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To the memory of my father Andrew Oseyi Ebadan
LASER WELDING OF SELECTED AEROSPACE ALLOYS

Abstract

The literature relating to the laser welding of a number of alloys was studied, this revealed gaps in the current state of knowledge. It was found that no indepth parametric or microstructural investigations had been carried out into the laser welding of non-ferrous alloys. Although problems associated with the laser welding of aluminium alloys had been identified, there was little in the literature to suggest an indepth study into the effect of surface modifications in enhancing the laser weldability of this material had been carried out. The laser welding of aluminium based metal matrix composites has also yet to be studied in detail. A programme was therefore set up to carry out an indepth investigation into the laser welding of a number of alloys namely; Nimonics C263, and PE11, aluminium 6061, and aluminium 6061 based boron fibre reinforced, and silicon carbide particulate reinforced metal matrix composites.

The study was aimed at developing an understanding of the microstructural effects of the laser welding process on the alloys, and assessing the structural integrity of the resultant welds. The effect of laser processing parameters such as laser power, laser beam traverse speed, lens focal length, and the manipulation of these parameters on the welding efficiency and weld area integrity was also investigated. Other tasks within the project included a study on the possibility of using an anodic film to enhance the laser weldability of Al 6061. Finally attempts were made to identify novel phases observed in the weld area of the composite materials.

Nimonics C263, and PE11 exhibited laser welds free of cracks and porosity. The difference in composition between the two alloys did not result in any significant dissimilarities in their response to the laser welding process. The welds in both alloys exhibited a fine columnar dendritic microstructure, and while carbides were observed in the interdendritic regions of the welds, electron optical analysis did not reveal any γ precipitates in this region. It was concluded that for the welding of thin gauge materials above a threshold laser power the resultant welding efficiency shows a greater dependence on laser beam mode, and laser spot size, than on laser power, and beam traverse speed. Aluminium 6061 was not easily welded with a laser in its as received form, and the welds showed some degree of porosity. Anodising was found to improve the welding efficiency in this material. While the presence of an anodic film on the metal surface increased the welding efficiency of the alloy, no relationship was found between the thickness of the anodic film and welding efficiency in the range of film thicknesses investigated in this work. Weld regions were observed to be cellular dendritic in structure, with narrow heat affected zones. No precipitates or low melting point phases could be identified in the weld region. Melt zones were successfully produced in the composite materials, with the main problem encountered being that of porosity adjacent to the weld bead. It was shown that manipulation of the laser welding parameters resulted in a decrease in this porosity. In the weld beads, a number of novel phases were observed. These were identified with the aid of TEM, and SIMS analysis techniques.
CONTENTS

Preface

Title Page

Abstract

CHAPTER 1

1. Introduction 1

CHAPTER 2

2. REVIEW OF THE RELEVANT LITERATURE
AND INTRODUCTION TO THE PRESENT WORK 6

2.1 Introduction to lasers and a review of the carbon
dioxide laser 6

The Carbon Dioxide Laser 8

Population Inversion and the role of nitrogen in carbon
dioxide lasers. 10

Depopulation and the role of helium in carbon dioxide
lasers 12

Heat disposal 12
Laser beam modes 13
Types of carbon dioxide lasers 15
Methods of operation 15
Laser beam characteristics 17
Energy, power, and intensity 17
Laser beam diameter 19
Beam traverse speed 20
Effective spot size 21
Material properties 22

2.2 Laser welding 24
Transmission 26
Blanketing 29
Previous work on laser welding 30
Advantages of laser welding 31
Factors affecting the laser weldability of materials 32
Reflectivity 32
Thermal diffusivity 32
Melting point 32
Boiling point 33
Viscosity 33
Previous work on the laser welding of metals 33

2.3 Review of the metallurgy of relevant Nimonic alloys, aluminium 6061, and metal matrix composites 42
Metallurgy of the Nimonic alloys 42
Nimonic C263 42
CHAPTER 3

EQUIPMENT AND MATERIALS

Introduction 52

3.1 Description of laser systems used in the experimental programme 53

3.1.1 Description of the Welding Institute 5kW continuous wave CO₂ laser 53
The blower 53
Laser optics 54
Voltage supplies 56
The laser head 56
The x-y table 56

3.1.2 The 2kW continuous wave CO₂ laser 56
Safety 56
3.2 Operation of the equipment

3.2.1 The 5kW laser

3.2.2 The 2kW Laser

Effect of discharge current
Beam alignment and mode structure
Determination of focal point
The blue flash test
The x-y table

3.3 Anodising procedure

Measurement of oxide coating thickness

3.4 Post laser processing analysis equipment

Metallographic preparation
Optical microscopy
Electron microscopy
SIMS investigation
The Quantimet 800
Microhardness testing

3.5 Materials

Nimonic C263
Nimonic PE11
Aluminium 6061 70
Aluminium 6061/boron fibre reinforced metal matrix composite 70
Aluminium 6061/silicon carbide particulate reinforced metal matrix composite 71

CHAPTER FOUR

4. LASER WELDING OF NIMONICS C263 AND PE11

4.1 Laser operating conditions 72
4.1.1 The 5kW laser 72
4.1.2 The 2kW laser 76

4.2 Results 79
Nimonic alloy C263 79
Effects of laser power on weld penetration and microstructure 80
The effect of laser beam traverse speed on weld penetration and weld bead top width 81
The Effect of Lens Focal Length on Weld Penetration 82
Derived results relating to the laser welding of Nimonic alloy C263 with a 2kW and a 5kW laser 83
Nimonic alloy PE11 85
Effect of laser power on weld penetration and microstructure 86
The effect of laser beam traverse speed on weld penetration and weld bead top width 87
Derived results relating to the laser welding of Nimonic alloy PE11 88

4.3 Discussion 89
Comparison between the welding with the 5kW and 2kW lasers 89
Comments on the effects of the laser welding process on the microstructure and mechanical properties of the welds in alloys C263 and PE11 98
Porosity and cracking 102
The effect of incident laser power, laser beam traverse speed, and lens focal length on welding of Nimonic alloys C263 and PE11 104

4.5 Summary 111

CHAPTER 5

5. LASER WELDING OF ALUMINIUM BASED ALLOYS AND METAL MATRIX COMPOSITE MATERIALS

Introduction 113
5.1 Comments on the optical properties of aluminium
and the effects of alloying elements on the laser
processing of aluminium alloys

5.2 Comments on the specific experimental conditions
used in the welding of the aluminium based materials

5.3 Results

5.3.1 As received aluminium 6061 welds
Effect of laser power on weld penetration and
microstructure
Description of weld structure
The effect of laser beam traverse speed on weld penetration
and weld bead top widths
Microhardness

5.3.2 Anodised aluminium 6061 welds
The effect of laser power on weld penetration and
microstructure
Description of weld microstructures
The effect of laser beam traverse speed on weld penetration
and weld bead top widths
The effect of anodic film thickness on weld penetration
and weld bead top widths
Microhardness
Summary
5.3.3 Aluminium based metal matrix composites

Al 6061/boron fibre reinforced MMC

Effect of laser power on weld penetration and microstructure
SIMS investigation
Electron microscopy
Summary

Al 6061/silicon carbide particulate reinforced composites

Effect of laser power on weld penetration and microstructure
Electron microscopy
Summary

5.4 Discussion

5.4.1 As received and anodised Al 6061 welds

The effect of welding parameters on weld penetration microstructure, and mechanical properties
The effect of optical properties on the welding of as-received Al 6061
The effect of sulphuric acid anodising treatment on the laser weldability of Al 6061
Porosity
Cracking

5.4.2 Aluminium 6061 based metal matrix composites

Effect of laser welding parameters
Al/B fibre reinforced MMC
CHAPTER 6

6. SUMMARY, CONCLUSION, AND SUGGESTIONS FOR FUTURE WORK

6.1 Summary

Nimonic C263 and PE11
As-received and anodised Al 6061
Al 6061 based metal matrix composites

6.2 Conclusions

Nimonic C263 and PE11
As-received and anodised Al 6061
Al/B fibre reinforced MMC
Al/ SiC particulate reinforced MMC

6.3 Suggestions for future work

References

Acknowledgements

Appendix 1
CHAPTER 1

1. INTRODUCTION

Since 1960(1) when the first lasing action was recorded lasers have been making an increasing contribution in areas of scientific and technological activity. The power output of a particular laser determines to a large extent the area of application to which it is suited. For example helium-neon lasers provide a source of coherent light for low power applications such as in alignment tools used in the surveying and construction industry, and in high speed printers used for data processing. On the other hand the CO₂ laser is used extensively where higher powers are required as in machining, cutting, and welding. The versatility of the laser is so great that it is not possible within the scope of this thesis to give a comprehensive survey of all its applications. Such details can be found in papers by Scrivener(2), Duley(3), (4), Saltonstall et al(5), and Laser Applications journal 1971(6). It would suffice here to list some of the most prominent fields of application and examples as follows.

- Communications, such as optical fibre links in laser diode transmitters for digital data transmission; optical data processing, in high speed printers; and in new video disc technology.
- Medicine and biology, such as in retinal surgery; cell irradiation and cauterization; and the study of biological systems by laser irradiation.
- Physical and chemical experimentation, such as isotope separation; and Raman spectroscopy studies.
Atmospheric measurements, such as measurements of clear air turbulence; and
the study of fog particles.

Materials processing.

Laser materials processing is the area of major interest for the purpose of this
thesis, therefore it is considered in greater detail here. The development of
multi-kilowatt continuous wave carbon dioxide lasers around 1970 greatly
increased the feasibility of using lasers for technical and commercial operations(2).
Laser materials processing can be divided into three basic areas which are surface
treatment, cutting, and welding. Laser surface treatments encompass operations
such as surface hardening, surface alloying, and cladding. These processes are
presently being applied in industry. There are however other aspects of laser
surface treatments which are still in the laboratory stage of development. These
include processes such as surface melting, particle injection, laser chemical vapour
deposition, laser physical vapour deposition, enhanced plating, and shock
hardening. Excellent reviews of these processes are given in reports by Steen(7),
and Steen and Powell(8).

Laser cutting can be regarded as an extension of the welding process, because it
involves the melting of the material being irradiated by the laser beam, but whereas
in welding the melt is allowed to resolidify, in cutting it is either vaporised or blown
away with the aid of a gas jet resulting in a cut. In the cutting of ferrous materials,
the cutting rate has been greatly accelerated by the use of oxygen, as the resulting
exothermic reaction contributes to the energy input to the process. Laser drilling
and machining rely on the same principle as laser cutting, and can be regarded as
intermediate stages in the cutting process. Available literature on these processes
include papers by Duley(4), Powell and Menzies(9), and Kovalenko(10). However,
since the present research is concerned with laser welding this area will be reviewed
in detail in the following chapter.
Much of the work on welding has however involved ferrous alloys, probably due to the fact that these materials are widely used in industry, and responded well to the laser welding process. While a certain amount of work has been carried out on non-ferrous alloy welding, surprisingly little has been reported on the laser welding of Nimonic alloys. For this reason there is very little information available on the response of these materials to the laser welding process and variations in the laser welding process parameters. The effect of these process parameters on the resultant microstructure of the weld zone, and hence the final properties of the joint have also to be considered.

In the course of this work Nimonic C263 which is an aerospace alloy was studied as was Nimonic PE11 which finds its uses more traditionally in the nuclear industry. The work in this thesis concentrates on dealing with a detailed study of a number of laser welding parameters such as; incident laser beam power, laser beam traverse speed, and lens focal length on the weld penetration achieved in these alloys. Microstructural aspects, such as the response of precipitation characteristics, dendrite arm spacings, dimensions of the heat affected zone, (HAZ), and the structural integrity of the weld zone to variations in laser welding parameters were also studied.

While a certain amount of work has been carried out on the laser welding of aluminium, and its alloys due to the inherent high reflectivities of these materials to the 10.6μm radiation of the CO2 laser, very little has been reported on the laser welding of Al 6061, and Al based metal matrix composites. It was undertaken in this work to carry out a detailed study on the effect of variations in laser welding parameters, such as laser beam power, and laser beam traverse speed on the weld penetration achieved in the welding of Al 6061. The effect of anodic surface coatings on the absorptivity of the Al 6061 for the 10.6μm laser radiation, and
hence the weld penetration observed during welding was also studied. This work included a detailed investigation into the effect of anodic coating thickness on the weld penetration achieved for a given set of laser welding parameters. It was possible to compare the responses of the as received Al 6061, and the anodised Al 6061 to the laser welding process, and thus comment on the suitability of using an anodised surface finish to enhance the laser welding efficiency of Al 6061. The microstructures attained within the joint region for both surface finishes were also studied, and related to the laser welding parameters. Aspects considered included, HAZ widths, precipitation within the weld region, and dendrite arm spacings.

Very few reports are available dealing with the joining of metal matrix composites by fusion processes, and the final aspect of these welding studies involved investigations into the response of Al based MMCs to the laser welding process. The aim was to identify the effect of variations in laser welding parameters on the welds produced with a view to determining optimum welding parameters for the materials. A detailed study of the resultant weld regions in Al/B fibre reinforced MMC, and Al/SiC particulate reinforced MMC was carried out, and the phases formed within the fusion zone were identified.

The following chapters in this thesis set out the experimental procedures adopted to carry out the studies outlined in the preceding paragraphs. Details of the results, their interpretation, and finally a discussion of the observed trends and the implications thereof is given. Finally the conclusions drawn from the work are detailed, and suggestions made for future work. An outline of how the chapters are set out is given below.

Chapter 2 - review of the relevant literature and introduction to the present work.
Chapter 3 - equipment and materials.
Chapter 4 - laser welding of the Nimonic alloys.

Chapter 5 - laser welding of the aluminium based materials.

Chapter 6 - summary, conclusions, and suggestions for future work.
CHAPTER 2

2. REVIEW OF THE RELEVANT LITERATURE AND INTRODUCTION TO THE PRESENT WORK

This chapter is divided into three sections. The first of these deals with an introduction to lasers and a detailed review of carbon dioxide lasers. The second main section is concerned with the interactions between lasers and materials, and the processing of materials using lasers. The third section is devoted to a brief review of the metallurgy of the Nimonic alloys, and in particular those studied in this work, as well as the metallurgy of aluminium 6061, and aluminium based metal matrix composites. This is then followed by an introduction to the present work.

2.1 Introduction to lasers and a review of the carbon dioxide laser

The concept of stimulated as distinct from spontaneous radiative transitions was introduced by Einstein in 1917\(^{(1)}\). However, the first experimental demonstration of amplification by stimulated emission was carried out several decades later by Maiman\(^{(11)}\) who used ruby to investigate the belief that the rate of decay of fluorescence depended only on the material itself, following the publication of experiments by Gordon et al\(^{(12)}\). However Maiman's experiments showed that ruby when stimulated by a very bright light from a flash gun produced a special kind of fluorescence. By shaping the ruby to produce an optically resonant cavity, and oscillating under the proper conditions, coherent radiation was produced. Historically this was the first recorded laser action.
The word laser itself is an acronym of: "Light Amplification by Stimulated Emission of Radiation", coined by analogy to another acronym: Maser (Microwave Amplification by Stimulated Emission of Radiation). The most important characteristic of the laser light is that it is coherent. The individual waves are all of the same wavelength or colour, and are all in phase. Table (1) shows some of the most popular lasers and their applications(13). Of these only two are used for nearly all materials processing operations: the CO$_2$, (carbon dioxide), lasers, and the Nd:YAG, (Neodymium:Yttrium Aluminium Garnet), laser. The CO$_2$ laser and the Nd:YAG laser have wavelengths of 10.6µm and 1.06µm, respectively. The output power of the Nd:YAG laser ranges in continuous operation from a few tenths of a watt to about 1000 Watts, and up to 10kW(13) for short periods when operated in the pulsed mode. With the CO$_2$ laser, output powers up to 20kW(13) in continuous operation have been observed, and operation in the pulsed mode has resulted in peak powers of 10MW being recorded.

Several factors determine how the laser beam interacts with the material it irradiates. These include (a) the diameter of the beam at the material surface, (i.e focused or not focused; (b) the physical properties of the material; (c) distance from which the beam is projected onto the material surface; (d) laser beam mode, i.e the energy density distribution within the beam; (e) laser output power level; and (f) the duration of interaction. The manipulation of these conditions produces the required laser material interaction. At the present time, lasers have reached the stage in their development when they are making significant contribution in areas of scientific and technological activity such as: communications, optical data processing, medical, biological, and industrial applications. Of paramount interest for the purpose of this work are the industrial applications of lasers which include operations such as cutting, drilling, surface treatment, and welding. The purpose of this project is to study the application of the CO$_2$ laser in the welding of selected alloys. In order
for a greater understanding of the processes occurring to be possible, a detailed appraisal of the CO₂ laser system is given.

**The carbon dioxide laser**

The CO₂ laser is one of a class of lasers known as molecular gas lasers. It owes its existence to the fact that a triatomic molecule such as CO₂ is capable of existing in several metastable excited vibrational, rotational, and translational states under the correct energy conditions. Relaxation from these metastable higher energy levels results in the emission of energy in the form of photons or light quanta.

The CO₂ molecule, like most molecules has, in addition to electronic energy levels found in atoms, vibrational, rotational, and transitional energy levels. Thus for a given electronic energy level various transitions between vibrational levels can be identified, and for any two different vibrational levels transitions between the rotational levels can be ascertained. These transitions are governed by a quantum mechanical selection rule that states that only those transitions involving a change in rotational angular momentum equal to $\frac{h}{2\pi}$, (h is Planck's constant), can occur. The value of $\frac{h}{2\pi}$ is represented by J.

The CO₂ molecule has three degrees of vibrational freedom by virtue of its linear and symmetric configuration shown in Fig. 2.1a. These degrees of freedom are designated by three integral numbers $v_1$, $v_2$, and $v_3$. The first number is the quantum level of the symmetric stretch mode, Fig. 2.1b. The bending mode is represented by the second quantum number, Fig. 2.1c, while the third quantum number represents the asymmetric stretch mode, Fig. 2.1d. These mixed vibrational states are further subdivided into a series of levels due to the gross rotation of the vibrating molecule. These levels are also quantised and designated by J, the rotational quantum number.
Laser action is achieved when gas molecules in the laser cavity are excited to higher energy levels by collision with energetic electrons provided by an electric discharge\textsuperscript{(14),(15)}. Initially the energised CO\textsubscript{2} molecules spontaneously decay to lower energy levels emitting energy in the form of photons.

\[
1 \text{ energised CO}_2 = 1 \text{ lower energy state CO}_2 + \text{ photons (1 photon)}
\]

The emitted photon is then available for further collision with any excited CO\textsubscript{2} molecules in the discharge tube. Any such collision would result in a downward transition from the higher energy excited state to the ground state with the subsequent emission of a photon.

\[
1 \text{ energised CO}_2 + 1 \text{ photons} = 1 \text{ lower energy state CO}_2 + \text{ photons}
\]

This stimulated emission is the basis of the lasing action. The pair of photons produced by this process are of identical wavelength, spatially coherent, in phase, and travel in the same direction. In the early stages the photons are emitted at random and many are lost through the walls of the discharge tube, Fig. 2. 2b, and c. At some point in time B, Fig. 2. 2d, a particular photon may be emitted in such a way that its direction of propagation is perpendicular to the mirrors in the discharge tube. This photon is reflected back along the optical axis n, initiating the cascading effect of amplification by stimulated emission, Fig. 2. 2e. Not all the photons thus produced are parallel to the optical axis and available for further stimulation. However those favourably orientated stimulate the emission of more photons and lasing occurs.

In the non excited lasing medium. (in this case the CO\textsubscript{2} gas), the majority of particles are in the lower energy levels. In order to achieve optical gain in the
lasing reaction it is necessary that the inverse of this be the case. This is known as population inversion, as the population density of the particles in the higher energy levels exceeds that in the lower energy levels.

**Population inversion and the role of nitrogen in carbon dioxide lasers**

From thermodynamics it is known that, for a temperature $T$, the distribution of the population of gas molecules between the energy states $E$ and $E + \Delta E$ is given \(^{(15)}\) by

$$
P(E + \Delta E) = e^{E/RT}/P(E)
$$

Where $P =$ population of gas molecules. This allows for the lower energy levels having a greater population of molecules at a given temperature $T$. In the presence of an electric discharge it is possible to provide conditions such that the majority of molecules are in the higher energy metastable states.

In the pure CO$_2$ laser, the upper laser level is designated 001, and the lower laser levels 100 and 020, respectively, with the ground state being 000. Molecules decaying to the ground state do so via the 010 vibrational level. A clear illustration of this \(^{(14)}\) is given in Fig. 2.3. Most of the energy transfer between molecules within the system occurs by a resonant process, in the sense that there is a redistribution of energy between the molecules without any loss of the total internal energy by its conversion to kinetic or thermal energy. Hence it is possible both to excite the CO$_2$ to the required upper laser level efficiently by electron impact in a gas discharge, and for the molecules in the lower levels to decay to the intermediate 010 level. However the further decay of molecules from the 010 level to the ground state involves a non resonant energy transfer process via collisions of CO$_2$ molecules with foreign gas particles, the walls of the discharge tube, or other CO$_2$ molecules. This is a highly inefficient process and can lead to a drop in the efficiency and power of the laser.
Ideally the high quantum efficiency of the CO$_2$ system should allow a practical efficiency of 20 - 30% to be reached, but in the pure CO$_2$ laser system the efficiency only reaches 1%. This is because the electron impact excitation does not produce the required selective excitation of the CO$_2$ molecules to the upper laser levels. Such an improvement in the quantum efficiency of the CO$_2$ laser system is achieved through the introduction of nitrogen molecules into the laser cavity.

The role of nitrogen, (N$_2$) in the CO$_2$ laser cavity, is to improve the efficiency of excitation of the CO$_2$ molecules. N$_2$ being a homonuclear diatomic molecule does not decay radiatively, or through collisions when excited, so it is very long-lived and acts as an energy reservoir. The energy of excitation of the N$_2$ molecule to the $v = 1$ level, Fig. 2.3, is approximately equal to the energy of excitation of the CO$_2$ (001). Therefore, by transfer of energy through collision, the N$_2$ can raise the ground state CO$_2$ to the higher (001), or upper lasing level. The higher vibrational levels of the N$_2$ occur at about the same energy as the CO$_2$ (001) levels. Hence in collisions involving N$_2$ (v=1) and CO$_2$ (000) efficient vibrational energy transfer can take place. This results in efficient selective excitation of the CO$_2$ molecules to the upper laser levels and hence an increase in efficiency and output of the CO$_2$ - N$_2$ laser compared to the CO$_2$ laser.

In the lasing medium it is also important for the decay of the CO$_2$ molecules from the lower lasing levels to the ground state to occur in order to subsequently extract efficiently the energy stored in the excited N$_2$ molecules. This is known as depopulation.
Depopulation and the role of helium in the carbon dioxide laser

Increasing the rate of depopulation can be achieved by removing the 'bottle-neck' resulting from the inefficient non resonant energy transfer process in the decay of CO₂ (010) to CO₂ (000) previously mentioned above. By adding helium, (He), which has some 4000 de-exciting transitions per second at a pressure of 1 torr compared to 100 for CO₂ the efficiency of the CO₂ - N₂ interactions can be further increased(5). In a typical CO₂ gas laser, the three gases are usually present in volume concentration ratios of 0.8 CO₂ : 1 N₂ : 7 He(7).

Helium has also been found to increase the rate at which CO₂ molecules are excited to the (001) level by direct electron impact, and by increasing the excitation rate of vibrationally excited N₂. Helium is thermally excited by its interactions with CO₂, and at high discharge powers it becomes difficult to ensure that the gas temperature remains below about 500K, this is necessary to avoid excessive thermal population of the lower lasing levels, and thus a loss in efficiency. Helium however has a high thermal conductivity, and aids in the removal of heat from the laser system. Further heat disposal is adequately dealt with by incorporating heat exchangers into the laser system.

Heat disposal

Conventional CO₂ lasers basically consist of a water or oil cooled coaxial tube with mirrors at both ends through which the laser mixture flows and is excited by an electrical discharge. The water or oil is maintained at a constant temperature and heat transfer is by conduction. These are known as low axial flow lasers, (LAFL), and are used only where moderate powers are required, (below 1kW). Under optimised conditions powers up to 8Wm⁻¹ can be expected for short tubes, (< 3m),
and somewhat less than 50Wm\(^{-1}\) for long tubes. A representative diagram of a low axial flow laser is shown in Fig. 2.4.

In large industrial lasers where high power outputs are required, power is increased by having a long discharge tube which is folded, since the output is limited by the discharge tube length. The lasing action is kept aligned with the tube with the aid of appropriate mirrors. The longest and most powerful conventional CW CO\(_2\) laser was reported by Horrigan et al\(^{(16)}\), who developed a folded system with a total length of 750ft and capable of generating 8.8kW of CW power.

Another more effective approach to the heat disposal problem is convective cooling used in fast axial flow lasers, (FAFL). In these lasers gas is moved out of the laser cavity and through heat exchangers where it is cooled prior to being reintroduced into the laser cavity. Powers ranging from 20 - 27.2kW have been reported by Hill\(^{(17)}\), and Brown and Davis\(^{(18)}\). Fig. 2.5 shows a representative diagram of a convectively cooled fast axial flow laser. With the advent of these systems work has become centred on areas of high pressure, and high fluid flow dynamics as the physics of convective flow pointed to higher powers as mass flow rates increased.

**Laser beam modes**

To obtain maximum power density on a workpiece it is necessary for the laser output to be focused to a small diameter spot\(^{(19)}\). The ability to do this is determined by the optical quality of the beam, in other words the beam mode.

The laser beam mode depends on the configurations of the electromagnetic fields allowed in the laser cavity. The shape of the laser cavity itself in the direction transverse to the optical axis provides the boundary conditions governing the
presence of these electromagnetic fields in the laser cavity. The allowed laser beam
modes are known as transverse electromagnetic modes, (TEM). Individual modes
are distinguished by two subscripts, m and n, where m and n are the number of
nodes in two orthogonal directions of the beam cross section. e.g TEM$_{mn}$, where
mn can be 00, 01, 11, 12, and so on respectively, Fig. 2.6.

The TEM$_{00}$ or Gaussian mode has the highest degree of symmetry, and is the
optimum beam mode as can be seen from fig. 2.7. The other modes are known as
higher order modes. The origin of these different modes can be traced back to the
mirrors in the laser cavity. In the cavity there are two mirrors, one a
partially-transmitting mirror, and the other a shallow hemisphere. The properties
of the hemispherical mirror are such that when photons are normally incident on
this mirror, they are reflected along a line as shown in Fig. 2.8. This gives rise on
emergence from the laser cavity to a Gaussian energy distribution. If however
either one of the mirrors in the optical cavity is tilted slightly off-axis, destructive
interference occurs giving rise to the nodes characteristic of the higher order TEM
modes.

The type of metal working operation determines how the laser will be set up.
According to Ramos et al(20), for welding, the desired beam characteristics are
limited energy density to avoid vaporisation, and a higher order mode with good
symmetry. The disadvantages of operation in the higher order modes include beam
divergence, and thus lower efficiencies of energy transfer. Also instabilities in
output occur as the beam oscillates from one output mode to another.
Types of carbon dioxide lasers

Carbon dioxide lasers can be placed into four broad categories according to the arrangement of the electrical excitation unit, and the gas flow direction relative to the optical axis as follows.

The axial system. Here the electrical field and the gas flow in the same direction and are coaxial with the optical axis.

Transverse gas flow system. Here the electrical field is coaxial with the optical axis, but perpendicular to the gas flow.

Transverse discharge system. Here the electrical field and gas flow in the same direction, but are perpendicular to the optical axis.

Transverse gas and transverse discharge. Here the electrical field, gas flow axis, and optical axis are mutually orthogonal.

Methods of operation

There are two operating modes as described below.

Continuous wave

During continuous wave operation of a CO₂ gas laser, population inversion is permanently maintained in the system and stimulated emission prevails continuously. Cooling is carried out in order for efficient extraction of the energy stored in the nitrogen molecules to be maintained.
Pulsed operation

If the electrical excitation is supplied in the form of short pulses with an average length $t$, during the time interval $dt$ when the pulse is switched off decay of spontaneous emission occurs. Thus the laser only delivers an output when the electrical excitation is switched on. This pulsing can be obtained by mechanically switching the output beam of a CW laser, by modulating or pulsing the electrical or pumping system, or by electro-optically switching the optical cavity. When used in the pulsed mode any awkward cooling requirements are removed as the laser is running cold. It is also possible to reach higher peak powers than in CW operation as the 'cold' laser is more efficient, due to the electrical surge when the lasing arc is struck. Peak powers reached are determined by the pulse shape, pulse width, and the pulsing rate.

Typical values for pulsing a normally 5kW CW machine are\(^{(21)}\) power = $10^5$W peak, 100W average; pulse length = 10msec - 100msec; pulse rate = up to 100pps.

2.2 Laser material interactions and the laser processing of materials

The first part of this subsection is concerned with the fundamental aspects of laser material interactions, and the properties of materials which affect these interactions. The second part is a brief review of materials processing with particular reference to known studies of laser welding.

2.2.1 Laser material interactions

Traditional methods of processing materials have depended primarily on bringing various forms of energy, (mechanical, electrical, or chemical), into contact with the
workpiece. In laser materials processing there is no intimate contact between the energy source and the workpiece. This is true also of the electron beam and some ultrasonic devices. The laser differs from these devices as a consequence of its directionality and coherence. The directionality allows it to be focused with the aid of a lens to form a spot with a theoretical diameter of little more than one wavelength\(^{(21)}\)

The interaction of intense laser light with matter is a complex process, but the basic and dominant physical mechanism is rapid heating. When the laser beam impinges upon the material, it is absorbed by interaction with the electrons which respond by being accelerated, (raised to a higher energy state). These electrons are now in a position to re-radiate their kinetic energy, (reflection), or transfer it to the material via collisions with lattice phonons, other electrons, ionised impurities, and defect structures. Von Allmen\(^{(21)}\) reports the general features of linear interaction of light with solid matter by modelling the solid as a set of harmonic oscillators under the action of an external, periodically changing force field, (Lorentz model).

**Laser beam characteristics**

Of particular importance in determining the reaction of laser radiation with a substrate, are the laser beam characteristics which are discussed briefly as follows.

**Energy, power, and intensity**

The laser beam output power, and traverse speed, are two of the more basic parameters commonly used to characterise laser material interactions, as they determine the rate at which energy is delivered to the substrate. The intensity of a beam of radiation on a metal surface can be described by the following compound parameter.
\[
I = \frac{P}{vD}
\]  \hspace{1cm} (2.1)

where \( P \) = laser beam power; \( v \) = relative velocity of laser spot and material; \( D \) = incident spot diameter; \( I \) = intensity of the incident radiation.

The depth of penetration during laser welding has been found to be directly related to the power density of the laser beam and is a function of the incident beam power and the beam diameter. Generally for a constant beam diameter, penetration increases with increase in laser beam power. Locke et al(2) and Baardsen et al(23) report that penetration increases almost linearly with incident laser power. It has also been observed that a minimum threshold power is required to initiate welding in a material of given thickness. Banas(24) studied the effect of power up to 77kW on butt weld penetration in 304 stainless steel, reporting that the maximum penetration observed was directly proportional to power\(^{0.7}\).

**Laser beam diameter**

From equation (2.1), it is evident that the intensity at the point on the substrate irradiated by the laser beam is a function of the area irradiated by the beam at this point, (\( D \) in equation 2.1). For laser beams with a large degree of coherence, there is still a measurable degree of divergence, which can be represented for a beam of Gaussian distribution by the following equation;

\[
\theta = \frac{2\lambda}{\pi D_0}
\]  \hspace{1cm} (2.2)

where \( \theta \) = full angle divergence; \( \lambda \) = wavelength; \( D_0 \) = diameter of the beam leaving the laser cavity.
Focusing the output beam of the laser is required for the majority of applications except perhaps heat treatment. This is because the unfocused output of the laser generally does not provide sufficient power density to raise the temperature of materials above their melting or boiling points. Also the diameter, (typically at least a few millimetres), is too large for applications requiring the heat affected zones to be confined to a few microns.

Fig. 2.9 diagramatically represents the focusing of a spatially coherent laser beam by a simple lens. The focused beam diameter, $D_f$ is directly proportional to the laser wavelength $\lambda$, the lens focal length $f$, and inversely proportional to the unfocused beam diameter $D_0$.

$$D_f = \frac{2\lambda f}{\pi D_0} \quad (2.3)$$

Simplified this becomes $D_f = f\theta$, where $\theta$ is the angle of far field divergence which is a direct consequence of the measurable degree of divergence in laser systems. The above equation predicts the minimum spot size at the focal plane of the lens. Thus it can be seen that the focused beam diameter is both lens focal length and wavelength dependent.

Measuring the focused beam spot diameter of high power lasers poses certain difficulties. This is partly due to the nature of the beam diameter and partly due to the definition of what is to be measured. A Gaussian beam diameter may be defined as the diameter at which the power of the laser beam has dropped to $1/e^2$ or $1/e$ of the central value. The beam diameter defined on the basis of $1/e^2$ of the central value contains more than 80% of the total power of the beam, whereas the power defined by $1/e$ is only slightly over 60%\(^{(26)}\).
Many techniques have been employed to measure beam diameters, and these include charring paper, and drilling acrylic or metal plates. These are however unsatisfactory as they are both power and time dependent. The laser beam analyser provides a better measurement of beam diameter based on a fixed relationship between the time of passage of a thin highly reflective wire rotated in a plane perpendicular to the beam, and the width of the beam as shown on an oscilloscope. The time axis on the oscilloscope corresponds to the passage of the wire through the beam, as the speed of the wire is known the width of the beam can be determined\(^{(26)}\).

**Beam traverse speed**

Generally it has been observed that the penetration depth achieved in laser beam welding is inversely proportional to the beam traverse speed. Fig. 2.10 shows that for a particular power welding can be performed over a range of thicknesses. Beyond that range full penetration welding is not possible. Higher speeds may lead to incomplete penetration, and lower speeds to excessive melting, loss of metal, and weld perforation. The range of welding speeds, i.e the range between maximum and minimum welding speed for successful welding decreases with increase in material thickness\(^{(27)}\).

Of particular significance for processing applications is the fact that there exists a finite region of length \(d\), over which the focused spot remains fairly constant. The parameter, \(d\), is generally described as the depth of focus. The size of the focused spot and the depth of focus to a first approximation vary in direct relation to one another. Thus an extremely small focused spot is achieved at the expense of a small depth of focus. The particular application of the laser will dictate a compromise between the two. Control over lens focal length, and unfocused beam diameter provide a means of achieving the desired result.
The intensity of coherent radiation focused by a simple lens can thus be described as:

\[ I_f = \frac{P}{\pi(D_0^2/2)} \]  

(2.4)

Where \( I_f \) is the intensity at \( D_f \), and is assumed to be uniform throughout the focused spot\(^{25}\).

Effective spot size

\( D_f = f \theta \), (equation 2.4), defines only the optical spot diameter. This does not account for the laser material interaction phenomena at the material surface as the spot effected on the material is almost without exception larger than the optical spot size. This is largely due to the thermal properties of the material, but the mode structure of the laser beam also has some influence on the effective spot size.

Experiments on bulk phosphor bronze\(^{25}\), showed that the effective spot size is more than twice the optical spot size, Fig. 2.11. The effective spot size is conveniently described as the area of the resultant heat affected zone within the material being processed.

These parameters described, all have an effect on the intensity of the laser radiation incident on the substrate surface. As \( P/VD \) increases the effect of the laser on the material increases, and this is what governs the boundaries between the three basic laser material interaction regimes. A guideline for metals is as follows\(^{21}\).

(a) Power density \( 10^5 \text{ Wcm}^{-2} \) - heating without phase change.
(b) Power density \( 10^5 < 10^7 \text{ Wcm}^{-2} \) - heating with solid/liquid phase change.
(c) Power density >10^7 Wcm^{-2} - solid/liquid phase change with plasma formation.

For the purpose of this work it is the regime 10^5 < I > 10^7 that is of importance. A brief survey of all the regimes is given in reference\(^{(21)}\).

When sufficient heat is supplied to the material its temperature increases until it reaches the melting point. With the addition of the heat of fusion the solid is transformed to the liquid state. Welding occurs when the liquid formed at the junction of two separate pieces of material is allowed to solidify\(^{(25)}\).

Material properties

Several intrinsic material properties determine the ease with which melting and hence welding occur. These are as follows.

(a) Properties which govern the flow of heat in the material. In general materials with a high thermal diffusivity accept and conduct thermal energy readily. This favours weldability.

(b) Properties which relate the amount of energy required to cause a desired phase change. These include density, specific heat, and latent heat effects.

(c) Surface reflectivity with respect to the laser radiation.

Of these the reflectivity is of critical importance in laser welding. Many materials show high reflectivities at certain laser wavelengths. Table (2.2) shows the reflectivity of several metals at a 10.6\(\mu\) laser wavelength for the purpose of comparison\(^{(18)}\). It is obvious that only a small amount of the incident CO\(_2\) beam energy is available for melting. The infrared absorption of metals largely depends on the absorption of radiation by free electrons. Arata et al\(^{(29)}\) measured the absorptivity of polished surfaces of several metals and concluded that absorptivity is proportional to the square root of the electrical resistivity, closely agreeing with the formula;
where $A = \text{absorptivity}$, and $r = \text{electrical resistivity}$. Bramson derived a temperature dependent relationship between the electrical resistivity and emissivity of metals\(^4\).

$$E_\lambda(T) = 11.2(r_{20}(1+\gamma T))^{0.5} - 62.9(r_{20}(1-\gamma T)) + 174(r_{20}(1-\gamma T))^{1.5}$$

$E_\lambda$ = emissivity at a given wavelength $\lambda$, $r_{20}$ = resistivity at 20°C, $\gamma$ = coefficient of resistivity change with temperature, and $T$ = temperature. This formula is valid for the calculation of absorptivity as absorptivity equals emissivity. However, the formula would only be valid for metals heated in a vacuum without a surface oxide layer. The Drude electron model provides an alternative to Bramson's equation for the calculation of absorptivity, and is based on the free electron model of metals\(^3\).

However, it has been established that the surfaces of most metals reflect light during only a small portion of a laser pulse, and only during the initial part of a continuous irradiation. As the temperature of the surface increases, the reflectivity decreases with a corresponding increase in absorptivity. For metals at their boiling points, reflectivity is generally assumed to be zero with the metal behaving like a black body absorber due to the highly absorptive plasma, (ionised metal vapour), present above the melt\(^3\), and within the keyhole.

The reflectance of most metals also increases with wavelength, hence more power would be required when processing with a laser of long wavelength than with one of a shorter wavelength. For example, comparing the Nd:YAG laser with a wavelength of 1.06\(\mu\)m to the CO\(_2\) laser with a wavelength of 10.6\(\mu\)m, 1 - $R$ for steel at 1.06\(\mu\)m is equal to 0.35, compared to the value of 0.05 at 10.6\(\mu\)m\(^4\).

Hence steel would be more easily processed using the Nd:YAG laser. The ease with which various metals can be processed with respect to each other using a laser of given wavelength also depends on their $R$ values. This is demonstrated by copper and steel for a wavelength of 10.6\(\mu\)m. The 1 - $R$ value for copper being 0.02, and that for steel 0.05 \(^4\).
2.2.2 Laser welding

Welding is one major fabrication process where the laser has found a relatively active field of application. Welding is a process of uniting two materials by heating them to a molten state and fusing them together. The laser's ability to produce a coherent, highly directional beam of light which can be focused to an extremely small spot, by means of a simple lens, and be projected over long distances without losing its power unduly, makes it, given a high enough power output, a source of intense localised power which can be used for melting and fusing materials.

When a laser beam is focused on to a material surface the beam is partly reflected and partly absorbed. The absorbed heat penetrates the material via thermal conduction, and should the power density be high enough a molten pool would result at the point of focus which on solidifying would form a weld.

At low laser powers the surface heating of the material relies totally on its heat conduction properties, and as such any surface penetration is usually limited to less than 2mm\(^{(32)}\). This is the basis for one of the main differences between laser and electron beam welding. However with multi-kilowatt CO\(_2\) lasers the coupling of the laser radiation to the workpiece occurs via plasma absorption within a void known as a 'keyhole' formed by the impinging laser beam. This slightly modified technique of welding is known as 'keyholing', and allows deep penetration welds to be achieved.

During laser welding the impinging beam heats the substrate past its boiling point, resulting in the formation of a vaporised cavity or void, more commonly referred to as a 'keyhole', Fig 2.12. This vaporised cavity contains plasma, (ionised
metallic gas), which is an effective absorber, trapping up to 95% of the laser energy within the 'keyhole', which in effect behaves as a black body absorber. The conditions of energy and material flow during laser welding were investigated by Klemens(33). According to Klemens the 'keyhole' or cavity is formed only if the beam has a sufficient power density. Arata et al(34) observed the behaviour of the 'keyhole' in steel during laser welding. It was found that in the absence of an assist gas, i.e, a gas flowing coaxially with the laser beam on to the workpiece surface the beam hole is unstable, and as the gas flow rate increases the 'keyhole' becomes more stable. There was an optimum gas flow rate for maximising the depth of the 'keyhole', and hence weld penetration. Beyond this optimum value, the 'keyhole' opening lengthens but the depth does not increase. Temperatures within the 'keyhole' can reach 25,000°C(32). Heat transfer then takes place by conduction from the 'keyhole' directly to the thickness of the material, rather than from the surface inwards as occurs with conventional welding techniques. As the beam moves relative to the material, molten metal fills in behind the 'keyhole' and solidifies to yield a deep penetration weld. Any material lost by vaporisation shows up as a depression in the solidified weld bead, or porosity in the weld.

Due to the narrow profile of the focused spot, there is a relatively small amount of molten material generated in the welds, and little time for this material to flow. Therefore the process always requires a close joint fit-up(35). The high temperatures involved necessitate adequate shielding of the laser welds. It is reported(36) that in laser welding a good shielding gas should possess the following qualities; good transmission, and good blanketing capabilities.
Transmission

This is the ability to permit energy from the beam to reach the substrate in a concentrated form, as the laser beam must traverse the atmosphere above the substrate. Several gases have been investigated as shielding media, including argon, helium, oxygen, carbon dioxide, and a mixture of hydrogen and helium. The results of this research has produced conflicting reports as to whether helium is more effective as a shielding gas than argon. It is however true to say that the basic phenomenon involved is that of a plasma effect, with the shielding gas being used to produce a plasma. In the low intensity regime of laser processing the laser induced plasma helps to increase the coupling of the laser energy to the workpiece via inverse Bremsstrahlung, i.e. the absorption of a photon by an electron leading to a transition of the electron to a more energetic state, thus enhancing process efficiency. In the high intensity regime the laser induced plasma is said to shield the workpiece.[37]

Work reported by Seaman[36] suggests that the substrate appears to contribute to the plasma formation in the shielding gas. Observing the focal point of a beam in air showed no plasma formation, but when used in welding 50% less penetration was obtained compared to when helium was used as a shield gas. It is believed that vapour from the 'keyhole' in the substrate may add significant energy to the shield gas and produce a plasma reaction at marginal levels of beam intensity, i.e., levels at which this phenomenon would not normally occur. Several studies have been carried out into the effect of laser generated plasma on welding, and details of these can be found in work by Herziger[38], Moon[39], and Dixon and Lewis[40].

Skripchenko et al[41] reporting on the study of penetration welding with a laser, state that the upper part of the channel formed in laser welding with complete penetration is considerably greater than the diameter of the laser beam. This was
observed to be due to the fact that the upper part of the channel is heated by the laser generated plasma, (ionised metal vapour), whereas the lower part of the weld channel is heated by the impinging laser beam after passage through the plasma. The power of the radiation which passes through the substrate and out of the bottom of the weld was said to depend on the thickness of the substrate at \( d/h \) greater than or less than 0.4, where \( d \) is the thickness of the material, and \( h \) is the weld depth, (Figs. 2.13 and 2.14). In the case of thin specimens Fig. 2.13b it is suggested that the heat source is represented by the plasma and radiation does not impinge on the surface of the penetration channel. Similarly at \( d/h > 0.4 \) the radiation passing through the plasma directly heats the walls of the vapour gas channel, and is partially absorbed by these walls. In this case the metal is heated by two heat sources. A plasma source with a large effective diameter predominates in the upper part of the weld, whereas a beam source whose diameter is comparable with that of the laser beam operates in the lower part of the weld channel. This accounts for the characteristic "wine-glass" shape of laser welds. Skripchenko et al also plotted a graph of \( B_2/B_1 \), where \( B_2 \) is the reverse bead, and \( B_1 \) is the surface bead, against \( P_2/P_1 \), where \( P_2 \) is the transmitted radiation, and \( P_1 \) the incident radiation, Fig. 2.14. The authors suggest that the working point should be in region II, characterised by a weak dependence of \( B_2/B_1 \) on \( P_2/P_1 \). In region I there is a strong dependence of \( B_2/B_1 \) on the power of the transmitted radiation, while in region III the power of the transmitted radiation equals approximately 50% of the initial power. It was also reported that by monitoring and varying the power of the transmitted radiation, the process of complete penetration could be stabilised, and the shape of the penetration zone predicted and controlled, Fig. 2.15. Numerous investigators have carried out high energy laser welding studies incorporating plasma control as a means of increasing process efficiency. A number of these will be reviewed briefly in the following paragraphs.
Work by Davis et al(42) concentrated on the use of plasma control to improve the efficiency of welding of 316 and 321 austenitic stainless steel. This involved the use of a high velocity jet of helium directed at the beam-workpiece intersection point, Fig. 2.16. The optimum configuration was found to be a 1.25mm diameter jet aligned in the joint plane, and pointing in the direction of welding at an angle between 25 and 50° depending on the material thickness; thicker materials requiring larger angles. Flow rates were 5 to 10 l/min and had to be closely controlled. The use of this device is reported to have resulted in an increase in penetration from 10mm to 13mm, and the welding speed was increased by 30%.

Moon(39) reports the use of an alternative plasma suppression device which consists of a nozzle through which helium gas was used to blow the plasma from the workpiece surface. The relationship between penetration and nozzle position, and the effect of nozzle size on penetration was established. The weld penetration was also determined as a function of the gap between the shield and the workpiece at a given laser power level. Maximum penetration was achieved at 1.6mm, the closest distance of the nozzle from the beam. Penetration decreased as this distance was increased. Optimum nozzle size was found to be 2mm.

Other work on the effect of assist gas in suppressing laser generated plasmas are reported by Arata et al(43), and Crafer(44). Arata et al utilised a device similar to that of Davis et al(42) inclined at an angle of 40° to the horizontal. The role of the assist gas was determined by measuring the assist gas pressure, and the vapour pressure within the 'Keyhole', as well as by high speed photography. They observed that under optimised gas assisting conditions the use of helium, argon, nitrogen, and carbon dioxide resulted in no appreciable differences in penetration. Crafer(44) reported an increase in weld penetration from 5mm to 6.5mm in the
welding of stainless steel with a 2kW laser. The author also states that while in the welding of thin gauge materials, 2mm to 4mm the plasma control device is not necessary to achieve full penetration, it has a considerable effect on weld profile, reducing root porosity and resulting in welds with parallel sides. Further work on the role of plasma control in decreasing porosity in laser welds was reported by Estill(45) in a study on 304 stainless steel. Plasma control was obtained by using a helium cross jet ranging in pressure up to 175kPa through a 1.23mm diameter orifice. The helium jet was inclined at 45° to the horizontal. Microporosity and root porosity was observed to decrease with plasma control. The author however reports difficulties in reproducing the data, and suggests careful statistical analysis of the laser parameters involved.

Blanketing

This is the ability to displace air from above the weld rapidly before the laser beam melts the area and solidification occurs. This is important in laser welding as speeds are high, and there is little time to displace air from the weld zone. Seaman(36) reports that the denser the gas the better its blanketing properties. Argon should therefore exhibit better blanketing properties than helium.

The choice of shielding gas depends to a large extent on the end use of the welded part. In cases where oxidation is not a problem the blanketing properties of a gas would not be of great importance, and cost of the gas could be a deciding factor.

In the following sub-section previous work on the laser welding of metals is reviewed. Details of the compositions of the alloys considered in this review are contained in appendix 1.
Previous work on laser welding

The effect of dynamic beam focusing on penetration was studied by Ivanov et al(46). They showed that the penetration of 12Kh18N1OT corrosion resisting steel, PT-3V titanium, and AMg6 aluminium alloys may be increased by 30 - 40%, and is accompanied by a reduction in the degree of variation of penetration if dynamic beam focusing is used. They report an optical scanner being developed for this purpose. The position of the focus was oscillated as a result of a periodic displacement of the focusing lens along the vertical axis with specific frequency and amplitude, (frequency range; 0 - 15 Hz, and amplitude; 0 - 3 x 10^-3 m) as shown in Table (2.3).

Work by Bashenko et al(47) on 08Kh18N10T steel specimens coated with potassium nitrate demonstrated the effect of coatings and edge preparation on resultant weld penetration. It was showed that the penetration increased, and the shape of the welds improved. The width of the weld was also found to decrease in specimens coated with water glass, ferric chloride, and FS-71 flux. Preliminary edge preparation was shown to improve weld formation, and increase the penetration. Each particular mode structure was reported to have its own optimum type of edge preparation, with the angle tending to the angle of beam convergence.

The relatively low thermal efficiency of the laser welding process and the associated low penetration capacity of the beam is the main disadvantage of the process. This can be overcome to some extent by the 'keyhole' effect. In electron beam welding the electrons are not subject to absorption and reflection as is the laser beam, and enhanced penetration is possible compared to that achieved in laser beam welding. A quantitative comparison of laser and electron beam welding has been summarised by Ball and Banas(48). Broader based comparisons between laser beam welding
and conventional welding processes have been made and examples of these can be found in papers by Willgoss et al\(^{(49)}\), and Darchuck and Migliore\(^{(32)}\). Despite its limitations laser beam welding has certain distinct advantages which allow it to compete favourably with well established processes.

**Advantages of laser beam welding**

The advantages of laser beam welding over conventional welding processes according to Charschan\(^{(50)}\), and Mazumder\(^{(51)}\), are as follows.

1. High processing speeds.
2. High energy density.
3. Welding can be performed in many environments, e.g under atmospheric pressure or in a vacuum.
4. Distortion and shrinkage in welded structures is negligible.
5. It is not necessary to restrain the workpiece during welding.
6. Narrow welds are usually possible with a high degree of accuracy.
7. Welds with little or no contamination can be produced in materials highly susceptible to oxidation.
8. The heat affected zone adjacent to the weld is very narrow.
9. Using light deflection techniques intricate shapes can be welded at high speed.
10. The laser system can be readily adapted for close circuit TV viewing to ensure the quality of welds.
11. Dissimilar metals can be welded.

The ability of a laser to weld a material is related to the effectiveness of the interaction between input energy and material.
Factors affecting the laser weldability of materials

The laser weldability of materials is subject to all the usual metallurgical constraints on weldability, e.g. carbon equivalent etc. In this section however, the factors considered are laser specific factors as they are of major importance for the purpose of this project.

Reflectivity

This parameter is critical to laser welding when compared to other welding practices as essentially a light beam is being used to process the material. However it only applies during the initiation of the weld, when the cold metal without any surface preparation reflects over 90%, and in some cases up to 99% of the incident laser beam energy. Once melting has been initiated reflectivity drops to approximately zero and welding can proceed unhindered.

Thermal diffusivity

This is defined as $k/cp$, where $k$ = thermal conductivity; $c$ = specific heat capacity; and $\rho$ = density. Thermal diffusivity is an inherent material property. For laser welding the value is required to be low so that the energy of the beam is concentrated into a small area, thus making it more efficient.

Melting point

High melting point materials such as tungsten are more difficult to laser weld than low melting point ones like lead. This is because, to initiate absorption and the
'keyhole' mechanism, the melting point must be reached during the initial stages of heat transfer, or a lot of energy would be lost by conduction into the surrounding material, thus decreasing the efficiency of the process.

**Boiling point**

When welding materials with low boiling points large volumes of vapour are produced. Severe porosity in welds on materials like zinc and magnesium is due to the melt puddle disruption and ejection of liquid metal from the fusion zone by the vapour.

**Viscosity**

High viscosity tends to inhibit the escape of bubbles formed due to the incomplete collapse of the "keyhole" behind the moving laser beam, thus leading to porosity in deep penetration welds.

**Previous work on the laser welding of metals**

In this section it is proposed to review the available literature on laser welding with reference to the specific materials studied. Particularly work on nickel base alloys, aluminium base alloys, and aluminium based metal matrix composites will be considered.

Since the demonstration of penetration welding by Brown and Banas\(^{(52)}\) using a multi-kilowatt CW CO\(_2\) laser the scope of technically and commercially feasible laser welding applications has increased. A large variety of metals and alloys have
been welded, e.g ferrous alloys, nickel alloys, titanium alloys, and aluminium alloys. Most of the initial parametric laser welding research was carried out on stainless steels due to the importance of these materials in the power plant and chemical industries. A series of studies were carried out by Banas\(^{53}\) on stainless steels. Using a high powered laser and the 300 series stainless steels the author concluded that the weld penetration achieved at a constant laser power did not vary significantly with the welding speed. This is however a very surprising result, as most other parametric studies involving beam traverse speed have revealed that weld penetration decreased as the laser beam traverse speed was increased at a constant laser power. Studies on 304 stainless steel revealed that the maximum penetration increased as the laser power was increased. Tensile tests also showed that the joint strength can equal that of the parent metal if the appropriate welding parameters are selected.

The solidification microstructures of continuous wave carbon dioxide laser welds were studied in AISI 310S, and type 304 austenitic stainless steel by Katayama and Matsunawa\(^{54}\). Metzbower et al\(^{55}\) carried out an in depth study on the laser beam weldments in A36, HY130, and HY80 stainless steel. Properties considered were the mechanical properties which were determined by measuring the Charpy energy, and the dynamic tear energy. The microstructures in the weld region were identified, and the effect of stress relief treatments on mechanical properties investigated. Stress corrosion cracking resistance in HY130 was also investigated. In another study on high strength low alloy steels, (A633, A710/736, and A737), Denney and Metzbower\(^{56}\) studied the effect of laser welding on the yield stress, ultimate tensile strength, elongation, and fracture toughness of the base plate and weldments in these materials. An attempt was made to relate the observed microstructure and mechanical properties of the joint to the rapid solidification rates
associated with laser welding, as the properties of HSLA steels depend in most cases on the precipitates formed by the alloying elements during cooling. Some degradation of mechanical properties especially toughness of the weld metal was observed, however in some cases this decline in properties in the weld zone could be offset by alloy additions to this region during welding. The laser welds were reported to exhibit a comparable toughness to GMA, (gas metal arc) welds, the authors suggest that this indicates that the rapid cooling rates encountered in laser welding were not a negative factor in determining weld toughness.

Other studies on ferrous alloys include the evaluation of laser welding of rimmed steels for the automotive industry by Baardsen et al(24). High speed laser welding of tin plate, and tin free steel used in can making was demonstrated by Mazumder and Steen(57). Laser welding was found to be the only way of welding tin-free steel without auxiliary preparation. Arc augmentation of the welding process improved the efficiency in terms of welding speeds. Metzbower(55), and Banas(52) evaluated results in HY series low alloy steels, and Seaman and Hella(59) reported the welding of modified 4340 alloy. Several other alloys studied include X-80 arctic pipeline steel(59), tank construction steels D6AC(60), and HSLA steels(61).

Mazumder and Steen(62) studied the relationship between the laser welding parameters and the metallurgical and mechanical properties of laser welded Ti6Al4V. Ti6Al4V showed no cracks, porosity, or inclusions on welding with a laser, (Seaman(63), Banas(64), Mazumder and Steen(5')). Further work by Mazumder and Steen(62) involving tensile tests show that the strength of the laser weld was comparable to that of the parent metal, Table (2.4). Adams(65) reported that laser welds could be made that exhibited the same fatigue characteristics as the base material. A comparative study of electron beam, laser beam, and plasma arc
welds in Ti6Al4V was carried out by Banas(64). Welds produced by all three methods had tensile strengths equivalent to those of the base metal, but the laser welds were reported to be narrower and more uniform than the arc welds or electron beam welds.

The welding of titanium alloy 6211 was investigated by Fraser and Metzbower(66). Comparing the properties of laser weldments to those of the base metal, GMA, or GTA, (gas tungsten arc), welds, they were found to have equal or in some cases superior properties. Toughness and stress corrosion cracking resistance were found to be good. This could be attributed to the fact that oxygen and nitrogen contamination were low due to the high energy, low heat input nature of the laser welding process. Microstructure was predominantly equiaxed and fracture occurred by void coalescence.

Peng, Sastry, and O'Neal(67) also carried out extensive studies on Ti6Al4V, Ti-Er, and Ti-Y. Details of their investigations can be found in references (67), and (68). They concluded that the cooling rate for Ti6Al4V was 10² -10⁴°C/s due to the predominantly martensitic structure produced. The weld penetration dependence on laser power and traverse speed for Ti6Al4V is shown in Fig. 2.17.

Mazumder(69) in a review of welding studies on titanium and its alloys reports that X-ray radiographs of successful laser butt welds of Ti6Al4V and commercially pure Ti showed no cracks porosity or inclusions. Similar observations were reported by Seaman and Banas(64). The microstructural variations arising from laser welding of Ti6Al4V has also been extensively studied with the aid of optical, scanning, and transmission electron microscopy. More details of these microstructures can be obtained from a review by Mazumder(69).
Moon and Fraser(70) studied titanium alloy Ti6Al2Cb1Ta0.8Mo. Mechanical properties, fracture toughness, and stress corrosion cracking resistance for the alloy were determined. The results obtained were found to be satisfactory, and equal or better than base metal, GMA, or GTA weldment properties. The study extended to the microstructural variations in the joint which were characterised. The authors suggest titanium alloys be protected from atmospheric contamination during welding.

Darchuck(32) reports the welding of tantalum and zirconium with the main problem being that of oxidation. Welding in a vacuum or the use of adequate shielding gas was reported to have resulted in acceptable welds. Mara and Murphy(71) report the welding of uranium and uranium alloys, while Faulkner et al(72) have reported on the laser welding of beryllium in a deuterium atmosphere. The possibility of welding copper alloys once the problem of limited absorption of the laser energy had been overcome was also reported by Darchuck(32), except in the case of the brasses which are regarded as unweldable due to their high zinc content.

Nickel alloys are generally regarded as being weldable by laser. These alloys produce good coupling of the laser beam due to their low thermal diffusivities, and are not susceptible to oxidation. However most of the nickel alloys are said to show porosity in deep welds as a result of high melt viscosities. Hot shortness has also proved to be a problem in some cases. The laser welding of Ni alloys has been investigated by Duhamel and Banas(73), and Russell et al(74)(75). The Ni alloys were said demonstrate good responses to laser welding, and exhibit acceptable mechanical properties.
Mehta et al(76) carried out an investigation into the use of laser welding for butt joining of Inco 718 vane pairs currently being performed by gas tungsten arc welding. The process parameters chosen for study were power, beam traverse speed, lens focal length, shield gas, and beam focus characteristics. Successful welds were made under laboratory conditions. The development of an automated laser beam welding facility for high volume production of 0.2mm thick Inconel 625 recuperator was also reported by Miller and Chevalier(77). The automated laser beam welding facility was found to be suitable for high volume production applications, and the laser beam welds were observed to be of a similar structure to that of resistance welds used for this application.

A parametric study on the laser welding of Nimonic C263 and Jethete M152 was carried out by Russell et al(74). Process variables examined were laser beam traverse speed, power, intensity, shield gas composition, and the use of a high speed jet to disrupt the plasma on the workpiece surface. The mechanical properties in 1.6mm thick C263 alloy were reported to be comparable to those of electron beam and tungsten inert gas welds, measurements of fatigue and tensile properties were carried out at room temperature, 750, and 850°C.

While there are reports on work carried out on In 718, Nimonic C263 and Jethete M152, it is unclear as to the detail of these works in examining the microstructural implications of the welding process on these alloys. This is surprising as the mechanical properties measured are a direct implication of the weld region microstructure. The fast cooling rates encountered in laser welding is characteristic of the process, and certainly its effect on the resultant microstructure in the weld region has been studied in titanium and its alloys, as well as a number of stainless steels, and other ferrous alloys. Russell et al report on the mechanical properties of
Nimonic C263, it is believed that an in depth study of the microstructural response of this alloy to the laser welding process would aid in the understanding of the development of these mechanical properties, and allow the laser welding parameters to be manipulated to give the desired joint properties. No previous work on the laser welding of Nimonic PE11 was found in the course of this survey.

Aluminium and its alloys have proved to be very difficult to weld due to their high initial surface reflectivity for the 10.6μm radiation from the CO₂ laser and high thermal conductivity. Snow et al(78) carried out an investigation into the welding of Al alloys 5456, and 5086. They concluded that Al alloys were very sensitive to the intensity of input energy and the welding variables. Problems were encountered in the form of excessive porosity and 'drop-through' of the bead in full penetration welds. This was thought to be related to liquid metal viscosity and the interaction of the shielding gas or plasma with the beam and the workpiece.

Snow and Breinan(79) reported the successful welding of Al alloys 2219, and 5456 up to 9.5mm, although porosity remains a problem. Tensile tests revealed weld properties close to that of the parent metal. Metzbower and Moon(79) studied the mechanical properties of laser welded Al alloy 5456. Tensile tests were performed on a standard ASTM round specimen 6.4mm in diameter. All the welds were reported to have failed in the weld region. The ductility of the specimens was low, and the amount of porosity visible on the fracture surface was high. This is contrary to results obtained by Breinan et al(83) who recorded an average strength of 342MPa for weld metal compared to a UTS of 345MPa for the parent material.

Further work on the laser welding of Al 5456 was carried out by Huntington and Eagar(81). The effects of surface preparation and joint geometry on laser power
absorption was considered. The as-received surface of Al 5456 was mill finished, other samples were sandblasted with mesh glass beads, anodised to 1mm thickness, or electropolished in an alcohol-water-perchloric acid solution. They reported initial absorption varying from a few percents of the incident radiation to 25% depending on the surface preparation, with the fraction of absorbed power increasing dramatically upon formation of a 'keyhole'. As a result welds made with a sharp 'bevel-groove' preparation were larger and more uniform than those made with either bead on plate, or 'square-groove' preparations. Metzbower and Moon(79) suggest that the 'keyhole' phenomenon of laser beam welding in an alloy in which there is a considerable difference in melting and boiling points of the constituent alloys could lead to vaporisation of the low boiling point elements. Electron microprobe analysis, (EPMA), of the cross sections of laser beam welds was said to reveal a decrease in magnesium content from the base plate to the weld centre of about 0.5%, a decrease which the authors report has been shown to decrease yield stress. Metzbower and Moon(79) also studied the microstructures of the base plate, heat affected zone, and fusion zone in Al 5456.

Mazumder(69) observed that alloy 5182 was easier to weld than 2036 and 6009. In evaluating this observation laser irradiation parameters were found to be of secondary importance with the pronounced loss of magnesium in alloys 2036, and 6009 being the main problem. Alloy 6009 showed excessive precipitation of AlMgSi which was thought to contribute to its poor weldability. Although laser irradiation parameters were critical for successful welds, factors such as additional cooling of the material by means of a chill plate were reported to improve welding in this study. Mazumder(69) quotes an unpublished report stating porosity free welds in aluminium were obtained using a 10kW CO₂ laser, and gas shielding where the plasma formed was pushed into the 'keyhole'. 
Crafer\(^{(82)}\) reported that aluminium alloys could be welded up to a material thickness of 3.5\(\,\,\text{mm}\) with a 2kW laser without the need for absorptive surface coatings, provided care is taken in selection and maintenance of the mode structure of the laser beam. Welds in 6mm H15 alloy were also reported at 5.6kW, and 44mm/s. This report does not however contain details of the experimental procedure followed for this work. Ricciardi et al\(^{(83)}\) investigated the laser welding of Al 6061, the metallurgy, and mechanical properties of the joint were considered. Sandblasted, pickled, and chamfered edges were studied. The authors report an improved absorption of the laser beam radiation with the sandblasted and pickled surface finishes. Nitrogen was said to reduce effervescence of the melt zone, which appeared to be enhanced by both helium and argon. X-ray examination of the weld area was reported to reveal the presence of porosity and some irregularities of the seam, while tensile testing revealed the weld to be the weakest part of the joint. What is not clear from this report is whether the effects of the surface finishes on the final properties of the joint were studied. While the metallurgy of the joint was said to have been studied, details of this aspect of the work are not given. It is believed that in order to adequately assess the effect of prior surface preparation on a welded joint it would be beneficial to investigate the properties that would otherwise be obtained in a similar material without prior surface preparation under similar experimental conditions. The results obtained could then be used as a standard against which to compare any results obtained from further studies on the material while varying the surface finish.

It is evident from the literature that very little work has been reported on the laser weldability of A16061. Certainly no evidence has been found in the literature of studies into the effect of the fast cooling rates associated with the laser welding process on the precipitation characteristics of this alloy, and hence the final properties of the material.
2.3 Review of metallurgy of relevant Nimonic alloys, Al 6061, and metal matrix composites

In this section a brief introduction to the metallurgy of the alloys used in the course of the study is given. Standard heat treatment practices, their structural implications, and finally the structural effects on the development of the mechanical properties of the alloys are considered. The alloys will be dealt with in three sub-sections, considering: the metallurgy of the Nimonic alloys, i.e., Nimonic C263, and PE11, the metallurgy of aluminium 6061, and the metallurgy of the aluminium base boron fibre reinforced, and silicon carbide particulate reinforced metal matrix composites.

Metallurgy of the Nimonic alloys

Nimonic C263

Nominal composition

<table>
<thead>
<tr>
<th>C</th>
<th>Si</th>
<th>Cu</th>
<th>Fe</th>
<th>Mn</th>
<th>Cr</th>
<th>Ti</th>
<th>Al</th>
<th>Co</th>
<th>Mo</th>
<th>B</th>
<th>Zr</th>
<th>Ni</th>
</tr>
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<tbody>
<tr>
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<td>19.0</td>
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<td>bal</td>
</tr>
<tr>
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<td>max</td>
<td>max</td>
<td>max</td>
<td>0.6</td>
<td>21.0</td>
<td>2.4</td>
<td>0.6</td>
<td>21.0</td>
<td>6.1</td>
<td>max</td>
<td>max</td>
<td></td>
</tr>
</tbody>
</table>

Nimonic C263 belongs to the Ni-Cr-Co group of nickel base superalloys. The alloy consists of a uniform solid solution of Co in a Ni-Cr matrix. Co replaces Ni in the matrix, and reduces the solubility of Al, and Ti thereby maintaining strength at high temperatures. This alloy was developed by Rolls Royce to satisfy their specifications for an alloy having good weld ductility and fabricability, as well as
high strength\(^{(84)}\). Nimonic C263 is used in parts of the gas turbine that are subjected to high temperatures, and are highly stressed, e.g. in the turbine casings and exhaust ducts. It is also used in areas such as the thrust deflector systems for breaking.

The structural features arising from a standard heat treatment of Nimonic C263 essentially consists of inter, and intragranular carbides of the form M(C, N), and M\(_{23}C_6\). In the fully heat treated condition a fine discontinuous precipitation of carbides is evident at the grain boundaries with very few particles in the matrix. The \(\gamma\) phase cannot be resolved optically but electron micrographs of extraction replicas show spherical particles of mean diameter 22nm to be present in the matrix. Because of the high Ti to Al ratio present in this alloy, the \(\gamma\) which forms is metastable, and on prolonged exposure to temperatures in excess of 700°C can transform to \(\eta\) Ni\(_3\)Ti. The acicular \(\eta\) phase forms at the grain boundaries resulting in a Ti impoverished matrix, facilitating the dissolution of the \(\gamma\) in the vicinity of the needles.

Solution treatment is usually carried out at about 1150°C for 2 hours, to dissolve the precipitable phases, principally \(\gamma\) and carbides prior to their controlled precipitation during reheating, or in special cases during cooling. The solvus temperature for \(\gamma\) in alloy C263 is between 910°C and 925°C, while the M\(_{23}C_6\) phase is dissolved between the temperatures 1050°C, and 1070°C. An aging treatment is then carried out at 800°C for 8 hours to precipitate out the \(\gamma\) phase, precipitation of the carbides also occurs during this ageing treatment.

The mechanical properties of the alloys are derived through solution strengthening and precipitation hardening. Solution strengthening occurs by virtue of the dissolution of alloying elements such as Co, and Mo in the Ni-Cr matrix, while
precipitation hardening occurs through the precipitation of \( \gamma \) and the metal carbides. \( \gamma \) is based on the formula Ni\(_3\)(Ti, Al), and can have a variable composition depending on the Ti and Al contents of the alloy. There is further flexibility of the composition as the Ni can be replaced by Co, Cr, and Mo, while the Ti, and Al can be replaced by Mo, and Cr. When precipitated in a spherical or cuboidal morphology, \( \gamma \) has a face centred cubic crystal structure which is coherent with the face centred cubic matrix. It therefore acts to impede the movement of dislocations within the metal grains. The properties of the grain boundaries are determined to a large extent by the metal carbides formed in these regions. There is an optimum amount and distribution of grain boundary carbides to give the best high temperature properties, and it is the purpose of the heat treatment to achieve this amount and distribution. Carbides increase strength by preventing excessive grain boundary sliding, and growth of voids along grain boundaries at high temperature.

Nimonic PE11

Nominal composition

<table>
<thead>
<tr>
<th>C</th>
<th>Si</th>
<th>Cu</th>
<th>Fe</th>
<th>Mn</th>
<th>Cr</th>
<th>Ti</th>
<th>Al</th>
<th>Co</th>
<th>Mo</th>
<th>B</th>
<th>Zr</th>
<th>Ni</th>
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<tr>
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<td>0.5</td>
<td>0.5</td>
<td>bal</td>
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<td>17.0</td>
<td>2.1</td>
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<td>4.75</td>
<td>0.001</td>
<td>0.05</td>
<td>37.0</td>
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<tr>
<td>0.08</td>
<td>max</td>
<td>max</td>
<td>max</td>
<td>19.0</td>
<td>2.5</td>
<td>1.0</td>
<td>max</td>
<td>5.75</td>
<td>max</td>
<td>max</td>
<td>41.0</td>
<td></td>
</tr>
</tbody>
</table>

Nimonic PE11 belongs to the Ni-Cr-Fe base alloy system. The structure of the alloy in the forged and heat treated condition generally consists of irregularly shaped grains, a random dispersion of primary titanium carbo-nitrides, and an intergranular network of complex chromium carbides of the form M\(_{23}\)C\(_6\). These carbides form as discrete globular particles. The \( \gamma \) precipitate is only sometimes visible in fully heat treated material using optical techniques, but may be
distinguished if the material is cooled sufficiently slowly from the solution treatment temperature for coarse precipitation to occur. The $\gamma'$ phase occurs as a continuous dispersion of fine spherical particles of mean diameter 18nm. If the material is subjected over prolonged periods to temperatures in excess of 575°C sigma phase is precipitated along the grain boundaries, eventually forming continuous networks. Under extremely severe exposure conditions not normally encountered in service, plates of $\eta$ phase can occur as well as intergranular precipitation of sigma phase resulting in a depletion of $\gamma'$ from the structure.

This alloy unlike alloy C263 is usually subjected to a two stage heat treatment. The treatment at 1020°C for 2 hours is designed to take the $\gamma'$ phase into solution along with the $M_{23}C_6$. The solvus temperature for $\gamma'$ in alloy PE11 is between 900°C and 925°C, while the solvus temperature for $M_{23}C_6$ is between 975°C and 1000°C. On subsequent aging at 800°C for 2 hours the carbides reprecipitate in their optimum distribution, a final ageing treatment at 700°C for 16 hours precipitates the $\gamma'$ as a fine homogenous precipitate.

The mechanical properties of Nimonic PE11 are derived from a combination of solid solution strengthening, and precipitation hardening as described for Nimonic C263. These alloys have good welded ductility, and fabricability, but are limited to temperatures of operation below about 550°C for reasons of microstructural stability. For this reason the alloys are used in the areas of the gas turbine where the temperatures are relatively low.
The metallurgy of aluminium 6061

Nominal composition

<table>
<thead>
<tr>
<th>Si</th>
<th>Mg</th>
<th>Cu</th>
<th>Cr</th>
<th>Al</th>
</tr>
</thead>
<tbody>
<tr>
<td>0.4</td>
<td>0.6</td>
<td>0.25</td>
<td>0.25</td>
<td>bal</td>
</tr>
<tr>
<td>1.0</td>
<td>1.3</td>
<td>max</td>
<td>max</td>
<td></td>
</tr>
</tbody>
</table>

Aluminium 6061 belongs to the 6000 series aluminium, (Al), alloys. Alloys in this group contain magnesium, (Mg), and silicon, (Si), in proportions sufficient to form magnesium silicide, the presence of which makes the metals heat treatable. Wrought aluminium alloys are first cast, and then subjected to a series of mechanical and thermal treatments. Each treatment results in a change in structure, and the finished structure is a composite reflecting the various changes induced during fabrication. The structure of the cast ingot is an array of grains each having a dendritic structure. The dendritic structure consists of primary aluminium surrounded by heavily cored, solid solution regions. The dendritic boundaries usually contain a constituent particle network of soluble and insoluble phases. Preheating is carried out prior to hot working of the ingot. This treatment homogenizes the solid solution by reducing or eliminating coring, and increasing its alloying element content. If the preheat temperature is above the eutectic temperature, the structure frequently exhibits rosettes of resolidified eutectic. Sheet metal exhibits recrystallised or unrecrystallised structures depending on the composition, the type, amount, and temperature of working, and any thermal treatments applied. The grains are generally elongated in the direction of working, and flattened in the thickness dimension, while constituent particles tend to become arranged in planes parallel to the surface of the product.
Generally heat treatable Al alloys contain alloying elements which are soluble in considerable amounts at elevated temperatures, and to a much smaller degree at room temperature. The alloys also tend to contain one or more insoluble elements which add to the strength of the product and aids the control of grain structure.

In Al 6061, Mg and Si combine to form a compound Mg$_2$Si which in turn forms a simple binary eutectiferous system in Al with a solid solubility of 1.85% at the eutectic temperature and which has a very low solubility at room temperature.$^{(84)}$ The presence of Mg and Si in the ratio to form Mg$_2$Si results in the alloys having excellent corrosion resistance. When Si is present in excess of the ratio, higher strengths are developed in the artificially aged condition, but resistance to corrosion is lowered appreciably. Cr or Mn is usually present as an insoluble element to control grain structure, and the presence of Cu also results in an increase in alloy strength.

Typical thermal treatments for Al 6061 first involve a solution heat treatment, i.e. heating to between 468 and 523°C, but below the eutectic melting temperature. At the e temperatures the maximum amount of solute is taken into solid solution. By quenching from the solution treating temperature, a non equilibrium supersaturated solid solution is obtained.

During ageing, natural or artificial, a redistribution of solute atoms within the solid solution lattice occurs, to form clusters or zones, GP (Guinier Preston zones) that are considerably enriched in solute. It is these precipitate atom clusters, GP zones which provide strengthening in the alloy. This local segregation of solute produces distortion of the lattice planes, both within the zones and extending for several atom layers into the matrix. With an increase in number or density of zones, the degree of disturbance of the regularity and periodicity of the lattice increases. The
strengthening effect of the zones can be considered as resulting from the additional interference with the motion of dislocations through the lattice that is afforded by this increased irregularity.

Figs. 2.18, 2.19, and 2.20 show the Al-Mg-Si system, with Fig. 2.18 showing a projection of the liquidus surface, Fig. 2.19, a vertical section along the Al-(MgSi) quasi binary, and Fig. 2.20 a projection of the solvus surface in the Al corner of the diagram.

Al 6061 is a general aircraft alloy, and finds its uses in airframe construction, e.g in the main structural members, wing spars, fuselage, and landing gear. These parts are usually made from extrusions or forgings. Skinning of the wings, and fuselage is of thin clad sheet in similar alloys. Alloys in the 6000 series generally exhibit good formability, good corrosion resistance, and good mechanical properties. They can be welded by conventional welding processes, but generally require a post weld heat treatment if full strength is to be restored.

The metallurgy of aluminium 6061 base metal matrix composites

Aluminium 6061/silicon carbide metal matrix composite

The Al 6061/SiC MMC is comprised of a matrix of Al 6061 containing a uniform distribution of irregular SiC particles. The reinforcing particle size is in the range 3 - 10 µm and forms 30% of the matrix total volume. This particular class of MMC is commonly made in three stages.
1. Blending of the SiC particles with the matrix alloy powder.
2. Cold compaction of the blended powders.
3. Hot pressing of the resultant green compact at a temperature above the solidus temperature of the matrix alloy.

Hot pressing is carried to enhance the matrix - particle bonding. Finally a modified version of the heat treatment usually practiced for the matrix alloy is used to optimise the composite properties.

This class of MMC's show promise for applications requiring a high degree of stiffness, a high strength to weight ratio, adequate corrosion resistance(85), possible use at higher service temperatures, and controlled thermal expansion(86).

**Aluminium 6061/boron fibre reinforced metal matrix composite**

The boron filaments used in the production of this composite were made by deposition of boron on a thin tungsten wire, to produce a filament 1.4mm in diameter. These filaments were then laid down on thin aluminium foil, alternating the layers of boron and aluminium. The aluminium foil and boron fibres were then diffusion bonded. The material used in this series of experiments comprised nine alternating layers of boron and aluminium, (nine - ply). These materials possess very high stiffness and strengths, and as such find their uses in aircraft components requiring low deflection to be combined with high strengths(87).

**Mechanical properties of metal matrix composites**

Work on characterising the mechanical properties of Al 6061/SiC is still in the very early stages. The strengthening effects of SiC, or boron, including the influence of microstructural details such as the quantity of the strengthening phase, aspect ratio, and loading fraction are not well understood. It would also appear that the properties of the matrix itself contributes to the overall strength of the composite.
As far as fracture properties are concerned there does not as yet appear to be a conceptual framework within which the limited data on the fracture of MMC's can be understood.

2.3 Introduction to the current work

From the literature it is evident that detailed studies of the laser welding of Nimonics have not been made, there are problems still to be resolved due to the reflectance of the CO₂ laser radiation in the welding of aluminium alloys, and little work on the laser welding of aluminium based metal matrix composites has been reported.

The programme of work on Nimonics C263, and PE11 was therefore devised to include the following.

1. Studies of the effects of laser welding parameters; such as laser power, and laser beam traverse speed on weld penetration, weld bead top widths, melt volumes, and welding efficiency.

2. Studies of the effects of lens focal length on welding penetration and efficiency.

3. Identification of any synergistic effects between laser parameters on the welding process.

4. A comparison between results obtained from welding using two different lasers to identify the effects of individual laser welding systems on the response obtained during welding.

5. Detailed macro, and microstructural studies of the welds produced and the related heat affected zones.
Work on the aluminium based materials, encompassed an in-depth study into the response of as received aluminium sheet, and anodised aluminium sheet to the laser welding process. Within this framework the following main points were considered.

1. Studies of the laser welding parameters; such as laser power, and laser beam traverse speed on weld penetration, and weld bead top widths in as-received aluminium 6061 sheet.

2. Studies on the effect of the sulphuric acid anodising process on the response of the aluminium to the laser welding process, measured in terms of weld penetration achieved and weld bead top widths.

3. The microstructural response of the material to the laser welding process, both in the as-received and anodised surface finishes. The effect of the process on dendrite cell sizes, and heat affected zone widths were measured.

4. Finally microhardness measurements were used to provide an indication of the effect of the welding process on joint properties.

5. The effect of the laser welding process on the aluminium 6061 based boron fibre reinforced, and the silicon carbide particulate reinforced metal matrix composites were considered.

6. Detailed studies of the resultant microstructures in the fusion zones of these materials were carried out, and the novel phases observed were identified.

The following chapters give details of the equipment used in the work reported in this thesis, the materials studied, and the experimental procedure followed. A detailed account of the results obtained in the course of this work, their interpretation, and discussion is given in chapters 4 and 5 for the Nimonic alloys, and the aluminium based materials respectively. This is followed by a summary of the findings, conclusions drawn, and suggestions for further work in chapter 6.
## 2.1. Characteristics of Common Lasers and Their Applications

<table>
<thead>
<tr>
<th>Class</th>
<th>Type</th>
<th>Wavelength, µm</th>
<th>Average Output Power</th>
<th>Mode of Operation</th>
<th>Pulse Repetition Rate, per second</th>
<th>Pulse Length</th>
<th>Applications</th>
</tr>
</thead>
<tbody>
<tr>
<td>Neutral gas</td>
<td>Helium-Neon (He-Ne)</td>
<td>0.6328, 1.15, 3.39</td>
<td>0.5-50 mW</td>
<td>Continuous</td>
<td></td>
<td></td>
<td>Interferometry, alignment, educational demonstrations, display projection</td>
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<td>Argon</td>
<td>0.4880, 0.5145</td>
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<td>Graphics, display projection, holography, pattern generation, pollution detection, spectroscopy, semiconductor annealing, photochemical, dye-laser pumping</td>
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<td>Krypton</td>
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<td></td>
<td>Display projection</td>
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<td>Sodium-Cadmium-Neon (Na-Cd)</td>
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<td></td>
<td></td>
<td>Pollution detection</td>
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<tr>
<td>Molecular gas</td>
<td>Carbon-Dioxide (CO₂)</td>
<td>9.6, 10.6</td>
<td>50-100 W</td>
<td>Repetitively pulsed</td>
<td>100</td>
<td>100 µs</td>
<td>Materials welding, hardening, cutting</td>
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<tr>
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<td>Ca-neon-Dioxide (CO₂)</td>
<td>9.6, 10.6</td>
<td>(Up to 10³ W peak)</td>
<td>Transversely excited atmospheric (TEA)</td>
<td>Up to 100</td>
<td>10 µs</td>
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<td>Oxygen</td>
<td>0.3371</td>
<td>0.1 to 1 W (up to 10⁴ W peak)</td>
<td>Repetitively pulsed</td>
<td>100</td>
<td>10 ns</td>
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<tr>
<td>Solid state</td>
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<td></td>
<td>Materials welding, drilling</td>
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<td>Repetitively Q-switched, continuously pumped</td>
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<td>200 ns</td>
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<tr>
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<td>Pulsed</td>
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<td>0.5-10 ms</td>
<td>Materials welding, drilling, nuclear fusion</td>
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<td>0.1 µs</td>
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</tr>
<tr>
<td>Liquid organic dye</td>
<td>Various (tunable)</td>
<td>1-0.4  (depending on dye)</td>
<td>Up to 5 W</td>
<td>Laser pumped</td>
<td></td>
<td></td>
<td>Process control, photochemistry, Raman spectroscopy</td>
</tr>
</tbody>
</table>
TABLE 2.2 Showing the Reflectivity of Several Metals at 20°C for 10.6 μm Radiation, Calculated using Bramson's Equation.

<table>
<thead>
<tr>
<th>Metal</th>
<th>Reflectivity %</th>
</tr>
</thead>
<tbody>
<tr>
<td>Copper</td>
<td>98.02</td>
</tr>
<tr>
<td>Silver</td>
<td>97.98</td>
</tr>
<tr>
<td>Aluminium</td>
<td>97.3</td>
</tr>
<tr>
<td>Nickel</td>
<td>95.27</td>
</tr>
<tr>
<td>Steel (alloy)</td>
<td>94.93</td>
</tr>
</tbody>
</table>

Table 2.3 The effect of dynamic focusing on penetration efficiency.

<table>
<thead>
<tr>
<th>Material</th>
<th>Focusing</th>
<th>Total thermal efficiency of penetration</th>
<th>Relative increase of penetration efficiency, %</th>
</tr>
</thead>
<tbody>
<tr>
<td>12Kh18N10</td>
<td>Static</td>
<td>0.27</td>
<td>78</td>
</tr>
<tr>
<td></td>
<td>Dynamic</td>
<td>0.48</td>
<td></td>
</tr>
<tr>
<td>AMg</td>
<td>Static</td>
<td>0.16</td>
<td>87</td>
</tr>
<tr>
<td></td>
<td>Dynamic</td>
<td>0.30</td>
<td></td>
</tr>
<tr>
<td>PT-3V</td>
<td>Static</td>
<td>0.39</td>
<td>69</td>
</tr>
<tr>
<td></td>
<td>Dynamic</td>
<td>0.66</td>
<td></td>
</tr>
</tbody>
</table>
Fig. 2.1 Diagramatic representation of the degrees of vibrational freedom present in the CO$_2$ molecule.
Fig. 2.2  Diagramatic representation of lasing

(a) Ground State.
(b) Excitation and Spontaneous Decay.
(c) Spontaneous and Stimulated Emission.
(d) On-Axis Stimulated Emission and Amplification.
(e) Amplification.
(f) Lasing.
Fig. 2.3  Diagramatic representation of the lasing levels operational in a CO₂ laser.
Fig. 2.4 Schematic diagram of a low axial flow laser

Fig. 2.5 Functional diagram of a fast axial flow CO₂ laser.
Fig. 2.6 Some low order TEM$_{mn}$ cavity modes for cylindrical symmetry. The dark areas represent areas of enhanced output, although the distribution of intensity in these regions is not uniform.

Fig. 2.7 Diagramatic representation of the TEM$_{00}$ Gaussian mode.
Fig. 2.8  Diagramatic representation of the hemispherical mirror in the laser cavity.
Fig. 2.9 Schematic illustration of laser beam focusing by a simple lens.
Fig. 2.10 Welding speed against power for Culham/Ferranti, 5KW CO2 laser for Ti-6Al-4V.
Figure 2.11  Effective spot size as a function of optical spot size.
Fig. 2.12 Schematic illustration of the laser welding process, showing the 'keyhole' produced by the impinging laser beam.
Diagram for explaining the dependence of the power of transmitted radiation on metal thickness (a) and the mechanism of heating thin (d/h < 0.4) (b) and thick (d/h > 0.4) (c, d) metal.

The power balance of the laser beam in relation to the ratio of the thickness of metal to the maximum penetration thickness.
Fig. 2.15  Diagram of the process of recovery of the power of transmitted radiation (a) and the form of the penetration zone in various sections of a wedge-shape specimen (b) at a welding speed of 0.6 m/sec: 1 - without power recovery, 2 - with power recovery.
Fig. 2.16 Diagramatic representation of a plasma control 'setup'.

- laser beam
- focusing lens

helium jet, nozzle
diameter 1.25 mm

25-50°

workpiece
Fig. 2.17: Weld penetration dependence on laser power and traverse speed for Ti-6Al-4V.
Fig. 2.18 Projection of the liquidus surface of the Al-Mg-Si system

Fig. 2.19 Vertical section along the Al-(MgSi) quasi-binary
Fig. 220 Projection of the solvus surface in the Al corner of the diagram.
CHAPTER 3

EQUIPMENT AND MATERIALS

Introduction

Although there are currently many lasers on the market, only two are used for nearly all materials processing operations. These are the Carbon Dioxide, (CO₂) laser and, the Neodymium Yttrium Aluminium Garnet, (Nd:YAG) laser. For successful welding to be carried out one of the requirements is a continuous energy source of several hundred Watts. In the case of materials such as metals which are difficult to process, by virtue of their high reflectivities, or high thermal conductivities multi-kilowatt energy sources are required. Such high powers are only possible when using the CO₂ laser.

It was determined through a survey of existing literature and knowledge of materials properties, that in the laser welding of aluminium base alloys and nickel base alloys, lasers having at least a multi-kilowatt power capacity were required. For this work, two lasers were used, a 2kW Control laser, and the Welding Institute 5kW CO₂ laser. The object of this was to enable a comparison to be made of welds in similar materials using lasers with different characteristics, e.g different maximum power capabilities, and different focused beam spot sizes.
This chapter will be divided into five main sections. The first deals with a 
description of the laser systems used in the experimental programme. This is 
followed by a detailed account of the procedure followed in the operation of the 
equipment. Next details of the anodising treatment applied to the as received 
aluminium 6061 are outlined. The post laser processing techniques used in the 
analysis of the welds are then considered. Finally details of the materials used in 
the study are given.

3.1 Description of laser systems used in the experimental programme

3.1.1 The Welding Institute 5kW continuous wave CO\textsubscript{2} laser

The laser is made up of three distinct parts.
1. The blower.
2. The laser optics.
3. The voltage supplies.

The blower

The blower is a Roots type positive displacement blower driven by a 45kW AC 
electric motor via a quadruple V-belt arrangement. The blower is supported on the 
blower raft, which is constructed from steel box sections perpendicular to the 
blower shaft and connected by thick steel plates. The blower raft also supports the 
blower motor, and is floor mounted on anti-vibration mounts. The driver motor is 
connected to the blower raft via slides, while the blower itself is mounted on four 
legs attached to machine pods on the impeller casing, Fig. 3.1.

The low pressure gas ducting is of H30 aluminium. This is isolated from the 
effects of vibration resulting from the blower by the insertion of four anti-vibration
sleeves. Evacuation of the system was by a GKT150 vacuum pump. The use of two valving lines in parallel allowed two rates of evacuation to be possible. Two solenoids controlled the evacuation. One was large and coupled to a large bore pumping line. This controlled the initial evacuation. The smaller solenoid valve was used to maintain the system pressure during operation. Both lines were connected to an aluminium pellet oil trap. The blower gearbox, pumping lines and a 2° vacuum pump are situated between the valves and oil trap. The 2° vacuum pump maintains the system vacuum when it is non-operational. An automatic isolator and air admittance system protects the laser from power failure.

Both gas and water cooling are used. The main gas coolers are modified automobile radiators mounted in vacuum boxes. The centre box has all twelve cathodes mounted on its two end faces. Water cooling jackets on the top and the two sides cool the cathodes and limit heat transfer to the optical bench and hence distortion. The after cooler is constructed of thin copper tubing, the blower oil coolers are also of copper tubing, and are immersed in the oil filled ends of the blower to maintain the correct temperature and viscosity. The main water heat exchanger consists of an 8500 litre capacity water tank cooled by a 25kW refrigerator. Water from the tank is fed through a filtered pump to a main from which individual cooling circuits are supplied. The water returns to the main tank via a gravity drain. There is also a secondary water heat exchanger available to the system.

Laser optics

A stress relieved optical bench positioned on an all welded steel mainframe carries all the lasing tubes, optical adjustment and power monitoring facilities.
Anti-vibration mounts isolate the bench from any vibration of the mainframe. At each end of the optical bench are gas feeds each supplying six of the twelve discharge tubes. The gas feed box is rigidly attached to the optical bench frame. A central gas collector-box sits midway along the bench. Between the gas feed boxes and central collector are two orifice plates. The orifices consist of a 'top hat' outer case, a fixed ceramic insert and anode assembly, and an inner nozzle also made from ceramic. The gap between the ceramics forms the gas passageway.

The oscillator section consists of four discharge tubes arranged in two pairs and connected in series by a folding arrangement, Fig. 3.2. The oscillator output is directed via further folds through eight amplifier tubes arranged in two banks each having two pairs of tubes connected by a fold. The diameter of the tubes in successive layers is tailored to follow the natural diffraction of the oscillator beam as it traverses the amplifier. Each tube is joined at one end to the orifice assembly by an insulating flange, and at the other end to the cathode by a threaded brass nut.

Discharge tubes within each bank are connected vertically to adjacent banks by means of three pairs of mirror assemblies. Each mirror is mounted with its polished face in contact with a precision machined surface, and cooling is provided by water cooled hats. Finally, a KCl lens focuses the beam to the workpiece. It also serves as a vacuum seal reducing transmission losses to a minimum, (less than 10%).

Power measurements are effected by means of calorimeters mounted on the outside of the vacuum box. A pair of sliding mirrors operated externally allow both the oscillator and amplifier beams to be dumped into the calorimeters. Additional calorimeter positions allow measurements to be made at several points in the system.
Voltage supplies

Two three phase 415V, 50Hz transformers with variac control drive the oscillator and amplifier tubes independently. The rectification is provided by separate aircooled diode banks.

The laser head

The raw beam emerging from the laser is directed towards the workpiece by means of highly reflective mirrors, and is focused onto the surface of the workpiece by a KCl lens to a spot diameter between 0.5 and 0.7mm. During welding the emerging beam remains stationary and the material is moved by attachment to an x-y table. The mode structure of the beam emerging from the laser was checked by taking a 'Perspex char' print before any welding was carried out. The beam was observed to give a 'top hat' type of Perspex print, Fig. 3.3.

The x-y table

The x-y table was attached to a computer via which the required speed was programmed into it. The table was driven hydraulically, and capable of reaching a maximum speed of 7mmin^{-1}, i.e 117mms^{-1} in the x direction. It was also capable of movement in the y and z directions. Attached to the x-y table was the welding jig by which the workpiece was restrained.

3.1.2 The 2.5kW continuous wave CO2 laser

Fig. 3.4 shows the basic features of the 2kW Control laser, (formerly the BOC 901). These can be seen to consist of an optical bench mounted on a steel
mainframe. The laser cavity comprises a spherical total reflectance mirror and two plane gold plated beryllium-copper mirrors. The output window is made of gallium arsenide which has a reflectivity of about 35% for 10.6μm radiation. All the optical elements are water cooled, with the circulating gases being cooled by gas liquid heat exchangers. The final focusing of the beam is achieved with the aid of a potassium chloride lens.

The excitation of the gases to produce the 10.6μm radiation is facilitated by a combination of low current (maximum 700mA dc), and high voltage electrical discharges. The physics of the lasing action has been reviewed in chapter 2, and will not be considered in detail here. It will suffice to say that by taking advantage of the vibrational and rotational properties of the carbon dioxide, nitrogen and helium molecules, a number of transitions between vibrational levels can be brought about in the presence of an electrical discharge. These transitions occur over a band 9 to 11μm with a mean wavelength around 10.6μm (4).

The 2kW CO₂ laser is of the fast axial design, and the lasing gases (CO₂, N₂, and He), are circulated at high velocities through the plasma tubes, (up to 500ms⁻¹), by the Roots blower. Discharge stability is achieved through shock stabilisation. Typical gas compositions used in the laser are as follows (5)

<table>
<thead>
<tr>
<th>GAS</th>
<th>Wt%</th>
</tr>
</thead>
<tbody>
<tr>
<td>He</td>
<td>75-85</td>
</tr>
<tr>
<td>N₂</td>
<td>10-14</td>
</tr>
<tr>
<td>CO₂</td>
<td>5-10</td>
</tr>
</tbody>
</table>
The optical cavity is 8m long and folded in the middle. A pneumatically operated on-off shutter beyond the output window directs the beam into a water cooled calorimeter when the beam is not in use. The temperature of the water in the calorimeter is related to the power output of the beam. A digital display of the power level is taken directly from a calibrated thermocouple reading of this temperature.

Safety

Several safety devices operate to bring about automatic shutdown, which would come into operation should any of the following conditions prevail.

1. Insufficient amounts of one or more of the constituent gases in the lasing gas mixture.
2. Absence of water cooling at any point in the system.
3. Output window breakage.
4. Presence of moisture in the shutter box.

The laser head and lens

Fig. 3.5 shows a schematic diagram of the laser head. It can be seen to consist of a lens, a lens mount, lens height adjustments, lens heater, gas inlet, x-y nozzle movement, and the nozzle itself. The laser head is remote from the main laser structure and the laser beam is isolated from the laser head by the shutter when the laser is not in use. During operation the shutter is remotely moved out of the way of the laser beam which then passes through the lens assembly.
The lens mount is basically a means of holding the KCl lens through which the laser beam is focused. The lens height adjustment varies the position of the focal point of the lens with respect to the nozzle.

The heater prevents the hygroscopic KCl lens from reacting with moisture in the atmosphere and becoming 'fogged'. The gas inlet allows for a coaxial flow of gas into the laser head around the lens holder assembly. This is essential even if a shield gas is not necessary from a materials processing point of view. The main purpose of this is to prevent dirt from collecting on the lens as this could lead to hot spots and cracking of the lens.

The x-y nozzle movement is used to centralise the focused laser beam as it exits through the nozzle. The nozzle mainly serves to direct the gas flow onto the workpiece. This is more important for cutting operations than for welding. A biconvex KCl lens is used in the laser head. The main criteria for this choice was the durability and the cost of these lenses.

The x-y table

A hydraulically powered x-y table was attached to the laser workbench. The rate of traverse of the table was controlled by two valves which varied the rate of oil supply to the x and y cylinders. The y movement was used mainly for positioning the workpiece relative to the laser beam, while the x movement was used for processing. This was because of the faster and longer movement in the x direction. It was possible to use mechanically activated interlocks to halt the movement of the x-y table at any time, hence preventing it from reaching the end of its traverse at high speed. The speed of the table in the x direction was determined
by calibrating the table. This was done by determining the time taken to traverse a known distance.

3.2 Operation of the equipment

The operation of the 2kW laser is described in detail in the BOC SERIES 9000 2W CO₂ laser machine operators manual. A brief description of the procedure followed is outlined in the following paragraphs.

3.2.1 The 5kW laser

This machine was in constant use commercially by the Welding Institute, it was therefore not possible to participate directly in the setting up of the equipment as this was carried out by the Welding Institute's own staff. However the mode of operation was similar to that of the 2kW Control laser which is described here.

3.2.2 The 2kW laser

Fig. 3.6 shows the power unit front panel controls. Before any experimental work was carried out the laser was switched on and the discharge allowed to stabilise. This would prevent any fluctuation of the output power which could result in the presence of an unstable discharge.

Pressing the start button results in a fast pump down of the gas circuit to produce a vacuum. When a low vacuum level is reached the gas blower backfills the system with a mixture of helium, nitrogen, and carbon dioxide. At the end of the backfill sequence the ready indicator lamp lights up and the HT supply is ready to be switched on.
Effect of discharge current

It was possible to vary the output power of the laser simply by altering the discharge current. The maximum discharge current for operation was 150mA. Above this current a safety device triggered a system shutdown. Discharge instabilities at low currents placed limitations on the level of minimum discharge current allowed for laser operation. Below 50mA this instability resulted in extinction.

In the course of these experiments the power used for welding was within the range 1.4 to 1.8kW. It was possible to vary the laser power within this range by altering the discharge current.

Beam alignment and mode structure

The alignment of the laser beam within the optical cavity is a critical factor for successful operation of the equipment. Poor alignment can result in loss of power due to edge truncation of the beam at the nozzle exit, or loss of power due to the presence of high order mode structures and asymmetric mode patterns. Extreme cases of misalignment can result in the output beam clipping the 'O' rings in the output window assembly. Burning of the 'O' rings could then occur, or the output window could break due to increased absorption and overheating.

(a) Incorrect alignment of the optical bench during setup. If present it can be rectified.

(b) Checking for thermal warping of the optical elements. This can generally be avoided by ensuring that all the optical elements are free of dust particles, scratches, and cracks. Also the cooling system can be checked to ensure that it is operating correctly.
(c) Tuning of the mirrors on the output window by means of adjustment screws.

The beam mode structure was checked before the start of the welding experiments with the aid of a 'Perspex char' print. The observed structure was as shown in Fig. 3.7. Attempts were made in some cases to improve the mode structure by aligning the beam to ensure that no edge truncation is occurring. This is done with the aid of a helium-neon laser. Further improvements can also be achieved by adjusting the mirrors at the output window. It is not usual to carry out major cavity realignment as this is a very time consuming process.

**Determination of the focal point**

The focus position of the laser beam is another important factor in all laser processing operations, as the size of the focal spot determines the energy intensity available for processing. It was first necessary to determine the approximate focal position of the lens. This was to ensure that the focal position was below the nozzle otherwise it would be impossible to focus on the workpiece.

A distance of about 2mm below the nozzle was chosen, so that: (a) the material ejected from the workpiece would not contaminate the lens and lead to breakage, and (b) the nozzle was sufficiently far from the workpiece to prevent excessive heating.

The approximate focal position of the lens was determined by passing a block of 'Perspex' through the beam in such a way that the shape of the beam was imprinted on the perspex. The narrow waist indicates the position of the focus, Fig. 3.8. The distance, $h_f$ from the top of the 'Perspex' to the focal plane was compared with the height, $h_n$ of the nozzle. The position of the lens was then adjusted until

$$h_f - h_n = 2 \text{mm}$$
The blue flash test

Having determined the approximate position of the focus, the blue flash test was used to obtain the actual focus point. When a CO₂ laser is focused on a metal surface in the presence of an argon gas shield, the hot plasma causes the argon molecules to ionise. The ions thus formed give off emissions in the Argon spectrum resulting in a blue flash being observed in the plasma. This ionisation only occurs if the intensity of the focused beam approaches the focused value.

It is generally found that in using this method to determine the focus point a blue flash can be obtained over a range of lens height above the workpiece. The exact position of the focus therefore has to be gauged by listening to the plasma events, and the focus spot deemed to be that region in the blue flash range over which stable plasma events are occurring. This is of course a subjective way of determining the focus spot, but it is the best available method at this point in time.

By moving the laser head up and down with the aid of the adjusting screws till the blue flash conditions were satisfied it was possible to determine the focus spot for the laser.

The x-y table

The x-y table was mounted on an adjustable bench setting. The bench height was adjusted by means of screws on each leg. This was necessary to allow for flexibility in adjusting the position of the workpiece in relation to the laser head. The speed of the x-y table was determined using an optical switch and an electronic counter. A calibration curve showing the x traverse speed against the speed control setting was thus obtained. From this curve the welding speed could be determined
for known settings on the speed control. Welding speed was set in the x direction, and the y movement of the x-y table was used only to position the workpiece in relation to the laser beam.

3.3 Anodising procedure

Since anodising is a conversion process, the appearance and properties of the oxide film are completely dependent on the composition of the Al, and its surface condition. To ensure an even coverage of the samples, they were all electrolytically polished in 'Phosbrile', (Brace, A.; Technology of Anodising, Robert Draper Ltd, 1969). The 'Phosbrile' solution used in this work was of the following composition.

- 70% phosphoric acid.
- 20% sulphuric acid.
- 10% nitric acid.
- 5g/l cupric nitrate.

Having polished the samples, they were anodised in a plastic bath containing containing approximately 10% H₂SO₄. A Variac control was used to vary the current. The bath was agitated by means of compressed air being passed through perforated pipes in the bottom of the bath. The agitation was gentle as it was only needed to disperse the heat generated during the anodising process. A number of samples were anodised to coating thicknesses of 10, 12, 14, and 16μm. Fig. 3.9 shows a schematic representation of the structure typically obtained from the sulphuric acid anodising process. It can be seen to comprise of hexagonal columns each with a central pore which reaches down to a thin barrier layer which is continually transformed into the porous form during the anodising process, (Henley, V.F.; Anodic Oxidation of Aluminium and its Alloys, Pergammon Press, 1982).
Measurement of oxide coating thickness

The thickness of the oxide coating attained by anodising was determined quantitatively by using the Permascope EW8. The Permascope is a coating thickness meter which utilises eddy currents in the measurements of the thickness of non-conductive coatings on non-ferrous metal substrates. The equipment is completely non-destructive in operation and comprises two units.

1. A probe which is applied to the test piece, and converts the variable to be measured, in this case, the coating thickness into an electrical signal.

2. A meter unit which processes the probe output signal so that a meter indication proportional to coating thickness is produced. Both units are connected via a flexible cable. Figs. 3.9a, and 3.9b show front and rear views of the Permascope. Measurement is carried out by placing the probe on the surface of the test piece and the result is read off the meter.

3.4 Post laser processing analysis equipment

Metallographic preparation

Following the welding of the Nimonic and the aluminum based materials, the welds were sectioned transversely and prepared for metallographic examination. Both the Nimonic alloy welds and the aluminium 6061 welds were sectioned using a guillotine, and wet ground flat through a series of grit papers ranging from 240 to 1200 grade. It was not possible to section the welds in the aluminium based metal matrix composites with the guillotine, and these welds were sectioned using the spark erosion technique. After sectioning these welds were also wet ground on grit papers ranging from 120 to 1200 grit finish.
Polishing was carried out on 6μm and 1μm diamond paste to attain a mirror finish prior to metallographic examination. Difficulties were encountered in the polishing of the composite material welds, predominantly in the Al/B fibre reinforced materials, as the soft Al 6061 matrix had a tendency to smear over the hard B fibres. This was minimised by polishing for the minimum length of time required to attain a suitable finish for examination. In order to attain a mirror finish in the aluminium 6061 alloy, these welds were polished using 5μm, and 1μm alumina, and 10% oxalic acid as a lubricant.

The composite materials were examined in the as polished condition as the microstructure in the fusion, and the adjacent fusion line region was clearly visible. In the study on Al 6061 the welds were etched using a mixture of 1g ammonium oxalate in a solution of 15% ammonium hydroxide in 100ml of distilled water. The specimens were etched at 80°C in a freshly prepared solution for 5 minutes. The etchant developed the grain boundaries in both the fusion zone and the base metal, the Mg₂Si precipitates were also revealed in the base metal. Both Nimonics C263, and PE11 were etched in a solution of 20ml nitric acid, 20ml hydrofluoric acid, and 60ml glycerol. This particular etch was useful for revealing grain boundaries in both the fusion zone and the base metal. In order to study the variations in precipitation characteristic between the base metal and the fusion zone, an electrolytic etching technique was adopted. This involved the use of a solution of 85ml phosphoric acid in 100ml distilled water. A stainless steel cathode was used, and the specimens were etched at 3V for 25s at room temperature. This technique revealed the carbide distribution in both the fusion zone, and the base metal. Further details of the etchants used in this work can be found in Smithell's Metals Reference Book, and Metallographic Etching by Petzow.
Optical microscopy

The prepared specimens were examined optically using the Reichert MEF3 optical microscope with a standard 35mm camera attachment. This equipment was used to obtain details of the penetration achieved during welding, and the weld bead top width for the welds in all the materials studied using the micrometer scale attachment. Measurements of dendrite arm spacings were also obtained for Nimonics C263, PE11, and Al 6061 using this equipment. Several series of photo-macro, and micrographs were taken, and are included within the text.

Electron microscopy

Scanning electron microscopy was carried out using the Cambridge stereoscan 2a on polished and etched weld samples. This technique provided high magnification micrographs to supplement the optical micrographs obtained from the MEF3.

The scanning transmission electron microscope, (JEOL Model), was used to examine carbon replicas, and thin foil specimens from the welded samples. Thin foil specimens were prepared by chemical thinning in the Nimonic alloys, and by ion thinning in the Al base metal matrix composites. Selected area diffraction, (SAD), patterns were used to identify the phases observed in the MMCs, and to attempt to identify γ in the Nimonic alloy welds. The SADs were analysed using the camera constant method, (Smallman, R.E, and Ashbee, K.H.G.; Modern Metallography, 1956.)

Fig. 3.11 is a schematic diagram showing the relationship between the camera length, L, and the diffracted spot on the SAD. θ is small, and thus approximately equals sin θ. From the Bragg equation;

\[ \lambda = d \cdot 2\theta \]
Where \( d \) is the \( d \) spacing corresponding to the spot being indexed.

Substituting \( 2\theta = \tan 2\theta = r/L \)

\[ L\lambda = dr \]

Where \( L = \) camera length, and \( r \) the distance from the centre of the SAD as defined by the spot corresponding to the transmitted beam, and the spot to be indexed. The term \( L\lambda \) is equal to the camera constant, and is determined from measurements of \( r \) on the SAD pattern of a substance of accurately known lattice parameter.

The procedure to index an SAD pattern using the Camera constant method is to measure the \( r \) value for each spot, calculate its \( d \)-spacing, and compare this with the \( d \)-spacings listed in the standard ASTM cards of the lattice parameters of known substances.

**Secondary ion mass spectroscopy**

The SIMS analysis technique utilises a beam of low energy ions to bombard a sample, sufficient energy is supplied to the surface for atoms to be sputtered into the vacuum where the ionised material is analysed using a mass spectrometer. This technique is capable of detecting elements, and isotopes. In this work it was used in the ion imaging mode to analyse the surface of the welds in the metal matrix composite materials.

Electropositive elements produce positive secondary ions, whilst electronegative elements produce negative secondary ions. These secondary ion yields are enhanced by further bombarding the sample with oxygen, and caesium ions thus giving improved sensitivities. SIMS is capable of detecting most elements in the ppm to ppb range.
The Quantimet 800

The Quantimet 800 was used to obtain measurements of melt areas achieved in the laser welding experiments on transverse section through the welds.

Microhardness Testing

Microhardness traverses were carried out using a Reichert MEF2 microhardness tester with a 50g load.

3.5 Materials

The materials used for these welding experiments were:-
Nimonic C263.
Nimonic PE11.
Aluminium 6061.
Aluminium 6061/silicon carbide metal matrix composite.
Aluminium 6061/boron laminate.

Nimonic C263

Composition

<table>
<thead>
<tr>
<th>Element %:</th>
<th>Al (Element %)</th>
<th>Ag (ppm)</th>
<th>B (ppm)</th>
<th>Bi (ppm)</th>
<th>C (ppm)</th>
<th>Co (ppm)</th>
<th>Cr (ppm)</th>
</tr>
</thead>
<tbody>
<tr>
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<td>50</td>
<td>1</td>
<td>0.08</td>
<td>21.0</td>
<td>19.0</td>
<td></td>
</tr>
<tr>
<td>Cu</td>
<td>0.2</td>
<td>0.7</td>
<td>5.6</td>
<td>1.9</td>
<td>2.4</td>
<td>rem</td>
<td></td>
</tr>
</tbody>
</table>
Nimonic PE11

Composition

<table>
<thead>
<tr>
<th>Element%</th>
<th>C</th>
<th>Si</th>
<th>Mn</th>
<th>S</th>
<th>Al</th>
<th>Ag(ppm)</th>
<th>B(ppm)</th>
<th>Bi(ppm)</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>0.10</td>
<td>0.5</td>
<td>0.5</td>
<td>0.01</td>
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<td>5</td>
<td>50</td>
<td>1</td>
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</table>

<table>
<thead>
<tr>
<th>Co</th>
<th>Cr</th>
<th>Cu</th>
<th>Mo</th>
<th>Ni+Co</th>
<th>Pb(ppm)</th>
<th>Ti</th>
<th>Ti+Al</th>
</tr>
</thead>
<tbody>
<tr>
<td>19.0</td>
<td>0.5</td>
<td>0.04</td>
<td>10.0</td>
<td>42.0</td>
<td>2.5</td>
<td>3.5</td>
<td>5.75</td>
</tr>
</tbody>
</table>

Zr Fe
1.0 rem

Aluminium 6061

Composition

<table>
<thead>
<tr>
<th>Si</th>
<th>Mg</th>
<th>C</th>
<th>Cr</th>
</tr>
</thead>
<tbody>
<tr>
<td>0.4</td>
<td>0.6</td>
<td>0.25</td>
<td>0.25</td>
</tr>
<tr>
<td>1</td>
<td>1.3</td>
<td></td>
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</tr>
</tbody>
</table>

Aluminium 6061/boron fibre reinforced metal matrix composite

The material used in this series of experiments comprised nine alternating layers of boron and aluminium, (nine - ply). Boron filaments made by the deposition of boron on a thin tungsten wire, to produce a filament 1.4mm in diameter, were laid down on thin aluminium foil, alternating the layers of boron and aluminium. The
aluminium foil and boron fibres were then diffusion bonded to give an aluminium fibre reinforced composite.

**Aluminium 6061/silicon carbide particulate reinforced metal matrix composite**

The Al 6061/SiC MMC comprised a matrix of Al 6061 containing a uniform distribution of irregular SiC particles. The reinforcing particle size was in the range 3 - 10 m and formed 30% of the matrix total volume.
Fig. 3.1 Diagramatic representation of the blower assembly.

1. Bellows.
2. Post discharge cooler housing.
3. Steel main frame.
4. Vacuum ducting.
5. Anti-vibration mounts.
8. Blower raft.
Fig. 3.2 Layout of laser tubes and optics.
Fig. 3.3     Diagramatic representation of top hat beam profile.
Fig. 34 Diagramatic representation of the basic features of the 2kW Control laser
1. Laser beam
2. Mirror
3. Focusing lens
4. Focused beam
5. Gas assist
6. Workpiece
7. Nozzle
8. Lens height adjustment
9. Lens mount

Fig. 3.5 Diagramatic representation of the laser head and lens arrangement.
Fig. 3.6 2kW las r. front panel controls
Fig. 3.7 'Perspex char' print illustrating the energy distribution within the beam spot.
Fig. 3.8a  Diagramatic representation of 'Perspex-print' showing beam focal point.

Fig. 3.8b  Illustration of lens focal point adjustments.
Fig. 3.9  Schematic representation of the structure of the anodic film obtained from the sulphuric acid anodising process.
Fig. 3.10  Illustration of the front and rear views of the Permascope.
Fig. 3.11 Schematic diagram showing relationship between camera length and diffracted spot.
Chapter 4

Laser welding of Nimonics C263 and PE11

The chapter will be divided into four sections. The first two deal with the laser operating conditions, the experimental results, and their interpretation. The results are then discussed in relation to current knowledge.

4.1 Laser operating conditions

In this section an account of the operating conditions used for this research project will be presented in two sub-sections. In sub-section 4.1.1, an account of the procedure followed using the 5kW laser is given, while in sub-section 4.1.2 details the procedure followed for the 2kW laser are presented. This approach allows the differences in procedures to be highlighted as they have a major effect on the results observed.

4.1.1 The 5kW laser

The equipment and weld settings used in these experiments were as follows.

- Welding Institute 5kW continuous wave CO\textsubscript{2} laser.
- 300 mm focal length KCl lens.
- Focal setting '0', (i.e focal point on material surface).
- Full foot shielding using helium, (i.e top and bottom trailing shield).
d. Nimonic C263 and PE11 sheet of thickness 1.2 mm.

The arrangement and operation of this equipment is highlighted in the following points.

1. The laser was focused and set at a power of 4.0 kW.

2. A sheet of Nimonic PE11 approximately 30 cm by 10 cm was clamped on to the x-y table.

3. A piece of scrap metal was placed in front and behind the workpiece to prevent damage to the x-y table by an over run of the laser head.

4. The safety shields enclosing the laser heads were then shut.

5. The x-y table traverse speed was then set at 100 mm/s, the laser shutter opened and a melt track generated across the material.

6. The sheet was then unclamped, repositioned with respect to the laser to give a clearance of approximately 2 cm from the previous melt track.

7. The traverse speed was then reduced and a further melt track laid down.

8. Procedures 6 and 7 were then repeated until a full penetration weld had been obtained.

For the purpose of these trials a total of four laser powers were investigated; 4.0 kW, 3.5 kW, 3.0 kW, and 2.5 kW. The laser beam traverse speeds were as follows; 100 mm/s, 83 mm/s, 67 mm/s, 50 mm/s, and 33 mm/s. Not all speeds were considered for each laser power as the melt runs at a particular power were terminated as soon as a full penetration weld was obtained.

Having completed a series of runs, all the welds were examined both macro and microscopically. The following characteristics were noted.

a. Melt depth.

b. Melt top width.
c. Melt bottom width for full penetration welds.
d. Melt cross sectional areas.
e. Variations in dendrite arm spacings, and precipitation in the fusion zone.
f. Microhardness traverses were carried out across a number of welds to provide
   a guide to assessing the mechanical properties of the joint.

Photomicrographs of the melt tracks were also taken. All of the results obtained in
the trials are detailed in section (4.2).

4.1.2 The 2kW laser

In the experiments using the 2kW laser the equipment and the weld settings were
as listed below.

a. 2kW Control laser.
b. KCl lens - focal length 100 mm.
c. Argon top shielding.
d. Nimonic C263 - thickness 1.2 mm.

The above equipment was set up and operated in the manner highlighted briefly in
the following points.

1. The laser was switched on and the discharge allowed to stabilise before the
   experiments were commenced. This was done to prevent any fluctuations in
   laser power due to an unstable discharge.

2. "Perspex" prints were taken to determine the laser beam mode, and ensure that
   the beam was emerging from the nozzle without clipping it.

3. Using the "Perspex" print method described in section 3.2.2 a rough estimate
   was made of the beam focal point.
4. A sheet of Nimonic C263 was cleaned in acetone to remove any grease or marking paints which could lead to contamination of the weld. The cleaned sheet was then clamped onto the x-y table.

5. The laser beam was then focused on the surface of the workpiece using the approximate focus position determined from the "Perspex" prints. Fine tuning of the focus was carried out using the blue flash technique, (section 3.2.2).

6. Having determined the beam focus the x-y traverse speed was set in the x direction. The starting speed was set at 46.7mm/s.

7. The shield gas was then turned on and the extractor fan brought into close proximity with the workpiece to extract any fumes resulting from the welding process.

8. A piece of scrap metal was placed in front of the workpiece to prevent any damage to the x-y table if the beam were to over-run past the workpiece.

9. Finally the laser shutter was opened and the material traversed under the focused beam in order to generate a melt track across the workpiece.

10. At the end of the weld run, the sheet was unclamped and repositioned with respect to the laser to give a clearance of approximately 2cm from the previous melt track.

11. The x-y table speed was then increased and a further melt track made on the workpiece.

Steps 10 - 11 were repeated keeping the power constant at 1.4kW to obtain a relationship for the variation of weld depth and width with laser beam traverse speed.
Keeping the speed at 75mm/s the above procedure was repeated to give a relationship for the variation of weld depth and width with laser power. Finally a series of melt tracks were laid down using KCl lenses of focal lengths 75mm, and 100mm, at a power of 1.4kW. The laser beam traverse speed was varied for constant lens focal length and laser power. Measurements were taken to identify the variation of weld depth and width with laser beam traverse speed.

All the samples collected from this series of experiments were sectioned, polished and optically examined. Microhardness testing was carried out on the samples to provide a guide to the mechanical properties of the joint.

4.2 Results

Nimonic Alloy C263

Visual examination prior to sectioning revealed the welds to have an even top bead surface. All the welds were bright in appearance which would suggest that no oxidation had taken place during welding. This would indicate that the helium and argon shield gases both had adequate blanketing properties at the gas flow rates, laser powers, and welding speeds involved.

In welds where full penetration had occurred the bottom weld bead was uniform. In the welds made using the 5kW laser, there appeared to be a slight discolouration of the weld bottom bead. This is attributed to inadequate underbead shielding, which could occur if the gas flow rate to the underside of the weld was too low to displace air from the welds at the welding speeds involved. There were no visible signs of porosity or cracking in any of the welds.
Effects of laser power on weld penetration and microstructure

Fig. 4.1 clearly shows the effect of laser power on weld penetration, weld shape, and weld macrostructure of alloy C263. Fig. 4.2 shows similar trends obtained from welds made using the 2kW laser. It is evident from Fig. 4.1, and Fig. 4.2 that the weld depth is proportional to the laser power. The details are given in Figs. 4.3, and 4.4 for the 5kW and 2kW lasers respectively.

Increasing the laser power while keeping the beam traverse speed and the beam spot size constant results in an increase in the power intensity of the irradiating beam, which in turn results in an increase in the observed weld depth, (Figs. 4.3 and 4.4). However the results obtained for the 2KW laser, (Figs. 4.2 and 4.4), appears to indicate a decrease in weld penetration at the higher powers investigated, (ie 1.7KW to 1.9KW). This decrease in weld penetration can be accounted for in terms of an increase in plasma formation at high laser powers resulting in a corresponding increase in the absorption of the impinging laser radiation and subsequent dissociation of this plasma from the material surface. The consequence of this is a decrease in the power available for welding, hence the production of shallower welds.

To study the macro and microstructural changes in the heat affected zone, HAZ, and fusion zone of the weld, optical, scanning, and transmission electron microscopy techniques were used. The variation of the macro and microstructure for a typical weld in Nimonic C263 is shown in Fig. 4.5. Fig. 4.5a shows the overall weld macrostructure, while Fig. 4.5b shows the base metal structure, cold worked, and solution treated. From Fig. 4.5c showing the fusion line, it was obvious that the heat affected zone width was negligible, and that there was no noticeable grain growth in the base metal grains. However there is a clear epitaxial
relationship between the weld metal grains and the adjacent base metal grains, and the change in structure across the fusion line is quite abrupt. No grain boundary liquation or cracking was evident along the fusion line, although a certain number of grains appeared to have undergone a degree of grain boundary coarsening, (cf. Fig.4.5d, and 4.5e).

The fast cooling rates associated with the laser welding process facilitated the formation of a columnar dendritic structure in the fusion zone. A longitudinal macrograph, Fig. 4.6a shows columnar grains growing from the parent metal into the weld centre. The weld centre itself exhibited a columnar dendritic structure within which interlocking columnar dendrites were clearly evident. The welds in Nimonic C263 examined showed no signs of weld centreline cracking.

Measurement of the dendrite arm spacings revealed them to be between 2 and 5 microns, Figs 4.6c, and 4.6d clearly show the effect of laser power on dendrite arm spacings. The effect of laser power on carbide precipitation is shown in Fig. 4.7. Semi-quantitative EDS, (energy dispersive spectrum), analysis of the weld and adjacent regions produced results documented in Table 4.1. From these results it would appear that a slight degree of microsegregation of Ti to the interdendritic regions of the weld occurred during solidification. This would suggest increased precipitation of carbides and gamma-prime in these regions of the weld.

The effect of laser beam traverse speed on weld penetration and weld bead top width.

During the welding experiments the weld penetration decreased as the laser beam traverse speed increased. The welds made using the 5kW laser and the 2kW laser obeyed this general trend as can be seen from the results in Figs. 4.1, and 4.2. Figs. 4.8, and 4.9 show the same results expressed quantitatively. A higher beam
traverse speed limits the beam interaction time at any point on the workpiece resulting in a decrease in the energy input into the substrate, Figs 4.10a to d demonstrate the effect of laser beam traverse speed on dendrite arm spacings, and carbide precipitation in the fusion zone. It is obvious that the dendrite arm spacings increase as the laser beam traverse speed is decreased, similarly the amount of precipitate formed in the interdendritic regions of the welds is also observed to increase.

From the results obtained in the welding of alloy C263 using both the 2kW laser and the 5kW laser it became obvious that the weld top bead width decreased as the beam traverse speed increased, (Figs.4.11, and 4.12). A high beam traverse speed limits the beam interaction time, thus restricting the energy input into the substrate and resulting in an increase in the rate of solidification. As a direct implication of this fact there is less time available for conduction processes to result in lateral transfer of energy in the workpiece.

**The effect of lens focal length on weld penetration**

The relationship between weld penetration and lens focal length is shown in Fig. 4.13 as a function of laser beam traverse speed. The predetermined limiting factor in these series of experiments was the material thickness. During experimentation, for a given power the beam traverse speed was decreased until a full penetration weld was obtained. The limiting speed for welding 1.2 mm thick material was thus determined for the preset laser parameters (laser beam power, spot size and shield gas). A greater degree of weld penetration was observed in weld runs using the 75mm focal length lens.
Consider equation (4.1) for the focused beam diameter;

\[ D_f = \frac{4\lambda f}{\pi D_0} \quad \text{--------(4.1)} \]

where \( D_f \) = focused beam diameter, \( \lambda \) = laser wavelength, \( f \) = lens focal length, and \( D_0 \) = unfocused beam diameter. It is evident that for a constant beam wavelength and beam diameter exiting from the laser cavity the focused spot diameter would be proportional to the lens focal length. It therefore follows that the shorter lens focal length would result in a higher power density, and hence a higher energy distribution within the focused beam spot, leading to a greater degree of weld penetration. Fig. 4.14 shows the effect of lens focal length on weld bead top width.

Microhardness traverses were taken across the welds, and into the parent metal to give an indication of the mechanical properties of the joint. Representative traces obtained for Nimonic alloy C263 are shown in Figs. 4.15a and 4.15b respectively.

Derived results relating to the laser welding of Nimonic alloy C263 with a 2kW and a 5kW laser

The melt area achieved during the laser welding trials was measured using the Quantimet 800. From these results the volume of melt produced during a beam traverse was calculated. The results obtained from these calculations are presented as a function of laser power, and laser beam traverse speed in Figs. 4.16, and 4.17, and 4.18, and 4.19 for the 5kW and the 2kW laser respectively. It is evident that as the laser power is increased the melt volume increased. A decrease in beam traverse speed was observed to bring about a similar effect. This was true for both the 5kW laser and the 2kW laser. It was however observed that the trend was for
the melt volume produced by the 5kW laser to exceed that produced by the 2kW laser.

From the measured laser beam traverse speed, laser power, and weld penetration, it was possible to determine the welding efficiency achieved during these experiments for both the 5kW, and the 2kW laser.

Welding Efficiency = \text{Welding Speed} \times \frac{\text{Thickness of Material Penetrated}}{\text{Laser Power}}

(4.2)

This derived data is presented in a series of graphs shown in Figs. 4.20 to 4.22. Figs. 4.20, and 4.21 indicates the dependence of welding efficiency on laser beam traverse speed for the 5kW laser, and the 2kW laser respectively. These results show that for a given laser power the welding efficiency increases to a maximum at which sufficient energy is supplied to the workpiece to facilitate melting and not be available for general conduction into the bulk metal. Above this speed insufficient energy is supplied to the workpiece, and a 'fall-off' is experienced in the observed welding efficiency. With the 5kW laser the maximum efficiency was achieved with the laser operating at a power of 3.0kW, and a speed of 67mm/s. The results from the 2kW laser show the welding efficiency to increase sharply initially as the beam traverse speed is increased. This is followed by a slowing down of the increase at about 69mm/s. Efficiency peaks at 98mm/s after which a sharp drop occurs. Plotting both sets of data on a similar axis in Fig. 4.22 allows a comparison to be made between the welding efficiency achieved with the 5kW laser and the 2kW laser. Fig. 4.23, and 4.24 shows the welding efficiency as a function of laser power at a constant beam traverse speed.
By plotting values of weld penetration versus a derived compound parameter, \( I = \frac{P}{vD} \), the synergistic effect of the laser welding parameters; power, speed, and beam diameter can be quantified. It is immediately obvious that weld penetration increases as \( \frac{P}{vD} \) increases. This is shown for both the 2kW laser, and the 5kW laser, Figs. 4.25a, b, and c, and Fig. 4.26 respectively. Substituting values of \( D \) for the 2kW laser, and the 5kW laser, \( \frac{P}{vD} \) for these conditions could be calculated. Plotting this data for the 75mm focal length lens and the 100mm focal length lens on the same axis it is evident that a line of constant slope is obtained, Fig. 4.25c. Combining the data in Figs. 4.25c, and 4.26 on a similar axis in Fig. 4.27, it can be seen that the data from both lasers display a characteristic slope, unique to the two individual systems.

**Nimonic Alloy PE11**

As will be seen the results obtained for alloy PE11 were similar to those obtained for alloy C263. All the welds were made using the Welding Institute 5kW continuous wave CO\(_2\) laser. It was possible using the results obtained from this work to determine the effect of material composition on the laser weldability of an alloy, by comparing the results obtained in the welding of Nimonic C263 and PE11 using the same laser system.

The weld bead tracks produced in Nimonic PE11 displayed the same degree of integrity as those produced in C263. Most of the welds were bright but some top and bottom beads did show a slight degree of discolouration.
Effects of laser power on weld penetration and microstructure

Fig. 4.28 shows the effect of laser power on weld penetration, weld shape, and weld microstructure of alloy PE11. These results are represented graphically in Fig. 4.29. It is evident from Figs. 4.28, and 4.29 that the weld depth increases with increasing laser power. This is similar to the result obtained in the welding of alloy C263. Thus the observed trend can be accounted for in terms of the increasing the laser power intensity of the irradiating beam at a constant beam traverse speed while the beam spot size remains constant.

Fig. 4.30 shows the macrostructure, and the microstructural variation of the joint for Nimonic PE11. Fig. 4.30a shows the overall weld macrostructure. The base metal exhibits irregularly shaped grains, in the cold worked and solution treated condition, (Fig. 4.30b). From Fig. 4.30c showing the fusion line, no substantial heat affected zone can be distinguished. Weld metal solidification occurred epitaxially from the base metal grains, with the fusion zone exhibiting a columnar dendritic structure perpendicular to the direction of the heat source, Fig. 4.30d. The weld metal dendrite arm spacing was found to be between 2 and 5 microns, Fig. 4.31a, and 4.31b show the effect of laser power on dendrite arm spacings in Nimonic alloy PE11. The dendrite arm spacings can be seen to increase as the laser power is increased. Optical analysis of the weld zone revealed carbide precipitation in the interdendritic region of the welds, (Fig. 4.32), it is obvious that the precipitation in the PE11 welds exceeds that observed in the alloy C263 welds, (cf Fig. 4.7). The carbide precipitation was also observed to be greater the higher the incident laser power. There was no evidence of cracking or liquation in the interdendritic regions of the weld.
Certain microcracks were observed in the region of the fusion line of the PE11 welds, these microcracks propagated approximately perpendicular to the fusion line and did not extend into the fusion zone of the welds, (Fig. 4.33). SEM micographs of the fusion line revealed grain boundary coarsening compared to the base metal grain boundaries, (Figs. 4.34a, and 4.34b). The microstructure was observed to exhibit a higher number of titanium precipitates compared to that observed in alloy C263, these precipitates were approximately 5 microns in size, Fig. 4.35. Despite the increased presence of these precipitates SEM analysis of the fusion line did not reveal any evidence of grain boundary liquation.

The effect of laser beam traverse speed on weld penetration and weld bead top width

During the welding experiments the weld penetration increased with decreasing laser beam traverse speed, (Fig.4.28). As with alloy C263 a higher beam traverse speed limits the beam interaction time at any point on the workpiece resulting in a decrease in the energy input into the substrate. Fig. 4.36 shows the observed results expressed qualitatively. There was a bit of overlap in the data. It would appear that an increase in power from 3.0 to 3.5kW does not result in any great increase in weld penetration, suggesting that for this range of powers the laser beam traverse speed is the limiting factor in determining weld penetration. Figs. 4.37a to d demonstrate the effect of laser beam traverse speed on dendrite arm spacings, and carbide precipitation in the fusion zone. It is obvious that the dendrite arm spacings increase as the laser beam traverse speed is increased, similarly the amount of precipitate formed in the interdendritic regions of the welds is also observed to increase.
While the weld bead top width appeared to decrease with increasing laser beam traverse speed, this was only slightly evident for the welds made at 2.5kW. Certainly there is again no great difference in measurements for the welds made at 3.0kW, and 3.5kW, Fig. 4.38. As stated in Chapter 2, a high beam traverse speed limits the beam interaction time, thus restricting the energy input into the substrate and resulting in an increase in the rate of solidification. Thus there is less time available for conduction processes to result in lateral transfer of energy in the workpiece.

Microhardness traverses were taken across the welds, and into the parent metal to give an indication of the mechanical properties of the joints. Representative traces obtained for Nimonic alloy PE11 are shown in Figs. 4.39a, and 4.39b.

**Derived results relating to the laser welding of Nimonic alloy PE11**

Measurements from the Quantimet 800 gives an indication of the melt area achieved in the welding of alloy PE11. Figs. 4.40, and 4.41 show this data as a function of laser power, and laser beam traverse speed respectively. It was observed that the trend was for the melt volume produced in Nimonic alloy PE11 to exceed that produced in alloy C263.

From the measurements taken from the sectioned weld samples of Nimonic alloy PE11 it was possible to calculate the welding efficiency achieved in the welding of the alloy. The welding efficiency was calculated according to equation 4.2. Fig. 4.42 shows the welding efficiency as a function of laser power, while Fig. 4.43 shows the data presented as a function of laser beam traverse speed for all the laser powers studied. It would appear that for a laser power of 4.0kW, and 3.0kW welding efficiency increases as laser beam traverse speed increases, and the peak at
which a 'fall-off' in welding efficiency is observed, as was the case with Nimonic alloy C263 is not reached within the range of laser beam traverse speeds investigated in this work. At 2.5kW a maximum is reached above which welding efficiency starts to fall, while at 3.5kW the response was relatively constant with increase in beam traverse speed.

4.3 Discussion

The main aim of this work was to investigate the laser weldability of Nimonic C263 and PE11 with a continuous wave carbon dioxide laser. Having successfully welded the materials, an attempt was made to interpret the observed microstructure in the joint region in terms of known characteristics of the laser welding process. This microstructure, together with microhardness tests on the weld region were used as a basis on which to comment on the mechanical properties of the laser welds in these materials.

This discussion will be divided into three main sections. The first deals with a comparison between the welding response observed using the 5kW, and the 2kW laser. This is followed by comments on the effects of the laser welding process on the microstructure and mechanical properties of the welds in alloys C263, and PE11. Finally the effect of incident laser beam power, laser beam traverse speed, and lens focal length on the welding of the alloys is considered.

Comparison between the welding with the 5kW and the 2kW lasers

Plotting a graph of weld penetration vs P/vD revealed that weld penetration increases as P/vD increases, for both the 5kW, and the 2kW laser respectively,
Figs. 4.25a, b, c, and 4.26. This compound parameter allows the synergistic effect of the basic laser welding parameters on welding performance to be investigated. Presenting both sets of data obtained for the 2kW laser using the 75mm, and the 100mm lenses on a graph, Fig. 4.25c, it was observed that a line of constant slope was obtained. This suggests that for a particular laser altering the focal length of the lens does not alter the slope of the line for weld penetration versus P/vD. Plotting the data in Fig. 4.25c, and 4.26 on the same axis, two lines of distinct slope are obtained for the welds made using the 2kW laser, and those made using the 5kW laser respectively. Thus a laser can be characterised by the graph of weld penetration versus P/vD. In an ideal system where all lasers operated with the same characteristics, the graph of P/vD for all lasers would fit on the one line. This is not the case and individual lasers have a slope characteristic to their mode of operation.

Considering the welds made using the 5kW laser the following became evident for Nimonic C263.

<table>
<thead>
<tr>
<th>Power (kW)</th>
<th>Limiting speed for full penetration (mm/s)</th>
</tr>
</thead>
<tbody>
<tr>
<td>4.0</td>
<td>83</td>
</tr>
<tr>
<td>3.5</td>
<td>67</td>
</tr>
<tr>
<td>3.0</td>
<td>67</td>
</tr>
<tr>
<td>2.5</td>
<td>50</td>
</tr>
</tbody>
</table>

Above these speeds for the given powers incomplete penetration was obtained in 1.2mm thick material.
For a power of 1.4kW the limiting speed for full penetration using the 2kW laser was found to be 64mm/s. This would suggest that welding with the 2kW laser at 1.4kW was more efficient than welding with the 5kW laser at 2.5kW. This fact is clearly demonstrated in Fig. 4.22, which shows the welding efficiency achieved using both lasers as a function of laser beam traverse speed. A plot of P/vD at 2.5kW for the 5kW laser versus weld depth on a similar axis as the plots for the 2kW laser, Fig. 4.27, however appears to suggest otherwise. The apparent shift of the line for the 5kW laser to the left is a consequence of the higher laser power forming the P term in the parameter P/vD.

Comparing the results obtained from welds using both lasers the observations recorded from the welding studies must be considered in terms of the welding parameters used in both cases.

<table>
<thead>
<tr>
<th>LASER</th>
<th>5kW</th>
<th>2kW</th>
</tr>
</thead>
<tbody>
<tr>
<td>POWER</td>
<td>2.5kW</td>
<td>1.4kW</td>
</tr>
<tr>
<td>LENS FOCAL LENGTH</td>
<td>300mm</td>
<td>100mm</td>
</tr>
<tr>
<td>SHIELD GAS</td>
<td>Helium</td>
<td>Argon</td>
</tr>
<tr>
<td>WELDING SPEED</td>
<td>50mm/s</td>
<td>64mm/s</td>
</tr>
</tbody>
</table>

The consequence of using a 300mm focal length lens is that the size of the incident laser beam spot would be greater than that obtained with the 100mm focal length lens used with the 2kW laser. Assuming all other variables in equation 4.1 were constant, and substituting measured values of D_f as follow; D_f 2kW = 0.4mm, D_f 5kW = 0.7mm, the resultant power, and energy intensity obtained for the 5kW laser would be less than that obtained for the 2kW laser.

The width of the focused beam incident on a material surface can however be estimated by considering the top bead width of the resultant weld(3r), as the width
of the weld cannot be smaller than the size of the beam incident on the material
surface. Thus an estimate of the differences in laser beam widths for the 2kW and
the 5kW laser can be obtained by comparing welds produced at similar speeds on
the workpiece.

From the observations of welds made during the experiments the following
measurements were taken.

<table>
<thead>
<tr>
<th>SPEED (mm/s)</th>
<th>POWER (KW)</th>
<th>PENETRATION (microns)</th>
<th>BEAD WIDTH (microns)</th>
</tr>
</thead>
<tbody>
<tr>
<td>5kW</td>
<td>66.7</td>
<td>2.5</td>
<td>640</td>
</tr>
<tr>
<td>2kW</td>
<td>64</td>
<td>1.4</td>
<td>1200</td>
</tr>
</tbody>
</table>

It is believed that as the same material was used in these experiments, (nimonic
C263), the thermal conductivity, (11.7W/mK), specific heat, (461J/KgK), and
melting range, (55K) would be similar in both cases allowing a good estimate of the
difference in beam widths for both lasers to be made. The results documented
above give an approximate idea of the difference in the effective spot size of both
the 5kW and the 2kW lasers. It can therefore be deduced that the effective spot
size of the 2kW laser covers an area 78% that of the 5kW laser under the conditions
detailed above.

From the results obtained on the effect of lens focal length on weld bead top widths
using the 2kW laser, it was found that there did not appear to be a large difference
obtained in the weld bead top widths for the 100mm focal length lens compared to
the 75mm lens, (Fig. 4.14). However there was a marked increase in weld
penetration achieved with the 75mm focal length lens. As the incident power
intensity is proportional to \(1/D\), (cf. \(I=P/vD\)), the energy intensity which depends on the beam interaction time \(D/v\) is proportional to \(1/D^2\) for a similar power, where;

\[
I_e = \frac{P \times v}{vD^2} \quad \text{(4.3)}
\]

Thus the 300mm focal length lens could be expected to result in a power intensity reduction, \(I_P\), which is proportional to an inverse of the increase in beam spot size, and an energy intensity reduction, \(I_e\), which is proportional to an inverse of the square of the increase in beam diameter compared to the 100mm focal length lens. Experiments by Seaman(14) showed that the effective spot size, i.e the area of the resultant heat affected zone within the processed material can be more than twice the theoretical optical spot size, defined by;

\[
D_f = f \theta \quad \text{(4.4)}
\]

While the focus spot dimensions determine the energy density incident on the material surface, the mode structure of the laser has a large effect on the spatial distribution of this energy, thereby determining how it is available for interaction with the workpiece surface.

Beam modes were described in Chapter 2, and Powell et al (87),(88) in recent publications investigated the effects of laser beam modes on laser cut quality. The beam mode describes the energy density cross section of the laser beam. This can be easily but only approximately determined by exposing a sheet of acrylic to the unfocused beam for a few seconds. As the beam irradiates the material the acrylic is rapidly evaporated at a rate determined by the local energy density. Both the lasers used in these experiments were fast axial flow lasers of the stable resonator design, Fig. 2.5, which generate TEM\(_{00}\) modes, Fig. 4.44. In these lasers, a number of factors can affect the output beam mode. These include: vibration,
optical component quality, and local variation of gas pressure in the cavity\(^{(32)}\).

According to Powell et al\(^{(88)}\), a fast moving gas within the laser cavity is subject to variations in pressure by virtue of the fact that it is being pumped at high velocity, \((> 200\text{m/s})^{(37)}\), around a complex three dimensional system, changing direction rapidly in the laser cavity. Shock waves can also be set up corresponding to high and low pressure zones. The refractive index of a gas is related to its density which is in turn determined by its pressure, thus local areas of high or low pressure in the laser cavity act as areas of different refractive index. Thermal variations in the laser cavity can also result in variations in the refractive index. These local variations in temperature and pressure cause areas within the lasing gas to act as complex prisms to the light passing through them, distorting the original mode of the laser beam.

The theory of lasers shows that actual Gaussian modes are generated only at the lasing threshold, i.e at the internal energy condition at which lasing will only just take place. All industrial machines operate at much higher energy levels, it is therefore very unlikely that the laser beam mode used for welding was a pure Gaussian mode, and Fig. 4.45 showing a representative perspex beam print from the 2kW laser clearly reveals this. It is evident that the beam is not a pure Gaussian type. In the 5kW laser larger volumes of gas are present in the lasing cavity increasing the possibility of irregularities in gas pressure and temperature, the resulting distortion of the Gaussian beam mode is therefore expected to be greater. Many corrupted Gaussian beams have been shown to loose the central peak of energy, and therefore present a flatter profile compared to a pure Gaussian beam, Fig. 4.46. As a rough approximation this flatter profile can be taken to be a mixture of a Gaussian beam, and a uniform flat topped energy profile, i.e a beam of uniform average energy across its cross section, or a 'top hat' mode.
The general reduction in energy density caused by a lateral spread in the beam mode, and its resultant inferior focusing characteristics can result in a drop in the effectiveness of energy transfer from the laser beam to the workpiece accounting for the loss in efficiency observed in comparing the welding performance of the 5kW laser to the 2kW laser.

In view of the preceding facts it can be argued that the beam mode produced in the 2kW laser was more favourable for welding the 1.2mm thick Nimonic C263. This is supported by the measurements and observations made using both machines.

Another variable to be considered in these welding experiments is the shield gas, (cf. page26). A helium full foot shield was used with the 5kW laser which implies a top and bottom trailing gas shield protecting both sides of the weld from oxidation. Argon top shielding was used with the 2kW laser. The difference in the choice of shield gas used was a matter of coincidence as both lasers were set up to use these particular gases. Apart from protecting the weld from oxidation the shield gas has a direct effect on the welding process itself as the laser beam has to pass through the gas above the workpiece before welding can occur. Thus there are two requirements for a good shield gas, and these include good transmission properties and good blanketing properties(36).

Helium is a gas of low atomic number and low density, and high ionisation potential, and for this reason it is not easily ionised to form a plasma. It is believed that plasmas formed by shield gases absorb the laser radiation thereby resulting in a reduction in weld depths obtained for a given laser power. Helium is generally used for its high ionisation potential in order to reduce this plasma reaction and obtain better transmission of the laser radiation. Experimental work carried out by
Seaman on a study of various shielding gases concluded that helium had better transmission properties as compared to argon. However observations carried out on the welds made in the course of this work show a greater degree of welding efficiency using the 2kW laser. This suggests that the weld penetration shows a greater dependency on focused laser beam spot size, and beam mode than it does on the shield gas used during welding. This deduction is made in the light of the fact that despite the higher power available for welding, and the use of helium which is supposed to favour transmission of the laser beam energy, the welds produced with the 5kW laser operating at 2.5kW penetrated the full thickness of the Nimonic sheet at a lower limiting speed than that observed for the 2kW laser operating at 1.4kW. It is however not possible within the scope of this work to carry out a full comparison of the effects of helium or argon on weld penetration. During welding the helium gas flow rate appeared to give satisfactory shielding of the weld top surface, as oxidation of the weld bead was prevented. Making a choice of shielding gas, factors such as the susceptibility of the material being welded to oxidation as well as the reflectivity, specific heat, thermal capacity and conductivity of the material would have to be taken into consideration as these factors would determine how readily the shield gas would ionise forming the weld penetration limiting plasma.

In the welding with the 5kW laser a 300 mm focal length lens was used, and following careful consideration of the results obtained through the series of experiments, it is believed that for the material thickness being studied this was not an ideal choice. Before welding was carried out the focal point of the laser was set in a standard position used in operation. Setting the focal spot of the laser at the standard setting was feasible bearing in mind the fact that there exists a finite region, of length d, known as the depth of focus over which the focused spot remains nearly constant. If this point is set at the x-y table level the assumption can
be made that it will either lie on the surface of the material being processed or within it. Having the focal spot set just below the surface of a workpiece is another condition in which successful processing can take place (74). The size of the focused spot and the depth of focus, d vary in direct relation to one another, for this reason the use of a large focal length lens in a commercial machine facilitates a high tolerance to problems encountered during welding. These include non flat materials, focal position movement due to effects such as thermal distortion of the lens, and materials of varying thicknesses. However, the greater the focal length of the lens the larger the focused spot size. This leads to a corresponding decrease in the power, and energy intensity within the incident beam spot. A decrease in incident beam energy results in a corresponding decrease in weld penetration. In work by Powell et al (89) using a 127mm focal length lens, an effective depth of focus of approximately 5mm was recorded. It can therefore be assumed that the depth of focus for the 300mm lens would be in excess of this value. These authors suggest that in the cutting of steel in the thickness range 1 to 8mm no indications exist to focus the laser other than at the material surface. The loss in beam quality, and increase in beam diameter outways the benefits of increased tolerances afforded by a large depth of focus in the welding of thin gauge materials.

In these welding experiments bead on plate weld were made, for this reason the only implication of the focused beam spot size has been its effect on the power density available for processing. In the joining of two pieces of metal it might prove beneficial to use a slightly larger beam spot size to allow for greater joint 'fit-up' tolerances. This decision must however be taken after considering the effects of the lower power intensity in the beam spot, and the fact that a larger heat affected zone would be obtained with its consequent effects on the final properties of the welded structure.
In a machine regularly used to weld a variety of materials, of different thicknesses, the advantages of a large incident beam spot and the accompanying large depth of focus cannot be disregarded. The focal spot diameter varies only slightly within the constraints of the depth of focus, and allows for flexibility of operation. However where thin materials are being welded, the aspect ratio of the weld, i.e. depth to width ratio must be taken into consideration. Too low a depth to width ratio would result in a loss of one of the main advantages of the laser welding process, i.e. narrow heat affected zones.

Comments on the effects of the laser welding process on the microstructure and mechanical properties of the welds in alloys C263 and PE11

The fast cooling rates, up to $10^4$ °C s$^{-1}$, associated with the laser welding process facilitated the formation of a columnar dendritic structure in the fusion zone in alloys C263, and PE11 as opposed to the recrystallised structure of the base metal. A longitudinal macrograph, Fig. 4.6a. shows columnar grains growing from the parent metal into the weld centre. The weld centre itself exhibited a columnar dendritic structure within which interlocking columnar dendrites were clearly evident. The consequence of this solidification substructure could be the formation of a plane of weakness at the centre of the weld bead where two solidification fronts from opposite sides of the weld impinge. The welds in nimonic C263 and PE11 examined however showed no signs of weld centreline cracking. In Figs. 4.5b, and 4.28b the grains in the zone, (HAZ), adjacent to the fusion boundary are observed to be very similar to those in the base metal. This suggests the HAZ, where grain coarsening, or refining occurs is very narrow. Figs. 4.5b, and 4.28b also show an abrupt change in microstructure across the fusion line. It is not clear...
what effect this abrupt change would have on the strength of this region of the weld. The narrow HAZ would generally result in good mechanical properties in many metals and alloys (54), and should be an advantage in determining the mechanical strength of alloys C263, and PE11.

Dendrite arm spacings were small compared to those measured for conventional welding processes, and were similar for both alloy C263, and PE11. Measurements of the dendrite arm spacings in both alloys appear to indicate the dendrite arm spacings decrease with a decrease in laser energy input into the weld, Figs. 4.6c, 4.6c, 4.10a, 4.10b, 4.31a, 4.31b, 4.37a, and 4.37b. The very low spacings, ranging from 1.5 to 5μm, are evidence of the high cooling rates involved in the laser welding process. This is in contrast to a mean dendrite arm spacing of about 5 μm recorded by Jahnke (89) for GTA, (gas tungsten arc), welds. Katayama, and Matsunawa (54), studying the solidification microstructure of laser welded stainless steel used a correlation between the primary dendrite arm spacing and the cooling rate to estimate the rapid cooling rates associated with the laser weld. For dendrite arm spacings in the range 0.5 to 2.2μm the cooling rates were extrapolated to be approximately 5x10^4 to 5x10^6°C/s which are over 100 times faster than cooling rates characteristic of conventional welding processes.

It is clearly evident from the micrographs that the weld microstructure is similar in both alloys C263 and PE11. Microstructurally, during solidification γ and the metal carbides will precipitate out of solution. The fine grain structure and fast cooling rates result in a fine precipitate of carbides in the interdendritic regions of the weld, (Figs. 4.7, and 4.32). Cooling rates do not allow for coarsening of the precipitates. SEM, and TEM analysis of the weld regions did not reveal any γ in this region. γ usually precipitates out of solution in the temperature range between
700 and 800°C. At the high temperatures reached in the weld region all the \( \gamma \), and carbides are taken into solution, with the carbides precipitating out on solidification. As the welds were not given any post weld heat treatment, it is likely that all the \( \gamma \) would remain in solution. The carbide precipitation observed in the weld region is much finer than that present in the base metal. Carbides play an important role in the determining the mechanical properties of Nimonic alloys by preventing grain boundary sliding, and acting as obstacles to dislocation movement, and crack growth in the event of ductile failure. However the effect of these carbides on the material resistance to brittle fracture is not clear. As the particles are discontinuous they may not adversely affect the material behaviour as they would not provide a continuous path for crack growth. It is therefore proposed that while the \( \gamma \) may remain in solid solution, the carbides are capable of providing strength in the weld at least equal to that of the base metal.

A series of microhardness tests carried out across the weld and into the base metal for both alloys C263, and PE11 revealed that the trend was for the fusion zone to show a hardness value comparable to the base metal in alloy PE11, and greater than the base metal in alloy C263. It is believed that this is a consequence of the increased carbide precipitation in alloy C263 compared to the base metal, Figs. 4.7, while in alloy PE11 although the weld shows an increase in carbide precipitation, the base metal also exhibits a considerable amount of carbide precipitates, Figs. 4.32. As the strength of nickel base alloys, increases with increase in hardness, it is proposed that the welds in alloy PE11 would have a strength at least comparable to that of the base metal. Alloy C263 on the other hand could exceed the base metal in strength. Results by Russell(74), and Ricciardi(83) appear to support this indication. The lower degree of precipitation evident in Nimonic C263 is believed to be a consequence of the effect of Co in increasing the solubility of C in the alloy C263 matrix. There was also a higher incidence of titanium precipitates, Ti(C, N),
in alloy PE11, these precipitates were approximately 5\(\mu\)m in size.

While Russell et al\(^{(74)}\) carried out a study on the tensile and fatigue properties of alloy C263 at room temperature, 700°C, 800°C, and 900°C, and based their conclusions on the results of two tests at each temperature, the authors recorded that with one exception all of the specimens failed through a gauge mark. Although the sample tested at 900°C failed in one case at the weld edge it is not possible to determine from the evidence presented where the weakest portion of the joint is in these welds, and to accurately quantify the elongation at failure of the test pieces. The fatigue behaviour of the welds in alloy C263 was however reported to be comparable to that of welds made by established processes such as electron beam welds, and TIG welds in materials of similar thickness. While from the literature it is evident that the authors considered what they described as a good profile from a fatigue point of view, it is not clear if any attempt was made to quantify the weld microstructure, and precipitation characteristics in their study on tensile and fatigue properties.

The effects of incident laser beam power and laser beam traverse speed on the microstructures of the alloys was also investigated, and the results documented in micrographs in section 4.2. The laser welding process is characterised by a high energy low heat input similar to that encountered in electron beam welding. Heat is transferred to the workpiece via a 'keyhole' mechanism, and temperatures within the "keyhole" have been reported to reach 25000°C\(^{(32)}\). Thus the resultant cooling rates on welding are extremely high. Macrostructurally the most obvious effect of laser power on the welds is evident in the change in weld shape as the laser power is increased for a constant laser beam traverse speed, and incident beam diameter. The 'keyhole' nature of the welding process was evident for a number of welds in
both Nimonic C263 and PE11, (Figs. 4.1,4.2, and 4.28), and some welds clearly show the transition from conduction limited welds, i.e welds in which the mode of heat transfer is by conduction from the surface into the bulk metal, to the 'keyhole' welds which exhibit a characteristic 'wine glass' shape.

The actual temperatures reached within the melt zone during welding depends to a large extent on the combination of laser welding parameters; laser beam power, laser beam traverse speed, lens focal length, and shield gas. In the course of these studies the effect of shield gas on the transfer of energy to the workpiece, by virtue of its effect on the laser induced plasma above the workpiece surface, or within the 'keyhole' were not investigated in detail. Much work has been carried out on the subject, and it is known that the shield gas plays an important role in laser welding. As the laser welding parameters determine the energy input into the welds, they in turn would have an influence on the solidification of the weld metal, and ultimately the mechanical properties of the weld.

Porosity and cracking

The welds produced in both Nimonic C263, and PE11 showed no porosity. This was contrary to observations made by Russell et al(76), who reported small round pores in some C263 welds. Except for a couple of welds made with the 2kW laser, the non-full penetration welds in alloy C263, and PE11 were free of porosity at the root of the weld. In alloy PE11 a certain amount of intrusions into the base metal from the weld fusion zone was evident. These cracks propagated approximately perpendicular to the fusion line and did not extend into the fusion zone of the welds, Fig. 4.33. SEM micrographs of the fusion line revealed grain boundary coarsening compared to the base metal grain boundaries, Figs. 4.34a,
and 4.34b. It is possible that these microcracks occurred as a result of either one or a combination of the following factors; large shrinkage stresses during cooling, and the inherent low ductility and high strength of the grains compared to the intergranular properties at high temperature. Evidence of the high degree of shrinkage associated with the laser welding of alloy PE11 is believed to be found in the fact that a number of welds in this alloy appeared to display a sunken weld top bead, (cf. Fig 4.28). Shrinkage of the molten metal on solidification could have resulted in insufficient metal being available to fill the gap created by the passage of the laser beam, while the stresses generated by the contraction in the weld region on solidification caused the observed microcracking along the fusion line.

To provide an understanding of the different response of alloys C263, and PE11 to the laser welding process, consider Tables 4.3 and 4.4 which show the nominal compositions of the alloys. From these tables it is evident that both alloys contain a number of elements that are known to be detrimental to welding by virtue of their effect on the susceptibility of the final welded structure to cracking. Some of these elements are silicon, boron, zirconium, sulphur, lead, titanium, and aluminium. Si, B, Zr, and Pb are known to result in hot shortness and HAZ cracks. S present in the metal without the correct amount of Mg to counteract its embrittling effects can cause severe cracking in the weld metal or HAZ, the amount of Mg required to counteract the effects of S being determined by the following equation (84);

\[ Mg = (5*S) + 0.02 \]

Finally Ti and Al can increase cracking by virtue of their effect on precipitation hardening.

It was observed that while C263 and PE11 contained most of the elements detailed above the incidence of cracking was only recorded for PE11 welds. Referring back to Tables 4.1 and 4.2, it can be seen that PE11 contains a slightly higher
percentage by weight of Si, Ti and Al. While it does contain an equal amount of B, and a lower percentage of Pb, it does contain Zr and S which are not present in alloy C263. It is suggested that combined with the absence of any Mg which would limit the detrimental effect of S the above factors result in alloy PE11 being more susceptible to cracking than alloy C263.

It is believed that the cracking observed in these welding trials can be minimised by ensuring a slow thermal cycle exists in welding to provide enough time for the bulk structure to accommodate the increased strains involved on solidification. This can be achieved by decreasing the welding speed, failing any improvement, the use of preheating prior to welding\(^{(84)}\) may provide an alternative solution to the problem. An increase in the weld thermal cycle time would also result in more melt, and time available for any cracks that develop to heal by liquid flow.

**The effect of incident laser power, laser beam traverse speed, and lens focal length on the welding of Nimonic alloys C263 and PE11**

It was found that in both Nimonic C263 and PE11 weld penetration increased with increase in incident laser power. As no reported laser welding studies were found for alloy PE11, this work has proved the material to behave in a manner characteristic of other materials welded by a laser, Figs. 4.3, 4.4, and 4.29. Both alloys welded also showed a decrease in maximum depths of penetration recorded as the laser beam traverse speed was increased, Figs. 4.8, 4.9, and 4.36. This was again characteristic of previous observations noted in laser welding studies.

It was shown that as the laser beam traverse speed increased the general indication was for the weld bead top width to decrease in alloys C263. This is believed to be
a consequence of the shorter beam interaction times at any point on the material surface as beam traverse speed increases. The interaction time, $t$, in laser processing can be expressed as $D/v$, it is clear that as $v$ increases while $P$ and $D$ remain constant, a reduction in power intensity $I_p$, and interaction time, $t$, will occur. This reduces the time available for conduction processes to occur, and facilitates the observed decrease in weld top bead widths. The decrease in laser material interaction time at any position along the weld that occurs at the higher beam traverse speeds is also believed to be the governing factor leading to the production of shallower welds. The energy input into the workpiece would be limited as interaction times decrease resulting in shallow welds. Microstructurally the implication of an increase in laser beam traverse speed was similar to that observed with a decrease in laser power, with a refinement of the weld microstructure occurring in terms of dendrite arm spacings and the size, and quantity of the carbide precipitates. It would appear from the graph of weld bead top widths versus laser beam traverse speed for the 2kW laser using both the 100mm, and the 75mm focal length lenses that there was not much difference in the weld bead width, (Fig. 4.14). The increased penetration effected by the use of the 75mm focal length lens however, is a result of the fact that it is the incident energy density rather than the power density that determines the laser penetration characteristics, (see page 93).

For alloy PE11 while it is clearly shown that for a laser power of 2.5kW, and 3.0kW that the weld bead top width does not decrease greatly with increase in laser beam traverse speed, this is not quite so evident for the higher laser powers, 3.5kW, and 4.0kW. It is believed that the discrepancies observed in the behaviour of weld bead top width in alloy PE11 is a consequence of the fact that at the higher beam traverse speeds when welding with high laser powers, more of the laser
energy is trapped. At high speeds of about 100mm/s sufficient energy is still absorbed by the workpiece to result in full penetration. At 4.0kW all the welds are full penetration welds, and in this case the explanation for the discrepancy can be found in the weld profile. It was observed that at these high powers the sides of the welds appeared to be eroded, Fig. 4.28, making it difficult to determine the exact extent of the weld bead top width. This could be due to shrinkage of the solidifying melt or increased turbulence induced by the high speed of traverse of the laser beam.

Further examination of the welds produced in both these materials appeared to suggest a greater welding efficiency for Nimonic PE11. The general trend was for the differences in penetration to be quite small at the higher powers studied, and more significant at the lower power of 2.5kW. Plots of melt areas measured using the Quantimet 800 however confirmed the fact as a larger area of the base metal was melted in the case of alloy PE11, (Fig. 4.47).

Examination of the specific heats of both alloys showed that the amount heat required to raise 1 kg of alloy C263 by 1 Kelvin was marginally lower than that required for alloy PE11 (Fig.4.48). Fig. 4.49 shows the variation of thermal conductivity of alloys C263 and PE11 up to 1000ºC. These values were calculated using a formula proposed by Powell [84] which relates thermal conductivity to electrical conductivity using the following equation;

\[ K = 2.2 \times 10^{-6} \frac{T}{\rho} + 6 \]

where \( K \) is the thermal conductivity in W/(mK), \( T \) is the absolute temperature, and \( \rho \) is the resistivity in ohm cm.
From the graph in Fig. 4.49 there is no great difference in the thermal conductivity of both alloys therefore this could not account for the difference in melting efficiency observed. Considering then the approximate melting ranges for both C263 and PE11 as shown.

<table>
<thead>
<tr>
<th>Nimonic</th>
<th>Solidus</th>
<th>Liquidus</th>
</tr>
</thead>
<tbody>
<tr>
<td>C263</td>
<td>1300</td>
<td>1355</td>
</tr>
<tr>
<td>PE11</td>
<td>1280</td>
<td>1350</td>
</tr>
</tbody>
</table>

It is evident that the melting range for PE11 is larger than that of C263, this suggests a greater melting efficiency for alloy PE11, and is consistent with the observations noted in this study.

It was observed that while at the lower laser powers ranging from 1.4 to 1.6kW the welds made using the 2kW laser showed a direct relationship to the incident laser power, above 1.6kW the weld penetration decreased with laser power (Fig. 4.4). It is believed that this behaviour can be explained in terms of the mechanism of laser welding. At the high laser powers required for deep penetration welding, 'keyholing' occurs, this involves the vaporisation of the workpiece material to form a cavity which is kept open by the pressure of the ionised metal vapour, (metal plasma), within it. Heat is then transferred directly to the thickness of the material in contrast to heat transfer by conduction via the surface. When the power density within the beam impinging on the surface is increased, the amount of metal vaporised increases. Plasmas are strongly absorptive to the laser radiation, and one of two things can happen when they are present in laser materials processing. The plasma could remain within the 'keyhole' transferring energy to the bulk material by radiative processes, or the plasma could dissociate from the material.
surface effectively decoupling the bulk of the laser radiation from the surface. The results obtained from the 2kW laser welds suggest the plasma in this case has dissociated from the material surface at the higher laser powers studied reducing the energy input into the material. The observed drop in weld penetration above 1.6kW is a direct consequence of this. While the actual decrease in efficiency of a laser has not been reported in the literature, methods of controlling laser induced plasmas have been studied extensively as a means of improving laser welding efficiency(42),(43).

The observations noted from the laser welds made using the 2kW laser suggests that at the higher laser powers plasma control by disruption, or any other means, is more critical than at the lower laser powers. Therefore for a given laser, although the parameter P/vD would suggest that increasing the laser power, while keeping the beam traverse speed, and beam diameter constant should result in increased welding efficiency, this is not always the case. The effect of shield gas in plasma disruption is not accounted for in this equation, and this should be accounted for as the plasma effect clearly plays a major role in determining weld efficiency.

Altering the lens focal length revealed a strong dependency of weld penetration on this particular laser welding parameter with the weld penetration being strongly enhanced by the shorter focal length lens. In the welding studies using the 2kW laser both 75mm, and 100mm focal length lenses were used at similar laser powers and speeds. The results documented in Fig. 4.13 confirm that full penetration of the 1.2mm sheet was obtained at lower powers when the 75mm focal length lens was used. This result suggests that the use of a 75mm focal length lens would improve welding efficiency.

According to Seaman(36), it is known that the focused beam spot diameter exhibits a direct dependency on the lens focal length according to the following formula(36);
where \( f = \) lens focal length, \( D_0 = \) diameter of beam exiting from laser cavity, \( \lambda = \) wavelength, and \( D_f = \) focused beam spot size. Thus for a given laser system \( D_f \) is proportional to \( f \), it follows therefore that laser output energy intensity within the focused beam spot is inversely proportional to the lens focal length. As a larger lens focal length would result in a larger focused beam spot size, and hence a lower incident energy intensity.

In order to appreciate the degree of dependence of the power intensity on the focused beam diameter the exact relationship between laser power intensity and beam diameter must be considered. It must be understood that the variable in consideration is an energy density, rather than a power density. To clarify this particular point a brief definition of power density and energy density is given.

1. A power density can be defined as the sum total of the power available per unit area of beam diameter for interaction with the material. Therefore at a constant speed \( P \) is proportional to \( 1/D \).
2. In contrast an energy density is defined as the sum total of the power available per unit area of beam diameter per unit time for interaction with the material. As \( t = D/v \), it follows that the energy density is proportional to \( 1/D^2 \).

Where \[ I_e = \frac{P \times V}{VD \times D} \]

\( I_e = \) energy density. \( I_e = \) energy density. \( I_e = \) energy density. \( I_e = \) energy density. \( I_e = \) energy density. \( I_e = \) energy density.

It therefore follows that for a given increase in focused beam diameter the power density available for processing undergoes a corresponding decrease by a factor of 1, but the energy density decreases by a factor of 2. As laser material interactions are time temperature dependent, considering the effect of laser beam diameters in...
terms of power densities will not result in a complete appreciation of the observed results.

The microstructural implications of this increased energy input into the workpiece, is evident from the narrow deep nature of the welds made using the 75mm focal length lens compared to those made using the 100mm focal length lens. Dendrite arm spacings increased as the lens focal length increased, and carbide precipitation increased, with the precipitate being refined.

In chapter 2, it was stated that the depth of focus of the focusing lens is proportional to the focused beam diameter. In laser materials processing it is imperative that the beam is focused to its minimum spot size in order to obtain the optimum energy density within the focused spot. This becomes critical when a lens of small depth of focus, i.e. a short focal length is used, as the margins within which the focused spot size remains approximately constant is reduced. If the beam is not in focus, D increases, as the power density is inversely proportional to D, and the energy density inversely proportional to D² a decrease in welding efficiency will occur thereby offsetting the advantages of welding with a lens of short focal length.

Further implications of welding with a lens of short focal length include the joint 'fit-up' tolerances of the parts to be joined. This particular parameter is redundant where bead on plate welds are being studied. However where two parts are to be joined the joint 'fit-up' must be such that the incident beam can adequately melt both edges of the parts to produce a joint. As the focused spot size can be determined from equation (4.1), adequate joint 'fit-up' should be attained for a particular lens,
or a lens which produces a focused beam spot capable of accommodating the 'fit-up' of the parts being joined should be used.

4.5 Summary

The following features were established in these welding studies.

1. Optimum welding parameters for 1.2mm Nimotics C263, and PE11 have been identified.

2. It would appear that the welding efficiency of the 2kW laser was greater than that of the 5kW laser in the welding of thin gauge materials.

3. A parameter for the comparison of different lasers in the form of a plot of \( P/vD \) vs weld penetration was identified.

4. The resultant weld microstructure following welding was studied, and weld microstructure and precipitation characteristics were related to the laser welding parameters and characteristics of the laser welding process.

5. It has been shown that it is possible to produce porosity free welds, in both alloy C263, and PE11 at the laser welding parameter settings investigated.

6. A slight dependence of weld depth on material composition was observed on comparing welds at similar powers and speeds for Nimonic C263 and PE11. Weld penetrations were greater in Nimonic PE11.

7. While a certain amount of fusion line cracking was observed in Nimonic PE11, it is believed that this would not have a great effect on the resultant properties of the joint.

8. The results from a series of microhardness traverses across the weld in representative samples led to the conclusion that the final properties of the
joint should at least approximate to that of the base metal. However, microstructural analysis of the fusion line suggests that this area would form the weakest link in the joint as a consequence of the heat affected grains in this part of the joint.
Table 4.1  Analysis of dendrites and surrounding regions in laser welded Nimonic alloy C263.

<table>
<thead>
<tr>
<th>Element</th>
<th>Dendrites</th>
<th>Interdendritic regions</th>
<th>Fusion line</th>
<th>Base metal</th>
</tr>
</thead>
<tbody>
<tr>
<td>Ni</td>
<td>58.73</td>
<td>58.548</td>
<td>58.294</td>
<td>52.764</td>
</tr>
<tr>
<td>Fe</td>
<td>0.744</td>
<td>0.867</td>
<td>1.006</td>
<td>0.824</td>
</tr>
<tr>
<td>Ti</td>
<td>1.544</td>
<td>2.578</td>
<td>1.672</td>
<td>2.496</td>
</tr>
<tr>
<td>Al</td>
<td>0.424</td>
<td>0.594</td>
<td>0.832</td>
<td>0.637</td>
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</table>

Table 4.1  Analysis of dendrites and surrounding regions in laser welded Nimonic alloy PE11.

<table>
<thead>
<tr>
<th>Element</th>
<th>Dendrites</th>
<th>Interdendritic regions</th>
<th>Fusion line</th>
<th>Base metal</th>
</tr>
</thead>
<tbody>
<tr>
<td>Ni</td>
<td>40.478</td>
<td>41.012</td>
<td>1.358</td>
<td>40.765</td>
</tr>
<tr>
<td>Cr</td>
<td>18.854</td>
<td>19.218</td>
<td>0.984</td>
<td>19.244</td>
</tr>
<tr>
<td>Fe</td>
<td>39.070</td>
<td>37.151</td>
<td>1.237</td>
<td>37.437</td>
</tr>
<tr>
<td>Ti</td>
<td>1.607</td>
<td>2.619</td>
<td>96.426</td>
<td>2.558</td>
</tr>
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</table>
### Table 4.3 Composition of Nimonic alloy C263

<table>
<thead>
<tr>
<th>Element %</th>
<th>Al</th>
<th>Ag (ppm)</th>
<th>B (ppm)</th>
<th>Bi (ppm)</th>
<th>C</th>
<th>Co</th>
<th>C</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>0.6</td>
<td>5</td>
<td>50</td>
<td>1</td>
<td>0.08</td>
<td>21.0</td>
<td>19.0</td>
</tr>
<tr>
<td>Cu</td>
<td>Fe</td>
<td>Mo</td>
<td>Ti</td>
<td>Ti + Al</td>
<td>Ni</td>
<td></td>
<td></td>
</tr>
<tr>
<td>0.2</td>
<td>0.7</td>
<td>5.6</td>
<td>1.9</td>
<td>2.4</td>
<td>rem</td>
<td></td>
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</tr>
</tbody>
</table>

### Table 4.4 Composition of Nimonic alloy PE11

<table>
<thead>
<tr>
<th>C</th>
<th>Si</th>
<th>Mn</th>
<th>S</th>
<th>Al</th>
<th>Ag (ppm)</th>
<th>B (ppm)</th>
<th>Bi (ppm)</th>
</tr>
</thead>
<tbody>
<tr>
<td>0.10</td>
<td>0.5</td>
<td>0.5</td>
<td>0.01</td>
<td>1.0</td>
<td>5</td>
<td>50</td>
<td>1</td>
</tr>
</tbody>
</table>

<table>
<thead>
<tr>
<th>Co</th>
<th>Cr</th>
<th>Cu</th>
<th>Mo</th>
<th>Ni + Co</th>
<th>Pb (ppm)</th>
<th>Ti</th>
<th>Ti + Al</th>
</tr>
</thead>
<tbody>
<tr>
<td>0.5</td>
<td>19.0</td>
<td>0.04</td>
<td>10</td>
<td>42.0</td>
<td>2.5</td>
<td>3.5</td>
<td>5.75</td>
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</tbody>
</table>

<table>
<thead>
<tr>
<th>Zr</th>
<th>Fe</th>
</tr>
</thead>
<tbody>
<tr>
<td>1.0</td>
<td>rem</td>
</tr>
</tbody>
</table>
Fig. 4.1  Weld bead profiles showing the effect of laser power, and laser beam traverse speed on weld penetration, weld shape, and microstructure for Nimonic alloy C263, (5kW laser).
Fig. 4.2  Weld bead profiles showing the effect of laser power, and laser beam traverse speed, on weld penetration, weld shape, and microstructure for Nimonic alloy C263 (2.6 kW laser)
Fig. 4.3 Weld penetration as a function of laser power for Nimonic alloy C263, (5kW laser).
Fig. 4.4  Weld penetration as a function of laser power for Nimonic alloy C263, (2kW laser).
Fig. 4.5  Macro and microstructures of a typical weld in Nimonic alloy C263

(a) Overall macrostructure.
(b) Base alloy away from the weld zone.
(c) Weld fusion line.
(d) SEM of base alloy grains away from weld zone.
(e) SEM of fusion line showing slight degree of grain boundary coarsening in the region of the fusion line.
Fig. 4.6 Macro and microstructures in the fusion zone of a typical weld in Nimonic alloy C263.

(a) Columnar dendritic macrostructure with grains growing into the weld centre perpendicular to the direction of the heat source.

(b) Columnar dendritic microstructure in the fusion zone.

(c) Dendrites in nimonic alloy C263, welded at 4.0kW, 67mm/s.

(d) Dendrites in Nimonic alloy C263, welded at 2.5kW, 67mm/s.
weld centreline
Fig. 4.7 Typical carbide precipitation observed in the interdendritic regions of the weld in Nimonic alloy C263.

(a) Carbide precipitation in Nimonic alloy C263, welded at 4.0kW, 67mm/s.

(b) Carbide precipitation in Nimonic alloy C263, welded at 2.5kW, 67mm/s.
Fig. 4.8  Weld penetration as a function of laser beam traverse speed for Nimonic alloy C263, (5kW laser).
Fig. 4.10  Typical effect of laser beam traverse speed on weld microstructures.

(a)  Dendrites in Nimonic alloy C263, welded at 2.5kW, 100mm/s.

(b)  Dendrites in Nimonic alloy C263, welded at 2.5kW, 33mm/s.

(c)  Carbide precipitation in Nimonic alloy C263, welded at 2.5kW, 100mm/s.

(d)  Carbide precipitation in Nimonic alloy C263, welded at 2.5kW, 33mm/s.
Fig. 4.9  Weld penetration as a function of laser beam traverse speed for Nimonic alloy C263, (2kW laser).
Fig. 4.11  Weld bead top width as a function of laser beam traverse speed for Nimonic alloy C263, (5kW laser).
Fig. 4.12  Weld bead top width as a function of laser bead traverse speed for Nimonic alloy C263, (2kW laser).
Weld penetration as a function of laser beam traverse speed for Nimonic alloy C263 showing the effect of lens focal length in enhancing weld penetration, (2kW laser).
Fig. 4.14 Weld bead top width as a function of laser beam traverse speed for Nimonic alloy C263 showing the effect of lens focal length, (2kW laser).
Fig. 4.15a  Typical microhardness traverse across Nimonic alloy C263 weld.
Fig. 4.15b  Typical microhardness traverse across Nimonic alloy C263 weld.
Fig. 4.16  Melt volume as a function of laser power for Nimonic alloy C263, (5kW laser).
Fig. 4.17 Melt volume as a function of laser beam traverse speed for Nimonic alloy C263, (5kW laser).
Fig. 4.18  Melt volume as a function of laser power for Nimonic alloy C263, (2kW laser).
Fig. 4.19 Melt volume as a function of laser beam traverse speed for Nimonic alloy C263, (2kW laser).
Fig. 4.20 Welding efficiency as a function of laser beam traverse speed for Nimonic alloy C263, (5kW laser).
Fig. 4.21  Welding efficiency as a function of laser beam traverse speed for Nimonic alloy C263, (2kW laser).
Fig. 4.22  A comparison between the welding efficiency achieved for Nimonic alloy C263 using the 5kW laser, and the 2kW laser, as a function of laser beam traverse speed.
Fig. 4.23 Welding efficiency as a function of laser power for Nimonic alloy C263, (5kW laser).
Fig. 4.24 Welding efficiency as a function of laser power for Nimonic alloy C263, (2kW laser).
Fig. 4.25 Weld penetration as a function of $P/vD$ for Nimonic alloy C263.

(a) 2kW laser, 100mm focal length lens.

(b) 2kW laser, 75mm focal length lens.

(c) Combined data for the 2kW laser.
Fig. 4.25b
Fig. 4.25c
Fig. 4.26 Weld penetration as a function of $P/vD$ for Nimonic alloy C263, (5kW laser).
Fig. 4.27 Combined data of weld penetration versus $P/vD$ for the 5kW and the 2kW lasers. Note characteristic slope obtained for the lasers.
Fig. 4.28  Weld bead profiles showing the effect of laser power, and laser beam traverse speed on weld penetration, weld shape, and microstructure for Nimonic alloy PE11.
Fig. 4.29  Weld penetration as a function of laser power for Nimonic alloy PE11.
Fig. 4.30 Macro and microstructures of a typical weld in Nimonic alloy PE11.

(a) Overall macrostructure.
(b) Base alloy away from the weld zone.
(c) Weld fusion line.
(d) Columnar dendritic microstructure in the fusion zone.
Fig. 4.31  Typical dendritic structure observed in the fusion zone of the weld.

(a)  Dendrites in Nimonic alloy PE11, welded at 4.0kW, 67mm/s.

(b)  Dendrites in Nimonic alloy PE11, welded at 2.5kW, 67mm/s.
Fig. 4.32 Typical carbide precipitation observed in the interdendritic regions of the weld in Nimonic alloy PE11.

(a) Carbide precipitation in Nimonic alloy PE11, welded at 4.0kW, 67mm/s.

(b) Carbide precipitation in Nimonic alloy PE11, welded at 2.5kW, 67mm/s.
Fig. 4.33  Typical microcracks observed along the fusion line in Nimonic alloy PE11 welds.
Fig. 4.34  SEM micrographs of a typical Nimonic alloy PE11 weld region.
(a) SEM of base alloy grains away from weld zone.
(b) SEM of fusion line showing slight degree of grain boundary coarsening in the region of the fusion line.
Fig. 4.35 SEM of titanium precipitate observed in the weld region.
Fig. 4.36  Weld penetration as a function of laser beam traverse speed for Nimonic alloy PE11.
Fig. 4.37  Typical effect of laser beam traverse speed on weld microstructures.

(a)  Dendrites in Nimonic alloy PE11, welded at 2.5kW, 100mm/s.

(b)  Dendrites in Nimonic alloy PE11, welded at 2.5kW, 33mm/s.

(c)  Carbide precipitation in Nimonic alloy PE11, welded at 2.5kW, 100mm/s.

(d)  Carbide precipitation in Nimonic alloy PE11, welded at 2.5kW, 33mm/s.
Fig. 4.38  Weld bead width as a function of laser beam traverse speed for Nimonic alloy PE11.
Fig. 4.39a Typical microhardness traverse across Nimonic alloy PE11 weld.
Fig. 4.39b  Typical microhardness traverse across Nimonic alloy PE11 weld.
Fig. 4.40 Melt volume as a function of laser beam traverse speed for Nimonic alloy PE11
Fig. 4.41 Melt volume as a function of laser power for Nimonic alloy PEl1
Fig. 4.42  Welding efficiency as a function of laser power for Nimonic alloy PE11.
Fig. 4.43  Welding efficiency as a function of laser beam traverse speed for Nimonic alloy PE11.
Fig. 4.44  Diagramatic representation of a TEM$_{00}$ beam mode.
Fig 4.45 Representative Perspex beam print obtained from the 2kW laser.
Fig. 4.46  Diagramatic representation of a 'top hat' beam mode.
Fig. 4.47 A comparison between melt areas measured in for Nimonic alloys C263, and PE11 for welds made at 2.5kW.
Fig. 4.48 A comparison between the specific heats of Nimonic alloy C263, and Nimonic alloy PE11 as a function of temperature up to 1000K.
Fig 4.49 A comparison between the thermal conductivity of Nimonic alloys C263, and PE11 as a function of temperature.
CHAPTER 5

LASER WELDING OF ALUMINIUM BASED ALLOYS AND METAL MATRIX COMPOSITE MATERIALS

Introduction

In this chapter the laser welding of aluminium 6061 and aluminium 6061 based metal matrix composites is studied. The chapter will be divided into four sections. The first one comments on the effects of the basic optical properties of Al and its alloying elements on the laser welding process. The next section deals with the laser operating conditions. This is followed by the experimental results and their interpretation. The results are then discussed in relation to the current knowledge.

5.1 Comments on the optical properties of aluminium and the effects of alloying elements on the laser processing of aluminium alloys

In laser welding, heating is brought about by the interaction of photons with electrons, (both bound and free), in the surface of the workpiece\(^{(21)}\). The electrons are excited to a higher energy state upon absorption of the optical energy. By undergoing collisions with lattice phonons, other electrons, impurities, and defect structures these high energy state electrons give up their energy which is
subsequently converted to heat. The term 'coupling of energy to the workpiece' can be used to describe the absorption of incident energy by the workpiece. In the welding of aluminium alloys, poor coupling of the laser energy occurs. This can be explained by considering the optical properties of aluminium, and the physical properties of some of the alloying constituents present in aluminium alloys.

The CO2 laser, which is still the only practical laser for deep penetration welding of metallic materials is electromagnetic radiation of wavelength 10.6μm. At these wavelengths the beam interacts exclusively with electrons, as atoms are too heavy to respond to the high frequencies, (v \(>\) \(10^{13}\)Hz), of the CO2 laser beam. The constituent atoms of Al 6061 have electrons in their shell which are only loosely bound to the nucleus. A comparison of Al and Ge, both in period three of the Periodic Table, shows that the outer shell electrons of Ge are more strongly bound to the nucleus due to the higher nuclear charge present in Ge, with eighteen electrons in its penultimate shell as compared to eight in that of Al. As a result of this there is a higher density of free electrons present in the solid Al 6061 alloy. When the laser beam impinges on the solid alloy, these electrons are capable of extracting energy from the electromagnetic field associated with the laser beam. This extracted energy results in an increase in the kinetic energy of the electrons. As the external field is periodically changing the oscillating electrons re-radiate their kinetic energy unless they are in a position to undergo frequent collisions with phonons or defect structures in the lattice. It is the re-radiation of energy by a large number of free electrons that is the cause of the high reflectivity of aluminium alloys.

At high laser intensities optical properties can be modified. The mechanism of this modification will not be considered in detail here. However when the net gain of energy by the electrons exceeds that lost by re-radiation, a build up of high
energy electrons occurs resulting in a rise in energy of the substrate. As far as the coupling of energy to the substrate is concerned, the effect of the electron cloud is to limit the rate of local energy deposition in two ways: (1) at sufficiently high electron densities the beam is reflected away from the electron cloud; and (2) diffusion of electrons away from the absorbing zone effectively increases the heated volume, and accounts for the fact that the high thermal conductivity of Al 6061 makes it difficult to laser process.

The effect of the physical properties of the alloying elements can be demonstrated by considering the fact that most Al alloys contain low melting point constituents which are easily vapourised from the surface of the workpiece forming a plasma of metal vapour, e.g magnesium present in Al 6061. Metal vapour plasmas shield the workpiece surface from the laser radiation.

In order to successfully weld Al 6061, and the Al based metal matrix composites a means must be found of coupling the laser beam to the workpiece by minimising the effects outlined above.

5.2 Comments on the specific experimental conditions used in the welding of the aluminium based materials

The welding Institute 5kW continuous wave CO$_2$ laser was used for all the welds on Al 6061, at the following fixed settings.

a. 300mm focal length KCl lens.
b. Focal setting '0', i.e the focal point was set at zero on the x-y table level.
c. Full foot shielding using helium, i.e trailing top and bottom shield.

The Al 6061 sheet was 15cm × 30cm × 2mm. Half of the samples were subjected to the sulphuric acid anodising process, details of which can be found in Chapter 3,
to produce oxide coatings of varying thicknesses ranging from 10μm to 16μm on the metal surface, while the other half were welded in the as received surface finish.

The following is a summary of the procedures used in the operation of the equipment.

1. The laser was set at a power of 4.5kW.
2. The as received sheet of aluminium was clamped on to the x-y table.
3. A piece of scrap metal was placed in front and behind the workpiece to prevent damage to the x-y table by an 'over-run' of the laser head.
4. The safety shields enclosing the laser head were then shut.
5. The x-y table speed was set at 100mms$^{-1}$, the laser shutter opened, and a melt track generated across the material.
6. The sheet was then unclamped, and realigned with the laser head to give a clearance of approximately 2cm from the previous melt track.
7. The beam traverse speed was then reduced, and a new melt track was made.
8. Procedures 6 and 7 were then repeated, until a series of welds had been produced on the sheet.

For the purpose of these experiments, a total of twelve melt tracks were made at laser powers of 4.5kW, 4.0kW, 3.5kW, and 3.0kW. The speeds investigated were: 100mms$^{-1}$, 83mms$^{-1}$, and 67mms$^{-1}$.

The procedure detailed for the as received sheet was then repeated for the anodised sheets of Al 6061. Having completed a series of runs, all the welds were sectioned, polished, and etched in a solution of 1g ammonium oxalate, and 15%
ammonium hydroxide in 100mls of distilled water. Specimens were etched for five minutes at 80°C to develop the grain boundaries in the weld, and parent metal of the alloy.

The equipment and materials detailed below were used in the study of the aluminium based boron fibre reinforced metal matrix composite.

a. 2kW Control CO₂ laser.
b. Lens focal length: 75mm.
c. Shield gas: argon.

The focal position was on the surface of the material.

Material: 2mm thick aluminium 6061/boron fibre reinforced metal matrix composite.

The procedure followed in utilising the equipment and materials was similar to that followed in the welding of alloy 6061, but a few unique features of this aspect of the study are outlined briefly in the following points.

1. The laser was focused and set at a power of 1.4kW.
2. A strip of the Al/B laminate was then clamped on to the x-y table.

For the purpose of these trials a total of three beam traverse speeds were investigated at the laser power of 1.4kW. These speeds were as follows: 46.7mm/s, 33.3mm/s, and 25mm/s. Melt tracks were made parallel to the boron fibres and across them.

A similar procedure was carried out for the aluminium 6061/silicon carbide material using the Welding Institute 5kW CO₂ laser. Two melt tracks were produced across the material at the following laser power and speed combination. Power 3.0kW, speed 100mm/s, and Power 2.5kW, speed 7mm/s. This material was welded with great difficulty as it had a tendency to produce a large amount of fumes, and particles on welding. Such fumes and particles resulted in
contamination of the KCl lens causing it to crack. It was for this reason that only a limited number of weld runs could be made in this material.

The melt tracks produced in both Al/B, and Al/SiC were sectioned and polished for microstructural examination. The welds were examined in the unetched condition.

5.3 Results

5.3.1 As received aluminium 6061

Effect of laser power on weld penetration and microstructure

Fig. 5.1 demonstrates the relationship between weld penetration, weld shape, and macrostructure to laser beam power in the welding of Al 6061 at a constant speed. These results are presented graphically in Fig. 5.2. This work showed that while it is possible to produce full penetration welds in 2.0mm Al 6061 using a 5kW laser, a threshold exists at 3.0kW at which no significant penetration of the workpiece was achieved in the range of laser beam traverse speeds investigated; 67-100mm/s.

Description of weld structure

Further detailed examination of the specimens whose macrographs are shown in Figs. 5.1 were carried out to determine the microstructural variations produced by laser welding, and the effect of the laser welding parameters on these variations. Optical, scanning, and transmission microscopy techniques were used.
The variations of a typical weld region macro and microstructure are shown in Fig. 5.3. Fig. 5.3a shows the overall weld macrostructure. The base metal displayed a precipitation hardened structure, Fig. 5.3b, with irregular grains elongated in the direction of rolling. Close examination showed the HAZ to be very narrow at approximately 60\( \mu \)m, Fig. 5.3c. The epitaxial relationship between adjacent regions on opposite sides of the fusion line is clearly evident from the continuous nature of the grain boundaries across the fusion line region, (Fig.5.3c). The grain boundaries in the HAZ appear to have undergone a certain amount of liquation during welding. However semi-quantitative EDS analysis of these regions failed to reveal any difference in composition as compared to that of the surrounding areas. The weld metal was observed to consist of a cellular dendritic structure essentially perpendicular to the direction of welding, Fig. 5.3d. Towards the middle and top sections of the weld labelled, (e), on the macrograph, Fig. 5.3a, the grains appear to display an equiaxed structure. This is the result of grains being sectioned transversely during preparation of the sections. In the later stages of solidification the growth direction of the grains is parallel to the welding direction. Fig. 5.4 shows the grains changing direction to follow the heat source.

Examination of the welds showed no change in alloy composition across the weld zone. It was however not possible to distinguish the \( \text{Mg}_2\text{Si} \) precipitate observed in the parent metal in the weld zone.

Macrographs of some of the weld seams showed some irregularity, along with the presence of craters and blowholes in some of the welds, Figs. 5.5a, and 5.5b. Longitudinal macrographs of some of the non full penetration welds also revealed the presence of non-uniform penetration along the length of the weld seam, Fig. 5.5c. Transverse macrographs of the sectioned welds in some cases showed porosity and cracking, examples of which are shown in Figs. 5.6a, 5.6b, 5.6c, and 5.6d.
The effect of laser beam traverse speed on weld penetration and weld bead top widths

Fig. 5.1 shows the macrostructural effect of laser beam traverse speed on weld penetration. The threshold at 3.0kW is clearly demonstrated. The general trend appears to suggest that the weld penetration achieved in the welding of Al 6061 increases with decreasing laser beam traverse speed. These results are shown graphically in Fig. 5.7. It was observed that the general trend was for the weld bead top width to decrease as the laser beam traverse speed increased, Fig. 5.8. It was also observed that as the laser beam traverse speed increased, the dendrite cell sizes decreased, with a corresponding decrease in dendrite arm spacings, Figs. 5.9a, and 5.9b.

Microhardness

Microhardness traverses were carried out across the joint in a number of welds. Figs. 5.10a, and 5.10b show typical plots of the observed results, which reveal the hardness of the weld metal to be marginally below that of the parent metal.

5.3.2 Anodised aluminium 6061 welds

The effect of laser power on weld penetration and microstructure

Fig. 5.11 demonstrates the relationship typically observed between laser beam power, weld penetration, weld shape, and macrostructure observed for the anodised aluminium welds, as seen in the material with a 12μm anodic film. It is evident that full penetration of the through thickness of the workpiece was obtained
at laser powers of 4.0kW, and 3.5kW. At 3.0kW and 2.5kW considerable penetration of the workpiece was achieved at all speeds investigated ranging from 67mm/s to 117mm/s, and for all the anodic film thicknesses ranging from 10-16μm. In the welding of non-anodised aluminium it was not possible to produce a significant weld bead at 3.0kW, and a laser beam traverse speed of 67mm/s. With the anodised aluminium, at a laser power of 2.5kW, and a beam traverse speed of 100mm/s significant melting of the base metal was accomplished. Fig. 5.12 presents these results graphically.

Description of weld microstructures

The variation of a typical weld macro and microstructure for anodised Al 6061 is shown in Figs. 5.13. Fig. 5.13 shows the overall weld macrostructure. The base metal was of a precipitation hardened structure, with irregular grains elongated in the direction of rolling, Fig. 5.13b. Fig. 5.13c reveals the HAZ to be about 35μm in width. This is very narrow compared to 60μm measured for the non-anodised material. The epitaxial relationship between the weld metal grains and the base metal grains observed in the non-anodised welds is also evident here. The weld metal structure was observed to be columnar dendritic in nature, Fig. 5.13d. Examination of the welds showed no change in composition in the weld zone. No Mg$_2$Si precipitates were observed in the weld zone, although they were clearly visible in the parent metal.

Figs. 5.14a, and 5.14b show the top and bottom weld beads in a representative weld plate for the anodised Al welds at 3.0kW. Note the full penetration weld at 67mm/s. The weld bead is narrow, and displays fewer craters and blowholes than those observed in the as-received Al welds. The non-uniform penetration is also clearly less pronounced than that observed in the as-received Al welds, Fig 5.14c.
The weld metal showed evidence of what appeared to be areas of gross porosity in the fusion zone, Fig. 5.15a. On closer examination it was revealed that these areas were formed from what appeared to be areas of non-adhesion between portions of the melt and the bulk weld metal, Fig. 5.15b, and 5.15c. These areas are thought to result from the incorporation of oxide from the anodised layer into the weld region during welding, creating what appeared to be globules of metal essentially surrounded by oxide. These areas are therefore unable to bond to the surrounding molten metal. In sectioning the weld transversely, it was fortuitious that some of these areas were sectioned through the middle, Fig. 5.15b. Fig. 5.15c shows where an area of non-adhesion has subsequently separated from the surrounding metal.

The effect of laser beam traverse speed on weld penetration and weld bead top widths

The effect of laser beam traverse speed on weld penetration in the anodised aluminium samples is shown in Fig. 5.11. As full penetration welds were obtained for most of the welding speeds investigated, the welds made at, 3.0kW, and 2.5kW give a most representative view of the effect of laser beam traverse speed on weld penetration. It can be seen, (Fig. 5.16), that weld penetration increases with decreasing traverse speed of the laser beam. The dendrite cell size, and dendrite arm spacing were also found to decrease with increase in laser beam traverse speed, Fig. 5.17.

The effect of laser beam traverse speed on weld bead top width at a constant laser power, and a constant anodic film thickness was investigated. It was generally found that, weld bead top width decreased as the laser beam traverse speed was increased, Fig. 5.18.
The effect of anodic film thickness on weld penetration and weld bead top widths

A study of the effect of anodic film thickness on weld penetration did not reveal any dependency of weld penetration on this parameter, Fig. 5.19. It was however found that the weld bead top width decreased as the anodic film thickness was increased, Fig. 5.20. It was also observed that porosity increased with increase in anodic film thickness for a given combination of laser power, and speed. Increase in heat input into the weld region at a constant anodic film thickness also resulted in an increase in the observed weld region porosity. This is thought to be the result of the larger energy input into the fusion zone facilitating the decomposition of the surface anodic coating on the workpiece.

Microhardness

Microhardness traverses were carried out across the joint in a number of welds. Figs. 5.21a and 5.21b show typical plots of the observed results, it is evident that the hardness of the weld metal is marginally below that of the parent metal.

Summary

Aluminium 6061 was welded in the as received and anodised surface finish conditions. Optimum welding conditions were determined for the material in these surface finishes, and it was shown that the anodising treatment enhanced the coupling of the laser radiation to the material surface, and hence the welding efficiency. The microstructural implications of the laser welding operation was considered, and hardness traverses were carried out to give a general indication of the joint properties.
5.3.3 Aluminium based metal matrix composites

Aluminium 6061/boron fibre reinforced metal matrix composite

Effect of laser power on weld penetration and microstructure

In these series of experiments, a number of welds were produced both parallel to and across the strengthening boron fibres. Figs. 5.22a, and 5.22b show the macrostructural variation in the welds produced in both these configurations respectively. Figs. 5.22c, and 5.22d show the parent MMC away from the fusion zone, with Fig. 5.22c giving an end view of the fibres, and Fig. 5.22d a view of the longitudinal section through the fibres. Adjacent to the fusion zone it is clearly seen that the fibres have melted and become intimately mixed in the fusion zone, Fig. 5.22e, and 5.22f for the welds parallel to the fibres, and the weld transverse to the fibres respectively. It is evident that the microstructure in the fusion zone is similar in both cases and differs from that of the base material, Fig. 5.22g. It would appear that on melting the boron and aluminium have reacted together and at the fast cooling rates involved produced a very fine structure. In order to analyse this structure, and determine the constituent phases the welds were examined by secondary ion mass spectroscopy, (SIMS). On obtaining the results of this study, which provided basic information about the constituent elements in the structures observed optically, a series of thin foil specimens were prepared. This was done to determine the exact composition of the observed phases. The results of these two studies are discussed in the following sections.

It was shown that as the laser beam traverse speed is increased, the weld penetration decreased. In the area adjacent to the weld region, a number of pores
were observed, increasing the laser beam traverse speed while resulting in the observed decrease in weld penetration, appeared to also result in a reduction of the associated fusion line porosity, Fig. 5.23a and 5.23b.

**SIMS investigation**

In the SIMS analysis two sources were used; (1) an oxygen duoplasmatron ion source to optimise the detection of electropositive elements, (e.g. boron, magnesium, aluminium, chromium, and manganese), and (2) a cesium thermal ionisation source, to optimise the detection of electronegative elements, (e.g. carbon, and oxygen). An area of the weld was selected, analysed, and the following trends became apparent. Fig. 5.22g a shows an optical micrograph of the melt zone, in which there are a number of 'needle-like' structures, (n), and irregular shaped particles, (s), clearly visible. Fig. 5.24a shows a secondary ion micrograph, (150 mm dia.), of the Al\(^{27}\) distribution in the melt zone. From this it was deduced that there was a variation in the aluminium content in the needles and the irregular shaped particles. Fig. 5.24b shows a secondary ion micrograph of the B\(^{11}\) distribution in the same area. From this micrograph it could be inferred that the 'needle-like, structures, (n), and the irregular shaped particles, (s), shown in Fig. 5.22g, contained aluminum and boron. The matrix in the melt zone, however showed no sensitivity for boron and was thought to be aluminium. From the results obtained in this study, no conclusions about the composition of the phases could be drawn as SIMS is a qualitative rather than quantitative analysis technique. To compliment the results obtained through SIMS, and conclusively identify the phases a number of thin foil specimens were prepared for electron microscopy, and associated electron diffraction measurements.
Electron microscopy

A series of electron diffraction patterns were obtained from the thin foil specimens prepared. These patterns were then analysed using the camera constant method. Fig. 5.25a shows a transmission electron micrograph of an irregular shaped particle, (s), and Fig. 5.25b the diffraction pattern taken from it. This pattern was found to correspond to that of the [100] plane of alpha AlB$_{12}$. Fig. 5.25c shows a micrograph of the diffraction pattern corresponding to the [010] plane of beta AlB$_{12}$. In analysing the selected area diffraction patterns obtained from a series of irregular particles, they were found to correspond to alpha AlB$_{12}$ as shown in Fig.5.25b, or to beta AlB$_{12}$, Fig. 5.25c.

Fig. 5.26a shows a transmission electron micrograph of a large needle, (bn in Fig. 5.22g). The diffraction pattern obtained from this particular needle is shown in Fig. 5.26b. From this pattern lattice parameters, (d-spacings), of 3.36Å and 2.71Å were calculated. These spacings were found to correspond to known d-spacings of AlB$_2$. A series of diffraction patterns were obtained on a number of these 'needle-like' structures. The d-spacings obtained were all similar to that obtained from Fig. 5.26b. Scanning Electron Microscope, (EDS), analysis of the needle shown in Fig. 5.26a produced the results shown in Fig 5.27. Peaks due to aluminium and tungsten are present. The presence of the Al was expected, as this element forms the bulk of the composite matrix. In Chapter 3 details of the composite, and manufacture are given. Suffice it to say here that the presence of tungsten in the melt region arises from the fact that the boron wires are produced by depositing boron on thin tungsten wires. Boron cannot be detected by the system used.
In the weld microstructure a series of smaller 'needle-like' structures, (sn), were also present, Fig. 5.28a. SEM analysis of these structures showed them to contain aluminium, silicon, tungsten, and possibly boron, Fig. 5.29. The diffraction pattern obtained for these structures is shown in Fig. 5.28b. The lattice parameters were found to correspond to known spacings for either $\beta$ or $\alpha$ AlB$_{12}$. Both these have very similar parameters, which would suggest a boron lattice with aluminium in the interstices.

Summary

Aluminium/boron fibre reinforced metal matrix composite was welded with a 2kW continuous wave CO$_2$ laser. It was determined that weld penetration decreases with increase in laser beam traverse speed, as did porosity in the weld region. Novel phases were observed in the weld region, and these were identified with the aid of SIMS, SEM, and TEM electron optical techniques.

Aluminium 6061/silicon carbide particulate reinforced composite

Effect of laser power on weld penetration and microstructure

Fig. 5.30a shows the microstructural variation of the weld in the Al/SiC particulate reinforced composite material. The fusion zone, (fz), can be clearly distinguished, as can the heat affected zone, (haz). Fig. 5.30b shows the parent MMC away from the fusion zone. In the welds area, adjacent to the fusion zone there appeared to be a denudation of reinforcing particles, Fig. 5.30c. This observed structural difference is thought to be the result of the aluminium matrix flowing around the silicon carbide reinforcing particles, with the end result being a coarsening of the
reinforcing particles in this region. The weld penetration was observed to decrease as the laser beam traverse speed was increased from 58.3 mm/s to 117 mm/s. There was a certain amount of porosity associated with both the weld region, and the adjacent HAZ. Increasing the laser beam traverse speed resulted in a reduction in this porosity, and associated reduction in weld penetration. Altering the beam traverse speed did not affect the microstructure in the weld region. The microstructure in the fusion zone which clearly differs from that of the base metal is shown in Fig. 5.30d. The constituent phases within this microstructure could not be readily identified, and were the subject of further analysis using secondary ion mass spectroscopy and transmission electron microscopy.

As in the analysis of the Al/B laminate welds, two sources were used: the oxygen duoplasmatron source, and the cesium thermal ionisation source. Fig. 5.31a shows a micrograph, (150 mm dia.), of the Al$^{27}$ distribution in the melt zone detected using the O$^{2+}$ source. From this micrograph it would appear that the needles, (n), and the bulk of the surrounding material contained a certain amount of aluminium. There were however particles corresponding to the irregular grey particles, (g), in the optical micrograph, Fig. 5.30a, which appeared to contain little or no aluminium. Fig. 5.31b shows the C$^{12}$ for the area analysed for aluminium. This micrograph appeared to show that the needles as well as a large area of the matrix contained carbon. Irregular areas corresponding to, (g), on Fig.5.30a, showed little or no sensitivity for carbon. Analysis for Si$^{28}$ from this area, revealed a Si distribution in these particles, with an even distribution of a small amount of Si in the bulk area. A small Si distribution was detected in the needles, Fig. 5.31c.
From these observations it is possible to deduce that the needles could be some compound of aluminium, and carbon with possibly a small amount of silicon, while the irregular particles were silicon.

**Electron microscopy**

Fig. 5.32a is an electron micrograph of a needle, (n), in the Al/SiC metal matrix composite, and Fig. 5.32b shows the diffraction pattern obtained from it. Calculation of the lattice spacings from this and patterns from other needle structures resulted in spacings corresponding to known values for alpha Al$_4$SiC$_4$. This confirms quantitatively the results obtained from SIMS which suggested that the needles contained all three elements aluminium, silicon and carbon.

Further analysis of the weld metal zone produced less definite results. Taking a number of diffraction patterns at random away from the needle structures revealed distinct areas of aluminium, and areas of silicon. Figs. 5.33a, and 5.33b show an X-ray pattern obtained for the silicon area and micrograph respectively. A series of EDS scans confirmed these results. The EDS scans also showed a number of the areas to contain both silicon and aluminium, Fig. 5.34. The presence of distinct areas containing silicon confirmed the results obtained using SIMS that suggested the irregular particles were silicon. It can therefore be deduced from the following results that the excess silicon after the formation of Al$_4$SiC$_4$, formed a solid solution with aluminium. Following these results it can be concluded that there was no carbon in the matrix in the weld zone.
Summary

Aluminium/silicon carbide particulate reinforced metal matrix composite was welded with a 5kW continuous wave CO₂ laser operating at a power of 2.5kW. Two laser beam traverse speeds were investigated; 58.3 mm/s, and 1167 mm/s. While there did not appear to be any difficulty in coupling the laser radiation to the metal surface, the large amounts of fumes and debry produced by the impinging laser beam made the material difficult to weld.

It was determined that the weld penetration decreased as the laser beam traverse speed was increased. Increasing the laser beam traverse speed also resulted in a decrease in weld porosity. Novel phases were observed in the weld region, and these were identified with the aid of SIMS, SEM, and TEM electron optical techniques.

5.4 Discussion

From the results documented in sections 5.3.2 it was evident that 2mm thick aluminium 6061 could be welded with a continuous wave CO₂ laser in both the as received and anodised surface conditions. It was also shown that it is possible to produce a solidified melt region of some integrity in aluminium based metal matrix composites. During the welding of the composite materials certain difficulties, such as the expulsion of dense black fumes and particles, which led to contamination of the lens, and ultimate breakage were encountered. This limited the number of welds which could be made in these materials. However sufficient samples were obtained to allow an in depth study of the resultant weld microstructures in the fusion zone and the adjacent HAZ.
This discussion will be divided into two main sections dealing with; the as received, and anodised Al 6061 welds, and the Al based composite material welds respectively. Within these sections two sub-sections will be considered. The first will deal with the effects of process parameters on the resultant weld structures, and the second will deal with a discussion of the porosity and weld cracking which was observed in certain welds.

5.4.1 As received and anodised Al 6061 welds

The effect of welding parameters on weld penetration, microstructure, and mechanical properties

The laser welding parameters investigated in the course of this work include the incident laser beam power, and the laser beam traverse speed. Their effect on both the as received, and anodised Al 6061 welds are considered. The effect of anodised coating thickness on Al 6061 on weld penetration was also investigated.

For the Al 6061 alloy, weld penetrations for both surface finishes increased as the laser beam power was increased at a constant beam traverse speed. Thus it can be said that the laser welding of Al 6061 obeys the general trend observed in parametric studies on the laser welding of metals(24). The weld penetration was also observed to decrease as the laser beam traverse speed was increased. Weld microstructure was cellular dendritic, and the laser welding parameters of power and speed were observed to have an effect on this structure. Figs. 5.9, and 5.17 show the effect of laser beam traverse speed on dendrite cell sizes. For a constant laser beam traverse speed the dendrite cell size is observed to increase with increase in laser power. Dendrite arm spacings range from 6µm at 4.0kW to 3µm at
2.5kW. Increasing the laser beam power also resulted in an increase in the dendrite cell size. These observations are all a consequence of the increased heat input into the material as the incident laser power is increased, or the laser beam traverse speed is decreased. Increased heat input into the material increases solidification times giving more time for metallurgical reactions to occur.

Optimum welding conditions defined as the combination of laser power and speed required to attain full penetration of the 2mm Al 6061 sheet was determined for both the as received, and anodised sheet, Figs. 5.35, and 5.36. During the welding studies full penetration welds were obtained at laser powers of 4.0kW, 3.5kW, at all speeds investigated for the anodised material, ranging from 67-117mm/s. Thus it is evident that a great increase in efficiency of welding is realised by anodising the workpiece prior to welding.

The effect of optical properties on the welding of as-received Al 6061

In the laser welding of the as received Al sheet a threshold laser power was identified at 3.0kW at which a significant melt zone could not be generated in the 2mm thick sheet, at the laser beam traverse speeds investigated. This is a direct consequence of the optical properties considered in section 5.1. The incident CO$_2$ laser radiation on the as-received Al 6061 surface interacts with the free electrons in the metal and is reflected. The net gain of energy by the electrons after reflection and conduction away from the area of beam incidence is insufficient to bring about significant melting of the workpiece. Heat transfer in this regime is by conduction, and the melt tracks do not display a profile characteristic of the 'keyholing' process seen in the welds made at the higher laser powers. Reflected radiation from the workpiece, caused complications during welding by being
cavity leading to instabilities in the discharge, and rendering the KCl lens prone to breakage. The use of a small nozzle during welding, or a shielding device between the lens and the workpiece could minimise this effect, allowing the beam traverse speed to be decreased. Increased interaction times afforded by slow beam traverse speeds, can allow the heat input into the material to be increased thus allowing the amount of energy retained in the beam interaction zone to be increased. This would facilitate 'keyholing', and hence deep penetration welding.

Huntington, and Eagar(81), in a study on the laser weldability of Al and Al 5456 concluded that increased power absorption does not occur on the onset of melting but rather with the beginning of 'keyhole' formation. It is evident from the welding of Al 6061 at 3.0kW that while some melting was achieved the energy density of the beam was not sufficient to initiate the formation of a 'keyhole'.

It is also possible that by using a lens of shorter focal length to further concentrate the energy of the laser beam, net gain of energy by the workpiece could reach a value sufficient to facilitate 'keyholing'. The threshold laser power could therefore be shifted to a lower value. Changing the surface condition of the alloy by applying the sulphuric acid anodising surface finish was found to eliminate the threshold at 3.0kW, and full penetration welds were achieved at this power for the anodised Al samples.

The effect of the sulphuric acid anodising treatment on the laser weldability of Al 6061

By anodising the Al 6061, the characteristics of the surface layers impinged by the laser beam are changed. (1) The free electrons present in the Al are involved in
forming a bond with the $O_2$ molecule, and as such their movement in the lattice becomes restricted. (2) The sulphuric acid anodising process introduces a porous film with a non-porous base, as shown in Fig. 5.37 onto the metal surface. The pores in the film serve to trap the incident radiation aiding absorptivity. The effect of the anodic surface finish on the absorption of laser energy and the subsequent enhancement of the welding efficiency is clearly demonstrated in this work for a range of laser beam powers, speeds, and anodised coating thicknesses. The threshold power at 3.0kW at which welding could not be effected for the as received material is removed, with significant melting being produced at the laser power of 2.5kW, and all the beam traverse speeds investigated.

Arata et al (29) initially investigated the effect of anodising on the absorption of laser radiation. While the applicability of the anodising treatment as a welding aid, and the accompanying effects of the oxide film on weld microstructure and integrity was not investigated, it was showed that the absorptance of 0.3mm Al increased up to 100% at an oxide coating thickness of 5μm. It is possible to use the theory of this work to account for the observations noted in these welding studies. These authors suggested that the absorption of the anodised surface finish changed with time, and they found that it decreased to a constant value of 55%. A number of anodic coating thicknesses were considered, and while coating thicknesses giving 100%, 64%, and 61% decreased to 55% not much change in absorption was recorded for the specimen having 48% absorptance at room temperature. This led the authors to suggest that the $Al_2O_3$ film thickness corresponding to 55% absorption can remain on the metal surface when irradiated with the 150W laser used for the experiments. To account for these observations it was suggested that during laser heating, the surface of the $Al_2O_3$ would reach its boiling point, while the bottom surface in contact with the Al surface would have a temperature similar
to that of the sheet. Therefore a step gradient in temperature should be produced in the film in the direction of the thickness, and the part of the film above its boiling point is removed by vaporisation. It was shown that the surface temperature of the base metal at about 1600°C is considerably lower than the boiling point of the oxide coating on its surface. Therefore the Al₂O₃ film can remain even after the oxidised specimen is heated, and melted by the focused laser beam, thus enhancing laser absorptance.

In these series of experiments the minimum oxide thickness measured using the permascope was 10µm, (details of this operation is given in Chapter 3). No significant difference was observed in the effect of anodic film thickness on the efficiency of the welding process. This is probably due to the fact that the film thickness at 10mm exceeded the minimum thickness that can remain on the metal surface during laser irradiation. The majority of the welds produced on the anodised sheet were full penetration welds. However welding at 3.0kW, and 2.5kW gave a clear indication of how the laser welding parameters interacted with the anodic film to affect the observed welding efficiency, Fig. 5.12.

Macrostructurally it was shown that for a constant laser power, weld penetration increased with a decrease in laser beam traverse speed. Where full penetration was achieved for both the lower, and higher beam traverse speeds the effect of a reduction in speed was to result in a broader weld. Similar effects were observed for a constant beam traverse speed, and increasing laser power. The laser welding process is capable of producing narrow welds, and advantage should be taken of this fact, as the minimum amount of material subjected to a non standard heat treatment the better the ultimate properties of the joint. It is therefore suggested the optimum welding parameters should be those which just penetrate the thickness of
the material being joined, and do not produce excessive broadening of the fusion zone.

Microstructural examination of the welds in the anodised material revealed the weld microstructure to consist of a uniform cellular dendritic microstructure which was very similar to that of the as-received material, while the base metal also consisted of elongated grains with a fine precipitate of Mg$_2$Si. These Mg$_2$Si particles could not be distinguished in the fusion zone, and a degree of grain coarsening was observed to have occurred in the HAZ along with coarsening of the Mg$_2$Si particles in this region. The laser power, and beam traverse speed were found to affect the dendrite cell sizes. Increasing the laser beam power at a constant speed resulted in an increase in the dendrite cell size, also decreasing the beam traverse speed at a constant laser power resulted in an increase in dendrite cell size, Fig. 5.17. This is a consequence of the higher energy input into the weld under these conditions.

A series of measurements carried out on polished and etched samples revealed dendrite arm spacings to be in the range 2 to 4$\mu$m at 4.0kW for the 10$\mu$m anodic film thickness material. This observation was consistent for all the film thicknesses studied. Microhardness measurements revealed the weld region to be of reduced hardness compared to the base metal. It is possible that this is due to the dissolution of the Mg$_2$Si precipitates in the weld region, and a disruption of the atom clusters forming the Guinier Preston, (GP), zones. In a study on welds in Al 5456 Altshuller and Ryvolla$^{(90)}$ recorded dendrite arm spacings of 5$\mu$m in 6$\mu$m thick material welded using a 5kW laser at 8mm/s. Small dendrite arm spacings could be the result of a large heat-sink from the 6mm thick material effectively quenching the weld. It is however not clear if welding was carried out at 5kW or if this was the lasers maximum power capacity. This dendrite arm spacing was reported to be smaller than that recorded for GTAW welds. These workers
suggest that in the light of the results obtained from GTAW and GMAW welds, the smaller dendrite arm spacings of the laser welds should produce similar if not superior mechanical properties in the joint compared to GTAW welds.

In laser welding by virtue of the fact that the heat input into the material is low, the solidification rates are high, $10^4$ to $10^6\text{s}^{-1}$. It is known that the final dendrite arm spacing in a casting is directly proportional to the solidification rate, this would account for the fine dendrite arm spacings recorded in these laser welding studies. The effects of the fast cooling rates, and low heat inputs associated with the laser welding process can therefore be summarised as follows.

1. An increase in solidification rate, and hence a small dendrite spacing, and an increase in mechanical properties.
2. Less melting in the weld region, resulting in a narrow weld bead and HAZ. A large aspect ratio leads to higher weld properties.
3. Less time at high temperatures for metallurgical reactions to occur.

**Porosity**

Both shrinkage and gas porosity are common defects encountered in the welding of aluminium alloys, and both seemed to be present in certain samples of the as-received Al welds examined. In laser welding the problem of porosity and cracking is compounded by the 'keyhole' mechanism, fast welding speeds, and associated fast cooling rates characteristic of the process, however the use of filler alloys and preheat could result in a reduction in this observed porosity. The weld made at 4kW, and 67mms$^{-1}$, is an example of a weld in which porosity is completely absent, Fig. 5.35. The pore in Fig. 5.6a appears to feature melt tracks
within it, and this is believed to be a feature of the non-uniform penetration observed in the longitudinal macrographs shown in Fig. 5.5c. These macrosections of the non-full penetration laser welds, showed pores to occur in the vicinity of the root of the weld. This suggests a lack of fluidity of the melt in this narrow region of the weld. Meleka\(^{(91)}\) studying the instability of the 'keyhole' in EB welding, reported an abnormally high incidence of weld root porosity, caused by instability of the 'keyhole'. As both processes rely on similar mechanisms, it is likely that this also applies in laser beam welding. Instability of the 'keyhole' could occur as a consequence of the high reflectivity of the workpiece affecting the lasing process. Certainly fluctuations in power were observed in the welding of the Al 6061. This was more pronounced in the welding of the as received material in which the incidence of non-uniform penetration was greatest. This implies that careful control of the energy input into the workpiece could reduce the incidence of this sort of porosity. Attention should therefore be paid to focusing, plasma control, and the shielding of the laser cavity from reflected radiation.

The backfilling of cracks by low melting point eutectics can also result in voids in the fusion zone, where there is insufficient melt to completely fill all the cracks, and compensate for solidification shrinkage in the melt region. Metzbower \textit{et al}\(^{(55)}\) report observing porosity in welds of Al 5456, and suggest that while this porosity appears to reduce elongation, and UTS, it does not seem to affect the yield strength of the alloy. It is usually overcome by using filler materials, which are less prone to cracking to reduce the weld metal susceptibility, and the use of preheat prior to welding to reduce the thermal shock to the material.

The welds in anodised aluminium, unlike those in the as received material displayed porosity distributed evenly within the fusion zone. This porosity was thought to
be the result of gases being trapped in the solidifying metal. As the solidification times are too fast for pore growth and coalescence to occur, leading to the observed uniform distribution of pores. However, closer examination of the pores appeared to suggest that within the fusion zone, certain areas of the melt had become surrounded by oxide possibly due to the decomposition of the surface oxide coating. On solidification, these regions failed to adhere to the surrounding weld metal. Figs. 5.15b, and 5.15c show an area thought to be caused by this oxide film boundary on an area of melt, and an area where this oxide bounded area has detached from the surrounding weld metal. There was a noticeable absence of shrinkage porosity in the anodised Al welds. The larger heat input into the material, by virtue of the high melting point of the Al₂O₃ coating would result in a decrease in thermal gradients within the material, thereby reducing the thermal shock to the material on solidification.

The role of the anodic film in contributing to these areas of porosity is evident from Fig. 5.38 showing two welds in Al 6061 anodised to 10mm, and 16mm respectively. The increased amount of porosity in the 16mm anodised material is clearly evident. However, it was possible to produce certain welds free of porosity in the anodised material as shown in Fig. 5.36. The implication of this is that, as the anodic film thickness did not appear to have any great effect on the resultant weld penetration, Fig. 5.19, the minimum thickness required to increase weld penetration should be used.

Cracking

Fig. 5.6c, and 5.6d, show examples of the cracks observed in certain samples of the as received Al 6061 welds. Ricciardi et al.(83), also reported incidences of
cracking in Al 6061 welds made by a laser, although it is not clear if these workers observed cracking in all their welds. It is evident that the crack in Fig. 5.6c is an intergranular crack. This manner of cracking is regularly observed in Al welding, and has been extensively studied and well documented. According to Kou et al(92), intergranular cracking occurs in the terminal stages of solidification, when the stresses developed in the grains adjacent to the last portions of melt to solidify exceed the strength of the solidifying metal. Such stresses can be induced by solidification shrinkage of the weld metal, and thermal contraction of the workpiece. Solidification is intergranular, and most theories of solidification cracking embody the concept of a coherent interlocking solid network, separated by essentially continuous thin films of liquid which are ruptured by contraction stresses. Fig. 5.39 shows the relationship between crack sensitivity, and % Mg$_2$Si for AlMgSi alloys. Al 6061 contains about 1% Mg$_2$Si, and as such is highly crack sensitive.

It is known that by refining the grain structure of welds solidification cracking can be reduced, as coarse columnar grains are more susceptible to cracking than fine equiaxed grains. This is because equiaxed grains can deform to accommodate contraction strains more easily. Liquid feeding, and therefore healing of incipient cracks can also be more effective in fine grained materials. The structures produced in laser beam welding are relatively refined columnar grain structures. However the cooling rates associated with laser welding are 10 to 20 times shorter than those encountered in arc welding processes. The consequence of this is that the material is subjected to high thermal stresses due to steep temperature gradients present in the structure. Also abrupt changes in growth occurs towards the weld centreline, Fig. 5.4 shows the grains changing direction to follow the heat source. These characteristics generally result in structures that are weak under stress, and as such result in cracking.
Preheating during welding could alleviate the effects of the steep temperature gradient, reducing the thermal stresses in the solidifying weld metal. This may result in a reduction in the incidence of cracking in these welds.

5.4.2 Aluminium 6061 based metal matrix composites

Effect of laser welding parameters

One of the main problems usually associated with the processing of composite materials using conventional processes arises from the differences in the physical properties of the matrix and the reinforcing material. These differences are shown below for the present systems.

<table>
<thead>
<tr>
<th></th>
<th>Aluminium 6061</th>
<th>SiC</th>
<th>B</th>
</tr>
</thead>
<tbody>
<tr>
<td>Melting point(°C)</td>
<td>660</td>
<td>3100</td>
<td>2180</td>
</tr>
<tr>
<td>Boiling point (°C)</td>
<td>1275</td>
<td>-</td>
<td>3800</td>
</tr>
<tr>
<td>Coefficient of Expansion(10⁻⁶K⁻¹mm⁻¹)</td>
<td>23.0</td>
<td>3.0</td>
<td>-</td>
</tr>
</tbody>
</table>

From the observations noted it was evident that it would be possible to produce melt tracks on the composite materials using the CO₂ laser, and that this would result in discrete microstructural changes in the area impinged by the laser beam. It can therefore be construed that despite the large differences in properties such as the melting and boiling points of aluminium, magnesium, silicon carbide, and boron it was possible to bring about melting in the higher melting point constituents without total vaporisation of the aluminium content of the weld zone. It is thought that magnesium may have been totally lost from the weld zone during welding as it was not detected in any composition scans during this study.
Al/B fibre reinforced MMC

In the Al/B fibre reinforced composite traverse of the laser beam results in melting of the aluminium matrix, and the reinforcing boron fibres. The variation of the laser beam traverse speed during welding did not have any direct effect on the weld region microstructure, except to decrease the weld penetration achieved as the beam traverse speed was increased. The development of the novel structures identified in section 5.3.3 are believed to be a consequence of the high temperatures, and fast cooling rates associated with the laser welding process. The solubility of boron in aluminium is low, and is 0.22 wt% at a eutectic temperature of 659.7°C. Although solubility increases with temperature reaching up to 1.5 wt% at 1500°C, the solubility at room temperature is negligible. The development of the phases observed in the Al/B weld region can be accounted in term of the Al-B phase diagram, Fig. 5.40. Fig. 5.40 reveals two phases to be stable at room temperature in the Al-B system, AlB₂, and AlB₁₂. These phases correspond to those identified from the diffraction patterns taken from the weld zone, alpha and beta AlB₁₂, and alpha AlB₂. The AlB₁₂, and AlB₂ phases are dispersed in the aluminium matrix in the form of irregular particles and 'needle-like' structures. AlB₂ forms from AlB₁₂ by peritectic reaction at 1250K. It is believed that the fast cooling rates associated with the laser welding process resulted in the retention of large amounts of AlB₁₂ in the fusion region. AlB₁₂ was observed more widely than AlB₂ confirming the fact that the time for metallurgical reactions to occur under conditions of fast solidification is limited.
Porosity

Examination of the joints in the boron fibre reinforced material, Figs. 5.22a, and 5.22b shows that in both welds, i.e parallel to the boron fibres and across them, there was evidence of porosity adjacent to the fusion zone. This porosity did not appear to be either shrinkage porosity, as identified in the as received Al welds, or gas porosity characterised by a 'round' shape. It is however thought to be the result of vaporisation of the matrix material, aluminium, adjacent to the fusion zone due to the development of hot spots in certain areas during welding. At the high temperatures reached preferential loss of elements such as magnesium, (boiling point 1092°C), from the fusion region is quite likely. This region reaches temperatures in excess of 2180°C to result in melting of the boron fibres, which exceeds the boiling point of magnesium at 1092°C by a long margin. When the laser beam traverse speed was increased while keeping the laser power constant, a decrease in the porosity was observed. The welds obtained under these conditions are shown in Figs. 5.23a, and 5.23b, from which the macrostructural variation of welds made at high speed can be clearly distinguished. This may be explained as resulting from a decrease in the interaction time of the laser beam at any point on the material surface, thus resulting in a reduction in the heat input into the material and the amount of aluminium, or magnesium which would be vaporised from the fusion line during the weld run.

Cracking

While in the Al/B fibre reinforced metal matrix composite material there was no cracking evident microstructurally in the fusion zone, some of the welds showed macrocracking. It was observed that while the welds made transverse to the fibres
did not show any macro-cracking, certain welds made parallel to the fibres were accompanied by a longitudinal crack running along the edges of the weld bead. The difference in the nature of the weld beads is believed to be a consequence of the nature of the composite material itself. In this particular material the strength lies along the direction of the fibres. Thus in welds transverse to the fibres although the aluminium matrix undergoes shrinkage on solidification the bulk structure is held together by the boron fibres. In the welds parallel to the fibres the shrinkage in the aluminium matrix on solidification results in a separation of the matrix from the fibres along the weld edge. This would suggest that if the laser welding of the fibre reinforced material was desirable, the joint edges would have to be forced together in order to maintain contact between the fibres and the matrix on melting, and cooling.

**Al/SiC particulate reinforced MMC**

In the Al/SiC metal matrix composite welds, as with the Al/B composite melting and mixing of the aluminium matrix and the silicon carbide reinforcing particles occurred resulting in the observed weld structure, Fig. 5.30a. The main effect of the increase in laser beam traverse speed from 58.3mm/s to 117mm/s was a significant decrease in weld penetration. The 'needle-like' structures in the weld zone also appeared to be larger in the weld made at 58.3mm/s. This is not surprising as the heat input into the weld is greater in this case, allowing time for diffusion mechanisms to result in growth of the weld structures. As documented in section 5.3.3, the 'needle-like' phases observed in the weld region were identified as Al₄SiC₄, while a certain amount of difficulty was encountered in conclusively determining the composition of the matrix surrounding the needles. The conclusions drawn in the identifying these phases can be explained in terms of
the Al-C, and the Al-Si systems, (Figs 5.41, and 5.42). Silicon is completely miscible with aluminium in the liquid state, while carbon is known to have very low solubility limits, which although not completely established are known to be less than for boron\(^{(84)}\). The solid solubility of silicon has been quoted at less than 10 at.% but greater than 1 at.%\(^{(84)}\). The analysis of the melt zone revealed carbon to be present in the needle-like structures, (n), while the grey particles, (g), were found to be residual silicon, with the remaining silicon thought to be present in the aluminium matrix in the form of a solid solution, as silicon is soluble up to 1.65 at.% in solid aluminium.

Barczak\(^{(92)}\) investigating the reaction of Al, Si, and C in a vacuum furnace observed the formation of alpha-Al\(_4\)SiC\(_4\) from the sublimation of the constituent materials in the furnace. The resultant Al\(_4\)SiC\(_4\) was reported to form prismatic needles, irregular forms and hexagonal plates. The needles observed in this study showed similar forms. It is believed that during laser welding at the high energy densities involved the region impinged by the laser beam is vaporised to form a 'keyhole', and in the vapour phase the components Al, Si, and C react to form Al\(_4\)SiC\(_4\).

A region adjacent to the melt zone in which a certain amount of redistribution of the reinforcing particles in the aluminium 6061 matrix is believed to have occurred was identified, Fig. 5.30c. The observed change in the reinforcing particles is thought to be the result of the aluminium matrix flowing around the SiC particles, allowing a certain amount of coarsening to occur. This situation could occur if heat is transferred by conduction from the molten weld region to the surrounding metal. The temperatures reached by this form of heat transfer would not be high enough to result in melting of the SiC particles but would probably facilitate diffusion, and hence coarsening of the particles. As little structural change has occurred in this
region its properties should resemble those of the base metal more closely than the weld region.

Porosity

Examination of the joint in Fig. 5.30 revealed a large amount of porosity in the fusion zone, and immediately adjacent to it. As was the case with the Al/B fibre reinforced material, it is believed that this is due to the vaporisation of high vapour pressure elements such as magnesium, (1092°C), and possibly some aluminium, (2174°C) at the high temperatures required to melt SiC, (3100°C). It is also possible that in this material made by the powder metallurgy route there was some residual gas present in the metal. An expulsion of this gas could account for the violent reaction of the material to welding and the observed porosity in the fusion zone. There is little data on the laser welding of powder metallurgy aluminium alloys, but Mosca et al (94) report gas evolution on laser welding of sintered aluminium alloys resulting in spongy welds. They suggest an excessive quantity of oxides which absorb the laser energy more than the metal matrix giving rise to over heated spots in the weld. Increasing the laser beam traverse speed and reducing the incident beam power was found to result in a decrease in porosity, although a corresponding decrease in weld penetration was obtained. A decrease in the laser beam interaction time at any point on the material surface would result in less heat being input into the material with a decrease in the vaporisation of the high vapour pressure constituents, and the incidence of hot spots within the weld region. A decrease in the laser beam power would bring about a similar effect.
Cracking

The Al/SiC particulate reinforced material showed no cracking either macro, or microstructurally. This is thought to be a consequence of the fact that the phases within the fusion zone have a lower thermal conductivity than the base alloy. The thermal gradient across into the base alloy is reduced, this in turn reduces the thermal shock to the material limiting contraction on solidification. The resultant effect is that the fusion zone is under less stress than is normally encountered in the welding of aluminium alloys, and consequently is less prone to cracking.
Fig. 5.1  Weld bead profiles showing the effect of laser power, and laser beam traverse speed on weld penetration, weld shape and microstructure for as received Al 6061
Fig. 5.2  Weld penetration as a function of laser power for as-received Al 6061.
Fig. 5.3  Macro and microstructure of a typical weld in as recieved Al 6061.
(a) Overall macrostructure.
(b) Base alloy away from the weld zone.
(c) Weld heat affected zone, (HAZ).
(d) Cellular dendritic microstructure in fusion zone.
Fig. 5.4  Weld metal grains changing direction to follow the heat source.
Fig. 5.5  Macrograph of typical weld beads in Al 6061.

(a)  Weld top bead.
(b)  Weld bottom bead.
(c)  Longitudinal macrograph showing non-uniform penetration along the weld seam.
Fig. 5.6 Typical pore and crack morphology observed in as received Al 6061 welds.

(a) Showing gross shrinkage, (s), and gas, (g), porosity within a non-full penetration weld bead.

(b) Melt tracks within a pore.

(c) Intergranular cracking along weld centreline.

(d) Crack in as received Al 6061 weld.
Fig. 5.7  The effect of laser beam traverse speed on weld penetration in as received Al 6061.
Fig. 5.8  Weld bead width as a function of laser beam traverse speed for as received Al 6061
Fig. 5.9 The effect of laser beam traverse speed on dendrite cell size, in as received Al 6061.

(a) Typical dendrite cells observed in as received Al 6061 4.5kW, 117mm/s.

(b) Typical dendrite cells observed in as received Al 6061 4.5kW, 83mm/s.
Fig. 5.10a Typical microhardness traverse across an as-received Al 6061 weld.
Fig 5.10b  Typical microhardness traverses across an as-received Al 6061 weld
Fig. 5.11 Weld bead profiles showing the effect of laser power and laser beam traverse speed on weld penetration, weld shape and microstructure for anodised Al 6061
Fig. 5.12  Weld penetration as a function of laser power for anodised Al 6061, (12μm anodic film thickness).
Fig. 5.13  Macro and microstructure of a typical weld in anodised Al 6061.

(a) Overall macrostructure.

(b) Base alloy away from weld zone.

(c) Weld heat affected zone, (HAZ).

(d) Cellular dendritic microstructure.
Fig. 5.14 Macrographs of typical weld beads in anodised Al 6061.

(a) Weld top bead.

(b) Weld bottom bead.

(c) Longitudinal macrograph showing reduced non-uniform penetration along weld seam.
Fig. 5.15  Areas of non-adhesion observed in anodised Al 6061 welds.

(a) Several areas of non-adhesion appearing as porosity in weld microstructure.

(b) SEM micrograph of areas of non-adhesion intact in weld region, and bounded by oxide film.

(c) SEM micrograph of area of non-adhesion dissociated from surrounding metal.
Fig. 5.16  Weld penetration as a function of laser beam traverse speed for anodised Al 6061, (12μm anodic film thickness).
Fig. 5.17 The effect of laser beam traverse speed on dendrite cell size in anodised Al 6061.

(a) Typical dendrite cells observed in anodised Al 6061, 4.0kW, 117mm/s.

(b) Typical dendrite cells observed in anodised Al 6061, 4.0kW, 83mm/s.
Fig. 5.18  Weld bead top width as a function of laser beam traverse speed for anodised Al 6061, (12μm anodic film thickness).
Fig. 5.19  Weld penetration as a function of anodic film thickness, for a constant laser power, and beam traverse speed (100mm/s).
Fig. 5.20 Weld bead top width as a function of anodic film thickness for a constant laser power, and beam traverse speed (100mm/s).
Fig. 5.21 Typical microhardness traverse across an anodised Al 6061 weld.
Fig 5.21b  Typical microhardness traverse across an annealed Al 6061 weld
Fig. 5.22  Macro and microstructural variation of a weld in Al/B fibre reinforced MMC.

(a) Overall macrostructure parallel to the reinforcing fibres.
(b) Overall macrostructure transverse to the reinforcing fibres.
(c) Parent metal away from the fusion zone transverse to the reinforcing fibres.
(d) Parent metal away from the fusion zone, longitudinal view of the reinforcing fibres.
(e) Weld fusion line parallel to the fibres.
(f) Weld fusion line transverse to the fibres.
(g) Weld fusion zone typical of welds transverse, and parallel to the reinforcing fibres.
Macrostructural variation of welds in Al/B fibre reinforced MMC, showing the effect of increased laser beam traverse speed in decreasing weld penetration, and porosity adjacent to the fusion zone.

(a) Weld made parallel to the reinforcing fibres.

(b) Weld made transverse to the reinforcing fibres.
Fig. 5.24 Secondary ion micrographs obtained from SIMS analysis of the Al/B fibre reinforced MMC.

(a) $\text{Al}^{27}$ distribution in the melt zone.

(b) $\text{B}^{11}$ distribution in the melt zone.
Fig. 5.25 Transmission electron micrograph of irregular shaped particle, (s), Fig. 5.22g.

(a) Transmission electron micrograph.

(b) Diffraction pattern taken from particle, (s), corresponding to [100] plane of α AlB₁₂.

(c) Diffraction pattern corresponding to [010] plane of β AlB₁₂.
Fig. 5.26  Transmission electron micrograph of large needle, (bn), in Fig. 5.22g.
(a) Transmission electron micrograph.
(b) Diffraction pattern taken from large needle, d-spacing corresponding to that of AlB$_2$. 
Fig. 5.27  Scanning electron microscope, (EDS), analysis of large needle, showing peaks due to Al. and W.
Fig. 5.28  Transmission electron micrograph of small needles, (sn), shown in Fig. 5.22g.

(a)  Transmission electron micrograph.

(b)  Diffraction pattern taken from small needles, corresponding to known d-spacings of either $\alpha$ or $\beta$ AlB$_{12}$. 
Fig. 5.29  Scanning electron microscope, (EDS), analysis of small needles showing them to contain Al, Si, W, and possibly B.
Fig. 5.30 Macro and microstructural variation of a typical weld in Al/SiC particulate reinforced MMC.

(a) Overall macrostructure.

(b) Parent Al/SiC metal away from the fusion zone.

(c) Heat affected zone adjacent to the fusion zone showing apparent denudation in reinforcing particles.

(d) Fusion zone.
Fig. 5.31  Secondary ion micrographs obtained from SIMS analysis of Al/SiC particulate reinforced MMC welds.

(a)  $\text{Al}^{27}$ distribution in melt zone. Note needle structures, (n), clearly outlined.

(b)  $\text{Cl}^{12}$ distribution for area analysed in 5.33a.

(c)  $\text{Si}^{28}$ distribution showing irregular particles, (g), to contain Si.
Fig. 5.32

Transmission electron micrograph of needle, (a), shown in Fig. 5.32d.

(a) Transmission electron micrograph.

(b) Associated diffraction pattern, with d-spacings corresponding to known spacings for Al$_4$SiC$_4$. 
Fig. 5.33

(a) X-ray map of Si area.

(b) Region from which map was taken.
Fig. 5.34  Scanning electron microscope, (EDS), analysis of Al/SiC particulate reinforced MMC weld region showing the presence of Al, and Si.
Fig. 5.35  Weld produced under optimum welding conditions for as received Al 6061. Note the absence of porosity, or cracking in the weld region.
Fig. 5.36  Weld produced under optimum welding conditions for as received Al 6061. Note the absence of porosity, or cracking in the weld region.
Fig. 5.37 Diagramatic representation of the structure of the anodic film produced by the sulphuric acid anodising process.
Fig. 5.38 The effect of anodic film thickness on porosity
(a) 10 μm film thickness
(b) 16 μm film thickness
Fig. 5.39  The relationship between crack sensitivity, and Mg$_2$Si for AlMgSi alloys.
Fig. 5.40  Aluminium-boron phase diagram.
According to Obinata and Komatsu

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Fig. 5.41  Aluminium-carbon phase system
Fig. 5.42  Aluminium-silicon phase diagram.
CHAPTER 6

SUMMARY, CONCLUSIONS, AND SUGGESTIONS FOR FUTURE WORK

This chapter is divided into three main sections. First a summary of the work reported in this thesis is given, together with a resume of the common trends observed in the welding of the different materials studied. This is followed by the conclusions which can be drawn from the observed results. Finally suggestions are made of areas of future interest which would supplement the work in this thesis, and further improve knowledge in the field.

6.1 Summary

Nimonic C263 and PE11

Nimonic C263, and PE11 were welded with the Welding Institute 5kW continuous wave CO₂ laser, Nimonic C263 was also welded with a 2kW continuous wave Control laser. Bead on plate welds were produced in 1.2mm thick material over a range of incident laser powers and speeds. It was found that for both alloy C263, and PE11 weld penetration increased as the incident laser beam power increased at a constant beam traverse speed. The weld penetration was also observed to decrease with increase in laser beam traverse speed for a constant incident laser beam power.
Microstructurally, the effect of laser beam power and speed on dendrite arm spacings, and carbide precipitation was determined. It was found that dendrite arm spacing increased with increase in incident laser beam power, and decreased with increase in laser beam traverse speed. The minimum dendrite arm spacing was measured at 2μm, and the maximum, 5μm. A number of microhardness traverses were made across the weld in Nimonic C263, and PE11. It was found that in alloy C263, the hardness in the weld region consistently exceeded that of the parent metal. In alloy PE11, the hardness in the weld region was observed to be marginally lower than that of the parent metal. In alloy C263, the precipitation in the weld metal exceeds that observed in the parent metal. As carbides act as obstacles to the movement of dislocations, and grain boundary sliding this is thought to account for the increased hardness observed in this alloy. In alloy PE11, while the observed precipitation in the weld metal exceeds that observed in the weld metal of alloy C263, it does not exceed that of the parent metal. As the difference in carbide density is small this results in the marginal difference recorded in the hardness of the weld metal, and the parent metal. The difference in the precipitation characteristics of alloy C263, and PE11 is a consequence of the fact that the alloys belong to different alloy systems. Nimonic C263 belongs to the Ni-Cr-Co alloy system, while Nimonic PE11 belongs to the Ni-Cr-Fe system. The presence of Fe decreases the solubility of carbon in the α-Ni matrix, leading to large degrees of precipitation in this group of alloys.

From the measured results obtained from the welded samples, a number of derived results were calculated; including the welding efficiencies of the alloys, and a compound parameter P/VD. The welding efficiency of Nimonic PE11 was observed to be marginally greater than that of Nimonic C263.
P/VD was identified as a compound parameter by which the synergistic effects of the laser beam welding parameters; incident beam power, laser beam traverse speed, and incident laser beam spot size could be quantified. It was established that weld penetration increased as P/VD increased for both the 2kW, and the 5kW laser in the welding of alloy C263. For both the 2kW laser, and the 5kW laser, a graph of P/VD for all the welding conditions studied produced a line of constant slope. This suggests that a laser can be characterised by its graph of weld penetration vs P/VD. Thus if the desired weld penetration were known, P/VD required to achieve it could be identified from the graph. Similarly from known values of P/VD, the achievable weld penetration could be identified.

Work on the welding of alloy C263 with both the 2kW, and the 5kW laser revealed a greater dependency of weld penetration on laser beam spot size, and beam mode quality. Certainly for the joining of thin gauge materials, it would not be desirable to use the highest laser power available. The loss in beam mode quality, and the increase in incident beam spot size results in a loss in efficiency of welding compared to that achieved with a laser of lower maximum power capacity, coupled with the associated higher mode quality and smaller incident laser beam spot size.

As-received and anodised Al 6061

Aluminium alloy 6061 was welded with the welding Institute 5kW continuous wave CO₂ laser in both the as received, and anodised surface finish. It was found that weld penetration increased as the incident laser beam power was increased, while an increase in laser beam traverse speed resulted in a decrease in weld penetration. For the as-received Al 6061 a threshold power was established at 3.0kW at which no significant melt zone could be produced in the 2mm thick Al
6061 sheet. It was found that sulphuric acid anodising of the sheet resulted in an increase in welding efficiency of the Al 6061. Evidence of this is found in the fact that while no significant melt zone could be produced in the as-received sheet at 3.0kW, in the anodised sheet at 2.5kW it was possible to penetrate up to 50% of the material thickness at a beam traverse speed of 100mm/s.

In both the as-received, and anodised Al 6061 welds it was observed that as the beam traverse speed increased the weld bead top width decreased. As the laser beam power was increased at a constant beam traverse speed, the weld bead top width was found to increase.

The effect of anodic film thickness on weld penetration was investigated by making welds on Al 6061 anodised to thicknesses ranging from 10µm to 16µm. At 4.0kW, and 3.5kW full penetration was achieved at most of the speeds studied. It was however clear from the lower laser powers of 3.0kW, and 2.5kW that the anodic film thickness did not appear to play any significant role in the weld penetration achieved at these powers. Anodic film thickness while not affecting weld penetration, appeared to have an effect on weld bead top width, with thicker coatings resulting in slightly narrower weld bead top widths.

The weld region exhibited a cellular dendritic structure in both the as-received and anodised welds, while the parent metal consisted of elongated grains in which the Mg2Si precipitates are clearly visible. It was not possible to distinguish these precipitates in the weld region. The welding process resulted in a small degree of grain coarsening in the heat affected zone, but there was no evidence of grain boundary liquation. The average dendrite arm spacing in the welds in both the as-received, and anodised Al 6061 was measured at 3µm.
Microhardness traverses across the joint region in the as received, and anodised Al 6061 welds revealed the hardness in the weld region was lower than that of the parent metal in both cases.

In the joining of Al based materials, porosity and cracking appears to be a major problem. It was however possible to produce a weld free from cracks and porosity at 4kW, and a beam traverse speed of 67mm/s in the 2mm thick as-received Al sheet. These conditions can be described as the optimum welding parameters for as received Al 6061 under the conditions of laser beam mode, beam spot size, and gas flow rate prevailing at the time of welding.

While cracking in the weld bead was not observed in the anodised welds, some of them did show the effect of the oxide film on the surface, in the form of areas of non-adhesion within the fusion zone. These regions are thought to be the result of oxide from the anodic film becoming incorporated into the melt, and forming a boundary around globules of melt preventing them from adhering to the bulk weld metal on solidification. These region coupled with areas of gas porosity resulting from dissociation of the oxide film in some cases led to a high degree of porosity. It is proposed that the use of a fluxing agent to disperse the oxide in the weld metal, or remove it from the weld may lead to improvements in the weld microstructure. The improvement in welding efficiency afforded by the sulphuric acid anodising process is great, and effort is needed to improve the microstructural integrity of the welded joint.
**Al 6061 based metal matrix composites**

Boron fibre reinforced metal matrix composite was welded using the 2kW laser, and welds were produced in the silicon carbide particulate reinforced material using the 5kW laser. While difficulties were encountered in the welding of the SiC particulate reinforced MMC due to the expulsion of fumes and debry from the workpiece surface during beam traverse, it was possible to produce melt zones in both materials. Discrete microstructural changes were observed in the fusion zone consistent with the melting and mixing of the matrix, and reinforcing fibres or particles to produce an alloy in the weld zone. The novel phases observed in the weld zone were identified with the aid of transmission electron microscopy, energy dispersive spectroscopy, and secondary ion mass spectroscopy.

In the boron fibre reinforced MMC the weld zone was found to consist of irregular particles of $\alpha$ and $\beta$-$\text{Al}_2\text{B}_12$, and small 'needle-like' structures of $\alpha$-$\text{AlB}_2$ dispersed in an Al matrix. The SiC particulate reinforced MMC weld zone was found to consist of 'needle-like' $\text{Al}_4\text{SiC}_4$, and blocky particles of Si precipitate dispersed in an Al-Si matrix.

In both the boron fibre reinforced MMC, and the SiC particulate reinforced MMC, porosity was observed adjacent to the fusion region. In the boron fibre reinforced MMC, although the fusion zone was porosity and crack free, the region adjacent to this zone showed loss of material. Optimising the laser welding parameters such that the amount of heat input into the material is insufficient to cause vaporisation of the matrix material could reduce this porosity, and was demonstrated in this work. The use of a filler alloy during welding to replace lost material could also prove helpful in reducing the porosity.
In the SiC particulate reinforced MMC the observed porosity extended through the entire joint; i.e., the fusion zone, and adjacent heat affected zone. In this material as was observed with the boron fibre reinforced composite, a decrease in heat input into the weld region appeared to result in a corresponding decrease in the associated weld porosity, although this was accompanied by a drop in weld penetration.

6.2 Conclusions

Nimonics C263 and PE11

1. The Nimonic alloys C263, and PE11 can be welded with a laser. The observed weld depth in alloys C263, and PE11 increases with increase in incident laser beam power at a constant beam traverse speed, while the observed weld penetration decreases with increase in laser beam traverse speed for a constant incident beam power. The use of a lens of short focal length serves to concentrate the laser beam power resulting in increased weld penetration.

2. The welding efficiency of Nimonic alloy PE11 exceeds that of alloy C263, and is evidenced in the larger melt volume produced in this alloy.

3. Weld microstructure is similar in both alloys, and the observed dendrite arm spacings increased as the laser beam power was increased, or laser beam traverse speed was decreased. The observed spacings were within the range 2 to 5μm.

4. Generally carbide precipitation within the weld region is more pronounced for alloy PE11 than for alloy C263, and was found to increase as the laser beam power increased, or laser beam traverse speed was decreased.
5. Hardness in the weld region is greater than that of the parent metal in alloy C263, while in alloy PE11 hardness in the weld region is marginally below that of the parent metal.

6. Laser welding parameters have a synergistic effect on the resultant welding efficiency, therefore they must be considered in this light.

7. In the joining of thin gauge materials above a threshold laser power the resultant welding efficiency shows a greater dependence on laser beam mode, and incident beam spot size than on the other laser welding parameters such as; laser power, laser beam traverse speed, and shield gas.

As-received and anodised Al 6061

8. In the joining of highly reflective metals such as Al 6061 with a laser, it is possible to reach a threshold at low laser powers below which no significant melting of the workpiece is achieved. Above this threshold weld penetration increases as the incident laser beam power is increased at a constant beam traverse speed, and decreases as the laser beam traverse speed is increased for a constant incident beam power.

9. Anodising Al 6061 by the sulphuric acid anodising process prior to welding results in an increase in laser beam absorption, and hence welding efficiency.

10. The presence of an anodic film on the metal surface prior to welding does not affect the weld microstructure, and both the as-received, and anodised welds showed a cellular dendritic structure in which the Mg₂Si precipitate could not be distinguished.
11. Weld region hardness in both the as-received, and anodised surface finishes is below that of the parent metal.

12. While it was possible to produce welds free of porosity and cracking in some cases in both the as-received, and anodised Al 6061, welds made under certain conditions were prone to porosity in both surface conditions, and certain welds in the as-received Al 6061 showed cracks. The use of a filler alloys less prone to cracking, or preheat prior to welding should therefore be considered as a mean of alleviating this problem.

Al/B fibre reinforced MMC

13. It is possible to melt Al 6061/boron fibre reinforced MMC with a laser.

14. It was not possible to produce welds parallel to the reinforcing fibres in this material without the use of a restraint forcing the two halves of the joint together, while it is possible to produce welds transverse to the reinforcing fibres.

15. The welding process disrupts the ordered nature of the composite in the fusion zone, producing an alloy in this region displaying novel structures. The use of the laser welding process in the joining of the Al/B MMC would therefore result in a deterioration of the unidirectional strength of the alloy, and should be considered only if this loss in strength does not render the material unsuitable for its end use.

16. The microstructure in the weld zone of the Al/B MMC was identified as a dispersion of irregular particles of a and b-AlB₁₂, and 'needle-like' a-AlB₂ in an Al matrix.
17. The porosity observed adjacent to the fusion zone in this material can be reduced by decreasing the heat input into the weld region during welding.

Al/SiC particulate reinforced MMC

18. While it was possible to produce a melt zone in the Al 6061/SiC particulate reinforced MMC welding is hampered by the expulsion of fumes and debris from the workpiece leading to contamination, and ultimate breakage of the focusing lens. The use of a nozzle or shielding mechanism, coupled with a strong flow of gas coaxial with the beam during welding should therefore be considered in all cases.

19. Welding disrupts the regular distribution of reinforcing particles to produce an alloy in the fusion zone. The resultant microstructure was identified as particles of Si, together with needles of Al₄SiC₄ in an Al-Si matrix.

20. A decrease of heat input into the fusion zone reduced the amount of porosity observed in this and the adjacent heat affected zone. The use of a filler alloy, and possibly the injection of particles into the fusion zone should probably be considered as standard practice.

6.3 Suggestions for future work

The synergistic effect laser welding parameters such as laser beam power, laser beam traverse speed, lens focal length, laser beam mode, and shield gas on welding efficiency has been identified. If it is to be possible to make adequate use of a parameter such as P/VD to characterise a laser, and determine the laser welding
parameters to achieve the required weld penetration from such a graph, a means
must be determined of standardising the laser beam mode. This would demand
reproducibility, and would probably require a photosensing device to allow the
laser operator to tune the laser up to a defined standard prior to operation of the
equipment.

The microhardness test has been used as an indication of the weld properties, and
has been used to interpret the effects of the observed microstructural changes in the
fusion zone. A series of mechanical tests would supplement this aspect of the
work, and allow it to be determined whether the parent metal, the heat affected
zone, or the weld metal forms the weaker part of the joint.

There is great scope for investigation into ways of reducing porosity in the anodised
Al 6061 welds. Areas for consideration include the fact that it may not be
necessary to anodise the bulk structure to be joined, and a start up area could be
anodised, and a keyhole initiated in this area which would sustain itself and
facilitate welding. After welding the starter area could be discarded, this could
remove the possibility of oxide contamination of the weld. The use of a fluxing
agent to remove oxide from the weld once melting has been accomplished should
also prove an interesting area of study. It may also be worth investigating the
effect of anodising processes other than the sulphuric acid process on welding
efficiency, and subsequent fusion zone integrity. Finally as it has been determined
that the anodic coating thickness does not affect the final weld penetration, the
minimum coating thickness to increase absorptivity, and minimise oxide
contamination may be considered.

The effect of the novel phases observed in the joining of the Al based MMCs
require further study. While it is taken for granted that this joining process would
lead to a degradation of the composite properties it would be necessary to quantify the degree of degradation. It may be possible to inject particles or fibres into the fusion zone together with a filler alloy to improve joint properties. Certainly an indepth study into the optimisation of the laser welding of powder metallurgy Al 6061 would aid in the understanding of the response of the Al 6061/SiC particulate reinforced MMC to the process.
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