input results in large thermal gradients; these thermal gradients along with the mechanical constraints cause the build up of residual stresses, distortions and hence out of tolerance parts. In spite of all the improvements and control of the welding process, a part produced by 3D welding alone does not provide satisfactory dimensional accuracy and surface quality. Because of complete melting, the accuracy as well as the surface quality of the parts is generally lower than that of machined parts. A post process (e.g. intermediate machining) is therefore required to address the precision issues.

Thus adopting a weld base RP system will result in facing of the following consequences which are conjugate to welding:

- Involves high heat input with large thermal gradients.
Large Substrate Re-melting thus transforming the microstructure.
High Temperature resulting in the Loss of Form.
Inadequate Surface Finish.
Large Deformation.
Residual stresses caused by thermal gradients.
Residual stresses induce warping.

Therefore keeping in view the above consequences the deposition parameters should be selected in consideration with the following criteria:

- depth of remelt should be shallow (for just reasonable metallurgical bonding).
- reasonable buildup height to remelt depth ratio.
- Low heat input imparted to the work piece to
  - prevent destroying underlying geometry
  - generate lowest possible residual stresses.

In view of the disadvantages and prototype layer requirements, it is quite logical to improve the performance of the process by studying the response of various process parameters so that the detrimental effect of the problematic issues can be minimized. In this regard the present research was divided in two main categories i.e.

- Structural investigation of the weld base RP part (comparison between direct deposition and deposition with intermediate machining).
- Study of different process parameters and their effects on deformations and residual stresses with the help of FEM.

However the following points further elaborate the outline of the present work:

- Development of a deposition system employing a welding machine and a conventional milling machine.
- Experimental Testing for Structural Investigation of RP Parts (investigation of the structural homogeneity on a microscopic level and its relation with hardness and tensile strength for 3D welding in comparison with hybrid welding / milling based RP).
- General Welding Simulation.
➢ Experimentation for Model Validation.
➢ Finite Element Analysis (Thermo-Mechanical analysis of metal deposition using FEM).
➢ Study of Process Parameters (Parametric analysis to study the effect of process variables).

2.7 Conclusion

Excessive energy transfer into underlying material by direct transfer of the welding arc causes penetration of the weld puddle and disrupts the geometry and microstructure of previous deposits. This is the major hurdle in producing dimensionally accurate parts by welding. Although VPGTAW and EBF$^3$ shows good prospects for future development, the hybrid system based on welding and CNC machining is a good approach towards producing structurally dense metallic parts with dimensional accuracy.
Electric arc welding process has been in use for steel fabrication since late nineteenth century. However the commonly used arc welding processes, GTAW and GMAW were brought in commercial use in early 1940s. The metal deposition by arc welding process is a highly non-linear and complicated phenomenon mainly due to non-linear thermal and structural material properties. The process involves temperature gradients of thousands of degrees, over a distance of less than a centimeter, occurring on a time scale of seconds, involving multiple phases of solids, liquids, gas and plasma [146]. The modeling of deposition process requires an integration of welding simulation and layer build-up by successive material addition. This chapter covers the theoretical background of the finite element modeling of welding. Various physical, metallurgical and numerical aspects such as governing equation of thermo-mechanical analysis, finite element (FE) matrix formulation, coupling (thermal–metallurgical-structural) between different fields, temperature dependent material properties, heat source modeling, addition of filler material, finite element analysis procedure and numerical aspect of simulation etc are discussed.

3.1 Introduction to FEM

Finite Element Method (FEM) is one of the most accepted and widely used tool for the solution of non linear partial differential equations which arises during the mathematical modeling of various processes. Originally developed for the analysis of aircraft structures FEM is now applicable in all fields of engineering and applied sciences like heat transfer, fluid dynamics, vibrations, magnetism etc. Figure 3.1 presents a block diagram for the procedural steps involved in FEM. The origin of the process is the mathematical model
Figure 3.1: Schematic Finite Element Methodology
from which everything originates. The mathematical model usually comprise of an ordinary or a partial differential equation. A discrete finite element model is generated from a variational or weak form of the mathematical model. The FE equations are processed by an equation solver, which delivers a discrete solution (or solutions). The concept of error arises when the discrete solution is substituted back in the “model” boxes. This replacement is generically called verification. The solution error is the amount by which the discrete solution fails to satisfy the discrete equations. This error is relatively unimportant when using computers, and in particular direct linear equation solvers, for the solution step. More relevant is the discretization error, which is the amount by which the discrete solution fails to satisfy the mathematical model. This error is reduced by remodeling the discretization process so that the FE model is as close as possible to the mathematical model.

3.2 Physics of Arc Welding

The phenomenon of arc welding is generally described by the flow of electrons due to the applied electric field between anode and cathode. The other two contributing factors are ionization gases surrounding the arc (shielding gases) and magnetic field around them. A thin layer of electrons from the surface of cathode accelerates towards the anode at the application of an electric field. This electron flow initiates an arc which results when an initial threshold level is attained. This threshold level is usually provided by some special arrangement such as short circuit between anode and cathode or through high frequency arc start. Temporary arc initiation arrangement increases temperature as well as current for a short time which is subsequently maintained under normal welding conditions. These flowing electrons collide with the atoms of surrounding gas and cause ionization by decomposing them into electrons and positive ions. The electrons obtained from decomposition of gas promote further ionization. The current of the electrically charged particles in the arc and the temperature are interrelated. A high temperature increases ionization and the increased ionization raise the temperature due to the released energy [148]. The ionization potential of the shielding gas is believed to contribute significantly towards the temperature of the arc. However, thermal conductivity of the shielding gas can alter the rate of heat transfer between anode and cathode. Similarly the reactive
thermal conductivity (due to dissociation of diatomic gases, e.g. Hydrogen, in the plasma and recombination in the gas boundary layer) can be another source of increased heat dissipation during welding.

The magnetic field around the arc directs charged particles to the center of the arc, causing localized arc spots on the anode and cathode and thus resulting in high temperatures at these spots. This heat dissipation is discussed in detail in [146]. The weld pool formation on the base metal is due to a combined effect of droplet formation of the filler metal and fusion of the base metal beneath the arc spot which melts instantaneously. The other associated factors such as shape of the weld bead, penetration depth and formation of crater etc. are based on the complex interdependence of welding parameters, plasma pressure and fluid flow in the weld pool. Convection is the primary mechanism for heat transport in the weld pool. The flow characteristics in the weld pool depend on several phenomena such as plasma pressure, electromagnetic forces, buoyancy forces and temperature dependant surface tension forces etc.

3.3 Computer Simulation of Welding

In the past considerable amount of work has been reported in field of welding simulation. Different aspects studied in this area are:

- Simulation technique - Efficiency and integration (computational strategies such as fully coupled, sequentially coupled etc. and their integration with optimized procedures)
- Heat source modeling (point heat source, multi-point heat source, line heat source, double-ellipsoidal power density distribution heat source model for arc welding and Gaussian heat distribution model for laser welding etc.)
- Numerical aspects (effects of element type, mesh size, time step size, integration technique, convergence criteria etc.)
- Material modeling of welding (micro-structure investigation and its dependence on temperature history, effect of thermal and mechanical properties on structural response etc.)
Investigation of welding residual stresses and distortions in structural components and implementation of mitigation techniques

The simulation of welding process first appeared in early 1970’s. A most recent and comprehensive review regarding the developments in welding simulation was reported by L.E. Lindgren and was compiled in three parts [149-151]. A brief description of the major contributions and current status of the research in this field is given here.

The model presented by D. Rosenthal [152] for a moving point heat source is considered the first step towards the simulation of welding. It presents an analytical solution of transient temperature distribution in arc welding. He presented a linear two and three dimensional heat flow in solid of infinite size or bounded by planes. The analytical model was validated with experimentally measured temperature during welding of plates of different sizes. A good agreement between calculated temperature and the experimental data was claimed. Transient temperature determined by using this model was applied as thermal load and structure analysis was performed directly. However, subsequent studies revealed that the model gave good approximation of temperature only in the far fields; whereas, in close proximity to the heat source the predicted temperature field was a bit higher. Other heat source models, for example multiple point heat sources by Rybicki et al. [153] and Debiacari [154], presented a better approximation of the transient temperature distribution. In another study, Seo et al. [155] used a line heat source to model heat input during welding. A predefined temperature at specified locations had also been utilized in some studies e.g. Goldak et al. [156]. A very valuable contribution for arc modeling is made by Goldak et al. [157-158] who presented a so-called “double ellipsoidal heat source model”, having Gaussian distribution of power density in space. In the present work of welding simulation the double ellipsoidal heat distribution scheme is generally utilized, however minor adjustments are made when required. In addition to the above mentioned heat source models a lot of other heat distribution patterns such as Sabapathy et al. [159] and Ravichandran et al. [160] are also in use.

Ueda and Yamakawa [161] and Hibbitt and Marcal [162] are among the pioneers who initiated application of FE technique on simulation of welding. The complexities they encountered during simulation restrained the other researchers from entering into this
field. Subsequently Friedman [163], Rybicki et al. [153] Andersson [164] presented their work which further explained the methodology of simulation. They presented a basic methodology of simulation and used the sequentially coupled analysis technique. Temperature dependant material properties were used in all these analysis and latent heat associated with liquid-solid phase transformation was also accounted for. Both Friedman [163] and Andersson [164] analyzed butt-welding of plates and used plane strain formulation with half sectional model for thermo-mechanical analysis of welding.

After the initial work, described above, there was a deluge of simulation work in which finite element technique was implemented. Towards the numerical aspect of simulation, numbers of experiments were made to evaluate the effect of modeling techniques, mesh intensity, element types, modeling of filler materials, numerical integration procedure and type of solvers etc. Similarly lots of studies were also dedicated to material modeling in which the effects of different material properties on thermal and structural response were studied. Regarding the efficiency and integrity of the computational technique most significant contribution is made by McDill at el. [165-169], who developed graded element for dynamic adaptive meshing. Automatic mesh refinement was achieved only at the places of high thermal or stress gradients. This Gradient dependant adaptive meshing scheme substantially reduced the computational expense by decreasing the total number of elements in the model. Her scheme for adaptive remeshing was further improved by Runnemalm et al. [170-171] and Lindgren et al. [172] and Hyun and Lindgren [173].

Few very useful recommendations regarding welding simulation by Lindgren [149-151] are presented below:

- The degree of the finite element shape functions for the displacements should be one order higher than the thermal analysis due to the fact that the temperature field directly becomes the thermal strain in the mechanical analysis.

- Usually the same elements are used in the thermal and mechanical analysis. For the determination of transient heat distribution, transient and residual strain and stresses a fine mesh with linear elements is preferable than fewer elements of higher order because linear quad, in two dimensions, and brick, in three dimensions, elements are the basic recommendation in plasticity.
• Using a linear element requires that the average temperature should be used to compute a constant thermal strain in the mechanical analysis. It is also important to under integrate the volumetric strain (as it gives constant volumetric strain) when using linear elements in order to avoid locking.

• The axisymmetric model corresponds to welding the whole part at once and gives larger deformation in axial direction than a moving heat source.

• Models where the arc is traveling in the mesh, usually have larger elements and it is recommended that one take smaller time steps than the length of the elements along the weld divided by the welding speed. Shorter time steps are required at the start and finish of the weld.

• Regarding addition of filler material, quiet or inactive elements can give the same results if implemented carefully.

• There is no longer any reason to use the analytical solutions for heat distribution in the model because the numerical simulation of the thermal field is quite straightforward.

• If the region near the arc is studied in detail, the double ellipsoidal model for prescribing the heat input may be better. This should then be combined with about ten elements across the axis of ellipsoid area of heat input.

• The microstructure evolution is important to include in the material modeling of ferritic steels since their properties can change a lot due to the phase transformations. The major influence is on the thermal dilatation and yield stress.

• The computational model needs at least some kind of experimental results in order to determine net heat input. Using thermocouple to measure temperature is straightforward. It is also wise to do more experiments in order to verify the computational model before one uses the model to study the effect of different changes in the design and/or the welding procedure.
3.4 Mathematical Model of Metal Deposition by Gas Metal Arc Welding

The modeling of metal deposition in gas metal arc welding is a highly non linear coupled thermo-mechanical phenomenon. The moving heat source results in localized heat generation and large thermal gradients. The non uniform temperature distribution results in thermal stresses and distortions. The basic theory describing the behavior of continuum combines the theory of heat conduction and Elasto-Plasticity, which is developed in the next section.

3.4.1 Thermo-Mechanical Formulation

For the formulation of thermal model, the basic equation used is the *first law of thermodynamics* which states that energy of the system and its surrounding is conserved. Applying this to a differential control volume, $V$, the heat conduction equation (ignoring the heat of deformations) is represented as;

$$ \rho(T)c(T) \frac{\partial T(x, y, z, \tau)}{\partial \tau} = \nabla \cdot q + Q(x, y, z, \tau) $$

(3.1)

$q(x, y, z, \tau)$, is the heat generation per unit volume.

The constitutive equation which relates the heat flux and the temperature distribution is the *Fourier law of heat conduction*.

$$ q = -\kappa(T).A.VT(x, y, z, \tau) $$

(3.2)

Substituting equation (3.2) into equation (3.1)

$$ \rho(T)c(T) \frac{\partial T(x, y, z, \tau)}{\partial \tau} + \nabla(\kappa(T).A.VT(x, y, z, \tau)) = Q(x, y, z, \tau) $$

(3.3)

Considering a constant thermal conductivity

$$ \rho(T)c(T) \frac{\partial T(x, y, z, \tau)}{\partial \tau} + \kappa(T).A.V^2T(x, y, z, \tau) = Q(x, y, z, \tau) $$

(3.4)

The temperature distribution is governed by equation (3.4).

According to the *law of equilibrium*, sum of all the forces and moments acting on a body is zero. Mathematically this is written as;
\[ \frac{\partial^2 u_i}{\partial t^2} = \frac{1}{\rho} \frac{\partial \sigma_{ij}}{\partial x_j} + F_i \]  \hspace{1cm} (3.5)

Where \( i = 1, 2, 3 \)

For linear thermo-elastic problems, the stress-strain relationship in terms of Lame’s constant is given as

\[ \sigma_{ij} = \delta_{ij} \lambda \varepsilon_{kk} + 2 \mu \varepsilon_{ij} - \delta_{ij} (3 \lambda + 2 \mu) \alpha T \]  \hspace{1cm} (3.6)

The strain-displacement relationship is

\[ \varepsilon_{ij} = \frac{1}{2} ( \frac{\partial u_i}{\partial x_j} + \frac{\partial u_j}{\partial x_i} ) \]  \hspace{1cm} (3.7)

Substituting equations (3.6) and (3.7) into equation (3.5) and simplifying

\[ \frac{\partial^2 u_i}{\partial t^2} = (\lambda + \mu) \frac{\partial \varepsilon_{kk}}{\partial x_i} + \mu \nabla^2 u_i - (3 \lambda + 2 \mu) \alpha \frac{\partial T}{\partial x_i} + F_i \]  \hspace{1cm} (3.8)

The term \((3 \lambda + 2 \mu) \alpha \frac{\partial T}{\partial x_i}\) provides coupling between equation (3.4) and (3.8). The temperatures are calculated from equation (3.4) and are applied as body loads through \((3 \lambda + 2 \mu) \alpha \frac{\partial T}{\partial x_i}\) in Equation (3.8). The calculated displacements are used to get stress and strain using equation (3.6) and (3.7) respectively.

### 3.5 Finite Element Formulation

For isotropic material, temperature distribution given in equation (3.4) can be rewritten in expanded form as;

\[ \rho C \frac{\partial T}{\partial t} = \frac{\partial}{\partial x} (K \frac{\partial T}{\partial x}) + \frac{\partial}{\partial y} (K \frac{\partial T}{\partial y}) + \frac{\partial}{\partial z} (K \frac{\partial T}{\partial z}) + Q \]  \hspace{1cm} (3.9)

Matrix form of equation (3.9) is;

\[ \rho C \frac{\partial T}{\partial t} = [L]^T ([D][L]T) + Q \]  \hspace{1cm} (3.10)

Where
\[ L = \begin{bmatrix} \frac{\partial}{\partial x} \\ \frac{\partial}{\partial y} \\ \frac{\partial}{\partial z} \end{bmatrix} \quad \text{and} \quad D = \begin{pmatrix} K & 0 & 0 \\ 0 & K & 0 \\ 0 & 0 & K \end{pmatrix} \]

L is the gradient vector while D is the material stiffness matrix.

Employing convective boundary conditions at the surface enclosing the volume “V”, it gives;

\[ [q]^T \eta = h_f (T_B - T) \quad (3.11) \]

Where \( \eta \) is a unit vector perpendicular to the surface.

Taking the boundary effects into account, the Equation (3.10) becomes;

\[ \rho C \frac{\partial T}{\partial t} = [L]^T ([D][L]T) + Q + h_f (T_B - T) \quad (3.12) \]

The differential Equation (3.10) when multiplied by \( \delta T \) and integrated over the control volume using the boundary conditions yields

\[ \int_v (\delta T \rho C \frac{\partial T}{\partial t}) dv + \int_v (\delta T [L]^T [D][L]T) dv = \int_v (\delta T Q) dv + \int_v \delta T h_f (T_B - T) dA \quad (3.13) \]

Let for any element \( E \), the temperature is represented as

\[ T = [N]T_E \quad (3.14) \]

and

\[ \delta T = [N]\delta T_E \quad (3.15) \]

Where \( T_E \) is the nodal temperature and \([N]\) is the matrix of element shape functions. The equation is valid for all permissible \( \delta T_E \). If

\[ B = [L][N] \quad (3.16) \]
Where B is the derivative of shape functions

Substituting Equations (3.14), (3.15) and (3.16) in Equation (3.13), yields

\[ \rho \int_v (C[N][N]^T \{T\}) dv + \int_v ([B]^T[D][B][T_E]) dv = \int_v [N]Q dv + \int_A [N]h_f (T_B - [N]^T \{T_E\}) dA \]  

(3.17)

Equation (3.17) containing nodal temperatures can be written in a condensed form as

\[ [C][T_E] + [K][T_E] = \{F_E\} \]  

(3.18)

Where;

\[ [C] = \rho \int_v (C[N][N]^T) dv \], Specific heat matrix

\[ [K] = \int_v ([B]^T[D][B]) dv + \int_A h_f [N][N]^T dA \], Thermal conductivity matrix

\[ \{F_E\} = \int_v Q[N] dv + \int_A h_f T_B [N] dA \], Heat generation and convection matrix

Equation (3.18) is an elemental equation with vector \{T_E\} containing unknown nodal temperatures. A system of equations is obtained by assembling the individual elemental equations. The system of equation is then solved using appropriate solution technique e.g. Newton-Raphson for the unknown nodal temperatures. The finite element form of equation (3.8) can be derived using the principle of virtual work. The principle of virtual work states that a virtual (very small) change of the internal strain energy must be offset by an identical change in external work due to applied load, mathematically this is written as

\[ \delta U = \delta P \]  

(3.19)

Where

\[ U = \] internal strain energy or internal work

\[ P = \] external work, like due to inertia effect.

\[ \delta = \] virtual operator
The virtual strain energy is given as

\[ \delta U = \int \{ \delta \varepsilon \}^T \{ \sigma \} d\{ V \} \]

(3.20)

\( \varepsilon \) = Strain vector

\( s \) = Stress vector

\( V \) = Volume of element

From the theory of basic solid mechanics

\[ \sigma = D \varepsilon^e \]

(3.21)

and

\[ \varepsilon = \varepsilon^e + \varepsilon^th \]

(3.22)

When

\( \varepsilon \) = Total strain

\( \varepsilon^e \) = Elastic strain

\( \varepsilon^th \) = Thermal strain

\( D \) = Material stiffness matrix

The thermal strain vector for an isotropic medium with temperature dependent coefficient of thermal expansion is given as

\[ \varepsilon^th = \Delta T \alpha(T) \]

(3.23)

\( \Delta T \) is the difference between the reference temperature and actual temperature.

Substituting equations (3.20) and (3.21) in equation (3.19) yields

\[ \delta U = \int \{ \delta \varepsilon \}^T [D] \{ \varepsilon \} - \{ \delta \varepsilon \} [D] \{ \varepsilon^th \} dV \]

(3.24)

The strain is related to nodal displacement by the following relations

\[ \{ \varepsilon \} = [B] \{ u \} \]

(3.25)

For a constant displacement, virtual straining energy is given as;
\[ \delta U = \{\delta u\}^T \int \! [B]^T [D] [B] \, dV \{u\} - \{\delta u\}^T \int \! [B]^T [D] \{\varepsilon^h\} \, dV \]  
\hspace{1cm} (3.26) 

The external virtual work due to inertia forces is formulated as;

\[ \delta P = -\int \! \{\delta w\}^T \frac{\{F^a\}}{v} \, dV \]  
\hspace{1cm} (3.27) 

Where

\[ w = \text{displacement vector of a general point} \]

\[ \{F^a\} = \text{acceleration force vector} \]

According to Newton second law of motion

\[ \frac{\{F^a\}}{v} = \rho \frac{\partial^2}{\partial t^2} \{w\} \]  
\hspace{1cm} (3.28) 

If the displacement within the element is related to nodal displacement by

\[ \{w\} = [N]\{u\} \]  
\hspace{1cm} (3.29) 

Then equation (3.27) can be re-written as

\[ \delta P = -\{\delta u\} \rho \int \! [N]^T [N] \, dV \frac{\partial^2}{\partial t^2} \{u\} \]  
\hspace{1cm} (3.30) 

Substituting equations (3.26) and (3.30) in equation (3.17), yields

\[ \{\delta u\}^T \int \! [B]^T [D] [B] \, dV \{u\} - \{\delta u\}^T \int \! [B]^T [D] \{\varepsilon^h\} \, dV = \]

\[ -\{\delta u\} \rho \int \! [N]^T [N] \, dV \frac{\partial^2}{\partial t^2} \{u\} \]  
\hspace{1cm} (3.31) 

\{\delta u\}^T \text{ vector is a set of arbitrary virtual displacement common in all terms, the condition required to satisfy equation (3.31) reduces to}

\[ [K_e] - \{F^h\} = [M_e] \{\ddot{u}\} \]  
\hspace{1cm} (3.32) 

When
\[ \begin{align*}
[K_e] &= \int_V [B]^T [D] [B] dV \quad \text{Element Stiffness Matrix} \\
\{f^\text{th}_e\} &= \int_V [B]^T [D] [e^{\text{th}}] dV \quad \text{Element Thermal Load Vector} \\
[M_e] &= \rho \int_V [N]^T [N] dV \quad \text{Element Mass Matrix}
\end{align*} \]

### 3.6 FE Modelling of Arc Welding

#### 3.6.1 Coupling Between Fields

The detailed modeling of metal deposition by welding, involving its complete physics and chemistry is a very complex phenomenon. This is because of the involvement of a number of diversified fields such as 3D heat flow, heat flow in multiple phases, phase transformation, complex weld pool dynamics, evolution of microstructure and overall thermo-mechanical response of the structure. Figure 3.2 presents the different fields and their couplings and table-3.1 further explain it. Further details can be found in references [148,174,175]. It is a laborious job to account for all the couplings that exist between these fields. In the computational mechanics of welding a number of simplifications are employed and therefore most of these couplings are ignored due to their week nature. To encompass the detailed weld pool phenomena may, however, be regarded as a separate research area because of the level of refinement required. Although other than that, the prediction of the temperature field and weld bead geometry cannot be subsequently accounted for in a macroscopic context [176]. However, if primary interest lies in the geometrical changes close to the weld, modeling of fluid flow becomes essential.

Due to the weak nature of some of the couplings and an insignificant effect on a macroscopic level of the weld pool physics, these can be ignored. Therefore all the bidirectional couplings associated with fluid flow in the weld pool (except 8 i.e. dependence of heat transfer on fluid flow) are ignored. Similarly, the other weak couplings such as microstructure stress dependency (coupling 4) and heat generation due to deformation (coupling 10) are also ignored. Hence the final form of field couplings, used in the present study, is shown in Figure 3.3.
Figure 3.2: Coupling between different fields

Figure 3.3: Selective coupling between different fields
<table>
<thead>
<tr>
<th>Coupling</th>
<th>Description</th>
<th>Nature</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>Microstructure depend on temperature</td>
<td>Strong</td>
</tr>
<tr>
<td>2a</td>
<td>Phase changes release latent heat</td>
<td>Medium</td>
</tr>
<tr>
<td>2b</td>
<td>Thermal conductivity and heat capacity depend on microstructure</td>
<td>Medium</td>
</tr>
<tr>
<td>3a</td>
<td>Plastic strain associated with phase change</td>
<td>Strong</td>
</tr>
<tr>
<td>3b</td>
<td>Thermal expansion depends on microstructure</td>
<td>Strong</td>
</tr>
<tr>
<td>3c</td>
<td>Elastic/Plastic material behavior depends on microstructure</td>
<td>Strong</td>
</tr>
<tr>
<td>4</td>
<td>Stress state affects phase changes</td>
<td>Weak</td>
</tr>
<tr>
<td>5</td>
<td>Structural deformations affect flow pattern</td>
<td>Weak</td>
</tr>
<tr>
<td>6</td>
<td>Fluid pressure causes deformation</td>
<td>Weak</td>
</tr>
<tr>
<td>7</td>
<td>Temperature affects fluid velocity</td>
<td>Weak</td>
</tr>
<tr>
<td>8</td>
<td>Fluid flow governs heat transfer</td>
<td>Strong</td>
</tr>
<tr>
<td>9</td>
<td>Temperature changes produce deformation</td>
<td>Strong</td>
</tr>
<tr>
<td>10</td>
<td>Mechanical deformation generate heat (due to elastic, plastic and thermal strain rate)</td>
<td>Weak</td>
</tr>
<tr>
<td>11</td>
<td>Flow pattern alters material behavior</td>
<td>Weak</td>
</tr>
<tr>
<td>12</td>
<td>Material behavior affect flow pattern</td>
<td>Weak</td>
</tr>
</tbody>
</table>

Table-3.1: Description of couplings in welding analysis
Since the coupling 10 has been ignored, the fully coupled (bidirectional) thermo-mechanical phenomenon of welding can be safely broken down into unidirectional coupled analysis. This simplified approach was implemented and a fully coupled thermal-metallurgical analysis was performed followed by a structural analysis. During the thermal analysis microstructure evolution can be modeled through either more sophisticated direct calculation procedure or by relatively simpler but common method of indirect incorporation of microstructure aspects into the material model. During the structure analysis, temperature and microstructure dependence of stresses and deformations are accommodated by invoking the results of thermal-metallurgical analysis into the structural (thermal-stress) analysis.

3.6.2 Material Model

The accurate implementation of computational weld mechanics depends mainly on the accurate modeling of temperature dependant material properties. But this is a difficult task due to insufficient material data available at elevated temperature. In the past, several research efforts have been dedicated to the investigation of material properties and their effect on the structural response under transient thermal loading during welding e.g. Zhu and Chao[177]. The evolution of microstructure and its effects on thermal and mechanical properties are the hard core issues in material modeling. In the computational weld mechanics microstructure evolution is addressed either by a direct or an indirect method. In the direct method calculations are made from thermal history, various phase fractions, properties of each constituent and deformation history. On the other hand in an indirect method calculations are made by considering the micro-structural dependency on the thermal and mechanical history. Although the indirect approach is relatively crude but is widely used in the welding simulation due to its relative simplicity.

Since the present work is related to the mechanical effects of weld based deposition and the study is comparative in nature with respect to various boundary conditions, process conditions and deposition patterns, therefore without going into the details of metallurgical investigations the indirect approach is adopted to account for the effect of microstructure evolution. The material used in this study is low carbon steel. The detailed properties of low carbon steel are given in the next section.
3.6.2.1 Material Model for Low Carbon Steel

The current study is limited to a single material both experimentally and in the material model i.e. low carbon steel. The filler metal was mild steel with AWS designation as ER70S-6 wire (for composition see table 3.2) while the substrate plate and the support plate are of mild steel. Complete ranges of temperature dependant material properties for this material are taken from Karlsson and Josefson [178].

Thermal properties i.e. specific heat and thermal conductivity as a function of temperature are shown in Figure 3.4. For specific heat, latent heat associated with low temperature solid-solid phase transformation is accounted for both the weld metal and base metal. Enthalpy formulation is employed to avoid numerical non-convergence, in the present work. In the enthalpy method, Voller et al. [179], avoids explicit tracking of the solid/liquid interface. Latent heat evolution during phase change is incorporated in the energy equation using the following definition of enthalpy. For each phase $\phi$, enthalpy is defined as

$$ h = \int_{0}^{\tau} c_{\phi}\rho dT + f_{l} L $$

(3.33)

Where $L$ is latent heat and $f_{l}$ is the local liquid volume fraction. For isothermal phase change the liquid fraction is determined by the melting temperature $T_{m}$:

For $T > T_{m}$, $f_{l}=1$

For $T < T_{m}$, $f_{l}=0$

As specified in [180], latent heat of 272 KJ/Kg for solid-liquid phase transformation is distributed over the melting/solidification range i.e. between solidus temperature 1480°C and liquidus temperature 1530°C. In order to model fluid flow (stirring) effect on the thermal field (coupling 8 in Figure 3.3) thermal conductivity is given an artificial rise to 230 KJ/mK at solidus temperature, as suggested by Andersson [164]. This results in a slightly larger weld pool with relatively uniform temperature within the pool.

Figure 3.5 presents the temperature dependence of Young’s modulus and Poisson. At elevated temperature material softens and this is reflected by a reducing Young’s modulus whilst on the contrary an increasing Poisson’s ratio. Young’s modulus reduces
<table>
<thead>
<tr>
<th>Sr. No.</th>
<th>Item</th>
<th>Percentage</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>Carbon</td>
<td>.06-.15 %</td>
</tr>
<tr>
<td>2</td>
<td>Manganese</td>
<td>1.40-1.85 %</td>
</tr>
<tr>
<td>3</td>
<td>Copper</td>
<td>.50 % max.</td>
</tr>
<tr>
<td>4</td>
<td>Silicon</td>
<td>.80-1.15%</td>
</tr>
<tr>
<td>5</td>
<td>Sulfur</td>
<td>.035 % max.</td>
</tr>
<tr>
<td>6</td>
<td>Phosphorus</td>
<td>.025 % max.</td>
</tr>
<tr>
<td>7</td>
<td>Nickel</td>
<td>0.15 % max.</td>
</tr>
<tr>
<td>8</td>
<td>Chromium</td>
<td>0.15 % max.</td>
</tr>
<tr>
<td>9</td>
<td>Vanadium</td>
<td>0.03% max.</td>
</tr>
<tr>
<td>10</td>
<td>Molybdenum</td>
<td>0.15 % max.</td>
</tr>
<tr>
<td>11</td>
<td>Iron</td>
<td>Balance</td>
</tr>
<tr>
<td>12</td>
<td>Others</td>
<td>Total .50 % max.</td>
</tr>
</tbody>
</table>

Table-3.2: Nominal Composition, ER70S-6 Carbon Steel.

Figure 3.4: Specific Heat and Thermal Conductivity (steel) as a function of temperature [178].
Figure 3.5: Young’s Modulus and Poisson Ratio as a function of temperature [178].

Figure 3.6: Yield Strength as a function of temperature (Depending on peak temperature reached) [178].
to a negligibly small value near melting, but an excessively low value results in numerical instability, due to extremely small values at the diagonal of the stiffness matrix. The chosen value in the present work is 12.4 GPa, however even lower value of 1 GPa is also reported in [181]. The values for both the bulk modulus and Poisson’s ratio are taken constant after 1200°C.

The material response is considered elastic perfectly plastic, without Kinematic hardening. It is argued in [178] that results from high temperature experiments on low-alloy steel show a nearly ideal plastic behavior at temperatures above 800°C and at the same time plastic strain accumulated before the final solid state phase transformation to a large extent relieved during the transformation. Hence elastic-perfectly plastic material behavior is considered by ignoring dislocation hardening. The effect of creep and transformation induced plasticity are not accounted for in the material model. The Von Mises yield criteria with associated flow rule has been used.

The temperature dependency of yield strength based on peak temperature reached is shown in Figure 3.6. During heating both the base metal and weld metal follow material characteristic curve for base metal, however during cooling different curves are followed. Thermal strain data from [178] is shown in Figure 3.7. FEM code ANSYS does not use thermal strain; instead it takes thermal expansion coefficient. Therefore, thermal dilatation is converted to thermal expansion coefficient using equation (3.34), taken from [183].

\[ \varepsilon^th = \alpha(T) \chi(T - T_{ref}) \]  

(3.34)

\( T_{ref} \) in above equation is the temperature at which thermal strain is assumed to be zero. Melting and ambient temperatures are used as \( T_{ref} \) for weld pool and rest of the material respectively. Further, to capture the effect of convective and radiative heat loss from the surface, temperature dependant equivalent heat transfer coefficient comprising of convection and radiation is calculated by using equation (3.31). Effect of radiation becomes more important at high temperatures, where it dominates over the surface convection and appears primary mechanism of heat loss from the surface. The value of
Figure 3.7: Thermal Strain as a function of temperature (Depending on peak temperature reached) [178].

Figure 3.8: Goldak Double Ellipsoidal heat source
emissivity \( (\varepsilon_{em}) \) is assumed to be 0.51, which is the average value of polished and rusted steel surfaces.

\[
h = \varepsilon_{em} b_{het} (T + 273)^4 - (T_{amb} + 273)^4 (T - T_{amb}) + h_{con}
\]  
(3.35)

3.6.3 Heat Source Model

It is believed the welded structures exhibit residual stresses and deformation because of the highly non-uniform temperature field applied during welding. It’s a widely known fact that both the residual stresses and welding deformations are highly sensitive to transient temperature distribution. Yet the transient temperature distribution itself is a function of the total heat applied and heat distribution pattern within the domain. Thus for the determination of realistic temperature profile in the target application, a very careful and accurate modeling of heat source is required. As described earlier that accurate modeling of weld pool with all its physics is quite complicated if not altogether impossible. Fortunately, the complexity of weld pool modeling is not a hindrance for modeling the macroscopic effect of welding [148]. If the main objective is to study the mechanical effects of welding, then the only critical requirement in modeling of heat source is the estimation of weld pool and HAZ (heat affected zone) dimensions. During the last four decades several developments in modeling of heat source have been reported in the literature, a very brief introduction of which can be found in [149].

In the present work the most widely accepted model for simulation of arc welding process, the so called doubled ellipsoidal heat source model, presented by Goldak et al. [157], was used. This model gives Gaussian distribution and has an excellent feature of power density distribution control in the weld pool and HAZ. The heat input is defined separately over two regions; one region in front of the arc center and the other behind the arc center, as shown in Figure 3.8. The spatial heat distribution in a moving frame of reference can be calculated with the governing equations (3.32) and (3.33).
Where,
\[ Q = VI \quad \text{and} \quad f_f + f_r = 2 \]

The origin of the coordinate system is located at the center of the moving arc. A user subroutine is used to calculate the centriodal distance of elements from the moving arc center corresponding to the arc position at any instant. The overall power density distribution obtained using double ellipsoidal heat source is shown in Figure 3.9, while figure 3.10 gives the heat distribution at the instant of heat flux application on the substrate plate.

For the three dimensional model discussed in this dissertation, the heat source is assumed to move through volume and calculated heat is applied to elements as volumetric heat generation. The added advantage of volumetric heat generation is that the elements lying on the surface can be used to model the surface heat convection which otherwise require additional two dimensional surface elements for this purpose.

### 3.6.4 Addition of Filler Material

Modeling of filler material has always been a hard core issue in computational weld mechanics because of its effects on the final weld bead geometry and computational expense. In general the deposition of filler material is modeled by using three different approaches, i.e. element movement, inactive element and quiet element techniques.

In the element movement technique, developed by Fanous et al. [185], elements of filler metal are modeled at some distance apart from the parent metal and are connected through gap elements. The filler metal elements move towards the base metal as the heat sources approaches to that particular location. To avoid abrupt heating of moving filler metal elements, the thermal conductivity of gap element is increased with decrease in
Figure 3.9: 3D Power Density Distribution

Figure 3.10: Heat distribution at the instant of heat flux application (W/m$^3$)
length. This technique is still at validation stage and is not widely reported in the literature.

In the inactive element technique, complete computational model is defined in the start of the analysis but the elements and nodes belonging to different weld passes are included in the FE model at the time when particular weld pass is laid. Thus the FE model is updated for each weld pass and is only accomplishable in the software which has the capability for model extension during the analysis. However, this technique is more realistic for multi-pass welding when the geometry of subsequent weld beads is strongly affected by the final (distorted) shape of already laid weld beads.

In the quiet element technique complete FE model, including all the elements and nodes of base metal and filler metal, is developed in the start. The elements belonging to filler metal are deactivated by assigning them a very low thermal conductivity (in thermal analysis) and very low stiffness (in structural analysis). The value of thermal conductivity and stiffness of deactivated elements should be low as these may not have any contribution in the rest of the model but should not be as low which may produce an ill conditioned matrix. The elements belonging to a specific weld bead are reactivated by “element birth” option at the start of the respective weld bead or when they come under the influence of welding torch. The material properties of reactivated elements are instated at the time of their activation.

Conventionally, inactive and quiet element techniques are used in welding simulation and both the techniques, if implemented properly, produce same results [186]. Quiet element technique is relatively straight forward and easy to implement by using inherent feature of element birth and death, available in most of the finite element software. However, for multi-pass welding this technique may give serious convergence problem during stress analysis as deactivated elements (from the beads to be laid yet) at the boundary of the previously laid weld bead may undergo severe shape distortion at the time of activation. The mechanism of element distortion has been explained by Troive et al. [187] who suggested that to get rid of this problem no structural constraint should be imposed on nodes belonging to deactivated element during structural analysis. Consequently deactivated elements deform along with the live elements and compatibility problem due
to deposition of undistorted elements (belonging to new weld bead) on the distorted elements (belonging to previously laid weld bead) can be reduced significantly. Wilkening and Snow [188] suggested that during structural analysis, all the nodes of deactivated elements should be maintained at so called softening temperature till the time of activation of respective element of the particular weld bead is reached. Except an axisymmetric analysis of multi-pass welding, the entire simulation work is pertaining to analysis of single pass welding, thus technique is opted with the above mentioned recommendation due its suitability for single pass welding and ease of application. Since a single layer is made up of multiple passes with each pass alters the stresses and distortions caused by previous passes therefore the buildup is modeled as a series of single pass, using quiet element approach and any lumping strategy as reported in Lindgren [149] have not been employed.

### 3.6.5 Analysis Procedure

The advantage of weak structural to thermal field coupling is employed to break down the complex coupled field thermo-mechanical analysis of welding into two parts. In the first part non-linear transient thermal analysis is performed to predict the temperature history of the domain for complete thermal cycle of single or multi-pass welding. The procedure for thermal analysis is indicated in a flow chart shown in Figure 3.11. As already described quiet elements technique with ambient temperature constraint on the nodes of deactivated elements is employed for modeling of filler material during this work. Nodal constraint is removed at the time of activation of respective element in thermal analysis.

For three dimensional models, time step is constant for each load step during heating and is based on the total heating time and number of elements in axial direction because heat source is supposed to stay on each element at least once, as recommended by Lindgren [149]. In the subsequent cooling, the time step progressively increased till the weldment cools to the ambient temperature.

In the second part, non-linear structural analysis is performed in which temperature history calculated during thermal analysis is applied as body load. Procedure for structural analysis is indicated in Figure 3.12. Load step time in structural analysis is kept
Figure 3.11: Analysis procedures for transient thermal analysis
Figure 3.12: Analysis procedures for transient structural analysis
the same as that of the respective thermal load step. In the structural analysis, element of a particular weld bead is activated at time when it cools down to solidus temperature. For this purpose a user subroutine is used in thermal analysis which tracks the averaged peak temperature of each element of filler metal and records the time for each element when it reaches the solidus temperature after peak temperature. In addition, initial strain history of each element is reset at the activation time.
CHAPTER 4

The Experimental Setup, 3D Finite Element Model and its Validation

In this chapter a 3D finite element model for GMAW based layered manufacturing is discussed. This model was developed and experimentally validated at The Faculty of Mechanical Engineering GIK institute. This chapter discusses the details involved in the model and the experimental equipment / procedures adopted for the validation of the model.

4.1 Introduction

3D finite element (FE) simulation of the welding process is computationally very expensive. This is because of a highly non-linear formulation and a transient solution. In the past due to limited computational power and data storage, different simplifications were adopted for the welding simulations. These included, solving the problem as a two dimensional case by using plain stress/strain or axisymmetric formulations. The main benefit of such simplifications was an excessive reduction in computational time with the results qualitatively acceptable. However some times this may cause erroneous results, because of over simplification. Therefore Karlsson et al. [189], Goldak et al. [158], Runnemalm and Lin [190] and Lindgren [191] advocated the use of 3D models to avoid any inappropriate results.

In 1999 Goldak et al. [192] analyzed the real time computational weld mechanics with respect to the rapidly growing computational power. He anticipated that computational weld mechanics will soon be faster than the real world welds. The time required for computation of thermal stresses depends on several factors which include part geometry, level of discretization, number of load steps, number of substeps in each load step and
The demand for increased computational power is continuously growing for the analysis of complex real world problems. However, the recent developments in the computer hardware / software had made it possible to solve 3D FE problems (at least for simple geometries and limited welding) with fair accuracy and affordable computational resources.

The modeling of metal deposition process is highly non linear due to the presence of different deformation modes. Therefore a complete 3D model is required to predict the temperatures, thermal gradients, residual stresses and deformations. A 3D model can give a detail insight into the physics of deposition process and can predict deformations and residual stresses in longitudinal as well as transverse directions. The present work outlines a 3D thermo-mechanical FE model for weld based LM using a moving heat source and successive metal deposition. The model has been limited to deposition of a single layer due to the requirement of large computational resources. For the experimental validation of the model a range of equipment was tailor made and experiments were performed. The validation was performed by comparing the depth of fusion Zone, temperature history, deformation and residual stress results from the model with the experimental results.

4.2 The Deposition System

The experimental setup constituted of a test bed facility for depositing of weld metal and removing of metal top by milling operation. It has the ability to build a single or multi layered plate shaped specimens with an optional intermittent machining after each deposited layer. In case of multi layered specimens the layers can be deposited either in same direction, alternating opposite directions or at 90° to each other. The single layered specimens were mainly used for the thermal and structural model verification while specimens from the multi layered deposit were used for microstructural examination, hardness testing and tensile strength investigations of the RP parts.

A semi-automatic in-house experimental facility was developed using a conventional milling machine and a Lincoln Power Wave 355 welding system. The milling and the welding machine were interfaced with each other for the control of deposition start up, ending, weld bead size and layer position / orientation. The automatic feed mechanism of
the milling machine was used to control the speed of the weld bead deposition. The
distance between the torch and the substrate (stick out) is adjusted using the z-axis feed
of the milling machine. A readjustment of the stick out was necessary before the
deposition of each commencing layer. The welding torch was attached to the milling
machine arbor with the feature to adjust the torch angle as per required for the welding
parameters.

The Power Wave 355 welding system is computer controlled using a waveform control
technology with the capability to produce one drop per pulse (ODPP) welds of high
quality. The welding parameters are maintained during the deposition to ensure constant
welding conditions. Figure 4.1 presents a picture of the welding system with the
computer interface and figure 4.2 shows the picture of the deposition system in which the
welding system is coupled with the conventional milling machine.

The logical flow diagram of the deposition system circuit is shown in figure 4.3. The
manual activation of a relay started the feed of the milling machine. This feed motion of
the bed activates a limit switch which is positioned in synchronization with the substrate
so as to start deposition at the required location. The limit switch initiates a timer, set at a
specified time \( t_0 \). This timer ensures the commencement and continuation of deposition
for the required time and length of weld pass. When \( t \geq t_0 \) the timer stops the welding
machine is deactivated and the deposition stops. The motion of the milling machine is
controlled by the adjustment / setting of the dogs of the machine. The welding torch is
offset after each weld pass by one half of the weld bead width for depositing the next
pass and the process continues until a complete layer was obtained. This offset was done
to ensure a proper overlap of the weld beads. A 1.2 mm diameter mild steel welding wire,
ER70S-6, was used to build a slab of 100 x 50 x 3.0 mm on the substrate using CO\(_2\) as the
shielding gas. The welding parameters used for deposition are shown in table-4.1. These
parameters were also utilized for the development and analysis of the FE model.

**4.3 Finite Element Model Description**

In the development of this simulation model [146] the commercial finite element
software ANSYS is combined with a user programmed subroutine to implement the
Figure 4.1: Power Wave 355 welding machine with computer interface

Figure 4.2: Welding Machine coupled with the conventional Milling machine
Figure 4.3: Flow Diagram of Deposition System

Table 4.1: Welding parameters

<table>
<thead>
<tr>
<th>Parameter</th>
<th>Value</th>
</tr>
</thead>
<tbody>
<tr>
<td>Average Welding Current (A)</td>
<td>232</td>
</tr>
<tr>
<td>Voltage (V)</td>
<td>23.9</td>
</tr>
<tr>
<td>Weld Speed (mm/min)</td>
<td>375</td>
</tr>
<tr>
<td>Wire feed (mm/sec)</td>
<td>8</td>
</tr>
<tr>
<td>Gas flow rate (SCFH)</td>
<td>35</td>
</tr>
<tr>
<td>Stick out (mm)</td>
<td>16</td>
</tr>
<tr>
<td>Weld bead to weld bead Offset (mm)</td>
<td>5</td>
</tr>
</tbody>
</table>
welding parameters like Goldak double ellipsoidal heat source, material addition and temperature dependent material properties. A sequentially coupled, 3D, thermo-mechanical FE analysis of metal deposition was performed in which the transient temperature distribution from thermal analysis was applied as body load to structural model. The model is limited to the buildup of a single layer of weld metal on a bolted substrate. The basic geometry of the model comprises of a rectangular 160×110×10 mm substrate plate bolted at the four corners to a 200×130×34 mm support plate. On the centre of the substrate plate an 100×50×3 mm weld layer was deposited in nine weld passes with an inter-pass time of 180 sec. The geometry of the model and the mesh are shown in figure 4.4 (a), with the mesh refinement at the deposition and substrate interface shown in the enlarged view. The heat distribution at the instant of heat flux application is shown in figure 4.4(b).

A raster deposition sequence is used for layer deposition as shown in figure 4.5. The deposition consists of nine successively deposited rows and in between rows; the model is allowed to cool for 180 seconds. The interpass time is in accordance with the actual experimental setup i.e. the time required for the deposition torch to be aligned to a successive weld start positions. The interpass time also makes sure that the temperature of the substrate is low enough to avoid excessive preheating. As for a zero inter-pass time the subsequently deposited weld bead experiences a high preheat.

Figure 4.6 represents the test plate on which a weld layer is deposited. The different lines marked on the plate represent the locations where temperature, deformation and residual stress distribution results are presented. Figure 4.7 shows all the boundary conditions, both thermal and structural, applied to the simulation model. For thermal analysis convection and radiation were applied at the surfaces exposed to the atmosphere while conduction was applied on the base of the substrate. The four corners of the substrate were completely constrained (all DOF = 0) in the structural model assimilating the bolting, while the base of the substrate was constrained in the negative Y direction.

The model is limited to deposition of a single layer. This may be sufficient to investigate and study the comparison of the effects of different deposition parameters. For the deposition of multiple layers, as is the case in RP, the effects (e.g. deformations) are
Figure 4.4(a): Meshed three-dimensional finite element model.

Figure 4.4(b): Heat distribution at the instant of heat flux application (W/m³)
Figure 4.5: Raster deposition sequence

Figure 4.6: Schematic of substrate plate and weld deposition (all dimensions in mm).
Figure 4.7: Side view of substrate showing the boundary conditions.

Figure 4.8: ANSYS Solid Element (Solid70 for thermal & Solid45 for structure).
expected to add up. Also the shape of the deposited layer (i.e. rectangular), considered in this study, is the simplest shape, which may not be common in RP. However the results from the current model may provide a good guide line for the general RP parts.

Following assumptions were made in the current FE model:

- The material model in this study was simplified by using a two material model, one for the weld deposition and other for the base plate.
- The support plate is not actually modeled; however a diffusion flux at the base of substrate plate was defined for conduction by applying an effective heat transfer coefficient. While in the structural analysis its effect is modeled by using a 2 noded contact element.
- The effect of convection in the weld pool was incorporated by increasing the thermal conductivity whenever any part of the model attained the liquidus temperature.
- There exists a delay in material addition between thermal and structural analysis, excluding the effect of molten metal expansion.
- The material model does not include strain hardening and the effect of creep.
- For simplification a rectangular cross-section for the weld bead was considered in place of a circular cord.
- The substrate plate is bolted at the four corners with the support plate with M12 bolts. In the simulation this bolting is simplified by constraining all the nodes present within the bolted area to have zero displacement.

These assumptions will be further discussed in the upcoming sections.

4.3.1 Thermal Model

A 3D thermal solid element type (Solid70) with eight nodes and with features to model material addition was used in the analysis (figure 4.8). Mapped meshing was employed to obtain a regular pattern with obvious rows of elements. Since the temperature gradients in the vicinity of the weld are much higher then those far from it, a fine mesh was used in areas near the deposited weld whereas the mesh gets coarser as one moves away from the welding zone.
A distributed moving heat source, referred as Goldak [157,158] double ellipsoidal heat source coupled with material addition, was used to develop the thermal model. The basic equations of the double ellipsoidal heat source for the distribution of heat in a three-dimensional model are given below, while the applied heat distribution is shown in figure 4.4(b):

\[
q_f = \frac{6\sqrt{3}\eta Q f_j}{\pi \sqrt{\pi} a_f b c} e^{-[-3\left(\frac{x^2}{a_f^2} + \frac{y^2}{b^2} + \frac{z^2}{c^2}\right)]} \\
q_r = \frac{6\sqrt{3}\eta Q r_j}{\pi \sqrt{\pi} a_r b c} e^{-[-3\left(\frac{x^2}{a_r^2} + \frac{y^2}{b^2} + \frac{z^2}{c^2}\right)]}
\]

\[Q = \eta V I\]  

and

\[f_f + f_r = 2\]

Where,

The description and numerical values for different variables in the power density distribution equation are given in table-4.2. The origin of the coordinate system is located at the centre of the moving arc and the movement of the heat source is achieved through a user subroutine. Another user subroutine is used to calculate the centroidal distances of elements from the instantaneous position of the moving arc centre. The spatial distribution of heat is calculated from equations 4.1 and 4.2 and is applied on elements as volumetric heat generation.

The FE model consists of 25000, 3D Solid-70 elements without mid side nodes. Since the heat flow is dominant in direction transverse to welding [149], therefore the mesh size was set 1mm across and 2 mm along the welding direction. The time step size is calculated by dividing the heat source length with welding speed. Substrate and the filler materials were modeled as mild steel and the properties are adopted from Karlson et al. [178], along with some extrapolations and simplifications as suggested by Abid et al. [193]. The material model in this study was simplified by using a two material model,
<table>
<thead>
<tr>
<th>Description</th>
<th>Symbol</th>
<th>Value</th>
</tr>
</thead>
<tbody>
<tr>
<td>Front ellipsoidal semi-axes length</td>
<td>$c_f$</td>
<td>5 mm</td>
</tr>
<tr>
<td>Rare ellipsoidal semi-length</td>
<td>$c_r$</td>
<td>15 mm</td>
</tr>
<tr>
<td>Half width of arc</td>
<td>$a$</td>
<td>5 mm</td>
</tr>
<tr>
<td>Depth of arc</td>
<td>$b$</td>
<td>4 mm</td>
</tr>
<tr>
<td>Fraction of heat deposited in front</td>
<td>$f_f$</td>
<td>0.5</td>
</tr>
<tr>
<td>Fraction of heat deposited in rare</td>
<td>$f_r$</td>
<td>1.5</td>
</tr>
<tr>
<td>Voltage</td>
<td>$V$</td>
<td>24 V</td>
</tr>
<tr>
<td>Welding current</td>
<td>$I$</td>
<td>232 A</td>
</tr>
<tr>
<td>Arc efficiency</td>
<td>$\eta$</td>
<td>75%</td>
</tr>
<tr>
<td>Welding speed</td>
<td>$v$</td>
<td>6.25 mm/s</td>
</tr>
</tbody>
</table>

Table 4.2: Double ellipsoidal heat source parameters.

![Heat Transfer Circuit](image)

Figure 4.9: Heat Transfer Circuit for support and substrate plate
one for the weld deposition and other for the base plate. This was done in place of the material being divided into 6 different material regions based on the peak temperature attained during heating phase, and than assigning different properties of each region during the cooling phase. These details have been discussed in chapter 3.

The material to be deposited was modeled using “quiet” element approach. In this approach, the part of structure which has not been deposited as yet is made passive or ‘killed’. These elements remain inactive until they are reactivated (made ‘alive’) at a proper time step. The thermal boundary conditions of the deposition, substrate and base plate, exposed to the atmosphere, comprised of combined convection and radiation. The combined heat transfer coefficient is temperature dependent and is calculated using the following relation

$$h_{\text{eff}} = h_{\text{conv}} + \varepsilon \sigma (T_s + T_\infty)(T_s^2 + T_\infty^2)$$ (4.5)

Where $h_{\text{conv}}$ is kept constant at 5.3 W/m$^2$ K, which is typical of free convection from a horizontal surface.

For conduction at the interface of substrate and base plate, a diffusion flux at the base of substrate plate was defined (figure 4.9). This is done by applying an effective heat transfer coefficient at the base of substrate plate, which consists of the conductive thermal resistance of the base plate and contact resistance between the base plate and the substrate plate (equation-4.6).

$$\frac{1}{h_b} = R_b = R_c + \frac{t_b}{k}$$ (4.6)

The bulk temperature for both $h_b$ and $h_{\text{eff}}$ is taken to be 300K. The fluid flow in the weld pool not only involves conduction but also includes convection. A common practice in analysis is to imitate this by increasing the thermal conductivity at higher temperatures [150]. The effect of convection in the weld pool, therefore, was incorporated by increasing the thermal conductivity from 34 to 100 W/m.K, whenever any part of the model attained the liquidus temperature [146].
4.3.2 Structural Model

The structural model used has the same geometry and similar features as the thermal model (figure 4.8). Solid45 element with large deformation capabilities and extra displacement shapes has been used. The mesh size and time step for structural analysis is similar to that of thermal analysis.

The method of material addition in structural analysis is similar to thermal analysis except for some time delay. At the start of the analysis all the deposition elements are “killed”. In the thermal model these elements are made alive at the time when the heat source reaches them. While in the structural model these elements are made “alive” when they achieve the peak temperature. In other words for the thermal model, elements become alive as soon as molten metal is deposited however, for structural model the elements become alive only when it solidifies and become structurally integral part of the substrate. This therefore causes a delay in material addition between thermal and structural analysis thus excluding the effect of molten metal expansion as it heats up and capturing the effect of contraction as it cools down.

The material properties used, i.e. Young Modulus, Yield Strength, Coefficient of thermal expansion and Poisson ratio are temperature dependent. However the material model does not include strain hardening and the effect of creep. All the properties of substrate and filler material are similar except for the yield strength of filler material, which is taken equal to the yield strength of welding wire used.

The support plate, being thick, acts like a rigid fixture for the substrate plate. Therefore it restraints the substrate from any downward movement (Y⁻ direction) into the support plate however allows any upward motion (Y⁺ direction), in the substrate unbolted condition. The support plate is not modeled as a separate geometry in the analysis rather its effect is modeled by using a 2 noded contact element Contac52. This element can simulate two surfaces which may maintain or break physical contact and may slide relative to each other. The element is capable of supporting only compression in the direction normal to the surfaces and shear in the tangential direction. A normal stiffness is defined for the element based on the stiffness of the surface in contact. The stiffness for the present case is calculated using equation-4.7.
\[ K = \frac{EA}{t_b} \] (4.7)

The effect of support plate is modeled by generating additional set of nodes, coincident, with the nodes at the bottom of substrate. These additional set of nodes are completely constraint during the entire solution, to simulate the rigid base. Contac52 elements are generated between these coincident set of nodes with one node attach to rigid base while the other node attached to the bottom of the substrate. Since these node sets are coincident therefore they represents an initial close gap condition or an initial perfect structural contact between the substrate and support plate. As the gap is initially closed therefore the nodes corresponding to substrate base cannot move down.

The substrate plate is bolted at the corners to the support plate. This structural boundary condition is simulated by constraining all degrees of freedom of all the nodes lying within a square of 20mm x 20 mm at each corner. As the deposition ends and the substrate cools down to room temperature the plate is unbolted by removing the constrains from these nodes.

4.4 Experimental Validation

The results from the 3D FE model were compared with the experimental data. Both the thermal and structural models were validated with two different experimental results each. The thermal model was verified firstly by comparing the experimentally obtained fusion zone with numerically predicted fusion zone of the weld bead. Secondly the numerical and experimental temperature histories were compared. For structural model validation, firstly the deformation results and secondly the residual stress results from experiments were compared with the FE simulation.

4.4.1 Validation of Thermal Model

The criterion for the validation of thermal model was based on correct estimation of the fusion zone and the temperatures. The methodology adopted is described below.
4.4.1.1 Fusion Zone Measurement

For thermal model validation, numerically predicted fusion zone is compared with experimentally determined one. A bead on plate experiment was conducted to measure the fusion zone with the deposition parameters similar to the ones used for FE simulation. The weld bead was deposited geometrically in the centre of the substrate plate to avoid unsymmetrical heat transfer conditions. A sample with rectangular cross section was cut using a wire cut machine for metallography. In order to ensure quasi-static welding conditions the sample was cut from mid-length of the weld bead. For metallographic sample preparation first of all it was sequential ground with silicon carbide (SiC) abrasive papers having grit size P320, P600, P800 and P1200 (European P-grade) chronologically. Secondly it was sequentially polished with diamond pastes; with particle sizes 9μm, 6μm, 3μm and 1μm respectively. Finally the sample was etched with 2% nital solution. Figure 4.10 shows the front view of the sample with the fusion zone and heat affected zone revealed with obvious outline.

The position of liquidus isotherm for experimental and numerical weld bead is shown in figure 4.11. It can be seen that the size and the shape of the fusion zone, from the FE model, are in conformity with the experimental results therefore the model can be used reliably for the prediction of remelting depth.

4.4.1.2 Temperature History

For further validation of the thermal model temperature history of the experimental and numerical study was compared. A K-type thermocouple was spot welded to the top of a grinded substrate plate and was used to measure the changing temperatures. The thermocouple was placed at a distance of 5mm (point-E in figure 4.6) from the first deposited weld bead in the central section of the substrate; and this was done to avoid excessive temperature rise and the danger of being affected by spatter. A single layered specimen was manufactured with deposition parameters similar to what were used in finite element modelling and bead on plate experiments. The substrate plate was placed over the support plate. The support plate was machined and ground so that a good thermal contact can be achieved between the top of support plate and the bottom of substrate plate.
Figure 4.10: Metallographic Sample for Fusion Zone Measurement

Figure 4.11: Experimental and Predicted Fusion Zone
Figure 4.12: Temperature history at $E$

Figure 4.13: Setup for Deformation Measurement
The experimentally determined temperature history along with the simulation results is shown in figure 4.12 for the first five weld passes. It can be seen that the predicted maximum temperature almost matches the experimentally determined maximum temperature and the rise in temperature with each pass is also reasonably similar. The cooling rates of both the simulation and experiment are also in reasonable agreement. The model therefore can be utilized for qualitative analysis and comparative study of the thermal effects of different deposition parameters.

4.4.2 Validation of Structural Model

As for the thermal model, the structural model was also validated in two phases. The distortions and residual stresses were validated separately but with same specimen. A 160×110 mm substrate plate was made out of a 12 mm thick CS plate. This substrate plate was ground on a surface grinding machine to a thickness of 10 mm. The plate was then taken to a furnace for stress relieving. The thermal cycle adopted for stress relieving is as stated below:

- heated up to 600 °C
- kept at 600 °C for 50 min.
- cooled at a rate of 75 °C per hour to 300 °C.
- cooling to the ambient temperature in the furnace with furnace turned off.

The substrate plate was then bolted at the four corners to a 200×130×34 mm support plate. The top and bottom surfaces of the support plate were also ground on a surface grinder to ensure a flat contact between it and the substrate plate. This was also required to ensure a flat reference surface when placed onto a surface plate for deformation measurements. A single layer of weld metal (100×50×3 mm) was deposited on the centre of the bolted substrate in nine weld passes with an inter-pass time of 180 sec. The welding parameters used are shown in table-4.1.

4.4.2.1 Deformation Results

The structural model was validated for deformation results by comparing the deflections obtained from the FE simulation with the experimental results, both in the substrate
bolted and unbolted condition. The coordinate measuring capability of a CNC milling machine was used to measure the out of plane distortion of the substrate plate when a single layer of metal was deposited. A dial gauge (least count of 1 micron) was used to measure the z-coordinate of predefined points on the surface of the substrate plate. The location of these points were selected clear of the weld area and bolts. Measurements were taken both in the substrate bolted and unbolted condition. Figure 4.13 shows the distortion measuring setup.

The same deposition parameters were used for both experimental deposition and FE simulation. The deflections obtained from FE simulation and experiments along Line MN and OP, for a bolted substrate, are shown in figure 4.14. The location of the lines is shown in figure 4.6. The simulation results are in excellent agreement with experimentally obtained deflections. The maximum difference for Line MN as well as for Line OP is not more than 6%. During deposition the thermal cycling results in compressive reaction forces at the plate edges. These reaction forces will cause the substrate plate to deflect upward. This upward deflection has a maximum at the mid-length of the plate and decreases on both the sides.

Figure 4.15, show the variation in deflections along MN and OP, when the substrate is unbolted. The numerical and experimental deflections are qualitatively in good conformity. The deformation pattern and the peak deflection values are predicted quite accurately. The predicted peak deflection value at start side (MN) is 0.48 mm as compared to an actual deflection of 0.5 mm, while on the end side (OP) it is 0.43 and 0.45 respectively. The error between measured and predicted values is very small at the deposition mid-length but increases progressively toward the edges, which were bolted during deposition. This increase in error is probably due to the simplified bolt model used. During simulation bolting was modeled by constraining all the nodes present within the bolted area to have zero displacement; in other words the simulation assumed bolts with infinite stiffness resulting in an infinite bolting force. However, in actual practice the bolts have finite stiffness therefore some of the reaction force will be taken up by the bolts; moreover the rise in temperature of the bolts will also reduce there stiffness. Hence the simulation underestimates the deflection, in unbolted condition, as the model experience over-constrained conditions in comparison to the actual substrate.
Figure 4.14: Comparison between experimental and FE deformation for Substrate bolted condition.

Figure 4.15: Comparison between experimental and FE deformation for Substrate unbolted condition
Figure 4.16: Comparison between experimental and FE deflection along Line AB.

Figure 4.17: Exploded View of Substrate Fixture for RS-200
The warping behavior of the substrate is presented in figure 4.16, by plotting the deflections along $AB$. The numerical results are compared for both bolted and unbolted substrate and are found to be in good agreement. When the substrate is bolted the weld end side ($OP$) gives higher deflection but this trend reverses on unbolting. This may be due to large thermal gradients created across the substrate thickness when the deposition starts.

### 4.4.2.2 Residual Stresses

For validation of stress results the experimentally determined axial and transverse residual stresses (along line AB figure 4.6) are compared with the numerical prediction. A center hole-drilling strain gage method was implemented for experimental determination of residual stresses. Since the welding residual stresses are known to be highly non-uniform through the thickness of the weldment, integral method for incremental blind hole-drilling was selected.

Hole-drilling is considered a semi-destructive technique for the determination of the residual stresses. In this technique a small hole is drilled at the centre of a special three-(or six-)element strain gage rosette, pasted on the surface of the stressed component. Strain is released because of the drilled hole which is measured. Using special data-reduction relationships, the principal residual stresses and their angular orientation are calculated from the measured strains. It is highly recommended that the stresses induced due to machining process involved in the hole-drilling should be as small as possible. In order to keep the machining stresses negligibly small, two techniques, namely high speed drilling and air abrasion are conventionally used. In the present work, the former technique was implemented and commercially available hole-drilling equipment, RS-200 milling guide by Vishay Measurement group was used.

If the size and shape of the component under investigation permits, the equipment is directly fixed over it. However it was not possible to directly attach the milling guide over the substrate to obtain the desired alignment accuracy. This was because of to its size and location of deposited weld on it. Therefore a special fixture was developed to hold the substrate plate. The fixture, shown in figure 4.17, and 4.18 was designed to provide a flat smooth platform for milling guide installation and to provide necessary but
Figure 4.18: Assembly of Substrate Fixture for RS-200

CEA-06-062UM-120

Figure 4.19: Strain gage rosette type CEA-06-062UM-120
adequate support to the work piece without introducing additional stresses in it. The accuracy of the method depends on the precision with which the various steps are carried out such as proper surface preparation, correct strain gage rosette selection, installation of strain gages, accurate alignment of the hole and correct selection of incremental hole depths. Experimental procedure used for residual stress calculations are as follows:

- **Strain Gauge Rosette Selection:** The choice of the strain gauge depends on the nature of the residual stresses, location and the depth to which the measurement is to be performed. In the present study the residual stresses were measured up to a depth of 1mm from the surface therefore 120 ohms, 3 gauge rosettes of the type CEA-06-062UM-120 was used. In addition this rosette type is specially designed to affix very close to the weld toe. The configuration of this rosette is given in figure 4.19.

- **Surface Preparation:** Strain gages can be satisfactorily bonded to almost any solid material if the material surface is properly prepared. The purpose of surface preparation is to develop a chemically clean surface having a roughness appropriate for the gage installation requirements, a surface alkalinity corresponding to a pH of 7 or so, and the surface should be clean enough so that the strain gauge lines for locating and orienting the strain gauge are visible. Abrasion during surface preparation can alter the near surface residual stresses, as reported by Pervy [194], therefore a very gentle abrasion was made according to the instruction laid in Anon [195].

- **Strain Gage Bonding:** The special purpose strain gage rosette CEA-06-062UM-120 manufactured by Vishay Micro-Measurements Group was used at the central section (along line AB figure 4.6), on both sides of the deposition 5mm and 20mm away from it and on the substrate top. This section is the most appropriate location for residual stress measurement to characterize general stress behaviour. Strain rosettes were pasted on the work piece according to the instructions laid in Anon [196].

- **Connecting Strain Gages:** Proper connection of lead wires to gauge tabs and connecting terminals is important for the fatigue life of strain gage, Anon [197]. Lead wire connections were made in the light of instructions given in [197-199]. After attaching lead wires, it was checked whether strain rosettes were properly bonded to
Figure 4.20: Substrate plate with strain gauges attached.

Figure 4.21: RS-200 System attached to Fixture Plate
the surface and whether strain gages were properly connected. The substrate plate with the strain gauges attached on it is shown in figure 4.20.

- **Alignment and Hole Drilling:** After placing the test specimen in the fixture (shown in figure 4.17 and 4.18), the milling guide assembly RS-200 was mounted on it and aligned with each rosette one by one. Figure 4.21 shows the substrate plate placed in the fixture with the milling guide aligned to a selected strain rosette. For each rosette, the lead wires of each strain gage were connected to a static strain indicator Model P3 (Vishay Micro-Measurements Group) making a quarter bridge circuit. After ensuring zero-balance of bridge circuits, the hole was drilled in successive increments and the released strain after each depth increment was recorded. The whole hole-drilling process was strictly in compliance with procedure laid down in Anon [200]. Total stress measurement depth is usually limited to 0.3-0.4 \( r_m \), where \( r_m \) is the mean radius of strain gauge rosette. As suggested by Schajer [201], complete hole depth (which comes out to be 1.02 mm for our application) was achieved in four steps and the hole depth increments were made progressively larger with the depth from the surface. The increments of 0.12, 0.2, 0.3 and 0.4 mm were used for successive steps in the hole-drilling.

- **Data Reduction:** The blind-hole geometry is sufficiently complex and no closed-form solution is available (from the theory of elasticity) for direct calculation of the residual stresses from the measured strains - except by the introduction of empirical coefficients [200]. The empirical coefficients for integral method, usually called calibration coefficients, depends on geometry of the specimen, strain gage rosette configuration, diameter of hole and depth of hole increment. Finite element technique is usually employed for determination of these coefficients. Numerically determined and experimentally validated calibration coefficients for a flat plate are available in the literature, as for example in [201,204]. In the present work, a commercially available data reduction software, H-Drill [202] developed by Schajer, was used.

The experimentally determined axial and transverse residual stresses are compared with corresponding simulation results in figure 4.22 and 4.23, respectively. The results are plotted at the top of the substrate along the central cross-section (line \( AB \) figure 4.6). The
Figure 4.22: Verification of axial residual stress

Figure 4.23: Verification of Transverse residual stress
finite element results are in qualitative agreement while quantitatively the finite element model can’t give reliable results. The deviation of experimental data from the calculated stress may be because of several reasons including approximation in mechanical properties of material involved, errors in calculation of calibration coefficients, possible eccentricity in the hole, probability of hole depth measurement error etc. It may also be noted that the incremental hole-drilling technique is more reliable up to about 70% yield stress i.e. approx 175 MPa for mild steel. Above this level, the stress calculated from the conventional analysis may be overestimated by up to 15% at full yield [205]; so to attempt to verify stresses of about 200 MPa by this method will produce errors. Another thing which may be noted is that the experimental stress results are for an unbolted substrate; and in conjunction the experimentally determined deformation results for the unbolted substrate also show more deviation in comparison with the FE deformation. This increased error once again may be attributed to the simplified bolt model used. During simulation bolting was modeled by constraining all the nodes present within the bolted area to have zero displacement; in other words the simulation assumed bolts with infinite stiffness resulting in an infinite bolting force. However, in actual practice the bolts have finite stiffness therefore some of the reaction force will be taken up by the bolts; moreover the rise in temperature of the bolts will also reduce there stiffness. Hence the simulation gives error in unbolted condition, as the model experience over-constrained conditions in comparison to the actual substrate. The comparison with experimental data shows that axial stress can be determined more correctly and since axial stresses are more dominant of the two, the model can therefore be used for fair prediction of axial residual stresses. In addition since the results are in qualitative agreement the FE model can be reasonably used in the studies where a comparison is made between different deposition parameters.

It is very evident from the plots that the axial stress is more dominant than the transverse stress. The axial stresses also vary between a tensile and compressive stress range while the transverse stresses are limited to the tensile range only. The magnitude and the distribution of axial stress is a strong function of ‘X’ co-ordinate as compared to the transverse stress. Therefore further results will be discussed with reference to the axial stress mainly. The peak axial stress are found in areas near the deposition end side (X =
80 mm), while lower residual stresses exists at the deposition start side (X = 30mm). This variation occurs due to the reheating affect of successive passes.

### 4.5 Conclusion

A finite element model was developed to predict the deformations associated with molten metal deposition as applied to layered manufacturing with specific application to process employing gas metal arc welding. Validation methodology followed in this dissertation is described and results of thermal and thermal-stress analysis are validated with the experimental measurements. The results show that the model can accurately predict the remelting depth. The substrate deflection for the bolted case is in excellent conformity with experimental data, while for unbolted case a qualitative agreement is concluded, due to a simplified bolt model. The finite element results are in qualitative agreement while quantitatively the finite element model can’t give reliable results. Since different finite element studies performed during the course of present research are parametric in nature and their main objective is to describe structural response under varying set of parameters. Therefore, a qualitative approach, which can sufficiently describe the general behavior, is considered sufficient to validate results. The axial residual stresses are much higher and a stronger function of position than the transverse residual stress.
CHAPTER 5

The Effects of Machining on Material Properties

In this chapter a comparison of the material properties is made between weld based prototype parts, produced both with and without intermediate machining after each deposited layer. Material properties were investigated both on a macro and microscopic level. The microstructure for the two deposition procedures were studied and compared. The hardness test results for the two procedures were investigated and the results were studied in the light of the respective microstructures. Tensile test samples were developed and testing was performed to investigate the directional properties in the deposited materials. The results presented in this chapter have been published in two international conferences and the International Journal of Materials and Product Technology [206-208].

5.1 Introduction

At present few RP machines are available commercially which can mainly produce parts that can be used either as models for visualization or for rapid tooling. The emphasis of the on going research in this field is to produce parts that can physically imitate and work like a component produced by a conventional manufacturing technique. Parts made by metals are of specific interest and welding based RP has good prospects in this regard [124,129,130,140]. Gas metal arc welding (GMAW) can be used in the development of a cost effective method for layered manufacturing (LM) of fully dense metallic parts and tools. However, involvement of high temperatures in welding causes accumulation of residual stresses thus resulting in warpage, delamination and poor surface quality [130,117,123]. Parts thus produced are of near net shape or out of tolerance. This therefore does not fulfill the criterion for fully functional parts. Research is being done to
improve and find welding parameters suitable for the RP process (i.e. parameters that impart minimum heat to the underlying geometry and produce good surface quality) [106, 117, 139]. Alternate welding techniques (GTAW, VP GTAW, EBF etc.) are also being explored in this regard [129,130,140,131,209,210]. Another option is the combination of GMAW as an additive and CNC milling as subtractive tool to manufacture parts with required tolerances [106, 126, 211, 212]. The drawback however, is the increased manufacturing time, material wastage, process complexity and cost. As the deposited layer height of deposition with intermediate machining (DWIM) will be less than that of deposition without machining (DWM), the level of thermal cycling of the underneath layers for the two buildups will be different. This may therefore result in different microstructures and thus different material properties.

For being fully functional it is important to know the material properties and the mechanical behaviors of the finally produced RP parts part. The layer by layer additive manufacturing nature results in non-homogeneous structures, porosity and anisotropic material properties. However these properties can be improved in some cases by post processing. Research work by Ashley [100] indicated that nearly fully dense part can be achieved in sintering via post processing. Kietzman et.al [213] reported anisotropic properties for castable thermoset resins in polymer shape deposition manufacturing (SDM), however an improved interlayer adhesion by post processing was also achieved. Ahn et al. [214] discussed the directional behavior of RP parts made by fused deposition modeling (FDM) and proposed a computer model to predict the anisotropic tensile failure. Directional behavior was also reported in microcast mild steel parts [113]. Song et.al [212] showed that the alternating deposition direction between the layers influences the tensile strength significantly. Parts made by GMAW on the other hand show excellent mechanical strength, reasonable isotropy and dense structures with minimal inclusions [116,117]. The above findings, however, are based on the experiments on a macroscopic level. Actual RP is complicated by the fact that each subsequent weld pass modifies the properties of the prior pass and may also alter the properties of other prior passes. Arc welding of steels causes major localized changes in microstructure due to the imposed thermal and stress cycles and the melting and mixing of filler metal with the base material. Considering the absence of macroscopic weld defects, such as porosity,
undercut, lack of fusion, and hot or cold cracking, the reliability of the weld deposit and therefore the RP part, under specified service conditions will depend on the mechanical (and sometimes chemical) properties of the solidified weld metal zone and the heat affected zone (HAZ) [215]. These properties in turn depend on the microstructures of the solidified weld bead and the HAZ.

The nature of RP buildup via welding is similar to multipass welding. Multipass welds in steels are generally metallurgically heterogeneous and consequently the mechanical properties can vary from region to region. The deposited metal is influenced significantly by the additional thermal cycles induced by the deposition of subsequent passes. Only the final pass deposited is un-tempered (i.e. has a primary microstructure). The remaining regions of the deposit undergo transient temperature rises high enough to cause partial or complete reverse transformation to austenite, which on subsequent cooling retransforms to ferrite, but not necessarily to the same microstructure as the primary regions. The regions which do not experience peak temperatures high enough to cause reversion to austenite, are tempered to an extent which is dependent on factors such as the starting microstructure and alloy chemistry. The evolution of microstructure in the weld metal is highly complex, and depends on the composition, composition gradients, and other variables such as cooling rate and peak temperature [216]. However, for a solidified weld bead subjected to a series of attenuating thermal pulses due to subsequent passes, only the first and second thermal cycle are likely to result in significant reaustenitisation [215]. Very low toughness values may occur in the HAZ which are mainly associated with small areas within the coarse grain HAZ called the ‘local brittle zones’ (LBZ). The main contributors to the low toughness LBZ’s are, upper bainite, precipitation, large grain size and martensite islands [217].

The multipass nature of the weld buildup, on the other hand, results in a few benefits which include ‘preheat’ from previous weld passes, annealing out of residual stresses due to previous weld thermal cycles, and structural refinement of coarse solidification structures [216]. As the microstructure of the buildup comprise of various zones, the overall mechanical properties of the weld metal are expected to be determined by the combination of the properties of all of the different zones. However, the relative importance of reaustenitised and un-reaustenitised weld metal in determining mechanical
properties of multipass welds remains unresolved. Different reported studies have conflicting results about the mechanical properties of the different weld zones [217-219]. The constituents of the microstructure of deposit may depend on a number of factors. Song et.al [211] reported the changes in grain size with changes in heat conduction rate. The effect of part geometry and therefore the change in heat sink capacity on the microstructure has also been presented by Jandric et.al [209]. However the hardness results of the two are contradictory with each other. Song et.al [211] reported hardness values to decrease gradually from bottom of deposit to the top, while Jandric et.al [209] reported a maximum hardness at the top deposited layer and a slight decreasing trend towards the middle and bottom layers.

In this work a comparison between the two approaches of buildup (i.e. DWM and DWIM) is done to identify the microstructural constituents and their arrangement in the deposit, and consequently to study the hardness and tensile strength behaviors of the resulting GMAW based RP parts.

5.2 Experimental Details

The study was conducted using Pulsed GMAW, manufacturing slabs of deposit, by following both DWIM and DWM procedures. Layers of weld were deposited on 12 mm thick mild steel substrate, which was bolted at the four corners to a 30 mm thick support plate. The buildup was done on a semiautomatic RP system developed by combining Lincoln’s Power Wave 350 GMAW machine with a milling machine. A 1.2 mm diameter ER70S-6 wire was used to build slabs on the substrate using CO₂ as the shielding gas. For DWIM samples, the wavy surface of each deposited layer, approximately 3 mm thick, was machined to 1.8 mm thickness and then the next weld layer was deposited on it. The pulsed welding parameters used for buildup are shown in table-5.1.

5.2.1 Metallographic Sample Preparation

Samples were cut out from both types of deposition across the weld direction for metallography. The samples were ground, polished and then etched with 2% nital solution for microscopic examination.
5.2.2 Hardness Test Samples

The same samples, used for metallography, were afterwards used for hardness testing. Vicker’s hardness testing was carried out on the deposited cross-section on a 1x1 mm grid both in the horizontal and vertical direction of the polished face. The testing was done on the DVK-2 Matsuzawa Vicker’s hardness testing machine with the following settings:

- Test Force (Indenter Load) = 5 kg
- Time of application of Test Force = 10 sec
- Loading Speed = 50 µm/s

The testing machine was also alongside tested for accuracy with standard hardness test block and an error of 0.98% was found with respect to it. This error is within the stipulated limit of 2% laid down by the ASTM standards [220].

5.2.3 Tensile Test Samples

Rectangular tension test specimens (figure 5.1), machined as per subsize specimen dimensions mentioned in ASTM E-8M [221], for both DWIM and DWM were cut out of the deposition longitudinal as well as perpendicular to the weld deposition direction (figure 5.2). The samples were taken from the center of the deposition so as to remove any material which was influenced by the substrate, or which had not undergone the full thermal cycling due to absence of any further layers on top of it. The samples were machined to a thickness of 5 mm. The testing was conducted on Instron 5567, 30 kN tensile testing machine.

5.3 Micrographic Analysis

HAZ in a single pass weld of structural steels consists of four distinct zones, i.e. coarse grain zone, fine grain zone, intercritical zone (IC) and subcritical (SC) zone. Each of
### Table 5.1: Welding parameters

<table>
<thead>
<tr>
<th>Parameter</th>
<th>Value</th>
</tr>
</thead>
<tbody>
<tr>
<td>Average Welding Current (A)</td>
<td>135</td>
</tr>
<tr>
<td>Voltage (V)</td>
<td>23.5</td>
</tr>
<tr>
<td>Weld Speed (mm/min)</td>
<td>375</td>
</tr>
<tr>
<td>Wire feed (mm/sec)</td>
<td>4.91</td>
</tr>
<tr>
<td>Gas flow rate (SCFH)</td>
<td>35</td>
</tr>
<tr>
<td>Stick out (mm)</td>
<td>16</td>
</tr>
<tr>
<td>Weld bead to weld bead Offset (mm)</td>
<td>4.5</td>
</tr>
</tbody>
</table>

Figure 5.1: Dimensional drawing of tensile test specimens made as per ASTM E-8M [221]. (All dimensions in mm).

Figure 5.2: Tensile test specimens cut longitudinal and perpendicular to the weld deposit direction.
these zones is formed by the local thermal cycle which is experienced during welding. In a multipass weld, some region in these zones undergo multiple thermal cycles, resulting in partial or complete alteration of the HAZ microstructures from the prior pass [215,216]. A typical multipass weld HAZ consists of localized and discontinuous zones with a mixture of numerous microstructural constituents. Some areas in the HAZ contain the required microstructures while others may comprise of unfavorable or unwanted microstructural features which have different properties than what are required.

The microstructure of samples from both types of deposition was comprehensively studied. It was observed that there is a good structural integrity and fusion between the weld beads. The microstructure of the entire body of the deposit can be divided into two main regions i.e. the un-re austenitised and the re-austenitised. The microscopy revealed that the microstructure of the un-re austenitised region comprises mainly of acicular ferrite, bainite, and ferrite-carbide aggregate in coarse grain boundary ferrite (figure 5.3). There are also some quantities of martensite, widmanstatten ferrite, intragranular polygonal ferrite, and spheroidal cementite (all not shown here). On the other hand the microstructure of re-austenitised region is mainly containing polygonal ferrite with some fine pearlite at grain boundaries (figure 5.4). To visualize the different zones made by the multipass deposit consider the schematic of figure 5.5, which shows the cross section of deposited weld layers. Figure 5.5(a), Shows the schematic for a DWM sample, while figure 5.5(b) shows the schematic for a DWIM sample. The columnar solidified weld section in figure 5.5 represents the un-re austenitised while the re-crystallized (coarse and fine) section represents the re-austenitised region of the deposit. It can be seen from figure 5.5(a) that the un-re austenitised region exists in each deposited layer, while it is absent in figure-5.5(b), except for top layer. With the reason being the lesser layer thickness of DWIM as compared to DWM samples. This is because, each deposited layer (in DWIM) is partly removed by machining, and the remaining is then re austenitised by the successively deposited layer. On the other hand, in the DWM samples, the effect of re austenitisation does not penetrate the entire depth of the layer thus resulting in some remaining unre austenitised structure at the bottom of the layer. Figure 5.5 also elaborates the re-austenitised region divided into the re-crystallized coarse and fine zones. It can be visualized from the schematic (figure 5.5) that the central body of the deposit without any
Figure 5.3: Weld microstructures of un-reaustenised region. 2% nital. 500X
Figure 5.4: Weld microstructures of reaustenitised region. 2% nital. 500X.
(a) re-crystallized coarse (b) re-crystallized fine.

Figure 5.5: (a) Schematic of the DWM (b) schematic of the DWIM, showing different sub zones.
intermediate layer machining will comprise of a more complex microstructure as compared to the one with intermediate machining. The existence of these different microstructure regions is discussed in the next chapter on the basis of peak temperatures reached during deposition.

A sequential variation is observed in the microstructure across the top weld layer of samples from both DWM and DWIM (along arrow direction-1 figure 5.5(a) and (b)). It is observed that the microstructure changes from predominantly un-reaustenitised (Figure 5.3) to predominantly reaustenitised (Figure 5.4) structure after every 4.5 mm. This is in perfect conformity with the weld bead inter spacing which is also 4.5 mm. The formation of the ferritic microstructure in the top layer is the result of reaustenitisation by each succeeding weld pass on the side. Microstructure of the top layer of DWIM (along arrow direction-1 figure 5.5(b)) is different from that of the layers underneath. This is because the top layer does not face any subsequent thermal cycle from another weld layer on it.

The microstructure of the central layers of the two deposition techniques, however, is different from each other. Microscopic examination of the central deposit of DWM reveals bands of reaustenitised and un-reaustenitised region appearing sequentially on each other. However, across the un-reaustenitised band (along arrow direction-4 figure 5.5(a)) there exists another sequential variation from reaustenitised to un-reaustenitised microstructure. The microstructure of the reaustenitised bands (along arrow direction-3 figure 5.5(a)), on the other hand, shows a sequential change in the grain size across the weld direction (Figure 5.4(a) & (b) shows the difference in grain size). The sequence is repeated every 4.5 mm, which once again is in agreement with the weld bead interspacing. The central layers of the DWIM, in contrast, constitute of the reaustenitised microstructure only, with a similar fluctuation in the grain size in bands placed alternately on each other (along arrow direction-3 figure 5.5(b)) (microstructure as in Figure 5.4). The sequence is also repeated every 4.5 mm. The change in grain size is the result of multiple reheating of preceding beads by subsequent weld passes. The net effect of reheating the HAZ therefore, is restructuring of the microstructure. Two sub-zones thus formed in the HAZ are the re-crystallized coarse (RC) and the re-crystallized fine (RF) (Figure 5.4). It may however be noted that the grain size in the HAZ is a function of peak temperature, which depends on the distance from the heat source [219]. The average
grain size of the RC sub-zone was calculated to be 10.8 µm while that for the RF sub-zone is 6.54 µm. Apart from these varying grain size bands, it can be seen that there are alternate layers which comprise of the RF zone only (along arrow direction-2 figure 5.5(a) and (b)).

The microstructure of successive layers deposited by microcasting for the same low carbon steel (ER70S-6), as reported by Merz [113], contains columnar martensite, in addition to the fine grain structure of heat effected zone (figure 5.6). The presence of martensite is due to the lesser depth of HAZ caused by the successively deposited layers by microcasting. Welding on the other hand causes much deeper HAZ, thus modifying the microstructure of the underneath layers and resulting in predominantly bainitic / ferritic microstructure.

As stated above, the HAZ of a weld consists of four sub-zones, namely the RC, RF, IC and SC. The IC and SC sub-zones are formed due to lower temperatures as compared to the RC and RF sub-zones. These are also created in very small thickness bands as compared to the other two. However these were only found in the HAZ of the first deposited layer i.e. in the substrate. The reaustenitised weld region of the remaining layers comprised only of the RC and the RF sub-zones. This is because the peak temperature at any sub-zone, either due to the first or any succeeding weld pass, dictates the final microstructure of the location. Therefore the microstructures of doubly reheated sub-zones are almost similar to the sub-zones of singly reheated weld regions, depending on the highest temperature reached. The IC and SC sub-zones are therefore, either modified or not created in the first place. This result is consistent with literature already reported by Dunne et.al [215] and Easterling [216].

The microstructure at the interface between the weld layer and the substrate plate mainly constitutes of a-ferrite with grain boundary pearlite and cementite and some fine martensite. As a result of pearlite-type reaction different complex cubic carbides are also visible. The grain size is quite variable, revealing areas of the base plate, weld material and heat effected zones (Figure 5.7). The IC and SC sub-zones can also be seen in the HAZ region of the substrate plate.
Figure 5.6: Droplet / substrate interface zone of microcast mild steel showing columnar martensite present in the droplet region (curtesy R. Merz [113])

Figure 5.7: Microstructure the interface of weld and substrate. 2% nital. 500X
Moving down in vertical direction from the top of the deposit to the substrate plate one passes through all the different structures of the different layers. Leaving the top layer and the interface layer, in the DWIM sample, the microstructure of the deposit shows a sequential change in grain size, from RC to RF, after approximately every 1.8mm and this is in good agreement with the deposited layer thickness. The reason for this sequential change is the same as discussed earlier. This microstructural variation is similar as shown in figure 5.4. On the other hand, in DWM samples, the microstructure oscillates between predominantly acicular and polygonal ferrite structures after every 3mm approximately, which is also in agreement with the layer thickness.

From the above discussion it can be concluded that the microstructure of the DWM deposit is much more multifarious as compared to the DWIM deposit, in which the central body of the deposit has a fairly homogeneous microstructure. This difference is due to the reduced deposited layer height, in the DWIM, caused by the intermediate milling operation. Neglecting or removing the top and bottom layers of the buildup by DWIM, will therefore result in a relatively homogeneous microstructure (mainly polygonal ferrite). Although the grain size will be sequentially fluctuating between coarse and fine. A homogeneous microstructure therefore may result in nearly isotropic material properties and higher toughness as compared to that of DWM.

This study was based on a single set of deposition parameters and simple part geometry (i.e. rectangular slab). The microstructure and hence the amount of reaustenitised and unreaustenitised areas is dependent on the heat input and the degree to which adjacent weld beads overlap, which in turn depends on a number of factors including the welding parameters, preheating, machined layer height, interbead spacing and geometry of the part. It is of interest to examine the case for DWM, where an underlying bead is completely reaustenitised by the deposition of molten metal. Reed and Bhadeshia [222] had suggested that when the weld beads are placed directly over the weld beads in the underlying layer, complete reaustenitisation occurs when the average reaustenitised thickness is greater than or equal to the average height of the deposited upper layer. It may therefore be possible to achieve completely reaustenitised microstructure in the DWM. Further studies therefore in this regard are required to be conducted to identify a relationship between the heat input, the extent of bead overlap and the amount of
reaustenitised region. Furthermore studies to identify the relationship between the heat input and the minimum height to which machining is done in a DWIM may be conducted, to achieve complete reaustenitisation or else the microstructure of choice.

5.4 Hardness Testing

The hardness values in the top layer of both DWM and DWIM samples (along arrow-1 figure 5.5(a) and b)) shows a fluctuating wavy profile with the peaks repeating after every 4 to 5 mm (figure 5.8). This is in accordance with the weld bead interspacing (i.e. 4.5mm). The fluctuating result is also in accordance with the microstructure of the top layer where the peak values represent the predominantly acicular portion and the lower values represent the predominantly polygonal portion of the microstructure. The average hardness of the layer is HV 202.98 for DWM sample and 194.1 for DWIM sample. The lower hardness value of the DWIM sample is due to the final machining of the top deposited layer, thus removing the tough top.

Three different hardness profiles are observed in the central deposited layers (designated as central layer 1,2 and 3 in figure 5.9). A similar profile as in the top of DWIM was observed in the different central layers of the DWM samples (central layer 1 figure 5.9), thus verifying the alternate presence of both reaustenitised and un-reaustenitised regions. This fluctuation was however not found in the central layers of DWIM samples, which verifies the presence of reaustenitised region only. Another fluctuating wavy hardness profile of relatively lesser magnitude was also observed on different alternate central layers of both DWM and DWIM samples (central layer 2 figure 5.9). This corresponds to the variation from RC to RF grains (along arrow-3 figure 5.5(a) & (b)). The higher hardness values correspond to the RF region and the lower to the RC. A third hardness profile observed in the central layers of both type of depositions is shown by central layer 3 figure 5.8. This represents the RF segment as shown along arrow-2 figure 5.5(a) & (b), and the profile shows narrowly scattered hardness values. (Note: central layer 2 and 3 figure 5.9 present hardness profiles for DWM sample; a very similar profile was observed for the DWIM sample)

A wavy behavior is also observed in the hardness profiles of both DWM and DWIM samples, near the interface of weld and substrate plate (figure 5.8). However the hardness
Figure 5.8: Hardness profiles for top and interface of DWM and DWIM

Figure 5.9: Hardness profiles for top and central layers of DWM
Figure 5.10: Hardness profile from top to interface for DWM and DWIM

Figure 5.11: Hardness profiles at different levels along the weld direction for DWIM
values are comparable to that of the top layer of DWIM. This may be due to the presence of fine martensite and α-ferrite microstructure in the heat affected zone. The average hardness for this region is HV 200.

Figure 5.10 shows the hardness profiles from top deposited layer to the interface of weld and substrate, for both types of deposition. The results represent an average of 20 contiguous locations (1mm apart) for the DWM and 13 contiguous locations (1mm apart) for DWIM samples. It can be seen that, for DWIM, the hardness at the top and near the deposit/substrate interface are higher, while these are lower in the central layers. However for DWM only the top layers hardness values are higher as compared to the central and interface regions. The hardness of the top and central layers is also lesser in DWIM as compared to the DWM. The higher value of hardness in the top layer of DWM is as explained previously. However the higher values in the central layers confirm the presence of un-reaustenitised microstructure. The average hardness of the central layers of DWM samples was found to be HV 192.2 while that for DWIM samples it was slightly lower i.e. HV 181.3. Moreover the hardness profile in the central portion is also fluctuating for both the cases and corresponds with the respective layer thickness of DWM and DWIM (i.e. 3mm and 1.8mm respectively).

Figure 5.11 shows the hardness profiles at the different levels for a DWIM sample cut along the weld direction. The deposition on its top shows little variation in hardness profile with no regular waviness. The average hardness of this layer is HV 195.6. Similarly the center of deposition shows no regular waviness with an average hardness of the layer about the same HV 178.3. This result is in accordance with the microstructure of these layers. Further more no wavy behavior is observed in the hardness profiles at the interface of weld and base plate with an average hardness of nearly HV 183.9. The hardness values in the substrate region are also shown in the figure 5.11 which does not show much variation with an average value of HV 154.9.

The hardness results discussed above are in good agreement with the microstructure of the corresponding layers. The main difference in DWM and DWIM is therefore found in the central layers.
5.5 **Tensile Testing**

Tensile tests of samples built by the two techniques, DWM and DWIM, were conducted on specimens produced both longitudinal and perpendicular to the weld deposit direction (figure 5.2). The minimum, maximum and average values for the tensile strength, yield strength and the elongation at break are shown in table-5.2. Visually, the tensile specimens show no evidence of any difference in the crystal structure on fracture, and the material strength is also almost the same for the two types of samples. However there is a slight difference in the average elongation of longitudinal and perpendicular samples built by both the techniques. The results however, reported by Merz [113] for the same material (i.e. ER70S-6) deposited by microcasting are quite different; clearly revealing anisotropic tensile properties with a very low value of elongation as well. On the other hand the weld based results, in general, show nearly isotropic material properties; which means that the bonding in the direction perpendicular to the orientation of the deposit is equally strong as in the longitudinal direction.

The tensile strength range for ER70S-6 weldments is specified as 551.58 - 586.05 MPa, and the yield point as 448.16 - 482.63 MPa, while the elongation is specified at 28.5% [223]. It can be seen from table-5.2 that the values are within the specified range, except elongation for the longitudinal samples which on the average are slightly higher (Figure 5.12). As already discussed, there are alternating coarse and fine grain regions with a slight difference in hardness across the weld direction. Chen *et al.* [218] carried out a comparative study on the toughness of specimens from multilayer C–Mn steel welds, which led to the conclusion that the weakest region is expected to be the reheated coarse region. Therefore it can be anticipated that the breakage will take place from this region for perpendicular samples. But in case of the longitudinal samples, no such segregation is there across the direction of the applied force, thus an average effect of the coarse and fine grain regions play role in the breakage. This therefore may be the reason for a slightly higher average elongation value in the longitudinal direction.

A check on the hardness values as related to the tensile strength is also within normal expectations. Converting the lowest Vicker’s hardness value to Brinell hardness and using the correlation for steel [224];
<table>
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<td>555.42</td>
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Table 5.2: Tensile test results

Figure 5.12: Stress-Strain plot of DWIM tensile test for both Longitudinal and Perpendicular specimens
Tensile strength $= 3.24 \times \text{HB} = 3.24 \times 180 = 583.2 \text{ MPa}$

This value is about 3.6% in excess from the minimum tensile strength of the DWM samples. Therefore it can be said that the hardness results are in conformance with the tensile strength. It may be noted that the lowest value of hardness is used here. This is because these tensile samples comprise of regions with different hardness values and yield will start in the region with the least hardness.

Figure 5.13(a) and (b) shows a representative stress-strain plot for the tensile test specimens of DWM and DWIM respectively. Both type of samples show similar results except that a sharp yield point (Lüders extension) is prominent in all of the DWIM samples (figure 5.13(b)). This however was not the case for DWM samples, as obvious from figure 5.13(a). In figure 5.13(b) the stress rises with insignificant plastic deformation to the upper yield point. At this point, the material begins to yield, with a simultaneous drop in the flow-stress to the lower yield point. After this occurs, there is little difference between the appearance of the stress-strain curves of figures 5.13 (a) and (b). In general, the grain size has a marked effect on the phenomena associated with the yield point [225]. The Lüders extension also increases as the grain size becomes finer. Since the tensile test samples were cut out from the central deposited layers, the DWM samples constitute of both reaustenitised and un-reaustenitised microstructure while the DWIM samples compose of only the reaustenitised microstructure. Therefore a higher population of RF grains in DWIM results in the formation of Lüders band. Furthermore the un-reaustenitised microstructure which is composed of mainly acicular structure with some martensite as well has therefore some population of body-centered tetragonal (bct) crystals. However, the reaustenitised region is mainly composed of polygonal ferrite having body-centered cubic (bcc) crystal structure. The bcc structure has a very low dislocation population relative to that of bct [226]. For a sharp yield point the initial density of mobile dislocations must be small [227]. Also strong yield points are observed in bcc crystals because of dislocations which are firmly locked by interstitial atom impurities such as carbon, oxygen and nitrogen [227]. Dislocation locking is also more effective at fine grain size. Finally it may be concluded that although there is difference in the microstructure and hardness of parts produced by the two deposition techniques but the effect on the tensile properties is limited to the formation of Lüders band.
Figure 5.13(a): Stress-Strain plot for DWM tensile test specimen

Figure 5.13(b): Stress-Strain plot for DWIM tensile test specimen
5.6 Conclusions

Reaustenitised and un-reaustenitised regions were found in the entire body of DWM while these are confined to the top layer of DWIM changing alternatively across the weld direction with intervals equal to the inter-bead spacing. The central layers of the DWIM deposit comprise only of reaustenitised region varying alternatively in grain size in both longitudinal and perpendicular direction. This sequential variation is in accordance with the inter-bead spacing in the across direction, and with the layer thickness in the perpendicular direction. The hardness results are in good agreement with the variation of the microstructure both for DWIM and DWM. The hardness values are higher at the top and interface layer while it is comparatively less in the central layers of DWIM samples. However, in DWM samples the hardness values are relatively higher in the top layer only. An RP part built by alternate welding and machining (removing the top and bottom layers) will result in relatively homogenous microstructure and uniform hardness. The tensile test results show no variation in the yield strengths of samples produced longitudinal and perpendicular to the deposition direction; however there is a slight difference in elongation. Moreover a sharp yield point was observed in the DWIM samples in contrast to the DWM samples.
Chapter 6

The Mechanical effects of different deposition parameters

This chapter presents a finite element (FE) based 3D analysis to study the thermal and structural effects of different deposition parameters and deposition patterns in welding based layered manufacturing (LM). Welding is one of those processes where high heat input results in large thermal gradients; these thermal gradients along with the mechanical constraints cause the build up of residual stresses. Residual stress induced warping is a major concern in a variety of LM processes, particularly those seeking to build parts directly without post processing steps. In order to reduce the residual stresses and deformation, the first step is to have a good knowledge and control of the thermal cycle associated with the deposition process. A commercial finite element software ANSYS is coupled with a user programmed subroutine to implement the welding parameters like Goldak double ellipsoidal heat source, material addition, temperature dependent material properties. Simulations carried out with various deposition sequences revealed that the thermal and structural effects, on the work piece, are different for different patterns. The sequence starting from outside and ending at the center is identified as the one which produces minimum warpage. The effects of interpass cooling duration were studied and it was found that an intermediate value of interpass time is suitable for a nominal level of deformations and stresses. A similar finding was made from the studies about different weld bead starting temperatures. The studies regarding different boundary conditions revealed that the deformations are least for adiabatic case while isothermal case produced the maximum deformations. The results presented are for deposition by gas metal arc welding but can be applied to other deposition process employing moving heat source. The results regarding deposition patterns, presented in this chapter, have been published

6.1 INTRODUCTION

Rapid prototyping (RP) is the process of building prototypes to establish the success of a proposed design [229]. Verification of a "successful" design has several aspects, including correct shape, correct dimensions and adequate strength etc. In RP, also known as "layered manufacturing" (LM) or "solid freeform fabrication" (SFF), artifacts are built in two-dimensional layers; which simplifies the manufacturing process and automation is made possible. Complex shapes are easy to build when broken down into simple 2D layers and there is no need of part specific tooling or fixtures. Rapid prototyping (RP) is widely accepted as a cost effective tool [230] for reducing marketing time of new products. Conventional RP techniques do not have the capability of directly producing fully functional parts. The present and future emphasis is to produce “form-fit-functional” parts rather than prototypes for visualization [231] thus ultimately leading towards the concept of rapid manufacturing [232]. Manufacturing of metallic parts with good accuracy and mechanical properties is one of the main aims of solid freeform fabrication (SFF). A number of techniques are under development like selective laser sintering, laser engineered net shaping (LENS) and shape deposition manufacturing (SDM) [113,233-235,113]. One such technique uses the capability of gas metal arc welding (GMAW) for metal deposition; with the promise of producing parts with high structural integrity and mechanical strength [236,115]. However, the big draw back of using welding as the deposition process is the large heat input to the substrate or to the previously deposited layers, thus causing high temperature gradients and resulting in deformations like bowing, shrinkage, angular distortions and warping [237]. Moreover, residual stresses also accompany these deformations which itself is a matter of concern. These stresses arise from the contraction of deposited material as it cools down, thus resulting in distortions and therefore out of tolerance parts. The buildup of these stresses may culminate into delamination of layers or part failure. In order to predict and minimize these problems, knowledge of thermal gradients and temperature history during manufacture is important. Previously it was identified that the thermal properties of deposited and substrate
materials, substrate thickness and the thermal boundary conditions does have effects on the thermal as well as structural behaviors of the deposit and substrate [141,142].

The patterns used to deposit a layer of material, significantly affect the resulting deformations, residual stresses, part geometry and mechanical properties. To study these effects of deposition patterns, a few efforts have been made in the past. Fessler *et al.* [101] reported that continuous bidirectional raster deposition of stainless steel and Invar produced significantly more deformations as compared to a tower technique (depositing in alternate lines then filling in between). Similar technique also termed as the spiral overlay welding was employed by Spencer *et al.* [117] at the University of Nottingham, UK, in which multiple beads were used to enable the production of parts wider than normally possible from a single bead. An improved imaging strategy was proposed by Zhang *et al.* [231]; first depositing the outline of the layer geometry and then filling the interior with a raster fill pattern thus resulting in lesser discretization errors apparent at the edges. In order to eliminate accumulated errors due to multiple layers, a change in the orientation of each layer was also suggested. However the alternating deposition direction between the layers influences the tensile strength significantly [212]. Grenestedt [238] has proposed a simple approach to find the weld path that minimizes the residual distortions for overlay coatings. This technique can also prove useful in layered manufacturing processes. Nickel *et al.* [137], developed a thermal model based on laser heating to study the effects of different deposition patterns on the subsequent deformations and stresses. It was reported that the spiral pattern scanned from outside to the inside produces low and uniform deflections. However a similar study of the patterns for weld based deposition [135,136] revealed that compared to the raster pattern, both the inside-out and outside-in deposition patterns result in substantially greater warping.

In RP finite element (FE) modeling is being used as a handy tool to investigate the effects of different parameters controlling the process. Starting from deposition at a droplet level [141,142,239], to full scale deposition 3D modeling had been employed to study different features related to RP [137,143,144,146,240]. A three-dimensional FE model [146] has been used to investigate the effects of different process parameters on the deformations and stress distribution within the substrate in welding based deposition of metal parts. Goldak double ellipsoidal moving heat source [157], temperature dependent material
properties and time dependent material addition were used in the development of this model. The model is limited to deposition of a single layer. This should be sufficient to investigate and study the comparison of the effects caused by different deposition parameters. For the deposition of multiple layers, as is the case in RP, the effects (e.g. deformations) are expected to add up. Although every deposited layer increases the stiffness of the plate, therefore these effects may not be same in magnitude for each subsequent layer. Also the shape of the deposited layer (i.e. rectangular) considered in this study, is a simple shape, which may not be common in RP. However the results of the current study may provide a good guide line for the general RP parts.

6.2 Results and Discussion

A complete 3D model of weld based deposition was developed by Mughal et al. [146] taking into account a 3D heat source, material addition and temperature dependent material properties. The model is limited to the buildup of a single layer of weld metal on a bolted substrate. The fusion zone, temperature history and deformations, predicted by the numerical model, show a good agreement with experimental data. The model can accurately predict the remelting depth. The substrate deflection for the bolted case is in excellent conformity with experimental data, while for unbolted case a qualitative agreement is found, which may be attributed to a simplified bolt model. It may be concluded that the finite element model can be used for reliable prediction and optimization of process parameters.

The commercial FE software ANSYS was used to develop the model [146] which, in this study, is used to investigate the effects of different process parameters on deformation and residual stresses of the substrate plate. A sequentially coupled, thermo-mechanical FE model was developed, in which the transient temperature distribution from thermal analysis was applied as body load to the structural model. The basic geometry of the model comprises of a rectangular 160×110×10 mm substrate plate, bolted at the four corners to a 200x130x34 mm support plate. The deposition in the original model [146] consists of nine successively deposited rows, with the length of each row equal to 100mm, and in between rows the model was allowed to cool down for 180 seconds (i.e. the interpass cooling time). To reduce the calculation time and storage required it was reduced
from nine to seven weld passes and from 100 mm to 80 mm weld pass length in the present study. Thus an 80×40×3 mm weld layer was deposited on the center of the substrate plate. The present study includes the comparison the effects of different interpass cooling durations, constant weld bead starting temperatures, thermal boundary conditions and deposition patterns. For the study of deposition patterns thermal analysis was performed for seven different raster deposition patterns, while for structural analysis, four of these patterns were short listed. The seven patterns are shown in figure 6.1; where the number shows the order of deposition and arrows show the direction of welding.

6.2.1 Thermal analysis of deposition

A full three-dimensional finite element thermal model with time dependent material addition was formulated. Earlier work sited in the literature, had been simplified by assuming the problem as symmetric, lumping the material deposition, or entirely ignoring the material addition thus only heating sequentially the substrate plate [135-137]. However as it is a transient case, it is neither geometrically nor thermally symmetric. Moreover making such simplifications regarding material deposition may not be realistic.

Figure 6.2 represents a schematic of the substrate plate with different sections marked where the simulation results are presented, and figure 6.3 shows the central cross-section (AB) with the seven deposited weld beads. The temperature history of the center points on the first, fourth and seventh weld pass (points F, I and L in figure 6.2) for the raster pattern a along center line AB on substrate top is shown in figure 6.4. The graph is plotted up to 600 sec, after which all the curves follow similar profiles to the room temperature. Peaks 1, 4 and 7 are the primary peaks caused due to the direct interaction of the welding heat source at points F, I and L respectively; while peaks 2, 3, 5 and 6 are secondary peaks caused when the heat source is positioned on the neighboring weld bead of the respective point (F, I and L). It can be seen that the maximum temperature reached (peak 1, 4 and 7) at each point (i.e. points F, I and L) is nearly same but the time at which the peak occurs is dissimilar. The reason being very obvious; as the heat source is in close proximity to any of the point under consideration, the temperature of that point will be higher. This supports the idea that the thermal cycle is not symmetric with respect to the geometry of the substrate, and is therefore dependant on the relative position of the heat source.
Figure 6.1: Raster deposition patterns

Figure 6.2: Schematic of substrate plate and weld deposition.
Figure 6.3: Cross-section of substrate along center line AB.

Figure 6.4: Time history temperature plot for pattern a at points F, I and L.

Figure 6.5: Directions of heat flow after deposition of (a) 3rd pass, (b) 5th pass
Considering the temperature history plot of point-\( I \), it can be seen that the peak temperature reached at this point due to the 3\(^{rd} \) weld pass (peak-3 i.e. approx. 1700K) is higher than the peak temperature due to 5\(^{th} \) pass (peak-5 i.e. approx. 1425K); although the two weld beads are in geometrical symmetry with respect to this point. This is because there was no weld material at point ‘\( I \)’ when the 3\(^{rd} \) weld bead was deposited while it was there when the 5\(^{th} \) weld bead was deposited (as shown in figure 6.5). This difference is the effect of transient material addition which results in a decreased thermal conductive resistance due to increased heat flow area. However this effect of increased heat flow area (i.e. the difference 1700K – 1425K) seems quite large. To confirm this finding, a simulation without any material deposition (i.e. only sequential heating) for pattern \( a \) was performed in which no such temperature drop occurs at point ‘\( I \)’ due to the deposition of 3\(^{rd} \) and 5\(^{th} \) weld pass (peak-3 & peak-5); as illustrated in figure 6.6. On the contrary, the temperature of peak-5 is slightly higher then peak-3, which is the result of added preheating. The question which arises here is that, why then the maximum temperature values for points \( I \) and \( L \) (peaks 4 and 7, figure 6.4) are not lesser in magnitude then that of point \( F \) (peak-1), for the same reason (i.e. increased heat flow area). A similar result for these peaks can be seen in figure 6.6, which therefore reveals that there is no substantial effect of material deposition on these primary peaks. These peaks (1,4 and 7) are actually caused when the welding heat source reaches directly on these points (\( F \), \( I \) and \( L \)), thus resulting in the high temperature required for melting. Another thing which raises question in figure 6.4 is why the primary peak temperatures for points \( F \), \( I \) and \( L \) aren’t different because of different preheating at the three points due to previous beads? To investigate the effect of preheating on these peaks, simulations with different initial preheating were performed. The results of these simulations for point \( F \) are presented in table-6.1and in figure 6.7. It can be seen that there is little effect of preheating on the primary peaks while the influence is sizeable on the secondary peaks. Hence this explains that preheating is the cause of the difference of temperature in peak 2 from 5 and similarly peak 3 from 6 (in figure 6.4). Whereas the temperature difference in peaks 2 and 5 from peaks 3 and 6 is mainly due to the increased heat flow area caused by material deposition. Hence finally it can be concluded that there is little effect of increased heat flow area and preheating on
Figure 6.6: Time history temperature plot for pattern $a$ at points $F$, $I$ and $L$ without any material deposition (only sequential heating).

Figure 6.7: Primary and Secondary peak temperatures reached at point $F$ for different preheating temperatures of substrate.
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</tr>
<tr>
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<td>1402</td>
</tr>
<tr>
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<td>2265</td>
<td>1467</td>
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<td>2280</td>
<td>1575</td>
</tr>
<tr>
<td>900</td>
<td>2304</td>
<td>1734</td>
</tr>
<tr>
<td>Difference (Max - Min)</td>
<td>600</td>
<td>56</td>
</tr>
</tbody>
</table>

Table-6.1. Primary and Secondary peak temperatures reached at point $F$ for different preheating temperatures of substrate.

<table>
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<tr>
<th>Plate Thickness, mm</th>
<th>Temperature of Primary Peak-4, K</th>
<th>Temperature of Secondary Peak-3, K</th>
<th>Temperature of Secondary Peak-5, K</th>
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</tr>
<tr>
<td>25</td>
<td>2234</td>
<td>1628</td>
<td>1331</td>
</tr>
</tbody>
</table>

Table-6.2. Primary and Secondary peak temperatures reached at point $I$ for different substrate thickness.
the primary peaks while on the contrary these two factors have a considerable effect on the secondary peaks.

These primary and the secondary peaks are also responsible for the microstructure of the deposited layer, produced finally. The primary peak causes an un-reaustenitised microstructure while the secondary peak causes reaustenitised microstructure. Figure 6.4 also show a third level of peaks i.e. the tertiary peaks. These too are responsible for the reaustenitised microstructure; however the difference between the secondary and tertiary is in the formation of re-crystallized coarse and fine grain regions.

The effect of increased heat flow area is further elaborated in figure 6.8(a) which shows the temperature profiles across the weld direction (along AB) at the end of each weld pass for pattern a. It can be noticed that the peak temperature after first weld pass is higher than that for the remaining passes. This peak value however increases slightly after every successive weld pass following the second pass. The reason for the drop in peak temperature after the first weld pass is an increased conduction path, available as a result of the material deposited in the first weld pass (as shown for 3rd and 5th pass in figure 6.5). After deposition of the first pass, the increased conduction area for the subsequent deposition remains the same. The cumulative preheating after each weld pass therefore causes a gradual increase in peak temperature value of the successive weld passes following the second pass. The results of simulation without any material deposition (i.e. only sequential heating) for pattern a confirm this finding, whereby no such temperature drop after the first weld pass existed; as illustrated in figure 6.8(b). This is therefore quite evident that the effect discussed above for the base case is due to an increased conduction path provided by previously deposited material on the side. Moreover the rise in the peaks after each weld pass in figure 6.8(b) is slightly more as compared to that in figure 6.8(a). It can therefore be inferred that this rise is due to larger thermal resistance as compared to the base case and thus resulting in more preheating.

Further more simulations with different substrate plate thickness were performed to investigate their thermal effects. Table-6.2 presents the results for the primary and secondary peak temperatures reached at point-I (Note that the peaks-3,4 and 5 corresponds to the one’s shown in figure 6.4). It can be seen that there is no substantial difference in
Figure 6.8(a): Temperature profile along $AB$ on substrate top at the end of each weld pass for pattern $a$ with material deposition.

Figure 6.8(b): Temperature profile along $AB$ on substrate top at the end of each weld pass for pattern $a$ without material deposition.
the primary peak temperatures. The reason for which is the same as explained above that the primary peak temperature is produced by the welding heat source. On the other hand, the secondary peak temperatures fall with the increasing thickness of the substrate. This is because of the increased heat sink available due to the increased thickness.

### 6.2.2 Interpass Cooling Time

The deposition in the initially developed and verified model [146] consisted of nine successively deposited rows and in between rows the model was allowed to cool down for 180 seconds (i.e. the interpass cooling time). The interpass time regulates the temperature of the substrate and therefore can control the preheating prior to the commencement of the next weld bead. Consequently, for a zero inter-pass time the subsequently deposited weld bead experiences a high preheat. On the contrary a longer interpass time results in lower preheating at the expense of increased production time and computational resources for simulation. Previously it was found that the preheating temperature reduces by about 41% if interpass time is increased from 0 sec to 60 sec while it reduces only by 7% (approximately) if interpass time is increased from 60 sec to 300 sec [241]. Also it has been identified previously in a 2D analysis that the deformations become independent of interpass time beyond 120 sec [141].

The effects of interpass cooling time were studied for five different cooling time values (i.e. 0, 16, 30, 60 and 120 seconds) during successive deposition (pattern a figure 6.1). The deformation plot, in bolted condition, along section AB is shown in figure 6.9 for the five interpass cooling times. A U-shaped (bell shaped) deformation profile is observed for the cases having more than 30 sec interpass time, whereas a W-shaped deformation profile is observed for 0 and 16 sec interpass time cases. In other words it can be seen that in general the deformation is minimum at points A and B for lower interpass time values while higher interpass values produce higher deformation values at these points. Whereas at the central section a reverse trend is observed, i.e. the 0 and 16 sec cases produce maximum deformations near point I. The same thing is further elaborated in figure 6.10. The change in deformation on points A, B and I is more prominent from 0 to 30 sec as compared to the change from 30 to 120 sec. Therefore it can be stated that there is lesser effect of interpass cooling time on the deformation above 30 sec interpass time.
Figure 6.9: Deformation plot along section AB for the five interpass time values.

Figure 6.10: Change in deformation for different interpass time values on points A, B and I.
Comparing the three higher value cases (i.e. 30, 60 & 120 sec) it can be seen that at points A and I the deformation is the least for the 60 sec case while at point B it is the 30 sec case. Furthermore it can be seen (figure 6.10) that for lower interpass time cases the deformation on point A is more as compared to that on point B, while this effect is reversed for longer interpass values. Although for the 30 sec case the deformation on the two points is nearly same. This therefore points towards a symmetric deformation result for the 30 sec case; but this result is only based on the end points (i.e. A & B), the slope on side B (figure 6.9) is quite fluctuating. The slopes of 60 sec case are much smoother on both the sides. Another thing to be noted here is that the effect of interpass time is more prominent on the slopes at central section and side B rather than on side A. This can be attributed to the fact that the first deposited bead in all the cases has undergone a similar starting procedure. For the lower interpass time (i.e. 0 sec), however, the affects of continuous deposition and thus resulting high temperatures dominates and thus influence the deformation on side A as well.

Figures 6.11 (a) & (b) further elaborates the above discussion. Once again it can be seen that in the longitudinal direction the curvature of the deformation plot is less for lower values of interpass cooling time. Minimum deformation results are being exhibited by the 0 sec interpass time results.

The axial stresses along section AB for the five interpass cooling time cases in bolted condition are presented in figure 6.12. The stresses for all the cases are compressive on the edges A and B while these are mostly tensile in the deposition area (with some exceptions for 0 and 16 sec). The compressive stresses on side A reduces with the increasing interpass time; however it is nearly the same for 60 and 120 sec cases. On the other hand, on side B there is not much difference in the compressive stresses for these interpass cases. This means that the interpass time has effect on the residual stresses at the deposition start side (side A) while it has little effect on the residual stresses at deposition end side (side B). A reverse picture can be seen in the deformation plots of figure 6.9. The deformations, for the five interpass time cases, on the deposition end side are more far apart as compared to the start side. A fluctuation in the plot for 0 and 16 sec interpass cases can be observed between the tensile and compressive stress region within the deposition area. This is the same region in which comparatively higher values of
Figure 6.11 (a): Deformation plot along section \( MN \) for the five interpass time values.

Figure 6.11 (b): Deformation plot along section \( OP \) for the five interpass time values.
Figure 6.12: Comparison of axial residual stresses along section AB for the five interpass time values.

Figure 6.13: Deformation plot along section AB for the five base temperature levels.
deformations were reported for these cases (figure 6.9), i.e. a W-shaped deformation profile. On the contrary the remaining interpass time cases remain in the tensile stress range within the deposition area and hence follow a U-shaped deformation profile. Hence it can be said that lower interpass cooling time values results in more compressive axial residual stresses on the deposition start side, and a dip from tensile into compressive zone in the deposition area.

Alongside the effects of preheating caused by the preceding weld beads, the succeeding weld beads stress relieve the substrate and the previously deposited material. This effect (stress relieving) however is dominated by the high preheating available in 0 to 30 sec cases. The affect on the other hand is visible in the 60 and 120 sec cases. It can be seen in the deposition area, marked in figure 6.12, that the stress profile in the first half (on side $A$) is lower than the second half (on side $B$); and this effect is more prominent in the 120 sec case as compared with the 60 sec case.

It can be concluded that a continuous deposition gives comparatively lower deformations at the edges and higher deformations at the center, however fluctuating residual stresses are produce in the substrate. In general as the interpass cooling increases so do the substrate deflection at the edges while it decreases at the center. For high interpass delay duration (120 sec), the substrate preheats available for the next row decreases, which results in larger temperature gradients across the substrate thickness at the time of next weld deposition and hence the deformation at the edges increases. On the other hand continuous deposition causes excessive preheating which results in high temperature areas and larger remelting of the substrate or of the underlying layers. In addition the area of interest in RP is the deposition area; 0 and 16 sec cases give the most deformations in this area (figure 6.9). The stresses are also fluctuating in between tensile and compressive range in this area. The surface finish of the build up is also affected as reported by Dickens et al. [115]. Hence 60 sec interpass case seems to be the best possible option amongst the studied cases with optimal deformation and residual stress results.

### 6.2.3 Constant Weld Bead Starting Temperature

A 3D welding process involving a simple temperature control technique, implemented by Dickens et al. [115] at the University of Nottingham, was reported to result in an
improved surface finish of the deposit. The same idea of temperature control was incorporated in the developed model to check the effects on the residual stresses and distortions. Simple raster pattern (pattern a figure 6.1) with weld beads deposited from time to time, side by side, was used to study this effect. Initially the substrate plate was preheated to a certain base control temperature and after each deposited weld bead it was allowed to cool to this base temperature before depositing the next bead. However, as different sections of the substrate plate are at different temperature levels after the deposition of each weld bead, the starting point of the next bead was taken as reference to compare temperature level with the control temperature for commencing the next weld bead.

Five different cases with base control temperatures 375, 450, 500, 650 and 800 K were studied. The deformation profiles along section AB, in substrate bolted condition, for the five temperature levels are presented in figure 6.13. It can be seen that the deformation reduces with the increasing base control temperature. This is further elaborated in figure 6.14, where the deformation at points A and B is plotted against the base temperature values. It can be noted here that other than the 800 K case the change in deformation is more prominent on side B, (figure 6.13 & 6.14) as compared to the side A. The profile of 650 K case on side B as compared to the cases with lower control temperatures is also a bit different and a beginning of reverse deformation at the central section is visible. This change is further augmented as the control temperature value is further increased (i.e. 800 K case); i.e. deformation at the central section is fully induced. In other words the deformation profile has changed from a bell shape to W-shape.

The axial stresses along section AB for the five control temperature cases in bolted condition are presented in figure 6.15. The stresses for all the cases are compressive on the edges A and B while these are mostly tensile in the deposition area (with some exception for 800 K). The compressive stresses on side A reduces with the increasing control temperature; while the converse is true on side B. However other than the 800 K case, the compressive stresses at the edges for the remaining cases are not much different. A fluctuation in the plot for 800 K case can be observed between the tensile and compressive stress region within the deposition area. This is the same region in which comparatively higher values of deformations were reported for these cases (figure 6.13), i.e. a W-shaped
Figure 6.14: Change in deformation for different preheating base temperatures on points A and B.

Figure 6.15: Comparison of axial residual stresses along section AB for the five base temperature levels.
deformation profile (this is very similar to the small interpass time cases i.e. 0 and 16 sec). On the contrary the remaining cases remain in the tensile stress range within the deposition area and hence follow a U-shaped deformation profile. Hence it can be said that a higher control temperature value results in more compressive axial residual stresses on the deposition start side, lower at the end side and a dip from tensile into compressive zone in the deposition area.

The effect of stress relieving by the weld beads on the substrate and previously deposited material is also visible from the figure 6.15; i.e. the stresses in the initial half (on side A) of the deposited area along AB are lower in magnitude than the stresses in the later half (on side B). This effect tends to reduce with the increasing control temperature. For higher control temperatures more heat is accumulated in the substrate and lesser time is available for complete thermal cycling for the stress relieving. Therefore the 375 K case present the maximum stress relieving amongst the discussed cases. Among all these cases the deformation plot for 450 K case produce the most symmetric profile along the section AB; but it gives high values of deformation and stresses.

Figures 6.16 (a) and (b) also present a similar picture of comparatively lower deformations for higher control temperature values. In addition the profile for the 650 and 800 K control temperature cases is much flat along MN and OP as compared to the other plotted curves; with 800 K case having the flattest profile and least deformation level. However the area of interest in RP is the deposition area; and 800 K case gives the most deformations in this area (figure 6.13). The stresses are also fluctuating in between tensile and compressive range in this area. Hence 650 K control temperature case seems to be the best possible option amongst the studied cases with optimal deformation and residual stress results.

Another thing which should be considered here is the deposition time. As the cooling time in between each deposited bead is increased so does the overall deposition time of the entire job. It was found that the 650 K case requires about 65% lesser time to deposit the same layer as compared to the 375 K case, thus pointing towards improved process efficiency. However as reported by Dickens et al. [141]; a higher control temperature results in a reduced weld surface quality. Further studies are required that will allow part
Figure 6.16 (a): Deformation plot along section $MN$ for the five base temperature levels.

Figure 6.16 (b): Deformation plot along section $OP$ for the five base temperature levels.
quality to be maintained without incurring excessive time penalties or adversely affecting positive process features.

6.2.4 Effect of thermal boundary conditions

When molten metal is deposited, the substrate at first acts as a large heat sink and for the initial part of the process the heat conduction is dominantly one dimensional into the thickness of the substrate. However, after this initial period the heat conduction becomes three dimensional. The boundary conditions alongside the substrate contribute towards this heat flow. In order to study the effect of heat sink, three different thermal boundary conditions were applied at the substrate base, simulating different heat sink capacities (i.e. isothermal, convective & adiabatic). Isothermal boundary conditions reflect an ideal heat sink while adiabatic boundary conditions exhibit an insulator. These extreme boundary conditions can provide a range of all the expected cooling rates and solidification times. In addition these limiting conditions were compared with the actual convective boundary conditions as described earlier. All these cases were simulated for a 0 sec as well as for 60 sec interpass cooling time.

As shown in figure 6.17 (for 60 sec interpass) adiabatic and convective bases give almost similar deformation results (adiabatic deformation slightly lesser on side A) along section AB. These are also quite lesser in magnitude than the isothermal base deformation result. Isothermal boundary conditions give higher thermal gradients; therefore the deformations are higher. An insulated base gives more uniform temperature distribution throughout the substrate resulting in an overall lower deformation. The convective case also has a similar behavior except a very slight difference at the start side. Figure 6.18(a) and (b) present a similar picture of deformation along MN and OP. The differences in the curvature of the profile along MN and OP for all these cases are also in accordance with the results in figure 6.17.

Figure 6.19 presents the plot of axial residual stresses along section AB for all these cases with a 60 sec interpass cooling time. As was the case for deformation, the stresses for the adiabatic and convective cases are quite alike. The stresses for isothermal case however, in contrast to the other cases completely lie in the tensile zone. The maximum tensile stresses are also higher in magnitude as compared to the other cases. Generally the profile
Figure 6.17: Deformation plot along section AB showing the effect of heat sink characteristics for 60 sec interpass cooling time.

Figure 6.18(a): Deformation plot along section MN showing the effect of heat sink characteristics for 60 sec interpass cooling time.
Figure 6.18(b): Deformation plot along section OP showing the effect of heat sink characteristics for 60 sec interpass cooling time.

Figure 6.19: Axial residual stresses along section AB showing the effect of heat sink characteristics for 60 sec interpass cooling time.
of the isothermal plot is similar to the other cases (although different in magnitude) except that on the side A. This difference on side A can be attributed to the higher thermal gradients that the isothermal case gives in the beginning. As the deposition carries on the substrate preheating reduces these thermal gradients. The effect of stress relieving by the weld beads on the substrate and previously deposited material is also visible within the deposition area; i.e. the stresses in the initial half (on side A) of the deposited area along AB are lower in magnitude than the stresses in the later half (on side B). This stress relieving is more prominent for the isothermal case as compared to the others. For the isothermal case the heat accumulation in the substrate is lesser as compared with the other cases which results in extra thermal cycling and therefore additional stress relieving.

Figure 6.20 presents the plot of deformations along section AB for all these cases with a 0 sec interpass cooling time. It can be seen that the adiabatic and convective cases results in a similar W-shaped plot, which is in accordance with the 0 sec interpass cooling time plot discussed in section 6.2.3. However the isothermal case produces a U-shaped plot which is contrary to a low (0 sec) interpass time. The reason to this once again is the speedy heat removal (higher thermal gradients) and therefore lesser heat accumulation in the substrate. High heat accumulation in the substrate causes more shrinkage forces at the deposition area thus resulting in an upward deformation in the central area of deposit and therefore a W-shaped plot.

Figure 6.21 presents the plot of axial residual stresses along section AB for a 0 sec interpass cooling time. As was the situation in the above plots, the stresses for the adiabatic and convective cases are quite alike. The stresses for isothermal case here again completely lie in the tensile zone. The fluctuation (as was observed in figures 6.12 and 6.15) between the tensile and compressive stress region within the deposition area can be observed in the plot for adiabatic and convective cases. This once again corresponds to the W-shaped deformation profile shown in figure 6.20.

The adiabatic and convective cases with 60 sec interpass seems to be the preferable choice amongst the discussed cases as these give the least deformations and comparatively lesser residual stresses.
Figure 6.20: Deformation plot along section AB showing the effect of heat sink characteristics for 0 sec interpass cooling time.

Figure 6.21: Axial residual stresses along section AB showing the effect of heat sink characteristics for 0 sec interpass cooling time.
6.2.5 Deposition patterns

6.2.5.1 Thermal effects of deposition patterns

In Layered manufacturing for raster deposition patterns, the deformations alongside the deposition are more dominant in direction parallel to the deposition direction as compared to across the deposition direction; whereby the maximum deformation occurring at the central section along the deposition direction [135-137]. Moreover, from the results of the previous study [146] maximum deformation occurs at the edges of the central cross section (i.e. at points A and B). Considering this factor, the material was deposited along Z-axis of the rectangular substrate plate and the study focused on results along lines MN, OP and AB (figure 6.2). In conjunction to the findings of the above simulation results (i.e 60 sec interpass time and convective boundary conditions being preferable) the effects of deposition patterns were studied. The patterns selected in this study are all raster with changes in the sequence of deposition as shown in figure 6.1. The spiral patterns were not considered because of complexity involved in the programming and the limitations of the computational resources especially in the 3D structural analysis. However this may be a feature to be studied in future.

Figures 6.22 (a-g) illustrate the temperature profiles along section AB, 60 seconds after the deposition of each weld bead for the respective seven deposition patterns. In addition figure 6.23 shows combined temperature profiles along section AB, 60 seconds after the seventh bead, for all the patterns. It can be seen from figure 6.22(a), 6.22(b) and 6.23 that for patterns a and b, a very similar profile is formed. Also the temperature history plot for bidirectional pattern b was very similar to the one obtained for base pattern a as shown in figure 6.4. Moreover, temperature history plots at points A and B shown in figure 6.24(a) are similar for both these patterns. However, there were some differences on locations away from the central section AB (i.e. at section XX and YY), which are of comparatively lesser significance. It can therefore be concluded that similar deposition sequences with different deposition directions (e.g. pattern a & b) does not cause a marked difference in the thermal profiles and temperature history of the substrate. The structural effects, however, are discussed in the next section which also upholds trends similar to that of thermal. As a result of the above discussion, only one pattern with bidirectional deposition
Figure 6.22(a): Temperature profiles along section $AB$, 60 sec after each weld pass (pattern $a$).

Figure 6.22(b): Temperature profiles along section $AB$, 60 sec after each weld pass (pattern $b$).
Figure 6.22(c): Temperature profiles along section AB, 60 sec after each weld pass (pattern c).

Figure 6.22(d): Temperature profiles along section AB, 60 sec after each weld pass (pattern d).
Figure 6.22(e): Temperature profiles along section $AB$, 60 sec after each weld pass (pattern $e$).

Figure 6.22(f): Temperature profiles along section $AB$, 60 sec after each weld pass (pattern $f$).
Figure 6.22(g): Temperature profiles along section AB, 60 sec after each weld pass (pattern g).

Figure 6.23: Temperature profiles along section AB, 60 sec after 7th weld pass for all patterns.
is included in this study. From the structural analysis as discussed in the next sections and also from the results of previous work [146], it is clear for pattern \( a \), that the deformation of the side where last weld bead was deposited (side with point \( B \)) is more than the side where first weld bead was deposited (side with point \( A \)). This non-symmetric behavior is also obvious from the temperature profiles of figure 6.22(a) and 6.23. Therefore it can be said that a non-symmetric temperature profile corresponds to a non-symmetric deformation and vise versa. It can be noted from figures 6.22(a-g) that the buildup of heating pattern from first to seventh weld pass for pattern \( d \) and \( e \) are more sequential and symmetrically distributed than that of any other pattern. The symmetric behavior of these two patterns can also be observed from figure 6.23. Moreover (from figures 6.22 a-g) it can be seen that the temperature profile after the seventh bead for all the patterns, except \( d \), is lower than the profile after the sixth bead on either one of the sides \( A \) or \( B \). This therefore reveals a non-uniform heating of the substrate for these patterns; while a nearly symmetric profile for pattern \( d \) along the section \( AB \), 60 seconds after deposition of the last weld bead. Due to the similarities in temperature profiles these seven patterns can be grouped into four sets. i.e.

1. Patterns \( a \) and \( b \)
2. Pattern \( c \)
3. Patterns \( d \) and \( e \)
4. Patterns \( f \) and \( g \)

It can be seen that the patterns of each set have similar build up of temperature profiles however profiles of pattern \( c \) resembles initially with first set and at the end with the fourth set (as obvious from figures 6.22(a-g) & 6.23). Therefore, for structural analysis only four patterns (i.e. \( a,c,d, \) & \( g \)) out of these four sets were considered ample to understand the effects of patterns. Moreover almost similar profiles, as along section \( AB \), were observed along sections \( XX \) and \( YY \) on the substrate plate for all the deposition patterns respectively (results not shown here). Hence uniformly distributed deformations and residual stresses can be expected for pattern \( d \).
The above stated findings can also be observed from the time history temperature profiles at points A and B, as shown in figures 6.24(a-f). The maximum temperature reached for pattern a (and b), at point A (522.4 K), is the lowest amongst all the patterns shown. While for pattern g it is the highest (598.5 K). On the other hand, the highest temperature at point B is reached for pattern a (and b) (628.2 K), while the least is for pattern d (551.1 K). The maximum and minimum temperature results are tabulated in table-6.3. It may be deduced from these results that the patterns a, b and c, has asymmetric thermal effects across the geometry of the substrate while patterns d, e, f and g has symmetric effects. However if one looks closely at the buildup of these temperature histories it is clear that the temperature buildup on both the sides (points A and B) is much even for patterns d then for any other pattern. The overall maximum temperatures reached are the least for pattern d as compared to others except that at point A for patterns a and b. Furthermore the plots of patterns d and g show similar progress of temperature history at the two points (A and B). However the difference of temperature between the two points after each weld pass decreases gradually from first to the last weld pass for pattern d, whereas the converse is true for pattern g. It may therefore be deduced that all the patterns contribute asymmetric thermal effects across the geometry of the substrate except pattern d which contributes nearly symmetric effects. Hence it can be concluded that the lower temperatures and symmetric temperature distributions in case of pattern d will result in least and symmetric mechanical effects on the substrate and thus on the finally produced RP part. Therefore minimum and symmetric deformations can be expected of pattern d; and this will be further observed in the coming sections as the structural results will be discussed. It can be further stated that considering the problem symmetric (especially for patterns other then d) and solving an axi-symmetric FE model will not be able to identify the variation in the results on the two sides of the deposition (i.e. side A and B).

Nickel et.al. [137] reported that maximum stresses are observed along the section where the last line is deposited. This therefore will be the central region for pattern d. It may be noted that the deposition of each weld pass cause a re-heating (stress relieving) effect on the neighboring / previously deposited material and this effect has been presented by Mughal et.al. [146]. This re-heating will also be symmetric for pattern d as compared to the other patterns. Therefore from the above discussion of sequential and fairly symmetric
Figure 6.24(a): Transient temperature history plots at points A and B for pattern a and b.

Figure 6.24(b): Transient temperature history plots at points A and B for pattern c.
Figure 6.24(c): Transient temperature history plots at points A and B for pattern d.

Figure 6.24(d): Transient temperature history plots at points A and B for pattern e.
Figure 6.24(e): Transient temperature history plots at points A and B for pattern f.

Figure 6.24(f): Transient temperature history plots at points A and B for pattern g.
Table-6.3. Maximum temperatures reached at points $A$ and $B$ for various patterns.

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<tr>
<td>$b$</td>
<td>523.4</td>
</tr>
<tr>
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<tr>
<td>$f$</td>
<td>568.2</td>
</tr>
<tr>
<td>$g$</td>
<td>598.5</td>
</tr>
</tbody>
</table>

Figure 6.25: Time history temperature plots at points $C$ and $D$ on substrate for pattern $a$.

Table-6.4. Deformation values at points $A$ and $B$ for the four patterns.

| Deposition Pattern | Deformation at point | Difference $|A-B|_i$ |
|--------------------|----------------------|----------|
|                    | $A$ (mm) | $B$ (mm) | (mm)   |
| $a$                | 0.251   | 0.3      | 0.049  |
| $c$                | 0.318   | 0.277    | 0.041  |
| $d$                | 0.282   | 0.262    | 0.02   |
| $g$                | 0.249   | 0.284    | 0.035  |
heating (figures 6.22 and 6.24), it can be anticipated that relatively more uniformly
distributed deformations and residual stresses can be expected for pattern \( d \) as compared
with the other patterns, and this will be further discussed in the next sections.

Figure 6.25 represents the time history temperature profiles at points C and D on the
substrate top for the pattern \( a \). An almost similar behavior is observed for the remaining
patterns, studied (results not shown here). This therefore imply that the deposition patterns
have a marked effect on the temperature distributions in the direction across to the
deposition direction, while there is very little effect on the edges with sides \( C \) and \( D \), and
this is consistent with earlier related work [135-137].

6.2.5.2 Effects of deposition patterns on deformation

Figure 6.26 presents the profile of deformation of the four short listed deposition patterns
(i.e. \( a, c, d, \) & \( g \)) on the mid section of the plate \((AB)\) in unbolted condition. It can be seen
that patterns \( a \) and \( g \) results in minimum deformation values on one side of section \( CD \), of
the plate, while pattern \( d \) results in minimum deformation on the other side. The same is
also obvious from figures 6.27 (a) and (b), where deformation profiles at the two edges of
the plate i.e. along lines \( MN \) and \( OP \) respectively are shown. (Note: These graphs are
showing the central portion which is the area beyond the bolts i.e. from 30mm to 130mm).
The difference in deformation of the two extreme sides (i.e. \( A \) and \( B \)) is the least for \( d \)
among the four patterns. This can be seen in table-6.4 as well.

Pattern \( d \) and \( g \) thus show overall minimum deformation values, with one having lower
values on one side and the other on the opposite side. However the change in deformation
along the line \( AB \) is more abrupt for all the patterns as compared to pattern \( d \) (i.e. the U
shaped curve of figure 6.26 is spread more widely for pattern \( d \) then any other pattern).
This change is the maximum for pattern \( g \). Hence it can be said that pattern \( d \) gives
minimum warpage among the four deposition patterns studied. In addition from the plot of
figure 6.26 and the values in table-6.4, it can be seen that pattern \( d \) shows comparatively
symmetric results. This was also obvious from the temperature results discussed in the
previous section.

Figure 6.28(a) illustrates the time history deformation plots at points \( A \) and \( B \) for pattern \( a \).
It can be seen that the profile is nearly similar to the time history temperature plot of
Figure 6.26: Deformation plot along section $AB$ of the four deposition patterns ($a,c,d,g$).

Figure 6.27 (a): Deformation plot along section $MN$ of the four deposition patterns ($a,c,d,g$).
Figure 6.27 (b): Deformation plot along section $OP$ of the four deposition patterns (a,c,d,g).

Figure 6.28 (a): Transient Deflection history on Thermal Cycling at points $A$ and $B$ for pattern $a$. 
Figure 6.28 (b): Transient Deflection history on Thermal Cycling at points A and B for pattern e.

Figure 6.28 (c): Transient Deflection history on Thermal Cycling at points A and B for pattern d.
Figure 6.28 (d): Transient Deflection history on Thermal Cycling at points A and B for pattern g.

Figure 6.29: Comparison of axial residual stresses along section AB for the four deposition patterns (a,c,d,g).
figure 6.24(a) (i.e. the rate of change of temperature corresponds to the rate of change of deformation on the two points A and B respectively). The same is true for the other three patterns (comparing figures 6.24(b-d) with figures 6.28(b-d) respectively). Therefore from these results it may be easily deduced that the temperature gradients over time (time history temperature plot) which cause the thermal stresses are the major contributors towards deformations as seen from the time history deformation plot. It is evident from figure 6.28(a) that the deflection is accumulative in nature as with each pass the deformation increases. When the first few rows are being deposited the deflection at point A as compared to point B, increases rapidly, but for the final few passes the rate of change of deflection for point A reduces and that for point B increases. The deflection of the substrate plate is due to the contraction forces arising from the shrinkage of the molten metal. As the effect of contraction forces, for initial few passes, is dominant on side with point A therefore the rate of change in deflection is larger, but for the final few passes the major effect of contraction forces shifts to the side with point B, thus resulting in a higher deflection on this side.

It is also clear from figure 6.28(a) that the deflection on point A is much more abrupt initially and later becomes nearly negligible whereas on point B it is quite gradual throughout the deposition. About fifty percent of the total deflection produced on point A is caused by the first deposited row, and the first two deposited rows produce large deflection on it, as compared to point B. This can be expected as the heat source is near the side with point A. For the central deposited rows the rate of deformation of point A falls down appreciably while that for point B it raises significantly. This rise continues for point B for the last two deposited rows while the change at side A is almost negligible. This behavior at start side is a bit different in comparison to that of end side, when the first two rows were deposited. The reason for this phenomenon is difficult to quantify exactly, as a number of different factors like reaction due to bolting, contraction forces, and change in stiffness of system as the deposition continues, are playing there roles. As the deposited molten material solidifies, it becomes part of the geometry. Thus with each deposited row the stiffness of the substrate increases and the heat source moves further away from side A. Before deposition the substrate is geometrically symmetric (with respect to CD figure 6.2) but after the central row is deposited the start side will be stiffer.
than the end side. The moment of inertia of the cross section \( AB \) increases by about 37% after the deposition of 5 weld beads. Therefore as the last two rows are deposited, warping due to contraction forces and distortions due to bolting forces have comparatively lesser impact on the stiffer start side. This consequently can be one of the reasons causing an insignificant deflection on the start side.

In a similar fashion the buildup of deformation (as shown in figures 6.28(b-d)) for the other three patterns can be explained. It is also clear that this buildup of deformation is more symmetric for patterns \( d \) and \( g \) as compared to the other two patterns. The above discussion shows that a major factor restricting the deflection is the increase in stiffness due to the previously deposited material. The location of this previous deposition with respect to the material being currently deposited is therefore of importance. Hence this justifies the importance for selecting of an appropriate deposition pattern.

### 6.2.5.3 Effects of deposition patterns on Stresses

Figure 6.29 shows the axial residual stress distribution across the middle section \((AB)\) of the plate after unbolting. For pattern \( a \), higher stresses are found in areas near the deposition end side (around 75 mm), while lower residual stresses exists at the deposition start side (around 35 mm). This variation occurs due to the reheating effect of successive passes. When molten metal is deposited, high tensile stresses appears due to restraint thermal contraction. The high heat input causes large compressive thermal stresses in areas adjacent to deposition, which changes to tensile thermal stresses as those areas cools down. As the next row is deposited the reheating causes compressive thermal stresses, once again, thus reducing the level of tensile stress. Therefore the areas which are reheated, give comparatively lower residual stresses. The stresses near the two end points \( A \) and \( B \) are compressive, with larger values near point \( B \). The difference once again can be explained due to the absence of reheating after the last deposited line near point \( B \).

Pattern \( c \) exhibits similar stress (figure 6.29) trends as that of pattern \( a \), except that the difference between the compressive stress levels of points \( A \) and \( B \) is lesser. The compression at point \( A \) is higher for pattern \( c \) as compared to pattern \( a \), and vice versa is true for point \( B \). The tensile stress portion is also shifted to left for pattern \( c \) as compared to \( a \). However this shift is little on side \( A \) and comparatively more on side \( B \). The shift on
side A (of pattern c) can be attributed to the instant of reheating of the first deposited line. For pattern a the first deposited line is reheated by the very next line, while for pattern c it is reheated by the fifth line (figure 6.1-a & c). The time lag involved in the second case (218.4 sec) results in more cooling of the first deposited line region and thus comparatively lesser stress relieving due to reheating. On the other hand the left shift on side B (of pattern c) can be attributed to the reheating of the fourth deposited line by the seventh (figure 6.1c). Note that the line at the same location for pattern a (the last deposited line i.e. 7th line figure 6.1a) is not reheated. Another factor which plays its role in this shift is the absence of preheating for the fourth line (pattern c), (i.e. there was no prior line deposited adjacent to the fourth line when it was deposited) whereas the seventh line (pattern a) is preheated by the sixth deposited line. The effect of preheating will be more evident when the shift of pattern g is discussed.

The left shift of tensile region for pattern d can also be explained in a similar manner. However this shift, as compared to pattern c, is less on side A and more on side B. This can be understood by considering the time lag involved in the reheating. In pattern d the first deposited line (extreme left line) is reheated by the third and the second deposited line (extreme right line) is reheated by the fourth (figure 6.1-d). Thus a time lag of 72.8 seconds for reheating is involved on both sides. Whereas for pattern c the time lag for reheating of the extreme left and extreme right line is 218.4 sec and 145.6 sec respectively (figure 6.1-c). The absence of preheating of second line (pattern d) is also contributing to the shift on side B. Once again it can be noted that only the stress profiles of pattern d are nearly symmetric among the four patterns and this is true for both axial and transverse stresses (results of transverse stresses not shown here). This is also in conformity with the thermal and deformation results.

The extreme left and right deposited line of pattern g does not undergo any reheating. This is the only pattern among the four in which two deposited lines are not reheated (6th & 7th lines figure 6.1-g). In the other three patterns only the last deposited line is not reheated. Due to the absence of reheating of 6th line (pattern g) the region of transition from compressive to tensile stress (side A) is shifted left, the most, as compared to the all other patterns (figure 6.29). Also another contributing factor to this shift is preheating. As this is the only extreme left deposited line (compared with other patterns), which had
undergone preheating (due to the 4th line figure 6.1-g). The region of transition from tensile to compressive stress, on side B, is also slightly shifted to left for pattern g as compared to pattern a (figure 6.29). Although in both cases the 7th line (patterns a & g figure 6.1-a & g) is preheated, while reheating is absent. The reason can be attributed to the time lag involved after preheating in the two cases. For pattern a the 7th line is preheated by the 6th deposited line while for pattern g, preheating is caused by the 5th line. Thus the region in this case is further cooled down for an additional 72.8 seconds.

It can be seen that the span of tensile stress distribution for pattern g is the broadest while that of pattern d is the narrowest among the four patterns. This can be related with the deformation profiles of figure 6.26, where it can be seen that pattern g has the narrowest shape while pattern d has the broadest. Hence the narrowness of stress profile corresponds to a lower curvature and similarly vise versa.

Figure 6.30 (a & b) shows the axial stress distribution along the two edges MN and OP of the plate. Little variation was observed among the transverse stresses along the edges EF and GH, which are also much lower in magnitude (not shown here). Therefore the axial stresses are the major contributor along the weld deposition direction as well.

6.3 Conclusions

In this chapter a 3D FE based thermal and structural analysis of different deposition parameters in welding based deposition process were presented. Commercial finite element software was used to implement the welding parameters like Goldak double ellipsoidal heat source, material addition and temperature dependent material properties. The thermal as well as structural results have shown that the process is not symmetric therefore any such assumption is not justified. Thermal analysis of seven different deposition patterns were carried out; which were short listed, on the basis of similarities, to four patterns for the structural analysis. These simulations revealed that the thermal and structural effects, on the work piece, are different for different patterns. The effect of deposition on the buildup of residual stresses and deformation can be controlled and possibly reduced by employing a suitable deposition pattern. The major effect of the patterns is along the central section across the deposition path. The pattern, in which the deposition sequence starts from the outer paths and ends at the center, was identified to be
Figure 6.30 (a): Comparison of axial residual stresses along section MN for the four deposition patterns \((a,c,d,g)\).

Figure 6.30 (b): Comparison of axial residual stresses along section OP for the four deposition patterns \((a,c,d,g)\).
the preferred one among the studied patterns. This preference is based on the least and near symmetric thermal and structural effects and minimum warpage caused by this pattern. The time history deformation plots at the substrate edges correspond to the time history temperature plots and this is true for all the deposition patterns. The effects of interpass cooling duration were studied and it was found that an intermediate value of interpass time (60 sec) is suitable for a nominal level of deformations and stresses. The deformation profile also changes from a U-shaped to a W-shaped deformation as the interpass cooling time approaches 0 sec. However, there is lesser effect of interpass cooling time on the deformations above 30 sec interpass time. A similar finding was made from the studies about different weld bead starting control temperature levels; with the 650 K case being the best possible option amongst the studied cases, producing optimal deformation and residual stress results. The studies regarding different boundary conditions revealed that the deformations are least for adiabatic case while isothermal case produced the maximum deformations. However the isothermal case does not change the deformation profile from U-shaped to a W-shaped with the reducing interpass cooling time (60 to 0 sec).
Chapter 7

Results, Conclusions and Recommendations

The ultimate objective of the ongoing research is to produce parts that can physically imitate and work like a component produced by a conventional manufacturing technique. Thus the idea is to produce fully dense, form-fit-functional, metallic parts and tools. The big drawback of using welding as the deposition process is the large heat inputs causing high temperature gradients and resulting in deformations, warpage, residual stresses, delamination and poor surface quality. In addition the layer by layer additive manufacturing nature results in non-homogeneous structures, porosity and anisotropic material properties. In order to predict and minimize these problems, knowledge of thermal gradients and temperature history during manufacture is important. Moreover, to overcome the problem of surface quality and out of tolerance parts a hybrid welding/CNC milling based RP system can be a good option. The problems associated with the use of welding as RP tool needs to be minimized by the proper investigation of the different deposition parameters and process conditions e.g. intermediate machining, deposition patterns, heat sink size, interpass cooling time, preheating and constant control temperatures on the material properties and mechanical behaviors of the finally produced parts.

This dissertation presented an analysis based on a numerical and experimental approach for the effects of different deposition and process parameters on welding based rapid prototyping process. The entire work is divided into two main parts. The first part is an experimental comparison of microstructure and material properties of the simple GMAW based layered manufacturing (LM) with the hybrid welding/milling based LM process. The second part presents a finite element (FE) based 3D analysis to study the thermal and
structural effects of different deposition parameters and deposition patterns in welding based LM.

7.1 3D welding verses hybrid welding / milling based RP

In this regard, as a first step, a comparison of the material properties was made between weld based prototype parts, produced both with and without intermediate machining after each deposited layer. Thus a comparison between the material properties of steel parts made by 3D welding based LM (DWM) and a hybrid welding / milling (DWIM) based RP was studied. The material properties were investigated both on a macro and microscopic level. Microstructure for the two deposition procedures were studied and compared. The hardness test results for the two procedures were investigated and the results were studied in the light of the respective microstructures. Tensile test samples were developed and testing was performed to investigate the directional properties in the deposited materials.

7.1.1 Micrographic Analysis

The microstructure of samples from both types of deposition was comprehensively studied. It was observed that there is a good structural integrity and fusion between the weld beads. The microstructure of the entire body of the deposit can be divided into two main regions i.e. the un-reaustenitised and the reaustenitised. The microscopy revealed that the microstructure of the un-reaustenitised region comprises mainly of acicular ferrite, bainite, and ferrite-carbide aggregate in coarse grain boundary ferrite. There are also some quantities of martensite, widmanstatten ferrite, intragranular polygonal ferrite, and spheroidal cementite. On the other hand the microstructure of reaustenitised region is mainly containing polygonal ferrite with some fine pearlite at grain boundaries. It was found that the un-reaustenitised region exists in each deposited layer of a DWM sample, while for DWIM samples, except for the top layer, it is absent in all the central layers. The reason being, the lesser layer thickness of DWIM samples (as compared to DWM) because of the intermittent machining involved in the hybrid approach. The extra layer thickness of the DWM samples, prevents the effect of reaustenitisation to penetrate the entire depth of the layer thus resulting in some remaining unreaustenitised structure at the bottom of the layer. The central body of the deposit for DWM case therefore, comprise of a more complex microstructure as compared to DWIM.
The central deposit of DWM has bands of reaustenitised and un-reaustenitised region piled sequentially on each other with the sequence corresponding to the layer thickness. A similar sequential variation is also observed across the band with un-reaustenitised structure, corresponding to the weld bead interspacing. On the contrary, across the reaustenitised bands the grain size of the microstructure varies sequentially, and this sequence too corresponds with the weld bead interspacing. A similar sequential grain size variation is also observed across the DWIM samples, which only comprise of reaustenitised microstructure. The change in grain size is the result of multiple reheating of preceding beads by subsequent weld passes.

It can be concluded that the microstructure of the DWM deposit is much more multifarious as compared to the DWIM deposit, in which the central body of the deposit has a fairly homogeneous microstructure. This difference is due to the reduced deposited layer height, in the DWIM, caused by the intermediate milling operation. Although the grain size in DWIM varies sequentially between coarse and fine. A homogeneous microstructure therefore may result in nearly isotropic material properties and higher toughness as compared to that of DWM.

This study was based on a single set of deposition parameters and simple part geometry (i.e. rectangular slab). The microstructure and hence the amount of reaustenitised and un-reaustenitised areas is dependent on the heat input and the degree to which adjacent weld beads overlap, which in turn depends on a number of factors including the welding parameters, preheating, machined layer height, interbead spacing and geometry of the part. It will be of interest to examine the case for DWM, where an underlying bead is completely reaustenitised by the deposition of molten metal. Further studies therefore in this regard are required to be conducted to identify a relationship between the heat input, the extent of bead overlap and the amount of reaustenitised region. Furthermore studies to identify the relationship between the heat input and the minimum height to which machining is done in a DWIM may be conducted, to achieve complete reaustenitisation or else the microstructure of choice.
7.1.2 Hardness Results

The hardness values along the different layers of both DWM and DWIM samples show a fluctuating wavy profile with the peaks repeating in accordance with the weld bead interspacing. The waviness is also in accordance with the sequential variation in the microstructure of the corresponding layer. However, layers in which predominantly reaustenitised and un-reaustenitised regions appear alternately, have hardness profile with higher amplitude as compared to those where RC to RF grains appear alternatively. Moreover there are also layers with almost uniform hardness profile, observed in the central layers of both type of depositions. This represents the segment containing only RF grain structure. The average hardness of the central layers of DWM samples was found to be HV 192.2 while that for DWIM samples it was slightly lower i.e. HV 181.3.

All the hardness results are in good agreement with the microstructure of the corresponding layers. The main difference in DWM and DWIM was found in the central layers. The results can be further refined by going for micro hardness testing of the samples. But as the variation in microstructure is on a millimeter scale, therefore the existing results can be considered reasonable.

7.1.3 Tensile Test Results

Tensile tests of samples built by the two techniques, DWM and DWIM, were conducted on specimens produced both longitudinal and perpendicular to the weld deposition direction. Visually, the tensile specimens show no evidence of any difference in the crystal structure on fracture, and the material strength is also almost the same for the two types of samples. The tensile strength the yield point and the elongation values are within the specified range for ER70S-6 weldments, except elongation for the longitudinal samples. These values on the average are slightly higher as compared to the perpendicular samples, and this is true for the samples built by both the deposition techniques. As the reheated coarse region is the weakest region therefore, the breakage for the perpendicular samples takes place from this region. But for the longitudinal samples, there is no such segregation across the direction of the applied force, and this therefore results in a slightly higher average elongation value. In general, it can be said that the results show nearly
isotropic material properties; which means that the bonding in the direction perpendicular to the orientation of the deposit is equally strong as in the longitudinal direction.

The correlation for hardness values as related to the tensile strength also holds within normal expectations. Another difference observed in the stress-strain plot is a sharp yield point (Lüders extension) prominent for all of the DWIM samples, which however, is absent for the DWM samples. The higher population of RF grains with body-centered cubic (bcc) structure and firmly locked dislocations are the reasons for Lüders extension in DWIM samples. Finally it may be concluded that although there is difference in the microstructure and hardness of parts produced by the two deposition techniques but the effect on the tensile properties is limited to the formation of Lüders band.

7.2 Finite Element Analysis

A 3D finite element (FE) based study of the thermal and structural effects of different deposition parameters and deposition patterns in welding based layered manufacturing (LM) was done. Welding is one of those processes where high heat input results in large thermal gradients causing the build up of residual stresses. The ultimate result of this is distortions, warping and hence out of tolerance parts. In order to reduce the residual stresses and deformations, the knowledge and control of the thermal cycle is of importance. FE simulations were carried out by means of a user programmed subroutine to implement the welding parameters like Goldak double ellipsoidal heat source, material addition, temperature dependent material properties. The model is limited to deposition of a single layer, which should be sufficient to investigate and study the comparison of the effects caused by different deposition parameters. For the deposition of multiple layers, as is the case in RP, the effects (e.g. deformations) are expected to add up. Although every deposited layer increases the stiffness of the plate, therefore these effects may not be same in magnitude for each subsequent layer. Also the shape of the deposited layer (i.e. rectangular) considered in this study, is a simple shape, which may not be common in RP. However the results of the current study may provide a good guide line for the general RP parts. Furthermore, the presented results are for deposition by gas metal arc welding but can be applied to other deposition process employing moving heat source.
The fusion zone, temperature history and deformations, predicted by the numerical model, show a good agreement with experimental data. The model can accurately predict the remelting depth. The substrate deformation for the bolted case is in excellent conformity with experimental data, while for unbolted case a qualitative agreement is found, which may be attributed to a simplified bolt model. The residual stress results are in qualitative agreement while quantitatively the finite element model can’t give reliable results. This may be due to some experimental errors associated with the incremental hole-drilling technique or due to the simplified bolt model. It may be concluded that the finite element model can be used for reliable prediction and optimization of process parameters especially for comparative studies.

7.2.1 Simulation Results

In contrast to the earlier works sited in the literature, adopting some simplified methodology, a more realistic approach was adopted. The following conclusions were drawn:

- It was established that the thermal cycle and the transient material addition can not be considered symmetric with respect to the geometry of the substrate, and is therefore dependant on the relative position of the heat source and the consequent material deposited.

- There is little effect of increased heat flow area and preheating on the peak temperatures reached at a point (when directly under heat source) while on the contrary these two factors have a considerable effect on the secondary peak temperatures reached at the same points due to deposition in the surroundings. The heat flow rate for the first deposition pass is lesser than the later passes which becomes almost uniform after the first pass. This is because of the one time increase in heat conduction area after the deposition of the first pass; causing an increased heat flow rate after the first pass.

- The increased heat sink available due to the increased thickness of the substrate plate does not have substantial effect on the primary peak temperatures reached on the substrate directly due to the heat source.
For the raster deposition patterns, the temperature distributions and deformations alongside the deposition are more dominant in direction parallel to the deposition direction as compared to across the deposition direction; whereby the maximum deformation occurring at the central section along the deposition direction.

A similar deposition sequence with different deposition directions does not cause a marked difference in the thermal profiles and temperature history of the substrate. The structural effects also uphold trends similar to that of thermal.

A non-symmetric temperature profile corresponds to a non-symmetric deformation and vise versa. The deposition pattern starting from outside and ending at the center show symmetric thermal, deformation and residual stress profiles; and similarly this is true for inside-out pattern.

The outside-in deposition pattern \(d\) gives minimum warpage among the four deposition patterns studied while it is the maximum for the inside-out pattern \(g\).

The profile of the time history deformation plots is nearly similar to that of the time history temperature plot for all the cases (i.e. the rate of change of temperature corresponds to the rate of change of deformation). Furthermore it was deduced that the temperature gradients over time (from time history temperature plot) which cause the thermal stresses are the major contributors towards deformations.

Before deposition the substrate is geometrically symmetric but after deposition the side with deposited material will be stiffer than the other side. For the present case the moment of inertia of the cross section increases by about 37% after the deposition of 5 weld beads. Therefore as the last rows are deposited, warping due to contraction forces and distortions due to bolting forces have comparatively lesser impact on the stiffer start side. A major factor therefore, restricting the deflection, is the increase in stiffness due to the previously deposited material. The location of this previous deposition with respect to the material being currently deposited is therefore of importance. Hence this justifies the importance for selecting of an appropriate deposition pattern.
• The axial stresses are the major contributor across as well as along the weld deposition direction.

• When molten metal is deposited, high tensile stresses appears due to restraint thermal contraction. The high heat input causes large compressive thermal stresses in areas adjacent to deposition, which changes to tensile thermal stresses as those areas cools down. As the next row is deposited the reheating causes compressive thermal stresses, once again, thus reducing the level of tensile stress. Therefore the areas which are reheated, give comparatively lower residual stresses.

• A continuous deposition gives comparatively lower deformations, however these produce fluctuating residual stresses in the substrate. In general as the interpass cooling increase so does the substrate deflection at the edges while it decreases at the center. For high interpass delay duration (120 sec), the substrate preheats available for the next row decreases, which results in larger temperature gradients across the substrate thickness at the time of next weld deposition and hence the deformation at the edges increases. On the other hand continuous deposition causes excessive preheating which results in high temperature areas and larger remelting of the substrate or of the underlying layers. In addition the area of interest in RP is the deposition area; 0 and 16 sec cases give the most deformations in this area. The stresses are also fluctuating in between tensile and compressive range in this area. The surface finish of the build up is also poor for lower interpass. Hence 60 sec interpass case seems to be the best possible option amongst the studied cases with optimal deformation and residual stress results.

• The deformation reduces with the increasing base control temperature however, these produce fluctuating residual stresses in the substrate. In general as the base control temperature increase so does the substrate deflection at the edges while it decreases at the center. In addition the deformation profile changes from a bell shape to W-shape as the base control temperature increases. However the compressive stresses at the edges, other than the 800 K case, are not much different. A fluctuation in the plot for 800 K case can be observed between the tensile and compressive stress region within the deposition area. A higher control temperature value results in more compressive
axial residual stresses on the deposition start side, lower at the end side and a dip from tensile into compressive zone in the deposition area. The stress relieving by the weld beads on the substrate and previously deposited material results in lower stresses in the initial half of the deposited area than the stresses in the later half. This effect tends to reduce with the increasing control temperature. The 650 K control temperature case seems to be the best possible option amongst the studied cases with optimal deformation and residual stress results.

- The cases with adiabatic and convective bases give almost similar deformation and stress results and are quite lesser in magnitude than the isothermal base case. Isothermal boundary conditions give higher thermal gradients and therefore higher deformations. An insulated base gives more uniform temperature distribution throughout the substrate resulting in an overall lower deformation. The convective case also has a similar behavior except a very slight difference at the start side. For a 60 sec interpass cooling time the stress profile of the isothermal plot is generally similar to the other cases (although different in magnitude), except that on the deposition start side. This difference is due to the higher thermal gradients that the isothermal case has in the beginning, which however reduces as the deposition carries on. The effect of stress relieving by the weld beads on the substrate and previously deposited material is more prominent for the isothermal case as compared to the others.

- For a 0 sec interpass cooling time the adiabatic and convective cases results in a similar W-shaped plot, while the isothermal case produces a U-shaped plot which is in contrast to the general behavior for small interpass time (0 sec). The reason to this once again is the speedy heat removal (higher thermal gradients) and therefore lesser heat accumulation in the substrate. High heat accumulation in the substrate causes more shrinkage forces at the deposition area thus resulting in an upward deformation in the central area of deposit and therefore a W-shaped plot. The adiabatic and convective cases with 60 sec interpass seems to be the preferable choice amongst the discussed cases as these give the least deformations and comparatively lesser residual stresses.
7.3 Recommendations for Future Work

The ability to rapidly produce the designs of parts and objects with complex 3D geometry is a basic necessity in today’s marketplace. RP has emerged over the last 20 years based on the principle of creating 3D geometries using computer aided design (CAD), directly and quickly with little human interaction. It is a technology used in facilitating concurrent engineering, thus it has started to change the way companies design and build products. The main goal is to reduce product development, manufacturing costs and lead times, thereby increasing competitiveness. In the last two decades impressive steps towards this goal have been made. However the field of RP is still developing, with much effort being expended on improving the speed, accuracy and reliability of RP systems and widening the range of materials for prototype construction. Another area for improvement is the costing, as most RP systems are currently too expensive to be affordable by any but the larger firms. However, RP technology is available to most companies via the services provided by different bureaus. The future is likely to see more user-owned RP machines as their costs are reduced. To date most RP parts are used for prototyping or tooling purposes; however, in future the focus may be to produce parts as end-use products. In this context the term ‘rapid manufacturing’ (RM) is used in which RP technologies are employed as processes for the production of end-use products. The introduction of non-polymeric materials, including metals, ceramics, and composites, represents a much anticipated development. These materials would allow RP users to produce functional parts. Today’s plastic prototypes work well for visualization and fit tests, but they are often too weak for function testing. More rugged materials would yield prototypes that could be subjected to actual service conditions. In addition, metal and composite materials greatly expand the range of products that can be made by RM.

On the basis of the present work following recommendations are made for further studies in the area of welding based RP:

- The affect of various deposition parameters on the weld quality and surface finish is suggested for future work; especially with respect to RP requirements (i.e. weld bead surface roughness, de-bonding, heat input and re-melting of previous layer, weld start and end effects etc.). Some work was done in this regard and various weld beads were
deposited varying different deposition parameters; but it was too preliminary to be reported. The samples so prepared and the related weld data can be utilized to carry on further research. It may be noted that the different testing facilities available in the Faculty of Materials and Metallurgy will be required in this regard.

- The microstructure and hence the amount of reaustenitised and un-reaustenitised areas is dependent on the heat input and the degree to which adjacent weld beads overlap, which in turn depends on a number of factors including the welding parameters, preheating, machined layer height, interbead spacing and geometry of the part. It is of interest to examine the case for DWM, where an underlying bead is completely reaustenitised by the deposition of molten metal. Studies in this regard are required to be conducted to identify a relationship between the heat input, the extent of bead overlap and the amount of reaustenitised region. Furthermore studies to identify the relationship between the heat input and the minimum height to which machining is done in a DWIM may be conducted, to achieve complete reaustenitisation or else the microstructure of choice.

- This study was based on a single set of deposition parameters and simple part geometry (i.e. rectangular slab). Working alongside with the deposition parameters study of different geometric shapes can be another area for further research. Affects of heat on buildup and surface quality of thin walls, corners and edges can be of interest.

- The patterns selected in this study are all raster with changes in the sequence of deposition. The spiral patterns were not considered because of complexity involved in the programming and the limitations of the computational resources especially in the 3D structural analysis. However this may be a feature to be studied in future.

- Another thing which may be considered is the deposition time. As the cooling time in between each deposited bead is increased so does the overall deposition time of the entire job. It was found that a constant weld bead starting temperature of 650 K requires about 65% lesser time depositing the same layer as compared to a 375 K case, thus pointing towards improved process efficiency. However as reported by Dickens et al. [117]; a higher control temperature results in a reduced weld surface quality.
Further studies are required that will allow part quality to be maintained without incurring excessive time penalties or adversely affecting positive process features.

- The present model is limited to the buildup a single layer therefore it is suggested that the model may be extended to the buildup of multiple layers. The multi layer model can be helpful to study and predict the de-bonding and distortions due to the orientation of different layers.

- It has been observed during the course of the analysis that the heat transfer to the substrate is predominantly across the thickness. Therefore it is suggested that the substrate made of various sections can be compared for better heat transfer conditions.

- For simplicity, the weld bead cross-sectional geometry considered in this study was rectangular. It would be interesting to make a more realistic improvement in the model by making the weld bead cross-section a circular segment. This may also help in studying the effects of weld bead to weld bead overlap.

- Converting the semi-automatic deposition system to CNC controlled fully automatic system for further experimental studies.

- Study of multi material interface.
Appendix A

Contour plots of Deformations and Residual stresses
Figure A-1: Contour plot of deformation (UY) for bolted Substrate, Raster pattern a and 60sec interpass time.

Figure A-2: Contour plot of deformation (UY) for unbolted Substrate, Raster pattern a and 60sec interpass time.
Figure A-3: Contour plot of transverse stress (SZ) for bolted Substrate, Raster pattern a and 60sec interpass time.

Figure A-4: Contour plot of transverse stress (SZ) for unbolted Substrate, Raster pattern a and 60sec interpass time.
Figure A-5: Contour plot of deformation (UY) for bolted Substrate, Raster pattern a and 0sec interpass time.

Figure A-6: Contour plot of transverse stresses (SZ) for bolted Substrate, Raster pattern a and 0sec interpass time.
Figure A-7: Contour plot of deformation (UY) for bolted Substrate, Raster pattern a and 16sec interpass time.

Figure A-8: Contour plot of transverse stress (SZ) for bolted Substrate, Raster pattern a and 16sec interpass time.
Figure A-9: Contour plot of deformation (UY) for *bolted* Substrate, Raster pattern \( a \) and 30sec interpass time.

Figure A-10: Contour plot of transverse stresses (SZ) for *bolted* Substrate, Raster pattern \( a \) and 30sec interpass time.
Figure A-11: Contour plot of deformation (UY) for bolted Substrate, Raster pattern a and 120sec interpass time.

Figure A-12: Contour plot of transverse stress (SZ) for bolted Substrate, Raster pattern a and 120sec interpass time.
Figure A-13: Contour plot of deformation (UY) for bolted Substrate, Raster pattern \( a \) and 375K preheat temperature.

Figure A-14: Contour plot of transverse stress (SZ) for bolted Substrate, Raster pattern \( a \) and 375K preheat temperature.
Figure A-15: Contour plot of deformation (UY) for *bolted* Substrate, Raster pattern a and 450 K preheat temperature.

Figure A-16: Contour plot of transverse stress (SZ) for *bolted* Substrate, Raster pattern a and 450 K preheat temperature.
Figure A-17: Contour plot of deformation (UY) for bolted Substrate, Raster pattern \( a \) and 500K preheat temperature.

Figure A-18: Contour plot of transverse stresses (SZ) for bolted Substrate, Raster pattern \( a \) and 500K preheat temperature.
Figure A-19: Contour plot of deformation (UY) for bolted Substrate, Raster pattern a and 650K preheat temperature.

Figure A-20: Contour plot of transverse stress (SZ) for bolted Substrate, Raster pattern a and 650K preheat temperature.
Figure A-21: Contour plot of deformation (UY) for bolted Substrate, Raster pattern a and 800K preheat temperature.

Figure A-22: Contour plot of transverse stresses (SZ) for bolted Substrate, Raster pattern a and 800K preheat temperature.
Figure A-23: Contour plot of deformation (UY) for bolted Substrate, Raster pattern a and adiabatic base.

Figure A-24: Contour plot of transverse stress (SZ) for bolted Substrate, Raster pattern a and adiabatic base.
Figure A-25: Contour plot of deformation (UY) for bolted Substrate, Raster pattern a and isothermal base.

Figure A-26: Contour plot of transverse stress (SZ) for bolted Substrate, Raster pattern a and isothermal base.
Figure A-27: Contour plot of deformation (UY) for bolted Substrate, Raster pattern b and 60sec interpass time.

Figure A-28: Contour plot of transverse stress (SZ) for bolted Substrate, Raster pattern b and 60sec interpass time.
Figure A-29: Contour plot of deformation ($UY$) for bolted Substrate, Raster pattern c and 60sec interpass time.

Figure A-30: Contour plot of transverse stress ($SZ$) for bolted Substrate, Raster pattern c and 60sec interpass time.
Figure A-31: Contour plot of deformation (UY) for bolted Substrate, Raster pattern \( d \) and 60sec interpass time.

Figure A-32: Contour plot of transverse stress (SZ) for bolted Substrate, Raster pattern \( d \) and 60sec interpass time.
Figure A-33: Contour plot of deformation (UY) for *bolted* Substrate, Raster pattern $f$ and 60sec interpass time.

Figure A-34: Contour plot of transverse stresses (SZ) for *bolted* Substrate, Raster pattern $f$ and 60sec interpass time.
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