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SOME ASPECTS OF THE LASER CUTTING AND WELDING OF A Ti-6Al-4V ALLOY

by

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A Doctoral thesis submitted in partial fulfilment of the requirements for the award of Doctor of Philosophy of the Loughborough University of Technology

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SOME ASPECTS OF THE LASER CUTTING AND WELDING OF A Ti-6Al-4V ALLOY

This thesis firstly contains a review of previous literature concerned with both the laser welding and cutting of titanium alloys. In addition a brief introduction to both laser physics and titanium metallurgy is also included in the literature review section. The experimental part of the thesis can be divided into three sections which are as follows.

(i) Measurement of focussed and raw laser beam diameter was carried out using a Laser Beam Analyser. In addition the laser mode used for both welding and cutting were evaluated.

(ii) 1, 1.7 and 2.7mm thick Ti-6Al-4V alloy was laser welded using both a 400W and a 2kW CO₂ laser in the continuous wave mode. The effect of variation in the main laser parameters upon weld morphology, microstructure, mechanical properties and oxygen contamination was evaluated. In addition further welding studies were carried out using the 400W laser in the pulsed mode. The effect of variation in pulse parameters upon weld bead morphology was studied. It was concluded from the laser welding section that using a 2kW CO₂ laser it was possible to produce full penetration welds in up to 2.7mm thick Ti-6Al-4V alloy sheet. The use of pulsed welding was found to enhance welding efficiency with a greater weld penetration being achieved than with an equivalent power continuous wave output.

(iii) Ti-6Al-4V alloy sheet up to 2.7mm thick was laser cut using a 400W CO₂ laser. The effect of variation in cutting speed and cutting gas pressure upon surface condition was studied. From detailed SEM examination of laser cuts a mechanism for inert gas assisted laser cutting was postulated. The effect of laser cutting upon the adjacent microstructure and oxygen contamination of the cut edge was also studied. As well as conventional laser cutting a series of laser cuts were also produced using the "Dross Jet". The effect of this device upon oxygen contamination and dross adhesion to the underside of the cut was studied. It was found that by using the dross jet in conjunction with inert gas assisted laser cutting that it was possible to produce cut edges with levels of oxygen contamination equal to that present in a guillotined cut edge. In addition the best laser cut edges were found to be as smooth as a typical guillotined edge. It was concluded that inert gas assisted laser cutting is an excellent process for cutting thin section Ti-6Al-4V alloy.

The final section of the thesis aims to compare the experimental results from the laser welding section with results generated by an existing mathematical model. The results show that the model can be used to predict weld penetration depths and HAZ widths to a reasonable degree of accuracy.
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APPENDIX 1. Listing Of "GLAZE 7" Programme
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CHAPTER 1. INTRODUCTION.

Over the past decade the CO₂ laser has established itself as a viable cost effective alternative to more traditional materials processing techniques in the areas of welding, cutting and surface treatment. This thesis presents an in depth analysis of the CO₂ laser processing of the most commonly used titanium alloy, Ti-6Al-4V. This alloy has found extensive use in both airframe and gas turbine manufacture, where its excellent specific strength to weight ratio can be utilized.

Laser profiling of sheet Ti-6Al-4V alloy is of obvious interest, particularly to airframe manufacturers, as Ti-6Al-4V is notoriously difficult to cut and machine by mechanical methods. Laser cutting offers potential cost reduction over traditional cutting techniques as a result of its speed, accuracy and reproducibility. However the cost advantages can only be realized once the major problems associated with laser cutting titanium alloys, which are oxygen contamination and dross deposition have been quantified and overcome. It is these two problems that are specifically investigated in the cutting research reported in this thesis.

The widespread use of Ti-6Al-4V alloy sheet in airframe manufacture necessitates that a reliable joining process is available. Electron beam welding has been traditionally utilized to produce high quality welds. Whilst this technique is used it entails the cost penalty of requiring high vacuum operating conditions. Previous work (1,2) has shown it to be feasible to weld Ti-6Al-4V alloy sheet using high power (2-10KW) continuous wave lasers. Few detailed oxygen analyses have been made in this earlier work. The aim of the welding research reported in this thesis is to optimize the main laser parameters with respect to weld quality and in particular to quantify the levels of oxygen contamination.
Previous laser welding of Ti-6Al-4V alloy sheet has concentrated exclusively on utilizing the laser in the continuous wave mode. Another objective of the present research is to assess the feasibility of producing welds using the laser in the pulsed mode.

The overall objective of the work carried out is to improve understanding of both laser cutting and laser welding. A detailed comparison of pulsed and continuous wave welding of Ti-6Al-4V alloy sheet is included. With this improved understanding of laser materials interactions, it should be possible to adopt the CO₂ laser as a flexible manufacturing system capable of consistent quality cutting and welding of Ti-6Al-4V alloy sheet, and other titanium alloys.
CHAPTER 2. REVIEW OF RELEVANT LITERATURE AND INTRODUCTION TO THE PRESENT WORK.

The following literature survey firstly contains a description of how a typical CO₂ laser works with an explanation of the physics behind the operation. In addition the literature review includes a section dealing with titanium alloy metallurgy. This is followed by a "state of the art" review of all current work concerned with either the cutting or welding of titanium alloys, using carbon dioxide lasers. Finally there is a critical appraisal of the literature and an introduction to the present work.

There are several different types of lasers being used in industry today. Table 2.2 summarises the characteristics and applications of a dozen types of lasers in common use. Although there are several different types of lasers available, approximately 90% of all industrial applications are carried out using either solid state Nd:YAG (Neodymium :Yttrium -Aluminium-Garnet), or gaseous CO₂ lasers. It is apparent from Table 2.2 that both types of laser can be used for welding and cutting. However each have inherent features which make them suitable in particular materials processing situations.

Nd:YAG lasers operate at a wavelength of 1.06 μm, which results in a greater proportion of the beam energy being absorbed by metals, than is the case with CO₂ lasers which operate at 10.6 μm. This effect is clearly shown in Table 2.1 which shows the variation in emissivity with laser wavelength. Thus when processing highly reflective materials such as copper, aluminium or brass, Nd:YAG lasers have a clear advantage over CO₂ lasers.

The major drawback of Nd:YAG lasers is that they can only produce a continuous power output of up to 1 KW, due to cooling problems within the lasing medium. Therefore for welding thick sections (greater than approximately 1mm) high power CO₂ lasers are
generally used. For this reason a 2 KW CO\textsubscript{2} laser was used during the present welding programme.

In general Nd:YAG lasers are not used for cutting due to the poor surface finish attainable. Therefore the present laser cutting study utilised a specialist cutting laser, the Coherent Everlase 400W CO\textsubscript{2} laser, which can produce good quality surface finishes. Since only CO\textsubscript{2} lasers were used in the experimental work for welding and cutting, it is useful to examine their operation, and in particular the laser physics behind them.

\textit{2.1 Principles Of The CO\textsubscript{2} Laser.}

The following in depth discussion of CO\textsubscript{2} lasers is best preceded by an introduction outlining the general principles.

Lasers of all types are devices in which a material is exposed to an intense source of energy (eg. an electric arc, flash lamps etc.) and then re-emits a proportion of this energy after quantising it by molecular or ionic transitions. Lasing itself is a signal amplification process. A molecule delays from a high to a low energy state releasing a photon (ie. the signal). When this photon then collides with another high energy state ion, or molecule it decays to a lower energy state, and a further photon is released. The two photons are in phase and travel in the same direction. Subsequent collisions generate identical photons which because of the similarity of the energy transitions involved are not only in phase but of the same wavelength. For this process to work it is necessary for the majority of the atoms or molecules to be in the higher (metastable) energy state, a thermodynamically unusual situation known as population inversion. Lasers employ mirror systems to contain this signal amplification (or stimulated emission) process and so only the original spontaneous decay events which resulted in photons
travelling parallel to the optical axis perpendicular to the mirror face are responsible for the eventually emitted beam of coherent light.

The CO₂ Molecule.

The first successful operation of a CO₂ laser was reported in 1964 by Patel (3). The emission process occurred on two vibrational-rotational bands of the CO₂ molecule in the 10.6 μm region. A large number of quantized possible transitions occurred over the band 9 -11 μm with a mean value of 10.6 μm. Fig 2.1 shows the energy level diagram for the CO₂ -N₂ system (4).

The CO₂ molecule is triatomic and has three main modes of vibration (5). These are as follows.

(a) The symmetric stretching mode
   This is shown in Fig. 2.2 and is known as vibration mode 1(v₁). It is characterised by an energy of \( \nu = 1337 \text{ cm}^{-1} \). The oxygen atoms are able to move in opposite directions along the plane of the molecule whilst the carbon atom remains stationary.

(b) The bending mode
   This is shown in Fig. 2.2 and is known as the second vibrational mode (v₂). It is characterised by an energy of \( \nu = 667 \text{ cm}^{-1} \). In this case the atoms are able to oscillate perpendicular to the internuclear axis. This mode is sometimes known as two fold degeneracy. The bending mode consists of two degenerate vibrations \( v₂a \) and \( v₂b \).

(c) The asymmetric stretching mode
   This is shown in Fig. 2.2 and is known as the third vibrational mode (v₃). The oxygen atoms are able to move together whilst the carbon atom moves in the opposite direction. The
vibrational state of excitation of the CO$_2$ molecule is described by a four figure notation ($n_1, n_2, n_3$) where $n_1$, $n_2$ and $n_3$ are the number of quanta of frequency $v_1$, $v_2$ and $v_3$ possessed by the molecule. The angular momentum of the bending mode is described by $l$, and takes the value $l = n_2$, $l = 2$ with each $l$ giving rise to a sublevel. When $n_2 = 1$, $l = 1$ and when $n_2 = 2$, $l = 2$ and 0, thus giving two sub levels. Therefore when only the bending mode of vibration is exited the energy state of the CO$_2$ molecule is written as (0110) if $n_2 = 1$ and as (0220) and (0200) if $n_2 = 2$. At the ground state there is no excitation so that the notation is (000). Each vibrational energy level can be further subdivided into rotational levels which are labelled by an additional quantum number $J$. Transitions can only occur between levels with $\Delta J = \pm 1$.

**Stimulated Emission**

A simple model may be used to represent the system in the lasing medium. The lasing medium can be considered as a mixture of atoms, ions and molecules. When the system interacts with an electromagnetic field transitions between the various energy levels will occur and energy will be emitted at a frequency characteristic to the material. Transitions can occur both between energy levels due to the quantised nature of the electron orbits and between vibrational or rotational levels of molecules.

Stimulated emission is the opposite process to absorption which occurs when a photon collides with either an atom or a molecule. For stimulated emission to occur a photon impinging on the molecule induces a downward energy transition in the molecule which liberates a further photon. This process is illustrated in Fig. 2.3.
The rate of stimulated emission is proportional to the energy density of the inducing field $p(v_{nm})$ at the characteristic frequency as well as to the number of atoms, or population of the excited state. The energy density of the inducing field is related to the induced power by the following equation.

\[
\text{Induced Power } P = h v_{nm} N_n B_{nm} p(v_{nm})
\]

where

- $N_n$ = Number of atoms
- $B$ = Population of the exited state
- $p(v_{nm})$ = Energy density of the inducing field
- $h$ = Plank's constant
- $v_{nm}$ = Frequency.

To increase the extent of stimulated emission it is necessary that the gas is in a thermodynamically unusual state referred to as population inversion. The distribution of the population of the gas molecules between two energy levels $E$ and $E + \Delta E$ is given by:

\[
\frac{P(E + \Delta E)}{P(E)} = \exp\left(-\frac{\Delta E}{kT}\right)
\]

From this the majority of the population of gas molecules are at the lowest energy states, at a temperature $T$. The effect of the electric discharge contained in the CO$_2$ lasing medium is to excite the molecules to much higher energy levels giving a high population of molecules in higher energy states. This process is known as population inversion. In order for the lasing action to take place population inversion must occur.

The Lasing Action

Fig. 2.4 is a diagramatic representation of how a laser actually works (6). In stage A the CO$_2$ molecules are in
their ground state. In stage B they have been excited by the electrical discharge in the cavity, and emission occurs as described in the previous section. This emission occurs in all directions sometimes stimulating the release of secondary photons as shown in stage C. Stage D shows a photon parallel to the optical axis being reflected back along the axis initiating the cascading effect which is known as Light Amplification by Stimulated Emission of Radiation. In stage E photons parallel to the optical axis stimulate more photons and the lasing action occurs. A percentage of the photons resonating back and forth are emitted through a partially transmissive mirror.

The Lasing Medium

The first successful operation of a CO₂ laser was reported by Patel(7) in 1964. The output power was only about 1mW. This was due to the inefficiency of the lasing process, which was around 1%. The gas used in this laser was pure CO₂. The output power was found to increase to about 10W, with a few percent efficiency when N₂ was added to the discharge tube, and when cooling of the tube was used.

Modern lasers now use CO₂, N₂ and He mixtures. The mixture used depends upon the particular operating conditions. The actual mechanism of how N₂ and He improve efficiency can be described as follows.

(a) The electrical discharge excites about 15% of the CO₂ molecules by electron impact, and also excites N₂ molecules into higher energy states.

(b) Some 85% of CO₂ molecules are pumped up into higher energy states by collision with excited N₂ molecules. There is an almost resonant energy exchange between CO₂ molecules in the ground state and N₂ molecules in the higher energy states, as
shown in Fig. 2.1. The upper N\textsubscript{2} levels act as a reservoir which couples energy into the CO\textsubscript{2} (001) lasing level via the first vibrational N\textsubscript{2} level, during the collision process.

**Cooling**

In order to extract the energy stored in the higher N\textsubscript{2} energy levels with a reasonably high efficiency it is important that that the (100) and the (020) lower lasing levels should not be heavily populated. The use of He in the lasing medium is to act as a coolant. When the CO\textsubscript{2} molecules give up their energy after collision with a photon they do not return to their ground state. Instead they retain a small amount of energy in lower energy levels. The He is able to extract this energy returning the CO\textsubscript{2} to the ground state where the CO\textsubscript{2} molecules can again be pumped up into higher energy levels by collision with exited N\textsubscript{2} molecules. In order to cool the He atoms an external cooling system is required. This takes the form of either an oil coolant in contact with the glass laser tubes, or by passing the gas mixture through heat exchangers.

The lasing gas is also constantly recycled through a series of filters. These remove any unwanted by products such as CO\textsubscript{2}, H\textsubscript{2}O vapour, and N atoms produced by the action of the electric arc on the lasing medium. If these unwanted gases were not removed they would steadily build up and the lasing action would become more and more inefficient. When these waste gases are removed new lasing gas is pumped into the laser cavity to replace them. Indeed some lasers do not recycle the lasing gas at all, and a constant stream of new gas is used. This form of laser has a slightly higher efficiency but the volume of gas, and thus the cost is very much greater than conventional CO\textsubscript{2} lasers.
Classification Of CO₂ Lasers

There are four main configurations of CO₂ laser which depend upon the arrangement of the electrical excitation unit and the gas flow direction relative to the optical axis. These are as follows.

(a) Transverse gas flow type
Fig. 2.5 shows a CO₂ laser used for multikilowatt operation. In this form of laser the gas flow is perpendicular to the electrical discharge and to the optical axes. This design provides for minimal dwell time of the lasing gas in the optical cavity, a condition necessary for higher power density and more power per unit length of lasing chamber.

(b) Axial gas flow type
Fig. 2.6 shows a typical axial flow CO₂ laser. In this case the gas flows rapidly through the lasing chamber coaxially with the direction of electrical discharge and optical output, which is about 50 Wm⁻¹ of lasing chamber.

(c) Transverse discharge type
Here the gas flow direction is the same as the electrical discharge direction which is in turn perpendicular to the optical axis.

(d) Transverse Gas and Transverse Discharge Type
Here the optical axis, the gas flow axis and the electrical discharge axis are mutually orthogonal.
2.2 Short Review Of The Metallurgy Of Titanium Alloys.

Titanium alloys are widely used in the aerospace industry where their high strength to weight ratios at elevated temperatures are required. Fig. 2.7 shows the comparative strength-density behaviour of two titanium alloys against that of three steels, where it can be seen that the titanium alloys have substantially better UTS/density values over a temperature range of 100°C to 650°C. More recently the excellent corrosion resistance of titanium alloys has resulted in many titanium components finding use in both the chemical industry and as protheses for implantation in the human body.

Major Features Of Titanium Alloys.

Titanium has several features which distinguish it from other light alloys such as aluminium, magnesium and zinc, which are as follows.

(a) At room temperature pure titanium has an hexagonal close packed (h.c.p) structure known as alpha titanium. On heating above 882°C the metal undergoes an allotropie transformation and body centred cubic (b.c.c) beta titanium is formed, which remains stable up to the melting point (1675°C).

(b) Titanium is able to form solid solutions with many substitutional elements such as Al, V, Mn etc. which are within ±20% of the atomic size of titanium. This effect is a result of titanium being a transition metal with an incomplete d shell in its electronic structure.

(c) Titanium has a high affinity for several interstitial elements, notably oxygen, nitrogen and hydrogen. Absorption of these gases takes place well below the melting point and can lead
to severe deleterious effects upon the mechanical properties of the alloy.

The Effect Of Alloying Additions Upon Titanium.

As was mentioned in the previous section titanium is allotropic and so is able to take the form of either alpha or beta phase, or a mixture of both. The nature of a particular alloy is dictated by the ability of any substitutional elements present to stabilize either the alpha or the beta phase. If the substitutional element has a number of bonding electrons less than four (eg Al) then it becomes an alpha stabilizer. However if more than four bonding electrons are present (eg V) then it becomes a beta stabilizer. Elements with four bonding electrons such as copper are neutral.

It is possible to classify the titanium rich alloy phase diagrams into three simple groups depending upon the proportions of substitutional elements present. Fig. 2.8 shows three such phase diagrams with the alloying elements favouring each being as follows,

- Fig 2.8a Al, O₂, N₂, C, Ga alpha stabilizers
- Fig 2.8b Mo, W, V, Ta beta stabilizers
- Fig 2.8c Mn, Cr, Fe beta eutectoid stabilizers

Classification Of Titanium Alloys.

Titanium alloys are classified according to their crystallographic structure. There are four main groups which are, full alpha alloys, near alpha alloys, alpha + beta alloys and beta alloys. Table 2.3 shows the various alloy groups with their respective mechanical properties and compositions.
Full Alpha Alloys.

The most widely used fully alpha alloy is commercially pure (CP) titanium which is essentially an alloy of titanium and oxygen, which acts as an interstitial hardening agent. The volume of oxygen present determines the mechanical strength of the alloy. CP titanium is generally available as wrought sheet or pipe, and has found extensive use in aerospace and general engineering, in particular as tubing for heat exchangers.

CP titanium (IMI 115) is extremely ductile with elongation values of approximately 25%, as shown in Table 2.3. This ductility can be attributed to the low c/a ratio of the h.c.p alpha phase structure which provides more slip planes for deformation than in other h.c.p metals such as Mg, Zn and Cd.

However CP titanium has a low tensile strength which limits its wide scale application. Recently the addition of 0.2% Pd has resulted in a new alloy (IMI 260) which has excellent corrosion resistance in hydrochloric and sulphuric acid environments.

Near Alpha Alloys.

This group of alloys was developed to operate in higher temperature environments than alpha alloys, and has the greatest creep resistance of any titanium alloy group at temperatures above 400°C (7). Near alpha alloys contain up to 2% of beta stabilizing elements introducing a small amount of beta phase into the microstructure, which improves forgeability.

Near alpha alloys can be divided into two groups according to their heat treatment. These are as follows.
(a) Alloys heat treated in the alpha+beta phase field.

Alloys such as IMI 679(Ti-11Sn-2.25Al-5Zr-1Mo-0.2Si) and Ti-8Al-1Mo-IV (an American alloy) are forged at temperatures within the alpha + beta phase field. In order to achieve the maximum creep strength the alloys are air cooled to form a microstructure which is a mixture of primary alpha and Widmanstatten alpha, which forms by nucleation and growth from the beta phase. If however a higher tensile strength is required the alloy can be quenched forming martensitic alpha, although this results in reduced creep strength.

(b) Beta heat treated alloys.

If near alpha alloys are forged in the beta phase field then deformation becomes easier due to the b.c.c structure and higher working temperatures. However grain growth can be a serious problem. The most commonly used alloy in this group is IMI 829 (Ti-5.5Al-3.5Sn-3Zr-0.25Mo-1Nb-0.3Si) which has good creep characteristics up to 600°C. Such alloys are heat treated in the beta phase field and then quenched producing a martensitic alpha structure surrounded by beta. Subsequent ageing at 500°C reduces the quenching stresses and the martensitic alpha transforms to alpha bounded by a dispersion of beta, which gives excellent creep properties.

Alpha + Beta Alloys.

This group of titanium alloys has the most commercial importance of any titanium alloy, with Ti-6Al-4V (IMI 318) making up more than 80% of the total titanium alloy sales in Europe and the USA. For this reason IMI 318 was chosen for detailed investigation during the present laser welding and cutting programme.
Alpha + beta alloys contain elements to stabilize and strengthen the alpha phase (eg Al) together with 4-6% of beta stabilizing elements such as vanadium. This allows substantial amounts of beta phase to be retained on cooling from the beta phase field as predicted in the phase diagram in Fig. 2.8b. Generally these alloys are then annealed at approximately 700°C to give an equiaxed alpha + beta structure.

Alpha + beta alloys have several advantages over near alpha and full alpha alloys. The major advantage is that alpha + beta alloys have a substantially higher tensile strength, as shown in Table 2.3. Indeed the Ti-6Al-4V alloy has a room temperature tensile strength of 1000KPa. The addition of beta phase (b.c.c.) to the microstructure also improves the formability of the alloy and consequently alpha + beta alloys are used as forgings eg. as fan blades in jet engines.

However the use of alpha + beta alloys instead of alpha alloys entails a sacrifice in creep properties above 400°C. Thus in higher temperature applications alpha alloys are generally used.

Beta phase alloys.

If sufficient beta stabilizing elements such as V and Cr are present in a titanium alloy then a fully beta microstructure can be produced at room temperature. These beta alloys have superior forming characteristics due to the fully b.c.c. structure of the metal. In particular cold forming is also a possibility. In addition beta alloys have a high tensile strength superior to that of both alpha and alpha + beta alloys, as shown in Table 2.3.

Beta alloys however suffer from two drawbacks. The first is that the addition of large quantities of alloying
elements increases the relative density in excess of 5. Secondly prolonged exposure above 200°C has resulted in precipitation of TiCr2 in certain alloys, which reduces the toughness of the alloy.

At the present time beta alloys have found few applications although a recently developed alloy Ti-10V-2Fe-3Al which has particularly good forging characteristics has found use for thick section forgings in the aerospace industry.

The Effect Of Cooling Rate Upon Titanium Alloy Microstructure.

The effect of cooling rate upon titanium alloy microstructure is of extreme importance, especially in the case of laser welding and cutting. This section aims to briefly show the effect of differing cooling rates upon titanium alloy microstructure. As approximately 95% of titanium alloy applications utilize alpha + beta alloys (7), or alpha alloys, these two alloy groups will be considered in detail.

(a) Full alpha alloys.

If a fully alpha phase alloy, which consists of equiaxed alpha grains, is heated into the beta phase field and subsequently cooled then the cooling rate determines the resulting microstructure. Rapid cooling (approximately $10^4$-$10^6$°Cs$^{-1}$) as would occur in laser processing would result in a martensitic transformation taking place. The transformation takes place as follows. When b.c.c beta phase is cooled rapidly there is insufficient time for complete molecular diffusion to occur and for the beta phase to transform to stable alpha. Instead a diffusionless shear transformation occurs and metastable martensitic alpha is formed, which has a hexagonal lattice structure (8). Slower cooling rate, such as furnace cooling,
is used then there is sufficient time for complete diffusion to occur and the beta phase fully transforms to alpha.

(b) Alpha + beta alloys

In alpha + beta alloys, such as Ti-6Al-4V, the effect of cooling rate upon microstructure is more complex due to the presence of beta phase in the room temperature microstructure. If we consider an equiaxed alpha + beta structure that has been heated into the beta phase field then depending upon the cooling rate there are three resulting microstructures that may occur.

Rapid cooling rates such as those encountered during laser welding and cutting result in a martensitic transformation similar to that described in the previous section. Air cooling would result in a Widmanstatten structure of alpha platelets interleaved with beta platelets, occurring in colonies. Furnace cooling would result in plate like alpha + beta within the prior beta grain boundaries.

2.3.1 Laser Cutting Of Titanium Alloys.

Potential Advantages Of Laser Cutting

Laser cutting of titanium alloys has many advantages over other conventional cutting techniques such as shearing and machining. The main advantages are:

(a) Rapid cutting speeds are possible.
(b) Laser cutting is less costly in batch production than machining, due to the high cost of dies and tools which wear out very rapidly when processing titanium alloys.
(c) Complex shapes can be profiled using computer controlled x-y tables without the production of costly press tools.
(d) a very smooth surface finish is attainable.

Disadvantages of laser cutting.

Laser cutting however has a few disadvantages, which are as follows.
(a) The initial capital cost of the laser is high.
(b) The cost of cutting gases is high.
(c) The thickness of material that can be cut economically is fairly low.

The Laser Cutting Of Titanium Alloys Using A 250W Laser.

A 250 W CO2 laser has been used in production to rough cut aircraft structural parts at the Grumman Aerospace Corporation since 1971. Huber(9) studied the equipment modifications and tooling requirements to adapt this process for the production of both flat and formed parts. The impact of higher laser powers on production was also studied. Part of the study was an examination of the laser cutting of Ti-6Al-4V. A summary of Huber’s results is given below.

(a) Oxygen assisted cutting of titanium did not improve cut quality at higher power levels. The higher incident power density led to a violent reaction in the oxidation prone titanium and also to very rough cuts.

(b) For thick titanium sections the primary cutting energy stems from the oxygen jet assist such that the cutting speed appeared to be independent of the power level.
(c) Inert gas jet assisted cutting of thin titanium sections gave a generally smooth cut surface with little apparent heat damage and easily removable dross at the lower lip of the cut.

(d) The use of \( \text{CO}_2, \text{N}_2 \) and compressed air as the gas jet produced a serrated edge. However compressed air provided a substantial increase in cutting speed over that obtained with inert gas assist cutting.

(e) When helium gas was used the surface irregularity of the cut was reduced. This was probably due to the higher jet velocity obtained.

Huber concluded that the surface irregularities of the cuts by control of the fluid dynamic characteristics of the jet. Thus helium was in most cases the best cutting gas for titanium. However Huber's work was of a very general nature with very little quantitative information.

The Laser Cutting Of Titanium Alloys Using A 1.5KW Laser.

Kovalenko, Arata, Maruo and Miyamoto (10) at Osaka University, studied the cutting of several different metals, including titanium, using a 1.5 KW \( \text{CO}_2 \) laser and argon was used as the cutting gas. The main results of their work are summarised below.

(a) With increasing argon pressure no change in kerf width was found to occur, but the maximum cutting speed was increased.

(b) If the amount of argon was large enough to shield the cutting region from oxidation and the cutting speed was fast enough to prevent overheating of the material adjacent to the kerf, the HAZ was found to be very small or even absent.
(c) Surface roughness results showed that there was only a slight improvement in quality with increasing argon pressure.

(d) Particular dross deposition was also observed when cutting 1 mm thick titanium sheet. At low cutting speeds this dross was mostly deposited on one side of the cut. An increase in cutting speed resulted in asymmetrical dross deposition.

Kovalenko et al. concluded that it was possible to laser cut titanium and achieve good quality results with the correct choice of cutting parameters.

One of the major problems associated with the laser cutting of titanium alloys is dross adhesion to the underside of the cut. Arata et al. (11) reduced this problem by using a tandem nozzle arrangement, which basically involved using a second nozzle following behind the cutting nozzle to increase the ejection velocity of the dross thus reducing adherence to the cut edge.
2.3.2 The Laser Welding Of Titanium And Its Alloys

There are several different processes used for the joining of titanium and its alloys. The most established processes in recent years have been as follows.

(a) Electron beam welding
(b) Plasma arc welding
(c) TIG welding
(d) Resistance welding
(e) Diffusion welding

However more recently laser welding has become increasingly used in the aerospace industry. The major competitor to laser welding is EB welding and much early work concentrated on comparing and contrasting EB welding and other welding processes with laser welding.

Potential Advantages Of Laser Welding

(a) The laser does not require a vacuum and can be transmitted over long distances. However there is a power loss of about 6% per Km \(U^2\).

(b) The laser is not affected by stray magnetic fields from the workpiece.

(c) The laser may be steered by mirrors to otherwise inaccessible locations.

(d) The laser may be time shared to supply a number of work stations thereby reducing idle time.

(e) The laser may be divided into several component beams of lower power to carry out different functions simultaneously.
Disadvantages of laser welding

The main disadvantages of laser welding as compared to EB welding are as follows.

(a) A higher capital cost for a power output of between 6 KW and 10 KW.

(b) A lower energy yield.

(c) A much greater consumption of water and gas.

(d) The scanning frequency of the laser beam is limited by mechanical oscillation.

(e) EB welding has a much greater depth of penetration than laser welding.

(f) Uncertain reliability of certain lasers is also a problem, but continued development is making this less and less of a problem.

Problems Associated With Welding Titanium

The welding of titanium involves several difficulties which can be overcome with the use of the correct welding conditions. Gurevitch (73) summarised these difficulties as follows.

The high activity of titanium and titanium based alloys towards gases, such as oxygen, nitrogen and hydrogen can result in gas being absorbed in the weld pool which embrittles the metal. In addition such impurities can raise the tendency to cold cracking in welded joints and and cause porosity. Impurities also greatly influence the character of phase and
structural transformations occurring under the effect of the weld thermal cycle. Serious difficulties can also arise due to a considerable grain growth tendency when heating above the alpha-beta transition temperature in the process of welding. Such grain growth problems during welding are aggravated by the low thermal conductivity and volumetric heat capacity of titanium. Grain growth in the areas of the heat affected zone adjacent to the fusion line involves a decrease in ductility and rupture strength.

Gurevitch concluded that in order to produce good welds in titanium the liquid weld pool should be protected from the atmosphere by either an inert gas shroud or by a vacuum chamber. He also advised that as little heat as possible should be used during the welding process.

The effect of oxygen contamination on the mechanical properties of titanium has been studied in detail by both Jaffee, Ogden and Maykuth (14), and Finlay and Snyder (15). Fig 2.9 shows that the UTS ,0.2% proof stress and hardness all increase with increasing oxygen contamination, whilst the elongation decreases. Carbon and nitrogen were also found to affect the mechanical properties of titanium, which can be seen in Figs 2.10 and 2.11. These results show that oxygen and nitrogen have a much greater hardening effect in titanium than carbon at corresponding atomic concentrations.

Comparison Of Welding Techniques

Banas (16) directly compared the EB welding, TIG welding and laser welding of Ti-6Al-4V, a very commonly used aerospace alloy, in thicknesses of 1, 1.5, 2.0 and 6.4 mm. Butt welded specimens of each thickness were prepared and subjected to metallographic analysis, and both destructive and non-destructive mechanical testing. The main results are given below.
(a) Radiographically sound welds could be produced in Ti-6Al-4V alloy by laser welding, EB welding and by TIG welding.

(b) Welds using all three processes were post weld stress relieved for 2 hours at 538°C. The tensile properties of both laser and EB welded specimens exceeded those of the base material. When TIG welding was used the tensile properties of the welded specimens were equal to those of the base material.

(c) The fracture toughness of all welds produced was lower than that of the base material. Plasma arc welds showed a 10% reduction, and both EB and laser welds a 40% reduction, in fracture toughness. However it was found that the fracture toughness of beam welds could be improved by additional heat treatment.

(d) Laser and EB welding have much lower specific energy inputs than TIG welding. Consequently more rapid cooling takes place in beam welds and therefore a much finer grain structure is produced which gives a harder weld structure.

(e) The fine grained structures produced by beam welding gave excellent fatigue properties and indeed laser welded samples could be made in Ti-6Al-4V which exhibited the same fatigue properties as the base material.

Banas concluded that the technique of laser welding was promising for joining thin (1 mm) to medium (6.4 mm) sections of Ti-6Al-4V.

Peng, Sastry and O'Neal (17) investigated the connection between microstructures and mechanical properties of a variety of titanium alloys when laser welded, EB welded and TIG welded. The alloys used are shown in Table 2.4. Butt welds were produced in each alloy by laser, EB and TIG welding. For laser welding a CW 1.5 KW CO₂ laser was used with a scanning rate
of 0.25 cm s\(^{-1}\). The incident beam power varied between 580 W and 1040 W. Helium was used as the shielding gas. The main results of the experimental work are as follows. Fig. 2.12 shows the variation in melt width and melt depth in laser melted Ti-6Al-4V and Ti-15V-3Al-3Sn-3Cr alloys at 1000 W incident power. Figure 2.13 shows the same alloys using 1500 W incident power. No significant differences existed in the melt widths of the different Ti alloys. However the melt depths at higher transverse rates are considerably lower in the beta alloy Ti-15V-3Al-3Sn-3Cr than in the near alpha alloy Ti-6Al-4V. The specific heats, thermal conductivities and melting points of the alloys studied did not differ by more than 10% and hence did not account for the observed difference in melt depth.

Metallographic examination of welds in near alpha and alpha + beta alloys showed that the base metal microstructure consisted of equiaxed grains surrounded by a small amount of beta phase and the fusion zone consisted of large transformed beta grains containing acicular martensitic alpha. However in the beta alloy the fusion zone was characterised by large columnar grains, and the HAZ consisted of smaller, almost equiaxed grains. A dendritic microstructure reflecting the first stages of the solidification process was observed only in the beta alloy. In other alloys the concentration of alloying elements was not sufficient to produce a segregation induced substructure and the solid state transformation of beta phase to martensite further masked the solidification substructures. The dendrite arm spacing was found to be the smallest in laser welded samples and the largest in TIG welded samples. Fig. 2.14 shows the dependence of dendrite arm spacing on cooling rate for laser melted Ti-15V-3Al-3Sn-3Cr. The cooling rates resulting from different laser beam densities were calculated using as a thermal model a Gaussian heat source moving at a constant speed, from which a plot of dendrite arm spacing (d) as a function of cooling rate, was constructed. It can be seen from the diagram that Ti-15V-3Al-3Sn-3Cr, because of its low thermal conductivity, has
a larger dendrite arm spacing than either a maraging steel or an aluminium alloy.

The tensile properties of laser welded titanium alloys were found to be comparable with EB welded and TIG welded specimens. In Ti-6Al-4V and Ti-6Al-2Sn-4Zr-2Mo fracture occurred in the base metal indicating that the welds were stronger in these alloys. In a Ti-8Al-1Mo-1V alloy however, fracture occurred in the welds, but in a Ti-15V-3Al-3Sn-3Cr alloy 50% of the specimens fractured in the parent metal, and 50% in the weld. The tensile properties of all the specimens were found to be generally 95 to 100% of the base metal.

The Welding Of Titanium Alloys Using Laser Powers Of Greater Than 10KW

Metzbower and Moon (2) examined the mechanical properties, fracture toughness and microstructures of laser welds in high strength alloys which included a titanium alloy, Ti-6Al-2Nb-1Ta-0.8Mo. The alloy was 12 mm thick, and 100 mm x 200 mm samples were laser welded to form 200 mm x 200 mm weldments. All welds were butt welds and helium gas was used as the shielding gas for all the welds. The laser parameters used were a power of 11 KW and a welding speed of 14.8 mms⁻¹. The main results of Metzbower's and Moon's work are summarised as follows.

Fracture toughness properties were measured on a subsize 12 mm dynamic tear (DT) test specimen that was notched in the weld. The titanium alloy fracture started in the weld and moved quickly into the base plate. There was also some evidence of porosity in the weld.

Metallographic examination of the weld was also carried out. The fusion zone had a microstructure consisting of
martensitic alpha and martensitic beta. The HAZ consisted of primary alpha, Widmanstatten alpha and beta phases. The size and number of the primary alpha in the HAZ decreased and the beta phase increased as the fusion zone was approached. The original base plate structure consisted of elongated and equiaxed alpha in a matrix of beta and Widmanstatten alpha phases.

A transverse cross section was cut from a weldment and tested for microhardness using a diamond pyramid indenter with a 300 gm load. The microhardnesses were measured from one side of the base metal. The results of a typical microhardness traverse are shown in Fig. 2.15. The microhardness exhibited little change across the three different zones with an average value of 36 Rc.

Metzbower and Moon concluded that it was indeed quite feasible to weld Ti-6Al-2Nb-1Ta-0.8Mo alloy using a laser. Porosity was a definite problem, although this could be overcome by careful selection of the shielding gas and effective shielding gas coverage.

Metzbower also carried out further work with F.W.Fraser (18). They again used 12 mm thick Ti-6Al-2Nb-0.8Mo alloy. Single pass weldments were fabricated using a continuous wave CO$_2$ laser at a power level of 11 KW and a welding speed of 14.8 mms$^{-1}$. Mechanical properties of the weldments were determined using transverse tensile specimens. Charpy "V" notch and dynamic tear testing were carried out to measure fracture toughness at 0C$^0$, and at 25 C$^0$. In addition hardness measurements were also taken in the fusion zone, HAZ and unaffected base metal. The resistance of the weldments to stress corrosion cracking was also determined. The main results of Metzbower's and Fraser's work can be summarised as follows.

During mechanical tests the tensile specimens fractured by microvoid coalescence. Porosity was observed on the fracture surfaces of some specimens. This porosity had no effect on
either the UTS, or the yield strength of the weldments. Values of elongation and reduction in area were lower in specimens where porosity was detected. It was found however that the amount of porosity could not be correlated quantitatively with the tensile properties.

Microhardness measurements across the weld, HAZ and base metal at the top, centre and bottom of a laser welded sample are shown in Fig. 2.16. The average hardness values of the base metal and the fusion zone were 32 Rc and 35 Rc respectively. The hardness of the HAZ was comparable to that in the fusion zone. Oxygen and nitrogen pick up in the weld were found to be small due to gas shielding with helium. This resulted in excellent stress corrosion properties. Another factor which relates good fracture toughness and resistance to SCC in these laser welds is fusion zone purification. The high energy density of the process reduces the inclusion content of the fusion zone. The size and distribution of the inclusions was not recorded although it was found in previous work that the inclusion content of the fusion zone was only about half of that of the base plate. In addition the rapid cooling involved in laser welding was found to result in a predominantly martensitic alpha microstructure.

Metzbower and Fraser concluded that in 12 mm thick plates of Ti-6Al-2Nb-1Ta-0.8Mo laser beam welds had mechanical properties equal to or better than the base plate.

Metzbower, Moon and Fraser (19) carried out further investigations into the laser welding of the Ti-6Al-2Nb-1Ta-1Mo alloy. A 15 KW CW CO₂ laser was used with a power of 11 KW and a travel speed of 15 mms⁻¹. The thickness of the alloy was 12 mm. The main results of their work can be summarised as follows.
The Ti-6Al-2Nb-1Ta-1Mo alloy was not susceptible to stress corrosion cracking when laser welded. A precracked cantilever beam was fractured in air prior to SCC testing. A $K_{ISCC}$ value of 90 MPa m$^{0.5}$ was obtained. The range of $K_{ISCC}$ for laser beam welds was 82-95 MPa m$^{0.5}$.

The fast cooling of the process resulted in a predominantly martensitic microstructure, which when fractured separated by microvoid coalescence. Metzbower et al. concluded that laser beam welds in Ti-6Al-2Nb-1Ta-1Mo were found to have mechanical properties at least equal to those of the base plate, and in many cases better providing that gas shielding was good.

The Laser Welding Of Titanium Alloys Using Lasers Powers Of Between 5 And 10KW.

Crafter and Pauley (20) used a 5 KW laser to produce welds in 8.9 mm Ti-6Al-4V. However no analyses of the results were carried out. Using the same 5 KW laser at The Welding Institute, Baker and Partridge (21) produced a series of welds in 4 mm thick Ti-6Al-4V. In this work the tensile, fracture toughness and fatigue properties of the welds, produced at welding speeds of 2 and 4 m/min, were examined, and the following conclusions drawn.

(a) The tensile strengths of welds produced were comparable with the parent metal, but the ductility values were lower (22% elongation and 40% reduction of area).

(b) Fracture toughness values were difficult to measure accurately, however $K_C$ values in the fusion zone were about 40% lower than those obtained in the parent metal.
(c) The fatigue strength of the as welded sheet was reduced by the presence of undercutting, which could be alleviated by surface machining.

(d) The presence of internal residual stresses in the welded material significantly reduced fatigue crack growth rates for cracks growing either parallel or normal to the weld.

(e) If undercutting could be minimized then laser welded Ti-6Al-4V may have mechanical properties as least as good as those produced by electron beam welding.

The Welding Of Ti-6Al-4V Using A 2 KW CW CO₂ Laser

Perhaps the most detailed study into the laser welding of titanium alloys was undertaken by Mazumder and Steen (1,22). Using a Control Laser 2 KW CW CO₂ laser autogenous butt welds were made in Ti-6Al-4V and CP Ti (annealed) sheet of different thicknesses, welding speeds and laser powers. Each specimen to be welded was carefully degreased with a special proprietary solvent from IMI (titanium). A sketch of the welding assembly is shown in Fig. 2.17. The assembly was shielded with helium above to remove oxygen and reduce plasma formation, and with argon beneath. The laser beam was focused on the surface using a 75 mm focal length KCl lens. The welds produced were examined in detail and the main results of Mazumder's and Steen's work can be summarised as follows.

Butt welds were X-rayed to reveal any discontinuity of the welds. All unacceptable samples had very obvious defects such as off centred welds. All other welds were found to be free of porosity, cracks and inclusions up to the limit of resolution (100-150 μm). These results showed that laser welding Ti-6Al-4V compared very favourably with EB welding, where porosity is a problem, suggesting that welding at atmospheric pressure is an advantage. Low porosity has also been found by Seamen (23).
However Morgan (24), when welding thin to thick sections, found wide scale porosity which may have been caused by surface tension effects at the high speeds he used.

Metallographic examination of welds (22,1) using optical and electron microscopy revealed that the HAZ consisted of a mixture of martensitic alpha and primary alpha. The fusion zone consisted of mainly martensitic alpha within the prior beta grain boundaries. The base metal was simply beta in an alpha matrix, the typical annealed structure for alpha-beta titanium.

Macrohardness traverses were made across laser welds. Fig. 2.18 shows a macrohardness traverse on a 1 mm thick specimen welded with a power of 1 KW and a speed of 23 mms⁻¹. The macrohardness for the weld zone was observed to be around 380-385 VHN, compared to 340-350 VHN for the parent matrix. Microhardness traverses could not reveal these differences due to the large hardness variation between the alpha and beta phases. For CP titanium the hardness of the weld ingot showed a slightly lower value compared to that of the parent matrix.

The variation of UTS, 0.2% proof stress and percentage elongation with both laser power and welding speed are shown in Figs 2.19 and 2.20. These results showed that within the limits of experimental accuracy neither the welding speed nor the laser power affected the resulting weld strength.

Fatigue testing of welded samples with the weld placed central and transverse showed that the endurance ratio (fatigue limit / UTS) was found to be 0.40-0.47, whereas for the unwelded specimens it was 0.5.

Any change in composition of the titanium alloy was found to affect the optimum welding conditions. Table 2.5 shows the welding speeds of CP titanium and Ti-6Al-4V of the same
thickness and welded with the same power. It can be seen from the Table that a higher welding speed could be achieved for material of higher conductivity for the same power and thickness.

Mazumder also produced a mathematical model and Fig. 2.21 shows the variation in titanium welding speed with power and compares this to his mathematical prediction. Fig. 2.22 shows that for a particular power welding can be performed over a range of speeds for a particular thickness, but that beyond this range full penetration is not possible. Mazumder also found that the width of the HAZ varied inversely with welding speed so that at lower speeds more energy was wasted in heating up the workpiece. Fig. 2.23 shows the variation of HAZ against the welding speed.

Fig. 2.24 shows the variation of titanium welding speed with thickness for laser powers of 1000W, 1500W and 2000 W. It can be seen that for the same power a slower speed is required for greater penetration. The results of oxygen analyses carried out on welded samples indicated that there was no significant oxygen contamination taking place during welding. Oxygen acts as an alpha stabiliser in titanium and can be added as a strengthening additive. Mazumder concluded that the laser welding process provides an opportunity to weld titanium at atmospheric pressure without serious contamination and deterioration of mechanical properties.

Additional Aspects Of Laser Welding.

The following section aims to review some recent literature associated with the laser welding of titanium alloys. In particular the effect of misalignment, gap size, solidification and dynamic beam focusing will be discussed.
Solidification Of Laser Welded Titanium Alloys

Gould, Baeslack and Williams (25) welded Ti-6Al-2Sn-4Zr-2Mo alloy and another titanium alloy "Corona 5", using laser, EB and TIG welding. They studied how conventional phase transformation variables, such as solidification rates and cooling rates varied with the welding process. Their results can be summarised as follows.

The Ti-6Al-6V-2Sn alloy developed hot cracks during pulsed arc welding. This was due to the large freezing range of the alloy. This could be described by the parameter, $-m_1(l-k)c_0$ where $m_1$ = liquidus slope, $k$ = partition ratio and $c_0$ = bulk composition.

Microprobe traces indicated that Fe and Sn were the only alloying additions to segregate appreciably. Since both of these formed low melting point eutectics they appeared to be responsible for hot cracking.

Segregation in titanium alloys was found to occur on two scales.

(i) Firstly interdendritic microsegregation which was responsible for shrinkage pores which was caused by pockets of solute rich liquid in the interdendritic regions.

(ii) Secondly banding or macroscopic segregation which is caused by variations in macroscopic growth rate, led to areas of solute depletion and solute richness. These were caused by variations in travel speed or arc intensity.
The Laser Welding Of Ti-Y And Ti-Er Alloys.

Peng, Sastry and O'Neal (26) studied the welding of three titanium alloys Ti-Y, Ti-Er and Ti-6Al-4V. Electron beam and CW CO$_2$ laser welding were used for all welds. A summary of the main results is given below.

(a) The fusion zone in laser welded Ti-6Al-4V and Ti-Er alloys consisted of mainly acicular martensitic alpha whereas the HAZ consisted of a mixture of martensitic alpha and primary alpha. The martensitic structure indicated that the cooling rate was in the order of $10^2$-$10^4$°C s$^{-1}$. The weld width and microstructures were similar to those in EB welded specimens.

(b) The tensile properties of laser welded alloys were similar to the base metal properties.

(c) The average size of Er disperiods in laser welded regions was smaller than in the base metal, and this refinement was beneficial for promoting ductile failure and increasing the fracture toughness of the welds.

The Dependence Of The Mechanical Properties Of Laser Welded Titanium Alloys On Misalignment.

One of the main advantages of laser welding is that no machined edge preparation is generally required, however the alignment of the components to be welded is critical. Grigoryants (27) working at the Bauman Institute studied the dependence of tensile strength, fracture stress and elongation on both the gap size and the misalignment in butt welded VT 3 and VT 7 titanium alloys. Grigoryants main results were as follows.

The edge preparation for laser welding without a filler wire should be examined carefully. An increase in relative gap
size (ie. butt gap / metal thickness) above 10-13% was found to result in a sharp reduction in the static strength of the welds.

The strength of welded joints produced by CO2 laser welding was equal to the strength of the parent metal over a wide range of conditions at a radiation power of 2-5 KW and a welding speed of 80-200 mhr⁻¹.

The impact properties of the weld metal and the HAZ of laser welded specimens was not inferior to either EB welded or TIG welded specimens. In fact in some cases it was found that the impact toughness of the weld metal was higher, caused by degasing of the remelted metal. This took place because of the large difference in intensity of absorption of radiation by metals and by non-metals which which resulted in preferential evaporation of impurities from the weld zone.

**Dynamic Beam Focusing.**

The relatively low thermal efficiency of laser beam welding, when compared to electron beam welding means that it is important to increase the efficiency of the process. Ivanov (28) increased the depth of penetration when laser welding titanium by between 30 to 45% by using dynamic beam focusing. Basically this process involved oscillating the lens, and thus the position of the focal spot, with a specific amplitude and frequency, until maximum penetration was achieved.

**Narrow Gap Laser Welding.**

Laser welds in 1.27 and 2.54 mm thick Ti-6Al-4V alloy separated by a 4 mm straight sided gap have been achieved with the use of a continuous addition to the weld gap of prealloyed
powder of the same composition, at the beam focus. A 5 KW CO₂ las er was used during the experimental programme. This process was repeated several times each time overlaying another fused powder layer until the gap was filled. Breinan and Snow (29) came to the following conclusions.

(a) The mechanical properties of narrow gap welds were comparable to close fitting laser butt welds in Ti-6Al-4V plate.

(b) Narrow gap welds in Ti-6Al-4V plate can be made by laser glazing with continuous powder feed. Improved beam control, to eliminate premature sidewall fusion above the gap floor is required to avoid side wall bridging and consequent macroscopic cavities.

(c) The fusion zone cooling rate was sufficiently rapid to produce a completely martensitic microstructure.

Summary Of The Welding Literature.

It is apparent from the welding section of the literature review that there has been a sizable amount of published work pertaining to the laser welding of titanium alloys. It is possible to highlight two major deficiencies of previous studies.

(a) In no case has any serious attempt been made to examine the level of oxygen contamination at the surface of laser welded titanium alloys. Only bulk values for the entire weld bead have been measured (22).

(b) All previous work has been conducted using the laser in the continuous wave mode and no attempt has been made to assess the feasibility of laser welding titanium alloys using the pulsed laser mode.
As a result of these two deficiencies the present welding programme will aim to investigate in detail both the problem of oxygen contamination and the use of pulsed welding, which have hitherto been neglected.

2.4 Critical Appraisal Of The Previous Literature And Introduction To The Present Laser Cutting And Welding Study.

It is apparent from the literature review that there has been little previous investigation carried out into the laser cutting of titanium alloys. Early work such as that of Huber (9) concentrated upon using oxygen as the cutting gas. However as has been pointed out oxygen results in deleterious effects upon the mechanical properties of titanium. It was therefore decided to use argon as the cutting gas in the present study to minimize the oxygen contamination problem.

Arata (10) studied the laser cutting of a wide variety of metals including a limited amount of work on CP titanium. Although the effect of variation in cutting speed upon kerf width, HAZ width, cut surface roughness and dross deposition was studied only a small number of samples were examined. In particular no attempt was made to examine, in detail, the laser cut edges and develop a mechanism for inert gas assisted laser cutting. As a result one of the major aims of the present cutting study was to use SEM analysis of laser cut edges and quantify the striation patterns that appear in order to develop views concerning the mechanism of inert gas assisted laser cutting.

In addition Arata made no attempt to quantitatively assess the amount of oxygen contamination that occurs during the laser cutting of titanium alloys. Due to the deleterious effects of oxygen upon the mechanical properties of titanium a great deal of effort should be devoted to examining this problem.
Arata both in his early work (10) and latter research (11) dealt with the problem of dross adhesion to the underside of laser cut titanium. He found experimentally conditions during which this problem did not occur and also developed a tandem nozzle cutting method which was only partially successful. The development of a new device to prevent dross attachment also seemed important. It was also desirable to evaluate the effects of using such a device upon cutting performance and oxygen contamination of titanium alloys during laser cutting.

There has been a great deal more research carried out in the field of laser welding titanium alloys, which is reflected in the relative lengths of the laser welding and cutting literature reviews. One of the earliest investigations was that of Banas (16) who conducted a comparative survey of electron beam, laser and TIG welded Ti-6Al-4V. Although metallography, non-destructive testing, mechanical testing and the effect of welding parameters upon both HAZ width and weld penetration were studied, no attempt was made to investigate the problem of oxygen contamination during the various welding processes. Thus one of the aims of the present laser welding programme was to compare the extent of oxygen contamination in both laser welded and electron beam welded Ti-6Al-4V alloy.

Indeed this lack of detailed investigation into the problem of oxygen contamination during the laser welding of Ti-6Al-4V has persisted throughout the literature review. Even Mazumder's detailed study (122) gave only bulk oxygen values for one laser weld. In order to deal with this deficiency it was important to determine oxygen gradients in laser welds as a function of the varying laser parameters.

Much of previous laser welding research has been carried out upon less commonly used titanium alloys such as Ti-6Al-2Nb-1Ta-0.8Mo (2, 19, 17). Although this is of some commercial
interest as over 80% of all titanium alloy sales are of Ti-6Al-4V (30) the present laser welding programme was restricted to an in depth study of the Ti-6Al-4V alloy.

Mazumder's work (1, 22) was the most detailed study undertaken into the laser welding of Ti-6Al-4V. In addition to the lack of research into the problem of oxygen contamination no attempt was made to explain the effect of variations in the laser parameters upon weld bead morphology. In particular no mention was made of keyholing which is a major feature of laser welding. These aspects of laser welding appeared worthy of further study.

A great deal of mechanical testing has been carried out upon laser weldments of Ti-6Al-4V (1, 22, 24) and of other titanium alloys (2, 17, 25). However all these mechanical tests were either tensile or fracture toughness investigations. To date there has been no attempt to assess the impact properties of laser welded Ti-6Al-4V.

It is also apparent from the literature review that all previous research work has been conducted using the laser in the continuous wave mode. However most CO₂ lasers have a facility by which their outputs may be pulsed. Pulsed laser welding has been used successfully to join stainless steels (31) In particular it seemed worthwhile to study the effect of variations in the main pulse parameters upon the weld bead morphology was studied. Use of a laser beam analyser would allow monitoring of the pulsed welding programme to monitor the pulsed laser output so that a comparison with the actual weld bead could be made.
Table 2.1. The variation in emissivity for various metals at different laser wavelengths. (12)

<table>
<thead>
<tr>
<th>Metal</th>
<th>Ar* (500 nm)</th>
<th>Ruby (700 nm)</th>
<th>Nd-YAG (1000 nm)</th>
<th>CO₂ (10 μm)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Aluminum</td>
<td>0.09</td>
<td>0.11</td>
<td>0.08</td>
<td>0.019</td>
</tr>
<tr>
<td>Copper</td>
<td>0.56</td>
<td>0.17</td>
<td>0.10</td>
<td>0.015</td>
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<td>Gold</td>
<td>0.58</td>
<td>0.07</td>
<td>—</td>
<td>0.017</td>
</tr>
<tr>
<td>Iridium</td>
<td>0.36</td>
<td>0.30</td>
<td>0.22</td>
<td>—</td>
</tr>
<tr>
<td>Iron</td>
<td>0.68</td>
<td>0.64</td>
<td>—</td>
<td>0.035</td>
</tr>
<tr>
<td>Lead</td>
<td>0.38</td>
<td>0.35</td>
<td>0.16</td>
<td>0.045</td>
</tr>
<tr>
<td>Molybdenum</td>
<td>0.48</td>
<td>0.48</td>
<td>0.40</td>
<td>0.027</td>
</tr>
<tr>
<td>Nickel</td>
<td>0.40</td>
<td>0.32</td>
<td>0.26</td>
<td>0.03</td>
</tr>
<tr>
<td>Niobium</td>
<td>0.58</td>
<td>0.50</td>
<td>0.32</td>
<td>0.036</td>
</tr>
<tr>
<td>Platinum</td>
<td>0.21</td>
<td>0.15</td>
<td>0.11</td>
<td>0.036</td>
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<tr>
<td>Rhenium</td>
<td>0.47</td>
<td>0.44</td>
<td>0.28</td>
<td>—</td>
</tr>
<tr>
<td>Silver</td>
<td>0.05</td>
<td>0.04</td>
<td>0.04</td>
<td>0.014</td>
</tr>
<tr>
<td>Tantalum</td>
<td>0.65</td>
<td>0.50</td>
<td>0.18</td>
<td>0.044</td>
</tr>
<tr>
<td>Tin</td>
<td>0.20</td>
<td>0.18</td>
<td>0.19</td>
<td>0.034</td>
</tr>
<tr>
<td>Titanium</td>
<td>0.48</td>
<td>0.45</td>
<td>0.42</td>
<td>0.08</td>
</tr>
<tr>
<td>Tungsten</td>
<td>0.55</td>
<td>0.50</td>
<td>0.41</td>
<td>0.026</td>
</tr>
<tr>
<td>Zinc</td>
<td>—</td>
<td>—</td>
<td>0.16</td>
<td>0.027</td>
</tr>
</tbody>
</table>

*At 20°C.
Table 2.2. Characteristics of common lasers and their applications (32)

<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 Repition 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, holography, pattern generation, pollution detection, spectroscopy, semiconductor annealing, photochemical, dye-laser printing</td>
</tr>
<tr>
<td>Ionized gas</td>
<td>Argon</td>
<td>0.4880, 0.5145</td>
<td>2-18 W</td>
<td>Continuous</td>
<td></td>
<td></td>
<td>Display projection</td>
</tr>
<tr>
<td>Ionized gas</td>
<td>Krypton</td>
<td>0.6471</td>
<td>0.5-1 W</td>
<td>Continuous</td>
<td></td>
<td></td>
<td>Pollution detection</td>
</tr>
<tr>
<td>Ionized gas</td>
<td>Helium-Cadmium (He-Cd)</td>
<td>0.4416, 0.3250</td>
<td>Continuous</td>
<td></td>
<td></td>
<td></td>
<td>Materials welding, hardening, cutting</td>
</tr>
<tr>
<td>Molecular gas</td>
<td>Carbon-Dioxide (CO₂)</td>
<td>9.6, 10.6</td>
<td>100 W/30 kW</td>
<td>Continuous</td>
<td></td>
<td></td>
<td>Materials welding, hardening, cutting</td>
</tr>
<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, drilling</td>
</tr>
<tr>
<td>Molecular gas</td>
<td>Nitrogen (N₂)</td>
<td>0.3371</td>
<td>0.1 to 1 W (up to 10⁹ W peak)</td>
<td>Repetitively pulsed</td>
<td>100</td>
<td>10 ns</td>
<td>Materials processing, dye-laser pumping</td>
</tr>
<tr>
<td>Solid state</td>
<td>Ruby</td>
<td>0.6943</td>
<td>50 W (up to 10⁹ W peak)</td>
<td>Pulsed</td>
<td>1</td>
<td>0.5-5 ms</td>
<td>Materials welding, drilling, satellite ranging, plasma diagnostics in fusion experiments, semiconductor annealing, marking</td>
</tr>
<tr>
<td>Solid state</td>
<td>Neodymium-YAG (Nd: YAG)</td>
<td>1.06</td>
<td>Up to 1 kW</td>
<td>Continuous</td>
<td></td>
<td></td>
<td>Materials welding, drilling, marking</td>
</tr>
<tr>
<td>Solid state</td>
<td>Neodymium-YAG (Nd: YAG)</td>
<td>1.06</td>
<td>75 W (up to 10 kW peak)</td>
<td>Repetitively Q-switched, continuous- ty pumped</td>
<td>25 000</td>
<td>200 ns</td>
<td>Materials welding, trimming, scribing, drilling, semiconductor annealing, dye-laser pumping, marking, component fabrication</td>
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<tr>
<td>Solid state</td>
<td>Neodymium-YAG (Nd: YAG)</td>
<td>1.06</td>
<td>500 W (up to 10 kW peak)</td>
<td>Pulse pumped</td>
<td>200</td>
<td>0.5-10 ns</td>
<td>Materials welding, drilling, marking</td>
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<tr>
<td>Solid state</td>
<td>Neodymium glass (Nd:glass)</td>
<td>1.06</td>
<td>Up to 10 W</td>
<td>Pulsed</td>
<td>1</td>
<td>0.5-10 ns</td>
<td>Materials welding, drilling, nuclear fusion</td>
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<tr>
<td>Solid state</td>
<td>Gallium-Arsenide (GaAs)</td>
<td>0.85-0.805</td>
<td>20 mW-1 W</td>
<td>Continuous</td>
<td></td>
<td></td>
<td>Range finding</td>
</tr>
<tr>
<td>Solid state</td>
<td>Gallium-Arsenide (GaAs)</td>
<td>0.85-0.805</td>
<td>6 mW</td>
<td>Current pulsed</td>
<td>Up to 5000 0.1 μs</td>
<td></td>
<td>Communications</td>
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<tr>
<td>Liquid</td>
<td>Organic dye</td>
<td>10.4 (tunable)</td>
<td>Up to 5 W (depending on dye)</td>
<td>Laser pumped</td>
<td></td>
<td></td>
<td>Process control, photochemistry, Raman spectroscopy</td>
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<table>
<thead>
<tr>
<th>Common designations</th>
<th>Al</th>
<th>Sn</th>
<th>Zr</th>
<th>Mo</th>
<th>V</th>
<th>Si</th>
<th>Other</th>
<th>Relative density</th>
<th>Condition</th>
<th>0.2% proof stress (MPa)</th>
<th>Tensile strength (MPa)</th>
<th>Elongation (%)</th>
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<tr>
<td><strong>α alloys</strong></td>
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<td></td>
<td></td>
<td></td>
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<td></td>
<td></td>
<td></td>
<td></td>
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<tr>
<td>CP Ti 99.5% IMI 115, Ti-35A</td>
<td>0</td>
<td>0</td>
<td>0</td>
<td>0</td>
<td>0</td>
<td>0</td>
<td></td>
<td>4.51</td>
<td>Annealed 675 °C</td>
<td>170</td>
<td>240</td>
<td>25</td>
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<td>CP Ti 99.0% IMI 155, Ti-75A</td>
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<td>0</td>
<td>0</td>
<td>0</td>
<td>0</td>
<td>0</td>
<td></td>
<td>4.51</td>
<td>Annealed 675 °C</td>
<td>480</td>
<td>550</td>
<td>15</td>
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<tr>
<td>IMI 260</td>
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<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td>0.2 Pd</td>
<td>4.51</td>
<td>Annealed 675 °C</td>
<td>315</td>
<td>425</td>
<td>25</td>
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<tr>
<td>IMI 317</td>
<td>5</td>
<td>25</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td>4.46</td>
<td>Annealed 900 °C</td>
<td>800</td>
<td>860</td>
<td>15</td>
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<td>IMI 230</td>
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<td></td>
<td></td>
<td>2.5 Cu</td>
<td>4.56</td>
<td>ST (α)*, duplex aged 400 and 475 °C</td>
<td>630</td>
<td>790</td>
<td>24</td>
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<td><strong>Near-α alloys</strong></td>
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<tr>
<td>B 1:1</td>
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<td>1</td>
<td>1</td>
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<td></td>
<td></td>
<td>4.37</td>
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<td>950</td>
<td>990</td>
<td>15</td>
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<td>2.25</td>
<td>11</td>
<td>5</td>
<td>1</td>
<td>0.25</td>
<td></td>
<td></td>
<td>4.82</td>
<td>ST (α + β) aged 500 °C</td>
<td>990</td>
<td>1100</td>
<td>15</td>
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<td>6</td>
<td>5</td>
<td>0.5</td>
<td></td>
<td>0.25</td>
<td></td>
<td></td>
<td>4.49</td>
<td>ST (β) aged 550 °C</td>
<td>900</td>
<td>1020</td>
<td>12</td>
</tr>
<tr>
<td>6 2:4 2</td>
<td>6</td>
<td>2</td>
<td>4</td>
<td>2</td>
<td>0.1</td>
<td></td>
<td></td>
<td>4.54</td>
<td>ST (α + β) annealed 590 °C</td>
<td>960</td>
<td>1030</td>
<td>15</td>
</tr>
<tr>
<td>Ti-11</td>
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<td>2</td>
<td>1.5</td>
<td>1</td>
<td>0.1</td>
<td>0.35 Bi</td>
<td>4.45</td>
<td>ST (β) aged 700 °C</td>
<td>850</td>
<td>940</td>
<td>15</td>
<td></td>
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<td>IMI 829</td>
<td>5</td>
<td>35</td>
<td>3</td>
<td>0.3</td>
<td>0.3</td>
<td>1 Nb</td>
<td>4.61</td>
<td>ST (β) aged 625 °C</td>
<td>-</td>
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<td></td>
<td></td>
</tr>
<tr>
<td><strong>α/β alloys</strong></td>
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<td></td>
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<tr>
<td>IMI 318, 6-4</td>
<td>6</td>
<td>4</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td>4.46</td>
<td>Annealed 700 °C</td>
<td>925</td>
<td>990</td>
<td>14</td>
</tr>
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<td>IMI 550</td>
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<td></td>
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</tr>
<tr>
<td>6 2:6</td>
<td>6</td>
<td>2</td>
<td>6</td>
<td></td>
<td>0.7 (Fe, Cu)</td>
<td>4.54</td>
<td></td>
<td>4.68</td>
<td>ST (α + β) aged 550 °C</td>
<td>1170</td>
<td>1275</td>
<td>10</td>
</tr>
<tr>
<td>6 2:4 6</td>
<td>6</td>
<td>2</td>
<td>4</td>
<td>6</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
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<tr>
<td>IMI 551</td>
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<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>Ti-8 Mn</td>
<td>4</td>
<td>4</td>
<td>4</td>
<td>0.5</td>
<td></td>
<td></td>
<td>8 Mn</td>
<td>4.62</td>
<td>ST (α + β) aged 500 °C</td>
<td>1200</td>
<td>1310</td>
<td>13</td>
</tr>
<tr>
<td>β- alloys</td>
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<td></td>
<td></td>
<td></td>
<td></td>
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<td></td>
<td></td>
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</tr>
<tr>
<td>13 11-3</td>
<td>3</td>
<td></td>
<td>13</td>
<td>11</td>
<td>11 Cr</td>
<td>4.87</td>
<td></td>
<td>ST (β) aged 480 °C</td>
<td>1200</td>
<td>1280</td>
<td>8</td>
<td></td>
</tr>
<tr>
<td>Beta III</td>
<td>4.5</td>
<td>6</td>
<td>11</td>
<td>15</td>
<td></td>
<td></td>
<td></td>
<td>5.07</td>
<td>ST (β) duplex aged 1315</td>
<td>1390</td>
<td>10</td>
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<tr>
<td>8 8-2 3</td>
<td>3</td>
<td>8</td>
<td>8</td>
<td></td>
<td>2 Fe</td>
<td>4.85</td>
<td></td>
<td>ST (β) aged 600 °C</td>
<td>1240</td>
<td>1310</td>
<td>8</td>
<td></td>
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<tr>
<td>Transagrace 129</td>
<td>2</td>
<td>11</td>
<td>11</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td>4.81</td>
<td>ST (β) aged 580 °C</td>
<td>1280</td>
<td>1400</td>
<td>6</td>
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<tr>
<td>Delta C</td>
<td>3</td>
<td>4</td>
<td>4</td>
<td>8</td>
<td>6 Cr</td>
<td>4.82</td>
<td></td>
<td>ST (β) aged 540 °C</td>
<td>1130</td>
<td>1225</td>
<td>10</td>
<td></td>
</tr>
</tbody>
</table>

* ST (α), ST (α + β), ST (β) correspond to solution treatment in the α, α + β, and β-phase fields respectively

▲ Annealing treatments normally involve shorter times than ageing treatments
Table 2.4. Chemical composition, alloy type and microstructures of laser welded specimens (17).

<table>
<thead>
<tr>
<th>Alloy composition</th>
<th>Alloy type</th>
<th>Alloy designation</th>
<th>Microstructure</th>
</tr>
</thead>
<tbody>
<tr>
<td>Ti-6Al-1Mo-1V</td>
<td>Near-alpha</td>
<td>Ti-6-1</td>
<td>Equiaxed alpha grains and small amount of intergranular beta</td>
</tr>
<tr>
<td>Ti-6Al-3Sn-4Zr-2Mo</td>
<td>Near-alpha</td>
<td>Ti-6-2-4-2</td>
<td>Equiaxed alpha grains and intergranular beta</td>
</tr>
<tr>
<td>Ti-6Al-4V</td>
<td>Alpha+beta</td>
<td>Ti-6-4</td>
<td>Equiaxed alpha and beta grains</td>
</tr>
<tr>
<td>Ti-6V-3Al-3Sn-3Cr</td>
<td>Beta</td>
<td>Ti-16-3</td>
<td>Equiaxed beta grains</td>
</tr>
</tbody>
</table>

Table 2.5. Welding speeds of c.p titanium and Ti-6Al-4V of the same thickness welded with the same power (1).

<table>
<thead>
<tr>
<th>Weld</th>
<th>Thickness</th>
<th>Material</th>
<th>Laser power</th>
<th>Welding speed</th>
<th>Thermal conductivity at 3000K</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>mm</td>
<td></td>
<td>Watts</td>
<td>mm/sec</td>
<td>W.M⁻¹.K⁻¹</td>
</tr>
<tr>
<td>2</td>
<td>2.032</td>
<td>CP Titanium</td>
<td>1140+10</td>
<td>9.5</td>
<td>20.93</td>
</tr>
<tr>
<td>4</td>
<td>&quot;</td>
<td>&quot;</td>
<td>&quot;</td>
<td>&quot;</td>
<td>&quot;</td>
</tr>
<tr>
<td>10</td>
<td>&quot;</td>
<td>Ti-6Al-4V</td>
<td>1110+10</td>
<td>6</td>
<td>7.33</td>
</tr>
<tr>
<td>12</td>
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<td>&quot;</td>
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</table>
Fig. 2.1. Energy levels diagram of the CO$_2$-N$_2$ system

- Symmetric mode
- Bending mode
- Asymmetric mode

<table>
<thead>
<tr>
<th>Energy (eV)</th>
<th>Symmetric Mode</th>
<th>Bending Mode</th>
<th>Asymmetric Mode</th>
<th>Symmetric</th>
</tr>
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<td>0.4</td>
<td></td>
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</table>

Fig. 2.2. The vibrational modes of the CO$_2$ molecule

- $\nu_1 = 1337 \text{ cm}^{-1}$, $n_1$ Symmetric
  
- $\nu_2 = 667 \text{ cm}^{-1}$, $n_2$ Bending (Twofold degeneracy)

- $\nu_3 = 2349 \text{ cm}^{-1}$, $n_3$ Asymmetric
Fig. 2.3. Diagramatic representation of stimulated emission (6).

Fig. 2.4. The lasing action in a CW CO2 laser (6).
Fig. 2.5. Transverse gas flow type CW CO₂ laser (6).

Fig. 2.6. Axial flow CW CO₂ laser (6).
Fig 2.7. Comparative strength-density characteristics for steels and titanium alloys. (34)

Fig 2.8. Basic types of phase diagrams for titanium alloys. The dotted phase boundaries in (a) refer specifically to the Ti-Al system. The dotted lines in (b) and (c) show the martensite start ($M_s$) temperatures. Alloying elements favouring the different types of phase diagrams are: a. Al, O, N, C, Ga; b. Mo, W, V, Ta; c. Cu, Mn, Cr, Fe, Ni, Co, H
Fig. 2.9. The effect of oxygen concentration on the mechanical properties of titanium.

Fig. 2.10. The effect of nitrogen concentration on the mechanical properties of titanium.
Fig. 2.11. The effect of carbon concentration on the mechanical properties of titanium.

Fig. 2.12. Dependence of (a) melt width and (b) melt depth in laser melted Ti-6Al-4V ($\text{Ti-6-4}$) and Ti-15V-3Al-3Sn-3Cr ($\text{Ti-15-3}$) at 1000W indicated power ($17$).
Fig. 2.13. Dependence of (a) melt width and (b) melt depth on laser beam traverse rate in laser welded Ti-8Al-1Mo-1V (○) and Ti-15V-3Al-3Cr-3Sn (○) at 1500 W indicated power (77).

Fig 2.14. Dependence of dendrite arm spacing on cooling rate for laser melted Ti-15V-3Al-3Sn-3Cr (77).
Fig 2.15. Microhardness traverse across a weld in Ti-6Al-2Nb-1Ta -0.2 Mo (2).

Fig. 2.16. Microhardness measurements across the weld ,HAZ and base metal at the top, centre and bottom of a laser welded sample (18).
Fig. 2.17. Welding assembly used by Mazumder and Steen (1).

Fig. 2.18. Macrohardness traverse for Ti-6Al-4V welds (1).
Fig. 2.19. Variation of UTS, 0.2% proof stress and percentage elongation with power for titanium welds (1).

Fig. 2.20. Variation of UTS, 0.2% proof stress and percentage elongation with welding speed for titanium welds (1).
Fig. 2.21. Variation of titanium welding speed with power (1).

Fig. 2.22. The range of welding runs in titanium alloys for various welding conditions (1).
Fig. 2.23. Variation of HAZ with titanium welding speed (f).

Fig. 2.24. Variation of titanium welding speed with thickness for laser powers of 1000+10 W, 1500+20 W and 2000+40 W (f).
CHAPTER 3. EQUIPMENT AND MATERIALS.

The following Chapter aims to review the equipment used both to produce and to analyse laser cuts and laser welds. The entire laser cutting programme, and a limited amount of laser welding, was carried out using a Coherent Everlase 525.2 laser. The laser cut edges were then analysed using a variety of techniques, such as surface profilometry, HAZ and kerf width measurement, optical and electron microscopy, and Auger analysis. Similarly the majority of the laser welding programme was carried out using a Control 2 KW laser, and laser welds were analysed using metallographic techniques in order to measure HAZ width and weld penetration depth. In addition chemical analysis was also carried out using Auger techniques.

3.1 Lasers And Ancillary Equipment.

The Coherent Everlase 525.2 Laser Installation.

This laser was used both for cutting, and some welding. A diagram of the Coherent Everlase 525.2 is shown in Fig. 3.1. This laser is a slow axial flow laser and can be operated either in the continuous mode (CW) or in the pulsed mode. The maximum power generated by this laser in the CW mode is 350 W, however when the pulsed mode is used higher peak powers can be generated repetitively.

The optical cavity, which is 12 m long, is constructed from glass tubing mounted on O-rings between metal blocks. These blocks hold the mirrors, electrodes and plumbing connections for both the gas and the coolant. The optical cavity is surrounded by another glass tube containing dielectric oil in a closed loop to act as a coolant.
The gas contained in the lasing cavity is a mixture of 13.66% N₂, 4.74% CO₂ and 81.60% He at a partial pressure of 20 torr. This gas is recycled during operation after being cleaned of waste products, by being passed through an oil trap and a series of filters, and catalytic converters. The laser discharge is initiated and supplied by a high voltage power supply, and the average discharge current is limited to 50 mA per section. [36]

The Laser Head

A photograph of the laser head is shown in Fig. 3.2. The head has five main functions which are as follows.

(a) To focus the laser beam which is achieved by using a 65 mm focal length zinc selenide lens.

(b) To introduce the cutting gas, in this case argon, coaxially into the laser beam.

(c) To align the laser beam with the lens using the X-Y movement on the head.

(d) To hold in place the cutting nozzle which in this case was 1 mm diameter outlet.

(e) To align the nozzle coaxially with the beam using the x-y movement on the head.

The x-y Table.

In order to move the titanium samples to be cut under the laser beam an x-y table, was used. This was supplied by Aerotech Ltd. The maximum speed of the table was 300 mms⁻¹. The
The control laser was driven by an Aerotech model 65 motor which has a maximum torque capability of 0.213 Kg m (peak) and a continuous capability of 0.046 Kg m. The motor torque constant is 0.0106 Kg m amp⁻¹ and the tach gradient 3 V per 1000 rpm. (37)

The Control Laser Continuous Wave (CW) 2 KW CO₂ Laser.

Fig. 3.3 is a schematic diagram of the Control Laser 2 KW CW CO₂ laser (Model 901 Mark 2). A roots blower circulates the lasing gas through the plasma tubes and the fast axial design makes it possible to realise high gas velocities, up to around 500 ms⁻¹. Discharge stability at such high gas velocities is achieved by shock stabilisation. Typical gas composition and consumption data for the CW 2 KW laser are given in Table 3.1.

A combination of low current (700 mA DC max) and high voltage (30 KW open circuit) electrical discharge results in the excitation of the gas mixture (as described in the literature survey) to yield 10.6 μm radiation. This radiation is strongly absorbed by many materials and so it is necessary to use specialised materials for the construction of the laser optics.

To contain and direct the beam, water cooled copper mirrors are used. Beam transmission or focussing is achieved by the use of 10.6 μm wavelength transparent media including gallium arsenide, cadmium telluride, zinc selenide and potassium chloride. The control laser 2 KW machine utilises an output window made of water cooled gallium arsenide, which has a reflectivity of about 35 % for 10.6 μm radiation. The lenses used for focussing the beam in this series of experiments were made of potassium chloride due to their low cost.
The total optical cavity is 8 m long folded in the middle and a pneumatically operated on-off shutter beyond the output window directs the beam into a water cooled calorimeter when not in use.

The laser used during the welding programme had also been specially adapted so that it was able to produce a pulsed output. Pulsing is achieved by simple electrical switching of the HT arcs. Prior to the ignition of the arcs the potential difference between the anode and the cathode is greater (35 KV) than that when the arcs are struck (25 KV). This drop in voltage on ignition is part of the design of the machine but does not happen quite instantaneously and so for a short time the arc energy is greater than nominal which boosts the laser output momentarily. Combining this with the fact that a cool lasing gas is more efficient than a steady state excited gas in producing laser light it can be seen that intermittent high energy pulses (approximately 3 x nominal at peak) can be emitted by the laser if the HT arcs are switched on and off. The response times of the electrical and gas systems limit the proximity of the pulses to each other and the average output will be, in general, lower than the nominal CW output for the same electrical conditions. The laser is shown in Fig. 3.4.

The Laser Head.

The laser head, shown in Fig. 3.5 was mounted on a screw thread that could be rotated in order to move the head up and down when focussing. In addition the head also had x and y movement controls so that the laser beam could be aligned directly in the centre of the nozzle. The laser head also performs two other functions, which are to act as a lens holder, and to introduce the shielding gas.
The x-y Control System.

Unlike the Everlase 525 laser used in the cutting programme, where an x-y table was used to move the workpiece, the actual laser head was moved using an NUM numerically controlled system, with the workpiece being stationary. The laser beam itself was directed into the moving head using a series of moving optics.

Laser Cutting Accessories.

The Experimental Arrangement for laser Cutting.

The experimental arrangement for laser cutting is shown in Fig. 3.6. This shows the laser head in position above the workpiece, with the dross jet below it. The same experimental arrangement was used for the laser welds produced using the Coherent Everlase, except that the dross jet was used purely to shield the underside of the weld, and not to blow dross, as in laser cutting.

The Dross Jet.

The dross jet, shown in Fig. 3.7 was used during the cutting programme to blow dross on to one side of the laser cuts. It basically consists of a series of eight nozzles inclined at an angle of 65°, connected to an argon gas supply. Each of the nozzles could be switched on and off by a series of electrical switches. However only one nozzle perpendicular to the cut was required when producing straight line cuts. The dross jet also performed another function which was to shield the underside of both cuts and welds from oxygen contamination.
Laser Welding Accessories.

Experimental Arrangement For Laser Welding.

The experimental arrangement for laser welding, using the Control laser is shown in Fig. 3.8. The photograph shows both the laser head above the workpiece, and the welding jig, which holds the workpiece in position.

The Welding Jig.

In order to protect the underbead of the solidifying weld from oxygen contamination a special welding jig, shown in Fig. 3.9 was constructed. The jig basically consists of a channel machined 5 mm deep out of mild steel, and 1 cm wide, with an argon gas supply fed from underneath the plate into the channel. Inside the channel are two sliding copper inserts which can be opened and closed depending upon the sample size to prevent argon escaping from either end of the channel. When the sample to be welded was placed on top of the channel an argon filled chamber was formed beneath the weld, protecting it from oxygen contamination. Small sliding clamps were used to keep the workpieces in position, and to firmly seal the argon channel. A small amount of argon gas was allowed to escape from two 1 mm diameter holes in the copper inserts to keep a steady flow of gas through the channel.

The Laser Nozzle.

In order to keep the top of the weld bead free from oxygen contamination a 10 mm diameter hole was drilled in the centre of the brass nozzle holder with a 1 mm slit parallel to the length of the weld bead. The 10 mm diameter hole allowed argon to protect the weld whilst it was being formed, whereas the
slit was made to allow a trailing argon shield to shroud the solidifying weld bead. A diagram of the nozzle holder arrangement is shown in Fig. 3.10.

The Laser Beam Analyser (LBA).

The LBA basically consists of a 1 mm diameter molybdenum wire, which is highly reflective, rotated at 1500 rpm by a synchronous motor in a plane perpendicular to the beam. When the wire passes through the beam a very small fraction of the beam is reflected on to the detectors. These are arranged in such a way that the distribution of intensity in both the x and y directions of the beam are obtained in a single scan. The signals are then separately amplified and displayed on oscilloscopes. The intensity of the displayed signal is directly proportional to the intensity of the laser beam. The time axis corresponds to the passage of the needle through the beam. A trigger pulse is also created by the LBA so that it can be synchronised with the oscilloscope. A schematic diagram (53) of the LBA is shown in Fig. 3.11.

The LBA can be used for a variety of different purposes as follows.

(a) To measure the raw beam diameter.
(b) To measure the focussed beam diameter.
(c) To display the laser mode.
(d) To measure temporal variations in laser power.

A full discussion of how the LBA operates is given in Chapter 4.
3.2 Physical Analysis Of Laser Processed Samples.

Metallographic Preparation And Examination.

Samples of the cuts and welds produced were mounted in Epomet mounting compound which is extremely hard and has good edge retention properties. The samples were then ground on rotary grinders using grades 240, 400, 600 and 1200 silicon carbide paper. Polishing was subsequently carried out on a high speed wheel using 1 and 0.5 μm alumina, with a 10% oxalic acid lubricant. The polished samples were finally etched in 1% HF, 20% HNO₃ (by volume) (84).

Optical And Electron Microscopy.

A Reichert MeF 3 was used for optical metallography, and for measuring both HAZ widths and Kerf widths. All electron microscopy was carried out using a Cambridge Stereoscan 2a.

Surface Profilometry.

A Talysurf 10 unit was used in conjunction with an Rₐ (roughness average) meter in order to give a quantitative measure of the surface roughness of laser cuts. The Rₐ value of the surface is the average height of the profile above and below the centre line and is given by

\[ Rₐ = \frac{1}{h} \int_{L} dh \]

where \( h \) = height of profile above or below the centre line at points at unit distances apart.

\[ L = \text{Sampling length} \]
The principle of the Talysurf is illustrated schematically in Fig. 3.12. The pick up is driven slowly across the surface and a sharply pointed stylus follows the profile of the surface irregularities. The pick up has an optical transducer and the vertical movements of the stylus are sensed photoelectrically. The signal is then produced either on a chart display, or as an $R_a$ value on the $R_a$ meter.

**Microhardness Examination.**

Microhardness traverses were carried out across both laser cuts and laser welds, using a Reichert MeF 2 microhardness tester, with a 60 gramme load.

**Tensile Testing.**

Tensile test pieces were produced in accordance with BS 18, the specifications of which are given in Fig. 3.13. The testing programme was carried out using a 50 Tonne Instron tensile tester (Model Number TT-PM-4), from which load-extension curves could be produced.

**Impact Testing.**

The dimensions of the impact test pieces are given in Fig. 3.14. Although the angle and radius of the notch was a standard Charpy dimension, the thickness of the material was only 3.2 mm, since this was the maximum thickness that could be welded. Impact testing was carried out on both the parent and the laser welded material, using a Lofthausen Charpy impact tester (Model Number PSW 30).
Radiography.

In order to detect porosity in the titanium welds radiographic analysis was carried out using a Siemens 70 KV X-ray set. Weld radiographs were produced on Kodak MX paper which has an ultra fine grain size and so can be enlarged.

ESCA (Electron Scanning Chemical Analysis).

ESCA or, XPS (X-ray photoelectron spectroscopy), as it is sometimes known, is an important technique for the chemical analysis of solid surfaces. ESCA was used to determine the chemical composition, of the topmost few atomic layers of the sample. The specimen to be analysed is mounted in an ultra high vacuum and irradiated with X-rays from a Mg $K_{\alpha}$, or an Al $K_{\alpha}$ source. Photoionization takes place in the sample surface, the resultant photoelectrons having a kinetic energy distribution which is determined by both the X-ray kinetic energy and the electron binding energy within the specimen. Only those electrons originating from within the top few atomic layers are able to escape without energy loss, thus the photoelectron spectrum provides an immediate elemental analysis of the specimen surface.

LIMA (Laser Induced Mass Analysis)

LIMA is a powerful and rapid technique for the analysis and characterisation of solid surfaces and particles. The technique combines the advantages of high sensitivity, high spatial resolution and full periodic table coverage with the ability to provide complex molecular information from fragmentation patterns. The maximum depth of analysis is about 10 $\mu$m.
LIMA utilises a pulsed laser at 0.53 or 0.26 μm wavelength, which is focused on to the area of interest on the sample, in an ultra high vacuum chamber. This intense pulse volatises a small volume of the sample, much of which is ejected into the vacuum as molecular ions. The ejected material is then mass analysed in a high resolution time of flight mass spectrometer. The entire process, from laser pulse to full mass analysis is performed in milliseconds.

At the present time however LIMA is at an early stage of development and is not fully quantitative. Further details of the LIMA technique are given in reference (38).

Auger Electron Spectroscopy (AES)

AES was used to analyse laser cut and welded samples of and Ti-6Al-4V in order to produce oxygen profiles of the cut edge. For AES analysis the sample, which is mounted in an ultra high vacuum chamber, is bombarded with an electron beam to induce secondary electron emission. Auger electrons, which contribute to the measured secondary electron spectrum, are the result of a particular rearrangement of electrons in the atoms perturbed by the incident beam. Auger electrons have energies determined by the energy levels in the parent atom and are characteristic of that atom. Since Auger electrons have energies typically in the range 0 to 2000 eV, the distance that they can travel in a solid before losing energy is limited to approximately 1-2 nm. Therefore in order to produce oxygen profiles at greater depths ion etching, which involves the bombardment of the metal surface with argon ions, progressively removes the surface layers of the metal.

Auger analysis was chosen to carry out oxygen profiling in both laser welds and cuts because this technique worked well over the range of profiling depths required. Further details of the Auger analysis technique are given in reference (39, 40).
3.3 Materials

Manufacturers Specifications.

The Ti-6Al-4V was supplied by IMI Titanium Ltd as hot rolled, annealed and pickled sheet. The annealing process was carried out at 720°C. The manufacturers specifications for the alloy are given in Table 3.2.

Analysis Of Material Supplied.

The manufacturers specifications were supplied with the materials. However detailed surface examinations of the Ti-6Al-4V sheet were carried out using both ESCA and LIMA techniques, in order to provide a detailed background to interpretation of the analytical results obtained for non-metallic elements in laser processed material.

ESCA Analysis Results.

An elemental analysis was carried out both on the surface of the samples and also on the samples after ion etching for two minutes. Ion etching involves bombarding the surface with argon ions to remove a few layers of atoms. This process also removes a large proportion of the surface impurities such as carbon (i.e. grease and dirt). However in most cases there were still high carbon and oxygen levels present so that for subsequent analyses the results were corrected for both carbon, and for carbon and oxygen. The overall results are given in Table 3.3.

The main features of the ESCA analyses were as follows.
(a) No vanadium was found to be present at the surface of the alloy even after ion etching for 2 minutes.

(b) An oxide layer was present on the samples and this consisted entirely of TiO$_2$ and Al$_2$O$_3$.

(c) In all but one case there was a greater ratio of Al:Ti in the surface region than in the bulk composition, as indicated by the data in Table 3.2.

LIMA analysis results.

LIMA results from the Ti-6Al-4V alloy surface are shown in Fig. 3.15, and results for the bulk composition, taken at a depth of a few μm are shown in Fig. 3.16. The results show the presence of both Ti and Al peaks in both cases. However the surface results show a much larger oxygen peak, than the bulk alloy, as would be expected due to the oxide layer. The major drawback of the LIMA technique, however, is that it is unable to give quantitative results. However research work is being carried out by Loughborough Consultants with the aim of calibrating the LIMA for a wide variety of materials. Another interesting feature of both Figs. 3.16 and 17 is that vanadium is present, which was not shown by the ESCA analysis. This is a reflection of the increased depth (1-10 μm) that is analysed by LIMA, compared to a few nm for ESCA. Further details of the LIMA technique are given in reference (39).
Table 3.1. Gas consumption data for the Control Laser. (36)

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<th>Gas</th>
<th>Composition Wt %</th>
<th>Typical gas consumption at full power with recirculator engaged (litre/hr)</th>
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<tr>
<td>Carbon Dioxide</td>
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Table 3.2. Manufacturers specifications for the Ti-6Al-4V alloy.

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Table 3.3. ESCA Analysis Results for Ti-6Al-4V.

(a) Uncorrected Results

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<th>Wt %C</th>
<th>Wt %Ca</th>
<th>Wt %Ti</th>
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(b) Results Corrected For Carbon

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(c) Results Corrected For Carbon and Oxygen

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Fig 3.1. Schematic diagram of the Coherent Everlase 525.2 laser.

Fig 3.2. The Coherent Everlase laser head.
Figure 3.3. Schematic diagram of the Control Laser 2 KW CW CO2 laser (model 901 mark 2)
Fig 3.4. The Control Laser.

Fig 3.5. The Control laser head.
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Fig 3.7. The Dross Jet.
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Fig 3.9. The Welding Jig.
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Fig 3.15. LIMA analysis of alloy surface.
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CHAPTER 4. THE MEASUREMENT OF BEAM DIAMETER AND INTENSITY DISTRIBUTION IN THE 2KW CONTROL LASER AND THE COHERENT EVERLASE 525.2.

A knowledge of the focused laser beam spot diameter \( (D_b) \) is vital during both welding and cutting since the energy density at the workpiece is directly related to the area of the focused spot. Thus it is important to have an accurate method of determining \( D_b \) in order to predict depths of penetration and maximum welding speeds. In the past many different techniques have been used with varying degrees of success, the following sections aim to briefly review some of these, and to compare these methods with the state of the art technique of using the laser beam analyser. Before doing this, however, it is necessary to discuss energy distributions in laser beams.

4.1 REVIEW OF PREVIOUS WORK.

The spatial distribution of energy across the cross section of a laser beam is described as its mode structure. This mode structure of a laser beam is determined by laser cavity design factors, such as the shape of the mirrors, the diameter of the tube, inhomogeneities in the lasing medium and pumping power. These only allow certain modes of vibration each of which has its own intensity distribution. Both the Control Laser and the Coherent Everlase 525.2 are designed to produce the TEM\(_{00}\) (or lowest order) mode structure, which has a Gaussian distribution of intensity. This can be mathematically described as follows.

\[
P_r = P_0 \exp \left(-\frac{2r^2}{r_0^2}\right)^2 \tag{4.1}
\]

where \( P_r \) = power intensity at a distance \( r \) from the beam centre
\[ P_0 = \text{power intensity at beam centre (i.e. peak intensity)} \]
\[ r = \text{radial distance from beam centre} \]

and \( r_0 \) is the Gaussian beam radius normally described as the radius at which the intensity is \( 1/e^2 \) (41), of the peak (central) power density. This value will be used in all future calculations referring to beam diameter.

The total power can be obtained by integrating the power intensity distribution, so that at the \( 1/e^2 \) beam radius

\[ P(\text{total}) = \int_0^\infty P_0 \exp \left(-2r/r_b\right) 2\pi r \ dr \quad (4.2) \]
\[ = \frac{P_0 \pi r_b^2}{2} \]

Therefore \( P_0 = 2P(\text{total}) / \pi r_b^2 \) \quad (4.3)

So substituting for \( P_0 \) in equation (4.1)

\[ P_r = \frac{2 \cdot P(\text{total}) \exp \left(-2r^2/r_b\right)}{\pi r_b^2} \quad (4.4) \]

**Single Isotherm Plotting**

This is the simplest method of measuring beam diameter and includes a variety of techniques such as charring paper and drilling holes in both Perspex and metal foils (42). These techniques are extremely inaccurate since the size of the isotherm recorded is dependent upon both the exposure time and the laser power, and the recorded beam diameter will appear greater with any increase in either. This is shown clearly in Fig. 4.1. It is impossible also to measure the beam diameter at \( 1/e^2 \) of the peak intensity, as is required.
Cohen and Epperson (43) developed a technique in which holes were drilled in titanium nitride, using a ruby laser, however the beam diameter was also found to increase with laser power, which is shown in Fig. 4.2.

**Photographic Techniques**

This is a multiple isotherm contouring technique and uses photographically sensitive paper and has been used with low power lasers (44). With high power lasers however a new paper would have to be developed and liquid nitrogen cooling would be required.

**Photon Drag Method.**

This method has been successfully used to determine the beam diameter of CW lasers (45,46,47). The detector basically consists of a germanium crystal connected to a signal amplifier, and to a storage oscilloscope. When the laser beam is fired on to the crystal a visual display of the power distribution falling on the crystal is shown on the oscilloscope. This is due to the transfer of photon momentum to mobile holes in the crystal, known as the photon drag effect. The incident beam is chopped by a rotating slit on the chopper so that the crystal is not damaged by excessive exposure to the laser beam. The slit must be wide enough to provide enough power to drive the detector, but narrow enough to give adequate resolution.

The main drawback of the photon drag method is that it gives a convoluted signal which is very difficult to interpret. This method may also only be used for Gaussian beam profiles. However most practical laser systems are affected by distortion of optical elements, producing slightly irregular nonaxisymmetric beams. This problem becomes greater with increasing laser power.
Thermal Rise Method.

This method assumes that any laser beam can be represented to a first approximation by a Gaussian source having an equivalent heating effect. Essentially the surface temperature of a laser heated spot is measured using an infra-red thermometer, from which the thermal history of the substrate can be constructed, and the time taken to reach 90% of the equilibrium temperature can be found. The value of 90% appears to be purely arbitrary. This time can then be compared with that predicted theoretically, by a finite difference mathematical model (49), from which it can be shown that regardless of heat loss, power, substrate or beam diameter ($D_b$)

$$D_b = (16 \text{ at } 90\% / 0^{*90\%})^{1/2}$$

where

- $a = \text{thermal diffusivity}$
- $T^* = \text{dimensionless temperature}$
  $$= \frac{T K D_b}{P(1-r_f)}$$
- $0^* = \text{dimensionless time}$
  $$= \frac{16 a t}{D_b^2}$$

$T = \text{Temperature}$
$K = \text{Thermal conductivity of the substrate}$
$P = \text{Laser power}$
$r_f = \text{Surface reflectivity}$
$t = \text{Time}$

Therefore to calculate the Gaussian beam diameter equivalent of any heated spot the time to 90% thermal equilibrium has to be measured at a known radial location i.e. known as $0^{*90\%}$. A typical infra-red thermometer trace in glass is shown in Fig. 4.3. There are 3 main problems involved in this method, these are as follows.
(a) The whole method depends upon there being no surface melting.
(b) The reflectivity does not remain constant with increasing temperature, which is assumed by the method.
(c) Complex heat flow analysis is required.

An accuracy of ± 19% was quoted for this thermal rise method.

Intensity Profile Measurement Using PMMA.

Miyamoto et al (49) measured the intensity profile and beam diameter of a focused CO₂ laser beam by evaporating a shape in PMMA (polymethyl methacrylate). The method utilized the fact that PMMA is not sublimated by the laser beam, and that there are negligible thermal conduction and reflection losses. Values of both the latent heat of evaporation and the evaporation threshold energy were found. The groove produced in the PMMA by the laser was scanned at various speeds using a T.V. camera, and the depth of the groove (z) plotted against \( W / \sqrt{Hv} \), where \( W \) = laser power, \( H \) = latent heat of evaporation and \( v \) = welding speed. The results are shown in Fig. 4.4.

Miyamoto et al. derived the expression below which describes the groove in terms of the latent heat of evaporation (\( H \)) in Jm⁻³ and the threshold energy for evaporation (\( G \)) in Jm⁻².

\[
W / \sqrt{Hv} = az + aG / H
\]

where \( a \) = beam radius and \( z \) = groove depth.

The beam radius \( a \) is thus equivalent to the gradient of the graph, since for a straight line graph \( y = mx + c \), where \( m \) is the gradient. This linear relationship appears to apply to groove depths of up to 0.4 mm whereafter the relationship does not hold, and so it is apparent that the method has limitations.
Theoretical Techniques For The Determination Of Beam Diameter.

The theoretical radius of the diffraction limited spot size was calculated by Ready (50) and is given by

\[ r_b = \frac{1.22 f \lambda}{D} \]

where

- \( r_b \) = radius of the diffraction limited spot size
- \( D \) = unfocused beam diameter
- \( \lambda \) = wavelength of laser radiation
- \( f \) = lens focal length

Substituting in values of \( f = 100 \text{ mm} \); \( D = 25 \text{ mm} \); \( \lambda = 10.6 \text{ m} \mu \text{m} \), for the Control laser, gives a value of \( D_b = 0.103 \text{ mm} \). Similarly for the Coherent Everlase 525.2 substituting in values of \( f = 63 \text{ mm} \); \( D = 10 \text{ mm} \); \( \lambda = 10.6 \text{ m} \mu \text{m} \), gives a value of \( D_b = 0.162 \text{ mm} \).

The above values are very small and theoretical techniques to calculate beam diameter do not take into account the following factors.

(a) Spherical aberration in the lens, which leads to an increased spot size. For an F1.5 lens an increase of spot size of 35% has been reported (51). However Steen (42) has reported that if the F number of the lens is greater than 7 spherical aberration is not a problem.

(b) Diffraction of the Gaussian beam caused by truncation at the aperture has been responsible for increases of up to 50% (52) in beam diameter at the 1/e² radius.

Thus the calculated focused beam diameter is always much smaller than the actual focused beam diameter.
4.2 Experimental Work And Analysis.

Determination Of Raw Beam Diameters, Focused Beam Diameters, Power Ripples And Mode Structures Using The Laser Beam Analyser (LBA).

In this series of experiments an LBA 2 was used. The LBA basically consists of a 1 mm diameter molybdenum wire, which is highly reflective, rotated at 1500 rpm by a synchronous motor in a plane perpendicular to the beam. When the wire passes through the beam a very small fraction of the beam is reflected on to the detectors. These are arranged in such a way that the distribution of intensity in both the x and y directions of the beam are obtained in a single scan. The signals are then separately amplified and displayed on oscilloscopes. The intensity of the displayed signal is directly proportional to the intensity of the laser beam. The time axis corresponds to the passage of the needle through the beam. A trigger pulse is also created by the LBA so that it can be synchronised with the oscilloscope. A schematic diagram of the LBA is shown in Fig. 4.5 (53).

An LBA can be used in three different ways as described below.

(a) The LBA is used with the detectors nearer the laser head than the reflecting rod, i.e. with the laser beam being reflected back towards the detectors at an angle of 67°. In this way it is possible to measure the raw beam diameter, and show the laser mode structure in both the x and y directions.

(b) The LBA is used as in method (a), with the needle stationary, so that the temporal, rather than the spatial variations in the laser beam are displayed. This method is used to determine the power ripple produced by certain lasers. The LBA can also be used in this way to accurately monitor pulse frequency and duration, when using the laser in the pulsed mode.
(c) The LBA is used with the detectors further away from the laser head than the rotating needle, as shown in Fig. 4.6, with the laser beam being reflected at an angle of 157°. In this way the focused beam diameter can be measured by rotating the needle through the focused spot.

Measurement Of Raw Beam Diameter And Mode Structure.

The LBA was aligned directly below the output mirror, of the Control Laser, with no lens present (method a). The laser was then switched on at a laser power of 1500 W, and the power intensity distribution was shown on a digital storage oscilloscope. The trace is shown in Fig. 4.7. The photograph was enlarged and thus the 1/e² beam radius could be measured. For the Control Laser it was found to be 25 mm.

The same experiment was repeated using the Coherent Everlase 525.2, at a laser power of 150 W. The mode structure is shown in both the x and y directions in Fig. 4.8. The mode structure is almost perfectly Gaussian which is ideal for laser cutting. As with the Control laser the raw beam diameter could be measured and was found to be 10 mm.

Measurement Of Focused Spot Size.

In this case the Control laser was set up as in method (c) with a 100 mm focal length KCl lens in place. The LBA was turned upside down with the rotating needle closest to the laser head. The laser was then switched on, at a power of 1500 W, and the beam roughly focused onto the LBA needle. As in the previous section the intensity distribution was displayed on the storage oscilloscope. The exact focus of the beam was determined by
moving the laser head until the peak heights displayed were at a maximum. The experiment was repeated using the Coherent Everlase 525.2 using a 63 mm focal length zinc selenide lens, and a power of 150 W. The results for the Control laser are shown in Fig. 4.9, and those for the Coherent Everlase 525.2 in Fig. 4.10.

There are two peaks present in Figs. 4.9 and 4.10 because as the rotating wire scans through the laser beam the x and y signals, from the detectors, are separated temporally because the scattering comes from opposite faces of the wire, this is shown in Fig. 4.6. Since the radius \( r \), of the wire is known it is possible to calculate the temporal separation of the peaks \( y \) by geometry. From Fig. 4.6 \( \sin \theta = 0.5 \), \( y/r \) and hence \( y = 2r \sin \theta \), where \( r = 0.5 \text{ mm} \) and \( \sin \theta = 0.92 \).

From this the temporal separation of the peaks \( y \) is 0.92 mm. This then allows calibration of the oscilloscope trace. The photograph shown in Fig. 4.9 and 10 were then enlarged and the beam diameter at the \( 1/e^2 \) value measured. For the Control Laser at a power of 1500 W, and using a lens focal length of 100 mm, \( D_b \) was found to be 0.446 mm, and for the Coherent Everlase 525.2, using a lens focal length of 63 mm, \( D_b \) was found to be 0.295 mm.

However there is also a correction factor \( (dx) \) that must be subtracted from these values and this is given in equation (4.5) (54).

\[
dx = 0.5r\cos \theta \left[ \frac{(d/L)^2 + (w/f)^2}{1} \right]^{1/2} \quad (4.5)
\]

where
- \( r \) = radius of measuring wire = 0.5 mm
- \( \theta \) = angle of reflection to detector = \( \cos 67^\circ = 0.39 \)
- \( L \) = distance detector-wire = 127 mm
- \( d \) = opening of the detector = 3.5 mm
- \( w \) = diameter of beam
- \( f \) = lens focal length
The first term results from the divergence of the radiation which is reflected into the detector, by the wire, and the second term is results from the divergence of the focused beam.

These values give \( dx = 0.026 \) mm for the Control laser, and \( dx = 0.031 \) mm for the Coherent Everlase 525.2.

The effective beam diameter \( (D_b) \) for the Control laser \( = x - dx \)
\[ = 0.446 - 0.026 \text{ mm} \]
\[ = 0.420 \text{ mm}. \]

The effective beam diameter \( (D_b) \) for the Coherent laser \( = x - dx \)
\[ = 0.295 - 0.031 \]
\[ = 0.264 \text{ mm}. \]

These values are much greater than those predicted theoretically, for the reasons given previously. The beam diameter of the Coherent Everlase is also much smaller than that of the Control laser due to the much smaller focal length lens and the superior quality mode.

Calculation Of Peak Power Intensity.

From equation (4.3), in section 4.1, the peak power intensity \( P_0 \) is given by

\[ P_0 = \frac{2P_{(\text{total})}}{\pi r_b^2} \]

Therefore substituting in values for the Control laser, at the maximum power

\[ P_{(\text{total})} = 2000 \text{ W} \]
\[ r_b = 0.5 \times 0.42 = 0.21 \text{ mm} \]

This gives the peak power intensity as \( 2.16 \times 10^{10} \text{ Wm}^{-2} \).

Similarly for the Coherent Everlase,

\[
P(\text{total}) = 350 \text{ W} \\
r_b = 0.264 \text{ mm}
\]

which gives a peak power intensity of \( 1.27 \times 10^{10} \text{ Wm}^{-2} \).

From these results it can be seen that although the Coherent Everlase produces only 20\% of the continuous power output of the Control laser, the Coherent laser has a peak power density of 60\% of that of the Control laser.

**Measurement Of Power Fluctuations**

In this case the laser was arranged as in method (b), except that the time base on the oscilloscope was increased, and that the rotating needle in the LBA was stationary under the beam. In this case the LBA detectors measure the temporal, rather than the spatial variations in the laser beam. Several peaks were displayed on the screen, as shown in Fig. 4.11 and 4.12. It can clearly be seen that there is a power intensity fluctuation. This was probably caused by incomplete smoothing of the HT power supply which feeds the discharge. This problem has also been shown in previous work (55).

From Figs. 4.11 and 12 it is also possible to identify the basic frequency of the power ripple, by measuring the distance between the peaks. For the Coherent Everlase laser it was found to be approximately 320 Hz, and for the Control Laser 300Hz.
Fig. 4.1. The error in using single isotherm techniques in measuring Gaussian beam diameter. (42)

Fig. 4.2. The apparent variation in beam diameter with power (43).
Fig. 4.3. Typical infra red thermometer trace for the thermal rise method, in glass (48).

Fig. 4.4. The depth of groove (z), in PMMA, against \( W / \sqrt[4]{Hv} \), using a laser power of 200 W. (49)
Fig. 4.5. Schematic diagram of the laser beam analyser (53).

Fig. 4.6. Cross sectional view of the wire cutting the laser beam.
Fig. 4.7. Oscilloscope trace for the raw laser beam at a power of 1500 W, in the Control laser.

Fig. 4.8. Oscilloscope traces of the raw beam in both the x and y directions in the Coherent Everlase 525.2, at a power of 150 W.
Fig. 4.9. Oscilloscope trace for the Control laser, focused using a 100 mm F.L. lens, at a power of 1500 W, showing the temporal separation of the peaks.

Fig. 4.10. Oscilloscope traces for the Coherent Everlase 525.2, focused using a 63 mm focal length lens at a power of 150 W.
Fig. 4.11. Power intensity fluctuation in the Control laser.

Fig. 4.12. Power intensity fluctuation in the Coherent Everlase 525.2.
CHAPTER 5. LASER CUTTING OF TITANIUM ALLOYS.

The cutting of titanium alloys in industry has long been a problem. Flame cutting is widely used to cut steels, but it cannot be used with titanium alloys due to the problem of oxygen contamination of the resulting cut edge. Guillotining, or machining, are the only methods at present available to profile titanium alloys. However the drawback of these is the high cost of dies and tools, which wear out very rapidly.

Recently the much more flexible process of laser cutting has been introduced to cut titanium alloys. In conjunction with a computer controlled x-y table a wide variety of complex shapes can be profiled. Two different types of laser cutting are commonly used.

(a) Oxygen assisted laser cutting.

In this case the laser beam is used to heat up the metal and then the coaxial oxygen jet exothermally reacts with the metal oxidizing it and forming a cut. (85)

(b) Inert gas assisted laser cutting.

Here the metal beneath the laser beam is melted and protected from atmospheric oxidation by the coaxial argon gas jet. This gas jet then physically shears the molten metal producing a cut.

When laser cutting titanium alloys inert gas assisted cutting is used in order to protect the cuts from oxygen contamination. This is a serious problem in titanium alloys since it reduces the fatigue life and embrittles the metal (8). A review of previous attempts at laser cutting titanium alloys, and the problem of oxygen contamination is given in the literature survey.
In this series of experiments laser cuts were produced in 1, 1.7 and 2.7 mm thick Ti-6Al-4V alloy sheet, using a Coherent Everlase 525.2 laser with a maximum power output of 350W. Argon shielding was used from above the cutting zone and a wide variety of cuts were produced using different gas pressures and cutting speeds. Very smooth cuts were produced, however in all cases there were two main problems. These were as follows.

(a) Dross adhesion to the underside of the cuts due to the high surface tension of liquid titanium alloy.

(b) Higher levels of oxygen contamination than in guillotined edges.

These problems were solved by the use of a dross jet which was positioned underneath the cut zone and simply blew argon at an angle of 90° across the underside of the cut blowing all the dross to one side of the cut and leaving the other side dross free. The dross jet also shielded the underside of the cut from oxygen contamination, and Auger analysis showed that laser cuts produced in 1 mm Ti-6Al-4V alloy sheet had levels of oxygen contamination equal to those in guillotined edges.

5.1 Experimental Procedure.

In this study a series of laser cuts were produced in 1, 1.7 and 2.7 mm thick Ti-6Al-4V alloy sheet specimens using a laser power of 300 W. The cutting speed was varied between 1 and 40 mms⁻¹, and the argon gas pressure between 70 and 270 KPa. Cuts were also produced using the dross jet, and the effect of the variation in both dross jet gas pressure (P_d), and dross jet-material distance (Y_{dm}) were studied. All the cuts produced were then examined using a variety of techniques including surface profilometry, electron microscopy, Auger analysis, kerf width measurement, optical metallography and HAZ width measurement.
All metallographic sections were prepared by first conventionally grinding and then polishing on a high speed wheel using a 1 μm alumina abrasive with a 10% oxalic acid lubricant. A final polish was carried out using 0.05 μm alumina. The polished samples were then etched in 1% HF, 20% HNO₃. (83)

The experimental arrangement of the laser cutting process is shown in Fig 5.1. The following nomenclature will be used in the remainder of this work.

\[ Y_h = \text{Laser head stand off.} \]
\[ Y_{dm} = \text{Dross jet-material distance.} \]
\[ Y_{ns} = \text{Dross jet nozzle separation.} \]
\[ P_h = \text{Laser head gas pressure.} \]
\[ P_c = \text{Gas pressure exerted on the cut.} \]
\[ P_d = \text{Dross jet gas pressure.} \]
\[ t = \text{Ti-6Al-4V sheet thickness.} \]

The relationships between the laser head stand off \((Y_h)\), the laser head gas pressure \((P_h)\) and the actual gas pressure exerted on the cut \((P_c)\) were determined experimentally and the results are shown in Fig 5.2.

5.2 Results and Discussion.

5.2.1 Laser Cutting Without Using The Dross Jet.

(a) Fundamental analysis of the laser cutting mechanism.

Although detailed analysis of the oxygen assisted laser cutting mechanism has been carried out (56) there has been little or no previous investigation into the mechanism of inert gas assisted laser cutting. One of the major aims of the cutting programme therefore was to conduct detailed analyses of laser cuts in order to postulate a mechanism by which inert gas
assisted laser cutting takes place. The following three types of observations form a satisfactory starting point for this analysis.

(a) Determination of the inclination of the cutting front angle ($\alpha$).

(b) Measurement of dross particle spacing.

(c) SEM analysis of cut edges.

**The Cutting Front Angle.**

When metals are laser cut, using either oxygen or argon as the assist gas, a cut front is formed the inclination of which is of critical importance when attempting to analyse the laser cutting mechanism. Fig 5.3 is a schematic representation of laser cutting, where $\alpha$ is the cut front angle (the angle of the cut front to the vertical).

The cut front angle $\alpha$ was measured experimentally by sectioning a series of laser cuts parallel to the kerf as shown in Fig 5.4. The sectioned cuts were then polished until the position at which laser cutting ceased could be seen. At this point there was in effect a frozen representation of the cutting front.

Fig 5.5 is an SEM micrograph of a sectioned laser cut produced using a cutting speed of 35 mms$^{-1}$. This shows the cutting front angle $\alpha$ to be approximately 6° to the vertical. Similar analysis of a laser cut produced using a cutting speed of 15 mms$^{-1}$ showed that the cutting front angle decreased slightly to approximately 4°. These results infer that decreasing cutting speed results in an decreased cutting front angle.
Having measured the cutting front angle $\alpha$ to be approximately $6^\circ$ (at the maximum cutting speed, 35 mms$^{-1}$) it is possible to calculate by simple geometry the depth to which the cut front is directly impinged upon by the incident laser beam. The focussed beam diameter was measured experimentally (using the laser beam analyser as described in Chapter 4) as 0.26 mm. It can be seen from Fig 5.6 that the incident laser beam impinges upon the entire depth of the cut front, when cutting 1mm thick material. It is also clear that there is a slight time lag between the beam impinging upon the metal and the cut front being established. The cause of this delay is that since the beam has a Gaussian energy distribution a certain amount of energy is required before melting can occur.

**Measurement Of Dross Particle Separation.**

Table 5.1 records the variation in dross particle separation with cutting speed for laser cuts produced using a power of 300W and a gas pressure ($P_h$) of 138KPa. It can be seen that the dross pitch is approximately constant at about 0.3 mm, which corresponds very closely to the measured focused laser beam diameter of 0.26 mm (as measured by the laser beam analyser in Chapter 4).

**SEM Analyses Of Laser Cut Edges.**

Micrographs of a series of laser cuts produced using cutting speeds of 15, 20, 25, 32.5 and 37.5 mms$^{-1}$, with a gas pressure ($P_h$) of 138 KPa are shown in Fig. 5.7. It can be seen quite clearly that there are two distinct regions present on the cut surfaces. These are a smooth region (the directly melted zone) above the line X-X and a rougher region (the fluid flow zone) below the line X-X. An explanation as to how these zones are created is given in the following section.
The Laser Cutting Mechanism.

From the SEM analyses, the cutting front angle (\(\alpha\)) determination, and the dross particle separation results it is possible to postulate a mechanism for inert gas assisted laser cutting. Fig. 5.8 is a schematic representation of the cutting process, which can be explained as follows.

Stage (A)
From Fig. 5.6 it is apparent that the whole of the cut front is irradiated by the beam during the cutting process. However this view is somewhat misleading as it implies that all the material is directly melted by laser irradiation. In fact only the topmost part of the melt is directly melted in this manner. This melt then flows over the surface of the remainder of the cut zone shielding it from direct laser impingment but at the same time absorbing the energy and melting by conduction. In addition a proportion of the beam has to pass over the solid top surface of the metal sheet before sufficient energy is absorbed to initiate melting.

Stage (B)
The molten metal created in the directly melted zone is then propelled by the pressure of the gas jet down the solid cutting front. As this molten material falls down the cutting front it melts the adjacent solid metal. This process creates the fluid flow zone which is characterised by a series of troughs gouged from the solid cutting front by the falling molten metal. Eventually the molten metal extends through to the bottom of the sheet at which point the material in the centre of the cut zone which is hotter and more fluid (due to the Gaussian energy distribution) is expelled from the bottom of the cut. The less fluid (cooler) metal adjacent to the cut edge, however, adheres to the bottom lip of the cut forming a small molten dross.
reservoir. This reservoir then grows in a spherical manner to achieve the minimum surface area to volume ratio.

Stage (C)
As the laser beam moves over the metal the directly melted zone advances ahead of the fluid flow zone creating a step, as shown in Figs 5.8 and 5.9. The depth of the step gradually decreases as the cutting front moves forward, until the depth of the step is so shallow that the incident laser beam is able to melt directly through it. This results in the flow of molten metal to the reservoir being terminated.

Stage (D)
The liquid metal reservoir then solidifies and acts as a partial barrier to the gas jet passing through the kerf. The gas jet is then deflected to the other side of the cut and initiates another molten metal reservoir. The process is then continually repeated until symmetrical dross deposition along the bottom edge of the cut is produced as can be seen clearly in Fig 5.10.

The Effect Of Commencing Cutting Above The Workpiece.

The majority of the laser cuts produced in this study involved the laser beam being switched on and then being moved across the edge of the sheet to be cut. However a few cuts were produced by switching the laser beam on with the material in position directly underneath the beam. It was found that either a short dwell time, or a greater laser power were required to produce a cut when the material was already in position underneath the beam, than when the laser beam was moved over the edge of the material.
This can be explained as follows. When a laser beam is moved over the edge of a metal sheet the top part of the sheet is initially melted, to form the directly melted zone. This molten metal then flows down the edge of the metal melting solid material and producing the fluid flow zone (as described in stage B above). However when laser cutting is commenced with the laser beam directly above the material to be cut in order for cutting to be initiated a directly melted zone must be produced which extends through the depth of the cut. This requires either a short dwell time, or a greater laser power. Once this full penetration melt has been produced then the cutting mechanism becomes the same as in conventional laser cutting.

(b) Analysis Of The Surface Roughness Of Cut Edges.

The surface roughness of a laser cut is composed of both random and cyclic events which create striations on the cut surface. In order to measure the total roughness of the laser cut surfaces a Talysurf 10 surface profilometer was used to analyse laser cut samples of Ti-6Al-4V alloy. \( R_a \) (roughness average) values could be read off a meter and chart traces of the surface profile could also be produced. Extensive SEM (Scanning electron microscopy) was also carried out in order to characterise the laser cut surfaces.

The Variation In Surface Roughness With Cutting Speed.

Fig 5.11 shows the variation in \( R_a \) with speed using a laser head stand off \( (Y_h) \) of 0.13 mm, and a head gas pressure \( (P_h) \) of 380 KPa producing a gas pressure \( (P_c) \) of 370 KPa at the cut. Positions 1, 2 and 3 correspond to \( R_a \) values 0.25, 0.5 and 0.75 mm away from the top edge of the cut. It can be seen that
the surface roughness is not uniform across the depth of the cut. At positions 1 and 2 the $R_a$ values are much lower than at position 3, which can be explained as follows. At low cutting speeds (below 15 mms$^{-1}$) the melt zone, which has low $R_a$ values, extends almost to the bottom edge of the cut so that positions 1, 2 and 3 fall within it. As the cutting speed is increased positions 1 and 2 are still included within the melt zone, but position 3 falls within the much rougher fluid flow zone. The $R_a$ values within the fluid flow zone then increase rapidly with cutting speed, which occurs because at low cutting speeds the removal of the melt by the argon jet is very efficient. However as the cutting speed increases the distance between the jet and the exit zone increases, as shown in Fig 5.12. Therefore ejection becomes less effective and more molten material adheres to the edge of the cut.

An SEM photograph of a cut edge produced using a cutting speed of 35 mms$^{-1}$ and a gas pressure ($P_c$) of 370 KPa is shown in Fig 5.13. The increased surface roughness near the bottom edge of the cut (the fluid flow zone) can be clearly seen below the line X-X. The positions 1, 2 and 3 are also marked illustrating how position 3 falls within the fluid flow zone.

The Variation In Surface Roughness With Gas Pressure.

Fig 5.14 shows the variation in $R_a$ with cutting speed for samples cut using a laser head stand off ($Y_h$) of 0.13 mm, and a head pressure ($P_h$) of 138 KPa, producing a gas pressure at the cut ($P_c$) of 130 KPa. The positions 1, 2 and 3 correspond to $R_a$ values 0.25, 0.5, and 0.75 mm away from the top edge of the cut in 1mm thick sheet. Fig 5.14 shows that the surface roughness ($R_a$) values are similar, at positions 1 and 2, to those shown in Fig 5.11, where a higher gas pressure ($P_h$) of 380 KPa was used. However the $R_a$ values at position 3 are much greater when a higher gas pressure was used because higher gas pressures result
in a shallower melt zone depth, and a larger fluid flow zone, for a given cutting speed. This occurs because an increase in gas pressure results in greater convective cooling by the gas jet, resulting in a shallower melt zone.

**Horizontal Striations.**

In addition to the vertical striations, there are also horizontal striations present in many of the cuts, and these are shown most clearly in Fig 5.15. These horizontal striations occurred both in the melt zone and the fluid flow zone of the cut surface showing that they were produced after the removal of most of the liquid material from the cut zone. Their cause can be explained as follows. Once most of the liquid metal has been removed from the cut zone a thin skin of molten material is left. As the laser beam moves on, the vertical gas jet then impinges upon the upper region of the solidifying melt, as shown in Fig 5.16. The gas jet creates a wave which is then convectively cooled by the gas jet, and solidifies creating a horizontal striation. The gas jet then is able to produce another horizontal striation below the first by the same mechanism. However it can be seen from Fig 5.15 that the striations are very closely spaced towards the top surface of the cut but the distance between each striation increases almost logarithmically towards the bottom edge of the cut. This can be explained by consideration of three factors.

(a) As the gas jet convectively cools and solidifies striations, the jet becomes warmer so that it becomes progressively less effective as a coolant.

(b) As the distance from the cutting nozzle increases the cooling effect of the gas jet decreases, due to a decrease in local flow rates as a result of "fanning out" of the gas jet in the kerf (see Fig 5.16).
(c) The melt thickness increases with distance from the top edge of the cut as a result of conductive heating and a decrease in fluid removal rates as the gas jet is dissipated. Therefore greater convective cooling by the gas jet is required to solidify a wave.

**The Mechanics Of Striation Formation In Laser Cutting.**

**Oxygen Assisted Laser Cutting Of Ferrous Metals.**

Vertical striations are produced along the cut edge during the laser cutting of most metals. The mechanism of striation formation in oxygen assisted laser cutting has been extensively studied\(^{(56,86)}\). A schematic diagram of the laser cutting of mild steel is shown in Fig 5.17. Basically the laser beam causes the metal to be heated up, with the position X reaching the highest temperature. As the beam encroaches over the metal the temperature at X exceeds that required for ignition of the metal in the flowing oxygen atmosphere. The combination of the exothermic ignition reaction and the continuing laser irradiation causes a highly energetic burning front to be produced which burns away radially from X, at a velocity greater than the cutting velocity. Therefore the burn front leaves the laser behind and subsequently cools down and self extinguishes. The beam then passes over the metal again and the process repeats itself generating vertical striations.

**Inert Gas Assisted Laser Cutting.**

The mechanism by which striations are produced in inert gas assisted laser cutting cannot be similar to oxygen assisted laser cutting since no exothermic reaction takes place. Instead a melt is produced and then blown away by the gas jet. However no previous theories have been postulated as to the cause of the striations.
Thus in order to analyse the mechanism of striation formation a series of laser cuts was produced, using a power of 300W each striation ($\lambda$= wavelength) was measured. From this the striation frequency ($f$) was calculated, where $f = v/\lambda$, and the results are given in Table 5.2.

From Table 5.2 it can be seen that the striation frequency is fairly constant at between 312-341 Hz. This frequency corresponds closely with the laser power ripple frequency of 320 Hz, as described in Chapter 4. Thus it seems probable that the vertical striations produced during inert gas assisted laser cutting are caused by a power ripple within the laser and occur most prominently within the directly melted zone of the cut.

The Effects Of Variation In Material Thickness On Cut Quality.

The Coherent Everlase was used to laser cut 2.7 mm thick Ti-6Al-4V alloy specimens. However cutting performance was found to be very poor. Although cuts could be produced the maximum cutting speed was only 0.7 mms$^{-1}$, due to the rather low power of 300 W available. Fig 5.18 also shows clearly the poor surface finish attainable, although even at such a low cutting speed Auger analysis indicated that the cut edge to be relatively free of oxygen contamination. Higher powers, and thus higher cutting speeds, could only be attained by using the 2 KW Control laser. The mode of this laser was so inferior (ie. not properly Gaussian), as shown in Fig 4.7 that high quality laser cutting was not possible. This is because a non-Gaussian beam increases the focussed spot size and therefore decreases the power density, for a given laser power. This has the effect of increasing both the kerf width and the HAZ width, and decreasing the maximum cutting speed. Also if, as in this case, the beam profile is non-axisymetric, cutting in different directions will produce
different cut finishes, which frustrates any detailed investigation. It was therefore decided to restrict cutting studies to the lower power superior mode machine.

Comparison Between Guillotined Edge And Laser Cut Edge Surface Roughness.

In order to compare the surface roughness of laser cuts and guillotined edges, surface profiles were taken of the smoothest laser cut, produced using a cutting speed of 25 mms⁻¹ and a gas pressure $P_h$ of 138 KPa, and of a guillotined edge. Fig 5.19 shows 6 surface profiles. 1, 2 and 3 correspond to profiles taken 0.25, 0.5 and 0.75 mm from the top edge of the laser cut, and profiles 4, 5 and 6 correspond to profiles taken 0.25, 0.5 and 0.75 mm from the top surface of the guillotined cut. It can be seen from Fig 5.19 that the laser cut is much smoother, all the way across the cut edge, than the guillotined cut.

Thus it can be concluded that laser cutting using the correct parameters can produce cuts with smoother surface finishes than conventional guillotining.

(c) Analysis Of Dross Adhesion To The Underside Of The Cuts.

When a metal is laser cut a melt is first produced, which is then sheared by the gas jet. The molten metal is then ejected from the lower edge of the cut zone. However due to the high surface tension of the ejected material some of the metal adheres to the bottom edge of the cut, and is known as dross. This dross is undesirable since it requires a subsequent machining operation to remove it, before further processing can be carried out.
Arata et al (10) reported in their cutting experiments that all the dross was deposited on to one side of the cut. However in this cutting programme it was found that the dross was deposited alternately on each side of the cut, as shown in Fig 5.10. This was due to the laser beam and the argon gas jet being aligned exactly coaxially, so that dynamic equilibrium was set up.

The Effect Of Cutting Speed On The Volume Of Dross Deposited Underneath The Cut.

In order to measure the volume of dross deposited during laser cutting a series of cuts produced using speeds of between 15 and 27.5 mms⁻¹ were weighed. The dross was then mechanically removed and the cuts weighed again, so that the dross volume could be calculated (a value of 4420 Kg m⁻³ was used as the density of Ti-6Al-4V). Fig 5.20 shows the variation in deposited dross volume with cutting speed, from which it can be seen that a greater volume of dross is deposited at higher cutting speeds. It could have been expected that at higher speeds less material would have been deposited, since the kerf width falls with increasing cutting speed, reflecting a drop in total melt volume. The results which are at variance with this simple analysis can be explained by referring to Fig 5.6, which shows that as the cutting speed increases the distance between the cutting nozzle and the exit zone increases leading to less efficient dross ejection. As a result of this a greater proportion of the dross is left adhered to the bottom of the cut.

(d) Kerf Width Analysis.

When titanium alloys are laser cut the laser beam melts an area of material, which is then blown away by an argon gas jet. The width of the gap left by the ejected material is known
as the kerf width. During laser cutting it is desirable to have the minimum kerf width possible so that smallest amount of material is lost and the minimum amount of energy wasted.

Kerf width measurements were carried out by using a calibrated graticule on the Reichert MeF 3 microscope. The upper kerf widths were measured as a function of speed, head pressure \( (P_h) \), dross jet-material distance \( (Y_{dm}) \) and head stand off \( (Y_h) \). The kerf widths on the bottom side of the cut could not be measured accurately due to dross adherence to one, or both sides of the cut.

The Effect Of Cutting Speed On Kerf Width.

Fig 5.21 shows the variation in kerf width with speed using a head stand off of 1.07 mm and a head gas pressure of 138 KPa producing a gas pressure at the cut of 130 KPa. The results show a decrease in kerf width with increasing cutting speed. Cut profiles produced at 3 different speeds are shown in Fig 5.22 illustrating this trend. These results are in agreement with a simple heat balance equation developed by Kamalu(57) which allows calculation of the process efficiency \( (n) \).

\[
n = \frac{vwtp \left( C_p T + L_f + L_t \right)}{P(1-r)} \quad (5.1)
\]

where

- \( n \) = Process efficiency.
- \( P \) = Incident laser power = 300W
- \( v \) = Cutting speed \( (\text{ms}^{-1}) \)
- \( t \) = Material thickness = 1 mm
- \( w \) = Kerf width
- \( p \) = Density = 4420 Kg/m\(^3\) \((58)\)
- \( r \) = Reflectivity
\( L_f = \text{Latent heat of fusion} = 4.36 \times 10^5 \text{JKg}^{-1} \) (8)
\( L_t = \text{Latent heat of transformation} = 90000 \text{JKg}^{-1} \) (8)
\( C_p = \text{Specific heat capacity} = 700 \text{JKg}^{-1} \text{C.} \) (58)

Using equation 5.1 it is thus possible to estimate the process efficiency by substituting in the above values and also using the experimental results of kerf width and cutting speed. The following assumptions were made.

(a) All the incident radiation was absorbed by the metal.

(b) A value of half the latent heat of transformation \((L_f)\) was used, since only 50% of the material was alpha phase and thus underwent the alpha-beta transition.

(c) A value of heat capacity \((C_p)\) was used at a temperature of \(0.5T_{\text{melting}}\).

Fig 5.23 shows the variation in process efficiency with cutting speed for laser cuts produced using a head pressure \((P_h)\) of 130 KPa. It can be seen that the process efficiency increases in a linear manner with increasing cutting speed. The major factors affecting this improvement in efficiency are as follows.

At low cutting speeds the beam is above the melt which is periodically ejected by the gas jet. Immediately after ejection the beam is free to travel through the recently formed cut and is therefore wasted to the cutting process until its movement across the sheet completely blocks this through passage. At very low cutting speeds it was possible to observe continuous leakage of part of the beam through the cut zone. This effect is progressively frustrated as the cutting velocity is increased in much the same manner as demonstrated in Fig 5.6. In addition the laser material interaction time decreases with increasing cutting speed and therefore there are greater losses due to
conduction to the HAZ with decreasing cutting speed. Finally the gas jet-material interaction time decreases with increasing cutting speed so that convective heat losses due to the gas jet are lower at higher cutting speeds.

The use of equation 5.1 to measure process efficiency however is not a true reflection of the actual efficiency of the process. This is because during laser cutting it is desirable to produce the minimum kerf width, and thus remove the minimum amount of material. Kamalu's approach however results in an increased process efficiency with both increasing cutting speed, and increasing kerf width as it is a calculation of melting efficiency. It would be better to express the cutting efficiency as a function of the material thickness, laser power and cutting speed, where the

\[
\text{cutting efficiency} = \frac{vt}{P} \tag{5.2}
\]

where \( v \) = cutting speed, \( t \) = material thickness and \( P \) = laser power. Using equation 5.2 it is possible to calculate the maximum cutting efficiency, for the 3 thicknesses of material (Table 5.3).

Table 5.3 shows the maximum possible cutting efficiencies that can be achieved using the Coherent Everlase with its maximum power output of 300W. It can be seen that the maximum cutting efficiency falls with increasing material thickness. This fall can be explained by examining the mechanism by which inert gas assisted laser cutting takes place. Initially a melt zone is created. The depth of the melt zone increases with increasing heat input (ie a low cutting speed). However the material thickness is also an important factor, since the thicker
the material the more heat is lost by conduction. This means that a greater heat input (lower cutting speed and higher dwell time) is required in thicker material to produce the same melt zone depth, which results in a lower cutting efficiency.

Once the melt zone has been created the molten material can then flow over the solid leading edge of the cut, melting some of this material. However the maximum thickness that can be laser cut is limited by the cooling effect of the leading edge, so that if the material is sufficiently thick to cool the flowing material below the melting point then the cutting process will cease.

The Effect Of Head Gas Pressure ($P_h$) On Kerf Width.

Fig 5.24 shows the variation in kerf width with cutting speed using a head gas pressure ($P_h$) of 380 KPa and a stand off ($Y_h$) of 0.13 mm producing a cut gas pressure ($P_c$) of 370 KPa. If these results are compared with those in Fig 5.21, where a much lower gas pressure ($P_h$) was used it can be seen that the kerf width values at a particular speed are almost identical, and the kerf width also decreases with speed in the same manner. Thus the head gas pressure seems to have little effect upon kerf width, once there is a sufficient gas pressure to remove all of the molten metal produced during cutting. A higher gas pressure is important however because it results in better fluid removal.

The Relationship Between Kerf Width And Laser Spot Diameter.

It can be seen from Fig 5.24 that the kerf width varies with cutting speed from a minimum value of 235 μm, at the maximum cutting speed, to a maximum of 270 μm, at the minimum cutting
speed. The maximum value corresponds closely to the focussed beam diameter of 264 μm (as measured by the LBA in Chapter 4, with all the other kerf width values being less. This is fundamentally different to the oxygen assisted laser cutting of mild steel (59) in which the focused beam diameter approximates to the minimum kerf width available. This is because the laser beam in the case of steels acts as an oxidation/combustion initiator, whereas during inert gas laser cutting the beam is the cut propagater.

The reduction in kerf width with cutting speed can be attributed to the energy distribution across the laser beam. The laser beam has a Gaussian distribution, as shown in Fig 4.10, so that the energy density falls with distance from the centre of the beam. In addition the faster the cutting speed the shorter the interaction time of the beam with the metal, so that only the more central regions of the beam can supply enough energy to melt the metal.

(e) HAZ Width Analysis.

When a material is laser cut some of the heat provided by the laser beam is lost by conduction to the material adjacent to the cut zone. This heat can alter the metallurgical structure of the material, and this region is known as the heat affected zone. It is important to have the smallest HAZ width possible in order not to affect the mechanical properties of the material unduly.

The widths of the HAZs produced on both the upper and lower sides of the cuts were measured using a graticule on a rotascope on a Riechert MeF 3 microscope. The variation in HAZ width with travel speed, gas pressures and with the use of the dross jet could then be examined.
The Variation In HAZ Width With Cutting Speed

Cuts produced without the use of the dross jet showed that the HAZ width on the top surface of the cut was smaller than on the bottom surface in all cases. This effect is illustrated in Fig. 5.25A, and can be explained as follows. As the metal is cut dross is produced which adheres to the lower surface of the cut. This accumulation of molten metal acts as a heat source during its solidification prolonging the thermal cycle and causing a higher heat input into the metal, resulting in a wider HAZ. The widths of the HAZs on both sides of the cut were identical.

The upper and lower HAZ widths decrease with increasing speed. This is because the heat input per unit area (P/vt) where P = laser power, v = cutting velocity and t = material thickness, decreases with increasing cutting speed.

The Effect Of Gas Pressure (P_h) On HAZ Width.

A series of laser cuts was produced in 1.7 mm thick Ti-6Al-4V alloy specimens, using a constant laser power of 300 W and a cutting speed of 10 mms\(^{-1}\). The argon gas pressure (P_h), however, was varied between 69 and 276 KPa and its effect on the HAZ width studied.

Fig 5.25B shows the variation in upper HAZ width with the gas pressure (P_h) exerted on the cut. The HAZ width was found to decrease with increasing gas pressure. This could be explained by the increased cooling effect of the gas at higher pressures. In addition a higher gas pressure results in faster melt removal so that less heat is lost by conduction to the HAZ. Previous work (J0) on the laser cutting of c.p titanium has also indicated that the HAZ width decreases with increasing gas pressure, until the HAZ width is negligible.
The Effect Of Convective Cooling On HAZ Width.

When laser cutting heat is not only lost by conduction to the metal substrate and by radiation to the atmosphere, but also by convective cooling by the argon gas jet. The nearest analogy to this is that of a vertically impinging jet on a flat plate in which Gordon and Cobonq {60} reported that at the stagnation point of the jet,

\[ \text{Nu}_0 = 13. \text{Re}^{0.5} \frac{D_j}{B} \]  \hspace{1cm} (5.4)

Then using the correlations given by Polhausen (17) and Steen (18) it can be shown that

\[ \text{Nu}_0 = 13 \text{Re}^{0.5} \frac{(D_j / B)}{\text{Pr}^{0.33}} \]

\[ = \frac{h_c D_j}{K_{\text{gas}}} \]  \hspace{1cm} (5.5)

or

\[ h_c = 13 \text{Re}^{0.5} \frac{\text{Pr}^{0.33}}{K_{\text{gas}}} \frac{D_j}{B} \]  \hspace{1cm} (5.6)

where \( \text{Nu}_0 \) = Nusselt number, \( \text{Pr} \) = Prandtl number, \( \text{Re} \) = Reynolds number, \( K_{\text{gas}} \) = thermal conductivity of the gas and \( B \) = jet plate distance = 0.5 mm.

Thus by substituting in values for \( K_{\text{gas}} \) and \( B \) the convective heat transfer coefficient \( (h_c) \) can be calculated, where

\[ \text{Re} = \frac{p_g V_g D}{M} \]  \hspace{1cm} (87) \hspace{1cm} (5.6)

\[ \text{Pr} = \frac{C_g M}{K_{\text{gas}}} \]  \hspace{1cm} (87) \hspace{1cm} (5.7)
and \[ V_g = \sqrt{\frac{2(P_1 - P_2)}{P_g}} \tag{5.8} \]

where \( P_g \) = gas density = 1.784 Kg m\(^{-3}\) \( (62) \)
\( V_g \) = gas velocity ms\(^{-1}\)
\( K_{gas} \) = thermal conductivity = 162 \( \times 10^{-4} \) W/mK \( (62) \)
\( M \) = gas viscosity = 2.2 \( \times 10^{-5} \) Ns m\(^{-2}\) \( (62) \)
\( D \) = nozzle diameter = 1mm.
\( C_p \) = specific heat capacity = 523 J/Kg K \( (62) \)

Using these values the convective heat loss \( Q_c \) can be calculated using equation (5.9), assuming that the gas jet impinges upon a circular area with a diameter of 1mm.

\[ Q_c = h_c A \Delta T \tag{5.9} \]

Table 5.4 shows the variation in convective heat loss with gas pressure \( (P_h) \), and it can be seen that the greater the gas pressure, the greater the convective heat loss, which is in agreement with the observed reduction in HAZ width. These results suggest that when utilizing a laser power of 400W between 10 and 12% of the available power is lost due to convective cooling of the gas jet.

These results cannot be completely accurate since they depend upon a number of assumptions.

(a) Most importantly the model was derived for laser welding, not laser cutting, and so there is no stagnation point of the jet.
(b) The constant multiplier may require modification with varying Re.
(c) The Re and Nu both depend upon the physical properties of the gas, which vary with temperature, of which no account is taken in the calculation.
(d) The variation in \( h_c \) with radial distance from the jet axis is ignored.
(e) Frictional losses within the cutting nozzle are ignored, giving a higher impinging gas velocity.
Metallographic And Compositional Analysis Of The Heat Affected Zone (HAZ).

The metallography of laser welded Ti-6Al-4V has been studied before (1), whereas that of cut edges has not, although the structures encountered are similar.

Nearest to the cut edge where the heat input is greatest the alpha-beta structure was heated well above the alpha-beta transition temperature into the beta phase field. Slow cooling from this region would have given an acicular alpha structure within the prior beta grain boundaries. However with laser cutting the cooling rates are very high, the order of $10^4 \text{°C} \text{s}^{-1}$, for the observed microstructure (63). This does not allow complete diffusion to occur and thus supersaturated non-equilibrium martensitic alpha phase is formed. The martensitic region produced in a typical laser cut can be clearly seen as the light region nearest the cut edge in the macrophotograph in Fig. 5.26. A higher magnification view of this region is shown in Fig. 5.27 where the characteristic martensitic plates can be seen within the prior beta grain boundaries.

Further away from the cut edge where the thermal cycle was less severe the metal was not been heated above the alpha-beta - beta transition temperature, but above the $M_s$ (martensite start temperature) as shown in Fig. 5.28. A mixed structure of primary alpha and martensitic alpha was produced, this structure is shown in Fig. 5.29. The amount of martensitic alpha present decreased further away from the cut. Near the martensitic zone there is a greater proportion of martensitic alpha, whereas near the parent metal there is a proponderance of primary alpha. This is the furthest extent of the HAZ, and beyond this point the parent metal structure of primary alpha and metastable beta is reached.
This pattern of microstructures was repeated in all of the cuts produced, although the extent of the purely martensitic zone decreased with speed until at speeds above 15 mms\(^{-1}\) the martensitic zone was absent and the mixed primary + martensitic structure extended up to the edge of the cut. This can be seen in Fig. 5.30. It was caused by the faster cutting speed giving a lower heat input into the metal adjacent to the cut edge, resulting in even the metal closest to the cut edge not being heated above the alpha+beta - beta transition temperature, so that a completely martensitic structure could be formed.

When very low cutting speeds, of around 10 mms\(^{-1}\) were used in conjunction with a low assist gas pressure, a very wide melt zone, and a correspondingly large volume of dross was produced and left clinging to the bottom edge of the cut. The very large heat input from the residual dross results in a very slow cooling rate adjacent to the bottom part of the cut, producing a basketweave beta structure, which can be seen as a dark region next to bottom of the cut edge in Fig. 5.31. This structure consists of alpha platelets with interleaved beta platelets occurring in colonies in a Widmanstatten structure, as shown in Fig. 5.32.

**Microhardness Examination**

Microhardness traverses were carried out on several of the laser cut samples, using a Reichert MeF 2 microhardness tester with a load of 60 g. Fig. 5.33 shows a microhardness traverse carried out on a laser cut produced using a cutting speed of 15 mms\(^{-1}\). It can be clearly seen that there is a region closest to the cut edge which has microhardness values of around 400 Kg mm\(^{-2}\). This value is much greater than that of the parent metal which is 350 Kg mm\(^{-2}\), because the region closest to the cut edge is martensitic in structure, caused by rapid cooling after cutting. The embrittlement could not have been caused by oxygen
contamination since Auger analysis indicated that oxygen only diffuses into the cut edges to a depth of approximately 200 nm.

A microhardness traverse carried out on a laser cut produced at a cutting speed of 25 mms$^{-1}$, is shown in Fig. 5.34. In this case the width of the embrittled zone is much less than in Fig. 5.33 because the lower heat input per unit length of cut, produced a narrower HAZ. The embrittled zone in this case also had lower microhardness values, around 400 Kg mm$^{-2}$ since the microstructure was a mixture of primary and martensitic alpha right up to the cut edge.

Thus it can be concluded that laser cuts produced with a mixed alpha and martensitic alpha microstructure had the least embrittlement close to the cut edge.

Chemical Analysis Of Laser Cut Edges.

Oxygen contamination is the most serious problem to be overcome when both welding and cutting titanium and its alloys. The effect of oxygen contamination on the mechanical properties of titanium has been studied extensively (8) and the results of this work are given in the literature survey. In general increasing levels of oxygen in titanium result in reduced UTS, fatigue life, 0.2% proof stress and increased hardness.

The mechanism by which oxygen, and indeed nitrogen, cause embrittlement in titanium alloys is known as interstitial solid solution hardening. Basically as the metal heats up, and melts under the laser spot it will tend to absorb oxygen according to Sievert's Law (64), which is given by

$$S = K p_g^{0.5}$$

where $S$ = gas solubility and $p_g$ = partial pressure of the gas. $K$ = constant.
In the case of laser cutting titanium alloys oxygen is absorbed into the melt, before that melt is removed by the inert gas jet. However the inert gas jet does not remove all of the molten material from the cut zone, and a small amount of the melt is left adhered to the sides of the laser cut. If proper care is not taken with gas shielding then this material can become contaminated with oxygen. The oxygen is then able to diffuse into the adjacent HAZ, even although it has not been melted.

Any oxygen absorbed interstitially by titanium alloys alters the lattice parameter, by increasing the c parameter and leaving the a parameter virtually constant. In this way the axial ratio increases approaching the ideal value for h.c.p materials. The effect of oxygen contamination upon the lattice parameters has been extensively studied (88) and the effect is clearly shown in Fig. 5.35. The increase in the c lattice parameter causes a strain (e) to be introduced, causing elastic distortion and thus strengthening of the material. The strain (e) is given by

\[
\frac{1}{a} \frac{da}{dc}
\]

where \( a = a \) lattice parameter and \( c = c \) lattice parameter.

Therefore to avoid any loss of mechanical properties during cutting oxygen contamination must be prevented by careful inert gas shielding, by using an argon jet coaxial with the laser beam. The analysis of oxygen contamination during cutting was carried out by Auger analysis 0.5 mm down from the top surface of the cut edge.

In order to assess how great the problem of oxygen contamination is during the laser cutting of titanium alloys the oxygen gradient in the guillotined sample was measured and used as the control sample with which laser cut edges could then be
compared. The oxygen gradient in a guillotined Ti-6Al-4V alloy edge is shown in Fig. 5.36. This shows that even at room temperature oxygen diffuses into the metal with the level of oxygen decreasing further away from the surface and reaching the bulk level of 0.2 wt% at a depth of 80 nm. This pick up of oxygen could be caused by absorption from the atmosphere during guillotining, where local heating could be high.

The Variation In Oxygen Contamination With Head Gas Pressure ($P_h$).

Oxygen contamination was found to decrease with increasing head pressure ($P_h$). Figs 5.37 and 5.38 show the oxygen gradients for 1mm thick Ti-6Al-4V alloy laser cut using gas assist pressures ($P_h$) of 110 and 380 KPa respectively, at a speed of 25 mms$^{-1}$. Fig. 5.37 shows extensive oxygen contamination with over 9 wt% oxygen being present 20 nm from the surface and the bulk oxygen level being reached at a depth of 700 nm. The improvement shown in Fig. 5.38 is dramatic, with the level at 20 nm being 1.6 wt% and the bulk oxygen level being reached at a depth of 120 nm. This much lower level of oxygen contamination is due simply to the increase in shielding gas (argon) pressure giving a greater exclusion of oxygen around the cut zone. However all the cuts produced without using the dross jet were found to have much higher levels of oxygen contamination than the mechanically cropped specimen.

Summary And Introduction To Laser Cutting Using The Dross Jet.

The aim of the previous sections has been to produce dross free cuts and optimize the laser cutting parameters with respect to both surface quality and oxygen contamination. It is apparent that it is impossible to produce dross free laser cut Ti-6Al-4V under these conditions. A device known as the dross
jet (described in Chapter 3, page 60), which blows all the
dross on to one side of the cut leaving one side dross free, was
used during the remainder of the cutting programme.

It was also apparent from the previous section that
oxygen contamination is a substantial problem during the laser
cutting of titanium alloys. Even when fast cutting speeds are
used in conjunction with high shielding gas pressures the level
of oxygen contamination in laser cut edges is greater than that
present in a guillotined edge. Although the dross jet was used
primarily to produce dross free cuts a secondary function was to
provide gas shielding from the underside of the cut zone. The
effect of using the dross jet upon oxygen contamination is
examined in the next section.

5.2.2. Laser Cutting Using The Dross Jet.

When a metal is laser cut using inert gas to blow away
the melt, some of the ejected metal sticks to the underside of
the cut, and is a particular problem when laser cutting titanium
alloys, due to the high surface tension of liquid titanium (60). The
removal of this dross during the cutting process this would
allow further fabrication without the need for an intermediate
machining process.

Previous attempts to remove this dross by the tandem
nozzle cutting method (11.) have only been partially successful,
and so for this series of experiments a new method of dross
removal was devised. In order to achieve dross free cuts a
dross jet was used. The dross jet was positioned directly
underneath the cut and argon gas was blown through one of the
eight nozzles perpendicular to the cutting direction. The effect
of this was to blow all of the molten metal produced during
cutting on to one side of the cut leaving the other side dross
free. This is a satisfactory condition in most practical applications where only one clean side of the cut edge is needed, on the component side of the cut.

This technique of blowing the dross on to one side of the cut has also been successfully used during the cutting of stainless steel. In this case the cutting gas used was oxygen, and so compressed air was used instead of argon to blow the dross to one side of the cut (65). A photograph of the dross jet can be seen in Fig. 3.7.

**The Effect Of The Dross Jet-Material Distance ($Y_{dm}$) And Dross Jet Gas Pressure ($P_d$) On Surface Roughness.**

Figure 5.39 shows the variation in $R_a$ with dross jet-material distance ($Y_{dm}$), in 1mm thick Ti-6Al-4V sheet specimens, where positions 1, 2 and 3 correspond to distances of 0.25, 0.5 and 0.75 mm from the top edge of the cut. It is apparent that as $Y_{dm}$ increases the cut surface becomes rougher. However this increase is most noticeable at position 3, which is closest to the bottom edge of the cut. One possible explanation for this effect is that the dross jet blowing argon onto the underside of the cut zone increases turbulence during solidification of the melt resulting in a rougher surface.

Fig. 5.40 shows the variation in $R_a$ with dross jet-material distance ($Y_{dm}$) for a gas pressure ($P_d$) of 620 KPa. If this is compared with Fig. 5.39 in which a gas pressure ($P_d$) of 414 KPa was used, it can be seen that both figures show a small increase in $R_a$ values with dross jet-material distance ($Y_{dm}$). However the $R_a$ values in all positions across the cut were higher when a gas pressure of ($P_h$) 620 KPa was used, probably as a result of increased turbulence caused by the intersection of the
jet from the laser head, and the dross jet. Figs 5.41 and 5.42 produced with dross jet gas pressures (P_d) of 414 and 620 KPa respectively and dross jet-material distances (Y_dm) of zero illustrate the increase in surface roughness with dross jet gas pressure.

Conversely a reduction in dross jet gas pressure (Y_d = 193 KPa), combined with a lower head gas pressure (P_h = 138 KPa), results in a much smoother cut surface, as illustrated in Fig. 5.43. The smoothest cut produced using the dross jet is shown in Fig. 5.44. This cut was produced using a dross jet pressure (P_d) of 193 KPa and a head pressure (P_h) of 138 KPa. This cut shows that there has been complete molten metal removal during cutting giving a very smooth dross free cut, with horizontal striations visible all the way across its surface. If Fig. 5.44 is compared with Fig. 5.15 which shows a cut produced under identical conditions, but without the use of the dross jet, it can be seen that the positions of the striations are identical. The only difference between the cuts is dross adhesion in Fig. 5.15.

It is also apparent from Figs. 5.39, 5.40 and 5.43 that the smoothest laser cuts are always produced when the dross jet-material distance (Y_d) is zero.

The Effect Of Using The Dross Jet On Dross Deposition Side.

The side of the cut on to which the dross was blown was also found to change when the dross jet-material distance (Y_dm) was increased above 7 mm. Fig. 5.45 shows the cuts produced using dross jet material distances of (Y_dm) 3, 7 and 9 mm. The dross appears firstly on the left hand side of the cut, then on both sides and finally on the right hand side of the cut. This change over in dross deposition is due to the inclination of the dross jet nozzles and can be explained as follows. When the
dross jet-material distance ($Y_{dm}$) is less than 7 mm a high pressure zone is formed on the right hand side of the cut blowing the dross on to the left hand side of the cut. However increasing the dross jet-material distance ($Y_{dm}$) above 8 mm forms a high pressure zone on the left hand side of the cut blowing the dross on to the right hand side of the cut. This was proved by a simple experiment which involved setting up a pitot tube with a pressure gauge attached, to the X-Y table, where the cut had been positioned. The X-Y table was then traversed in a perpendicular direction to the cutting direction at different dross jet-material distances ($Y_{dm}$). The results of this experiment are shown in Fig. 5.46. The position of the centre of the cut was found to occur in between the peak pressures for dross jet-material distances ($Y_{dm}$) of 7 and 8 mm. Fig. 5.46 also shows how the peak pressure exerted by the dross jet decreases with increasing dross jet-material distance ($Y_{dm}$). The pressures shown in Fig. 5.46 are only the vertical component of the dross jet gas pressure ($P_d$).

The Effect Of Using The Dross Jet On Kerf Width.

Figs 5.47, 5.48 and 5.49 show the variation in kerf width with dross jet-material distance ($Y_{dm}$) for three different sets of conditions. It can be seen that the kerf widths did not vary with increasing dross jet-material distance ($Y_{dm}$).

The Effect Of The Use Of The Dross Jet On HAZ Width.

The use of the dross jet was found to alter the HAZ widths dramatically. A cross section through a cut produced without using the dross jet (Fig. 5.50) shows that the HAZ widths on both sides of the cut are the same for both top and bottom surfaces. The cross section shown does not show dross on both sides of the cut because the dross formed in a pattern where
individual globules alternated from side to side of the cut, as shown in Fig. 5.10. An identical cut produced with the use of the dross jet is shown in Fig. 5.51. From this photograph it can be seen that although the upper HAZ widths are the same, the lower HAZ widths on either side of the cut are different. On the dross free side the upper and lower HAZ widths are identical unlike the situation in Fig. 5.50 where no dross jet was used. This is because there was no dross adhesion to the lower edge of this side of the cut, so that the cooling rate for both the top and bottom edges of the cut were the same. The side on to which the dross was blown (Fig. 5.51) shows a much larger HAZ width than either the dross free side of the cut, or the cut produced with no dross jet. The larger HAZ width was due to all of the dross normally deposited on to two sides of the cut, being deposited on to just one. This acts as a large heat input into the lower edge of the cut, producing a correspondingly larger HAZ width. It is apparent from Figs. 5.50 and 5.51 that the dross particles are martensitic in nature, which reflects the rapid cooling rate occurring.

The Effect Of Using The Dross Jet On Oxygen Contamination.

Oxygen has a very serious effect on the mechanical properties of titanium alloys, as discussed in the literature survey, and so it is important to shield the cut from oxygen contamination not only from the top by the coaxial argon jet, but from the bottom by using the dross jet. The effect of varying dross jet gas pressure ($P_d$), and varying dross jet-material distance ($Y_{dm}$) on oxygen contamination was studied.
The Effect Of The Variation In Dross Jet Pressure \( (P_d) \) On Oxygen Contamination.

The use of the dross jet on the underside of the cuts was found to not only blow all the dross to one side of the cut but also to shield the underside of the cut from oxygen contamination. Fig. 5.52 shows the oxygen gradient for the cut edge of a 1 mm thick sample cut with a head pressure of 380 KPa, at a stand off of 0.13 mm, producing a pressure of 370 KPa on the cut. The dross jet pressure used was 414 KPa. These results show a large improvement over cuts produced without the use of the dross jet, where the bulk oxygen level was reached at a depth of 40 nm, which is the same as that in the guillotined sample in Fig. 5.36.

An increase in dross jet gas pressure \( (P_d) \) to 620 KPa was found to decrease the level of oxygen contamination only slightly. Fig. 5.53 shows that the bulk level is again reached at a depth of 40 nm, but there is a decrease in oxygen levels nearer the cut surface from 2.1 wt%, when a gas pressure \( (P_h) \) of 414 KPa was used, to 1.8 wt% when 620 KPa was used. The oxygen profile shown in Fig. 5.53 is in fact identical to that of the guillotined sample.

The use of both a lower head pressure \( (P_h) \) of 138 KPa and a lower dross jet pressure \( (P_d) \) of 193 KPa, with a stand off \( (Y_h) \) of 0.13 mm producing a cut gas pressure \( (P_c) \) of 140 KPa was found to increase the level of oxygen contamination. This can be seen in Fig. 5.54 where although the level of oxygen contamination is less than in cuts without the use of the dross jet, it is more than in cuts produced using higher dross jet and head pressures.
The Effect Of Dross Jet Material Distance On Oxygen Contamination.

An increase in dross jet material distance ($Y_{dm}$) was found to increase oxygen contamination in the cuts. Fig. 5.55 shows the oxygen gradient in the dross free side of a cut produced with a dross jet material distance ($Y_{dm}$) of 6 mm. If this cut is compared to that in Fig. 5.53 produced under identical conditions, with the exception that a dross jet-material distance ($Y_{dm}$) of virtually zero was used, the level of oxygen contamination was found to be slightly greater. This effect was caused by the larger dross jet-material distance ($Y_{dm}$) allowing more oxygen incorporation into the jet. Also the larger the dross jet-material distance ($Y_{dm}$) the smaller the pressure exerted on the cut as shown in Fig. 5.2.

Summary of the effects of using the dross jet.

The initial aim was to optimize the dross jet parameters with respect to cut surface roughness, HAZ width, dross deposition and oxygen contamination. It was found that although the use of a dross jet had considerable effect, variations in dross jet parameters had virtually no effect upon kerf width and only a limited effect upon HAZ width. A dross free cut edge could be produced using many combinations of dross jet-material distances and dross jet gas pressures. In addition there was found to be an increase in cut surface roughness close to the bottom edge of the cut, which was probably caused by increased turbulence. However the most dramatic effect of using the dross jet was to provide a secondary shielding device on the underside of the cut zone. This work has shown that providing that the laser and dross jet parameters are optimized dross free laser cuts with levels of oxygen contamination equivalent to those in a guillotined edge could be produced.
5.3. Conclusions

A number of conclusions can be drawn from the work undertaken during the laser cutting program.

**Conventional Laser Cutting Conclusions.**

(a) The smoothest cuts were produced using speeds of between 25 and 30 mms\(^{-1}\), and a gas pressure \((P_c)\) of 138 KPa. Increasing the cutting speed increased the surface roughness of the cuts. An increase in gas pressure also increased the surface roughness of the cuts.

(b) The roughness of the cut was found not to be uniform across the cut surface, the smoothest region being nearest the top surface and the roughest nearest the bottom surface.

(c) The kerf widths of the cuts decreased with increasing cutting speed.

(d) Gas pressure was found to have no effect on the kerf widths produced.

(e) There was some oxygen contamination when cuts were produced using gas shielding from above only. The extent of oxygen contamination decreased with increasing gas pressure but there was still a much larger level of oxygen than in the guillotined sample.

(f) Cuts produced using speeds below 15 mms\(^{-1}\) had an HAZ consisting of a martensitic region close to the cut edge with a mixed primary alpha and martensitic alpha structure beyond this. Above this speed there was little or no martensitic region formed and the HAZ was totally a mixed primary alpha + martensitic alpha structure.
(g) Laser cuts produced with mixed primary alpha and martensitic alpha microstructures had microhardness values of around 400 Kgmm\(^{-2}\), close to the cut edge. Whereas those with totally martensitic regions close to the cut edge had values of over 420 Kgmm\(^{-2}\).

(h) Increasing argon gas pressure caused decreasing HAZ width.

(i) Laser cuts could be produced in up to 2.7 mm thick Ti-6Al-4V, with a laser power of 300 W.

(j) The vertical striations observed during the laser cutting of Ti-6Al-4V are possibly a result of the power ripple within the laser power supply.

(k) The horizontal striations observed were possibly caused as a result of the fanning out effect of the gas jet.

**Laser Cutting (With The Dross Jet) Conclusions.**

(a) Cuts with oxygen levels identical to those in the guillotined cut were produced when the dross jet was used from below the cut.

(b) The use of the dross jet was found to produce dross free cuts and also have little effect on the roughness of the cuts produced.

(c) The use of the dross jet was found to decrease the HAZ width on the dross free side of the cut and increase the HAZ width on the other side.
Table 5.1. The variation in dross particle separation with cutting speed. \( t = 1 \text{mm}, p_h = 138 \text{ kPa} \).

<table>
<thead>
<tr>
<th>Cutting Speed (mms(^{-1}))</th>
<th>Dross Particle Separation (mm)</th>
</tr>
</thead>
<tbody>
<tr>
<td>15</td>
<td>0.28</td>
</tr>
<tr>
<td>20</td>
<td>0.31</td>
</tr>
<tr>
<td>25</td>
<td>0.29</td>
</tr>
<tr>
<td>30</td>
<td>0.34</td>
</tr>
<tr>
<td>32.5</td>
<td>0.36</td>
</tr>
<tr>
<td>35</td>
<td>0.29</td>
</tr>
<tr>
<td>37.5</td>
<td>0.33</td>
</tr>
</tbody>
</table>

Table 5.2. Striation frequencies in laser cut Ti-6Al-4V. \( t = 1 \text{mm}, p_h = 138 \text{ kPa} \).

<table>
<thead>
<tr>
<th>Cutting Speed (mms(^{-1}))</th>
<th>Wavelength (mm)</th>
<th>Frequency (Hz)</th>
</tr>
</thead>
<tbody>
<tr>
<td>15</td>
<td>0.0375</td>
<td>341</td>
</tr>
<tr>
<td>20</td>
<td>0.0594</td>
<td>323</td>
</tr>
<tr>
<td>25</td>
<td>0.0801</td>
<td>312</td>
</tr>
<tr>
<td>32.5</td>
<td>0.1026</td>
<td>314</td>
</tr>
<tr>
<td>37.5</td>
<td>0.1130</td>
<td>332</td>
</tr>
</tbody>
</table>
Table 5.3. Maximum cutting efficiencies.

<table>
<thead>
<tr>
<th>Material Thickness (mm)</th>
<th>Cutting Speed (mms$^{-1}$)</th>
<th>Power (W)</th>
<th>Cutting Efficiency (mm$^2$/J)</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>37.5</td>
<td>300</td>
<td>12.5</td>
</tr>
<tr>
<td>1.7</td>
<td>12</td>
<td>300</td>
<td>6.8</td>
</tr>
<tr>
<td>2.7</td>
<td>1</td>
<td>300</td>
<td>0.9</td>
</tr>
</tbody>
</table>

Table 5.4. The variation in convective heat loss $Q_c$ with increasing gas pressure ($P_h$).

<table>
<thead>
<tr>
<th>Gas Pressure (KPa)</th>
<th>Gas Velocity (ms$^{-1}$)</th>
<th>$Re \times 10^5$</th>
<th>$Pr$</th>
<th>$h_C$ (Wm$^{-1}$K)</th>
<th>$Q_C$ (W)</th>
</tr>
</thead>
<tbody>
<tr>
<td>241</td>
<td>520</td>
<td>0.421</td>
<td>0.677</td>
<td>38593</td>
<td>50.2</td>
</tr>
<tr>
<td>207</td>
<td>482</td>
<td>0.389</td>
<td>0.677</td>
<td>37097</td>
<td>48.3</td>
</tr>
<tr>
<td>172</td>
<td>439</td>
<td>0.355</td>
<td>0.677</td>
<td>35439</td>
<td>46.1</td>
</tr>
<tr>
<td>138</td>
<td>393</td>
<td>0.318</td>
<td>0.677</td>
<td>33541</td>
<td>43.6</td>
</tr>
<tr>
<td>103</td>
<td>340</td>
<td>0.275</td>
<td>0.677</td>
<td>31191</td>
<td>40.6</td>
</tr>
</tbody>
</table>
Fig. 5.1. Experimental arrangement for laser cutting.

Fig. 5.2. The relationship between laser head stand off, head gas pressure and the gas pressure at the cut.
Fig. 5.3. Schematic representation of the laser cutting process.
Fig. 5.4. Sectioning procedure for cut front angle ($\alpha$) determination.

Fig. 5.5. SEM of a sectioned laser cut where ($\alpha$) is the cutting front angle.
Fig. 5.6. Schematic representation of the cut front showing what depth of the cut front is irradiated by the beam.
Fig. 5.7. SEM photographs of laser cut edges produced using cutting speeds of between 15 and 37.5 mms$^{-1}$.
Fig. 5.8. Schematic representation of the inert gas assisted laser cutting mechanism.
Fig. 5.9. Schematic representation of "step formation" in inert gas assisted laser cutting.

Fig. 5.10. Dross deposition onto the underside of a cut edge.
Fig. 5.11. The variation in $R_a$ with cutting speed using a head stand off of 0.13 mm and a head gas pressure of 380 KPa. ($t=1\text{mm}$)

Fig. 5.12. Schematic diagram showing how the increase in cutting speed affects the surface roughness of the cut.
Fig. 5.13. SEM photograph of a laser cut edge produced using a cutting speed of 35 mm/s and a head gas pressure of 380 KPa. (x60)

Fig. 5.14. The variation in $R_a$ with cutting speed using a head stand off of 0.13 mm and a head gas pressure of 138 KPa.
Fig. 5.15. SEM photograph of a laser cut edge produced using a cutting speed of 35 mm s⁻¹ and a head gas pressure of 138 KPa. (x60)

Fig. 5.16. Schematic representation of horizontal striation formation in inert gas assisted laser cutting.
Fig. 5.17. Schematic diagram showing the formation of striations during the oxygen assisted laser cutting of mild steel.

Fig. 5.18. Cut edge quality when cutting 2.7 mm thick material.
Fig. 5.19. Comparison of surface profiles of laser cut and guillotined edges. $t = 1 \text{mm}$, $v = 25 \text{mms}^{-1}$, $p_h = 138 \text{kPa}$.

Fig. 5.20. The variation in adherent dross volume with cutting speed. $t = 1 \text{mm}$, $p_h = 138 \text{kPa}$. 
Fig. 5.21. The variation of kerf width with speed using a head stand off of 1.07 mm and a head gas pressure of 138 KPa.

Fig. 5.22. Cut profiles produced using cutting speeds of 10, 25 and 40 mms$^{-1}$ (from left to right). (x50)
Fig. 5.23. The variation in process efficiency with cutting speed.

Fig. 5.24. The variation of kerf width with speed using a head stand off of 0.13 mm and a head gas pressure of 380 KPa.
Fig. 5.25A. The variation in upper and lower HAZ widths with speed using a head gas pressure of 110 KPa.

Fig. 5.25B. The variation in upper HAZ width with gas pressure using a cutting speed of 25 mms⁻¹.
Fig. 5.26. Macrostructure of a laser cut edge produced using a cutting speed of 20 mms$^{-1}$ and a head gas pressure of 110 KPa. (x70)

Fig. 5.27. The martensitic zone (x700).
Fig. 5.28. The titanium-vanadium phase diagram.

Fig. 5.29. The primary alpha+martensitic alpha zone (x700).
Fig. 5.30. Macrophotograph of a laser cut edge produced using a cutting speed of 25 mms$^{-1}$ and a head gas pressure of 380 KPa. (x700)

Fig. 5.31. Macrophotograph of a laser cut edge produced using a cutting speed of 10 mms$^{-1}$ and a head gas pressure of 380 KPa. (x70)
Fig. 5.32. The basketweave zone. (x550)

Fig. 5.33. Microhardness traverse carried out 0.5 mm from the top of a laser cut produced using a cutting speed of 15 mms⁻¹.
Fig. 5.34. Microhardness traverse carried out 0.5 mm from the top of a laser cut produced using a cutting speed of 25 mms\(^{-1}\).

Fig. 5.35. The variation in lattice parameter with oxygen contamination.
Fig. 5.36. The oxygen gradient in a conventionally guillotined sample.

Fig. 5.37. Oxygen gradient in a laser cut produced using a head gas pressure of 110 KPa and a cutting speed of 25 mms⁻¹.
Fig. 5.38. Oxygen gradient in a laser cut produced using a head gas pressure of 140 KPa and a cutting speed of 25 mms⁻¹.

Fig. 5.39. The variation in $R_a$ with dross jet-material distance using a cutting speed of 25 mms⁻¹, stand off of 0.13 mm, a head gas pressure of 380 KPa and a dross jet gas pressure of 414 KPa.
Fig. 5.40. The variation in $R_a$ with dross jet-material distance using a cutting speed of 25 mms$^{-1}$, a head stand off of 0.13 mm and a head gas pressure of 380 KPa, with a dross jet gas pressure of 620 KPa.

Fig. 5.41. SEM photograph of a laser cut edge produced using a cutting speed of 25 mms$^{-1}$, a head gas pressure of 380 KPa and a dross jet gas pressure of 414 KPa. (x60)
Fig. 5.42. SEM photograph of a laser cut edge produced using a cutting speed of 25 mms⁻¹, a head gas pressure of 380 KPa and a dross jet gas pressure of 620 KPa. (x60)

Fig. 5.43. The variation in $R_a$ with dross jet-material distance using a cutting speed of 25 mms⁻¹, stand off of 0.13 mm, a head gas pressure of 138 KPa and a dross jet gas pressure of 193 KPa.
Fig. 5.44. SEM photograph of a laser cut edge produced using a cutting speed of 35 mms$^{-1}$, a head gas pressure of 138 KPa and a dross jet gas pressure of 193 KPa. (x60)

Fig. 5.45. Cuts produced using dross jet - material distances of 3, 7 and 9 mm (from left to right) showing change of dross deposition side. (x15)
Fig. 5.46. The variation of gas pressure exerted on the cut with distance, head stand off and dross jet-material distance.

Fig. 5.47. The variation of kerf width with dross jet-material distance using a head gas pressure of 380KPa, a cutting speed of 25 mms$^{-1}$ and a dross jet gas pressure of 414KPa.
Fig. 5.48. The variation of kerf width with dross jet-material distance using a head gas pressure of 138 KPa, a cutting speed of 25 mms\(^{-1}\) and a dross jet gas pressure of 193 KPa.

Fig. 5.49. The variation of kerf width with dross jet-material distance using a head gas pressure of 380KPa, a cutting speed of 25 mms\(^{-1}\) and a dross jet gas pressure of 620KPa.
Fig. 5.50. Transverse section through a laser cut produced using a cutting speed of 25 mm s\(^{-1}\) and a head gas pressure of 140 KPa. (x70)

Fig. 5.51. Transverse section through a laser cut produced using a cutting speed of 25 mm s\(^{-1}\), a head gas pressure of 140 KPa and a dross jet gas pressure of 193 KPa. (x70)
Fig. 5.52. Oxygen gradient in a laser cut produced using a head gas pressure of 380 KPa and a cutting speed of 25 mms\(^{-1}\).

Fig. 5.53. Oxygen gradient in a laser cut produced using a head gas pressure of 380 KPa, a cutting speed of 25 mms\(^{-1}\) and a dross jet gas pressure of 620 KPa.
Fig. 5.54. Oxygen gradient in a laser cut produced using a head gas pressure of 140 KPa, a cutting speed of 25 \( \text{mms}^{-1} \) and a dross jet gas pressure of 193 KPa.

Fig. 5.55. Oxygen gradient in a laser cut produced using a head gas pressure of 380 KPa, a cutting speed of 25 \( \text{mms}^{-1} \) and a dross jet gas pressure of 620 KPa, with a dross jet-material distance of 6 mm.
CHAPTER 6. LASER WELDING.

At present electron beam welding is the most established method of joining titanium and its alloys. This process however suffers from the major drawback of requiring a vacuum chamber, which makes it very expensive and awkward to weld large components. The aim of this series of experiments was to study the feasibility of using CO₂ lasers to weld 1, 1.7 and 2.7 mm thick Ti-6Al-4V alloy sheet, and discover the effects of the various process parameters (welding speed, laser power, lens focal length and laser mode) on the following.

(a) Penetration depth.
(b) HAZ width.
(c) Porosity.
(d) Microhardness.
(e) Melting and joining rate efficiency.
(f) Mechanical properties.
(g) Oxygen contamination.
(h) Metallography
(i) Weld surface topography

In addition experimental data have been compared with data obtained from existing model simulations for welding processes.

6.1 Experimental Procedures.

The laser welding programme was carried out in two sections with welds being produced using both the continuous wave and the pulsed laser modes. The experimental arrangements used in each case are as follows.
(a) Continuous wave welding.

Welds were produced in the continuous wave mode using both the Coherent Everlase and the Control lasers.

The experimental arrangement utilized when welding with the Coherent laser was similar to that used during the cutting programme and shown in Fig. 5.1. However, a much lower argon jet pressure \((P_h)\) of 30 kPa was used, and instead of blowing argon from one nozzle of the dross jet, the two nozzles perpendicular to the weld were used to shield the underbead of the weld from contamination. The Coherent laser used a 63 mm focal length zinc selenide lens to focus the raw beam.

The second part of the continuous wave welding programme was carried out using the Control 2 KW laser. A schematic diagram of the experimental arrangement is shown in Fig. 6.1. In this case shielding was achieved by the use of a specially designed welding jig and nozzle, described in chapter 3, page 61. The Control laser was fitted with both 100 and 150 mm focal length potassium chloride lenses during the welding programme.

The samples were carefully degreased with acetone, before welding. Welds were produced in 1, 1.7 and 2.7 mm thick Ti-6Al-4V alloy sheet, using laser powers of between 300 and 2000 W, and welding speeds of between 2 and 120 mms\(^{-1}\). The effect of variation in welding speed and laser power on HAZ width, porosity, oxygen contamination, microhardness, melting efficiency and mechanical properties were studied.

(b) Pulsed mode welding.

The pulsed welding programme was carried out using the Coherent Everlase with the same experimental arrangement as that used
during the continuous wave programme. Welds were produced in 1mm thick Ti-6Al-4V using an average power output of 300W. Special attention was paid to the pulse parameters and their effect upon the weld bead morphology.

6.2 Results And Discussion.

6.2.1 Continuous Wave Mode Welding.

(a) Weld Bead Penetration And Profile Analysis.

The aims of this section is to examine the influences of the process parameters upon the condition of the finished weld and to give an improved understanding of the mechanisms of laser welding. Various aspects of the laser-material interactions could then be identified and explained. Of particular interest is the influence of an effect known as "keyholing" on the weld bead morphology. Laser melting can be divided into two distinct regimes.

(a) conduction limited welding; and (b) "keyhole" welding

Conduction limited welding is so called because the laser acts as a point source of energy on the surface of a molten pool and growth of that pool into the depth of the material is achieved by conductive and convective heating. Welds created by this interaction have a distinctive cross section which is approximately semi circular.

At very high incident power densities (approximately \(10^9\) Wcm\(^{-2}\)) the laser melting is carried out by a different mechanism than that of the simple conduction limited model outlined above. A melt is established which has a very severe thermal gradient from its centre (at the laser material interaction point) to its edge. This thermal gradient is associated with a reciprocal
surface tension gradient which has the effect of drawing the material away from the centre. This "dimpling" further enhances beam absorption, and combined with the vapourization of material at the laser material interaction point, causes the laser to drill a hole into the metal below the melt. This cavity, or "keyhole", is kept open by the vapour pressure of the boiling metal. The laser beam thus produces a weld bead with a distinctive cross section, with an aspect ratio (width:depth) of less than one.

A schematic representation of the "keyholing" process is shown in Fig. 6.2, where the change in welding mechanism from conduction limited to "keyholed", with increasing power density, is also shown.

The Variation In Depth Of Penetration With Welding Speed Using The Coherent Laser.

Fig. 6.3 shows the variation in weld penetration and bead shape with welding speed, for welds produced in 1mm thick Ti-6Al-4V at a laser power of 300 W, using the Coherent laser. It can be seen that both the weld bead morphology and the penetration depth are dependent upon welding speed. It is apparent from Fig. 6.3 that as the welding speed decreases the depth of penetration increases as a result of the increased energy density at the laser-material interaction point, which can be expressed as,

\[
\frac{\text{Laser Power}}{\text{Welding Velocity} \times \text{Focussed Beam Diameter}}
\]

The shape of the weld beads also vary with welding speed. At welding speeds above 20mms\(^{-1}\) the weld bead morphology is typical of that encountered during "keyhole" welding. Once full penetration has been achieved, in this case using welding
speeds of below 20mms$^{-1}$, and although the "keyhole" cavity remains open weld growth can only take place by sideways conduction. As a result welds produced using processing speeds less than the minimum required for full penetration do not exhibit the characteristic "keyhole" shape and appear to be conduction limited.

The minimum aspect ratio which was achieved when laser welding 1mm thick Ti-6Al-4V using the maximum laser power of 300W was 1:0.83. This result suggests that although "keyholing" is the predominant mechanism the welds are to a certain extent conduction limited and in order to produce fully "keyholed" welds with aspect ratios of less than one a greater laser power would be required.

The Variation In Depth Of Penetration With Welding Speed Using The Control Laser.

From the results in the previous section it can be seen that although it is possible to produce full penetration welds in 1 mm thick Ti-6Al-4V using the Coherent laser they are not fully keyholed, and have an aspect ratio of 1:0.83. Therefore in order to produce welds in both thicker material, and with a smaller aspect ratio a greater laser power is required and this was achieved by using the 2 KW Control laser.

Using a laser power of 1800 W (almost the practical maximum of the Control Laser) it was possible to produce full penetration keyhole welds, with aspect ratios of 1:1.25 (1mm thick material) and 1:1.1 (1.7mm thick material), as shown in Fig. 6.4. The welding speeds used were 120 and 60 mms$^{-1}$ respectively.
The variation in weld penetration with welding speed for laser welds produced in 2.7 mm thick Ti-6Al-4V is shown in Fig. 6.5. It can be seen that the results follow a similar pattern to those achieved using the Coherent laser, with the weld penetration increasing with decreasing welding speed. This effect is a result of the increased energy density at the laser-material interaction point.

Fig. 6.6 shows the variation in weld bead shape with welding speed for laser welds produced in 2.7 mm thick Ti-6Al-4V using a laser power of 1800 W. It is apparent from Fig. 6.6 that at high welding speeds (30 mms⁻¹) the weld bead is "keyholed" (Fig. 6.6a). However as the welding speed was decreased, although the "keyhole" still remained open, the "keyhole" effect was obscured by sideways conductive melt growth. Eventually this combination of "keyholing" and conductive growth produces a full penetration weld as shown in Fig. 6.6c, with an aspect ratio of 1:0.6. In order to obtain the desirable aspect ratios of approximately 1:1.25 achieved in thinner material higher laser powers would be required.

The Variation In Depth Of Penetration With Material Thickness

Fig. 6.7 shows the variation in depth of penetration with welding speed for 1.7 and 2.7 mm Ti-6Al-4V alloy sheet. It is apparent that for a given laser power and welding speed there is a greater weld penetration depth in the thinner section material than the 2.7 mm thick sheet. This effect can be attributed to the thicker material acting as a more effective heat sink. The variation in weld penetration with material thickness is illustrated in Fig. 6.8, which shows 2 welds produced under identical conditions in 1 and 2.7 mm thick material.
The Variation In Depth Of Penetration With Laser Power.

The effect of variation in laser power on penetration depth is shown in Fig. 6.9. It can be seen that increasing laser power results in an increased weld penetration depth. This result indicates that the penetration depth is directly related to heat input which is given by laser power/welding speed. Thus the larger the laser power (or the slower the welding speed) the greater the penetration depth.

It is also apparent from Fig. 6.9 that at laser powers of less than 700W the rate of increase in penetration depth is less than at higher powers. One possible explanation for this effect is that low laser powers are insufficient to rapidly produce a melt directly underneath the beam. Instead a small proportion of the beam energy is required to preheat the spot before melting takes place resulting in lag between the leading edge of the beam and the melt. At high powers melting is almost instantainious.

The Variation In Depth Of Penetration With Lens Focal Length

The effect of using both 100 and 150 mm focal length KCl lenses on weld penetration depth was studied using the Control laser. As can be seen from Fig. 6.10, which shows the variation of depth of penetration with welding speed for the two different lenses, the depth of penetration achieved with the 100 mm focal length lens was about 25% greater than that of the 150 mm focal length lens. This effect can be attributed to the shorter focal length lens producing a smaller laser spot size. Since the power density is given by the laser power divided by the spot size, a higher heat input per unit area is attained when using shorter focal length lenses. This result also indicates that when using a shorter focal length lens more accurate alignment of the workpieces is required during butt welding.
The focussed laser spot size can be calculated by using the following expression (50).

\[ R_b = \frac{1.22 F \lambda}{D_b} \]

Where \( R_b \) = Radius of the diffraction limited spot size, 
\( D_b \) = Unfocused beam diameter, \( \lambda \) = Wavelength of the laser radiation and \( F \) = Lens focal length.

However theoretical techniques to calculate spot size are influenced spherical aberration, and the effect of diffraction on a Gaussian beam. As a result of these effects the actual focused beam diameter is always greater than the calculated value. A full account of the problems encountered in measuring beam diameters is given in Chapter 4, page 85.

Although the depth of penetration which can be obtained with a shorter focal length lens is improved, it must be mentioned that this reduction in focal length also reduces the depth of focus of the beam. This effect can give rise to alignment problems during the welding of sheets that are not perfectly flat.

For the Gaussian mode, the depth of focus (\( Z \)), is given by (83):

\[ Z = \pm \frac{4 \lambda}{\pi D} \left[ \frac{2 \lambda f}{\pi D} \right]^2 \]

where \( f \) = lens focal length, \( D \) = unfocussed beam diameter, and where \( (f/D) \) is known as the Fresnel number. Thus a smaller Fresnel number results in a greater depth of penetration, however the correspondingly smaller depth of focus means that much more accurate focussing is required. Table 6.1 shows the calculated
depth of focus (Z) for both the Control laser and Coherent laser installations. The values used for the unfocussed beam diameter were 11mm (Coherent laser) and 24mm (Control laser) as measured using the laser beam analyser in Chapter 4, page 89.

The Variation In HAZ And Melt Width With Laser Power And Welding Speed.

Fig. 6.11 shows the variation in both melt width and HAZ width with welding speed in 1.7 mm thick Ti-6Al-4V alloy sheet using a laser power of 1500W. It can be seen that there is a gradual increase in both HAZ and melt width with decreasing welding speed. However once full penetration has been achieved (at speeds of below 25 mms⁻¹) both the HAZ and melt width increase substantially. This effect can be explained by analysing the weld thermal cycle. Initially the laser spot impinges upon the workpiece and directly melts a small volume of material approximately the same diameter as the focussed laser beam. This molten material is heated to well above its melting point and heat is transferred by conduction to the adjacent metal, which then melts. In this way a molten weld pool is established. The HAZ is created by conduction of heat from the weld pool during the cooling period of the thermal cycle. Therefore as the welding speed is reduced the laser material interaction time is increased resulting in both a wider and more voluminous weld bead, which in turn produces a correspondingly wider HAZ. However once full penetration has been achieved the weld bead is unable to grow downwards and instead grows sideways by conduction to form a massive weld bead. As a result of this large melt volume a wide HAZ is created by conduction from the weld zone.
Fig. 6.12 shows the variation in total HAZ width (including the melt width), in 2.7 mm thick material, with laser power. It is apparent that there is a gradual increase in both HAZ and weld bead width with increasing power. The effect of increasing the laser power is similar to decreasing the welding speed as both these parameters determine the energy density at the laser material interaction point, which is expressed as follows,

\[
\text{Laser power} = \frac{\text{Welding Speed} \times \text{Focused Spot Size}}{}
\]

This simple expression shows that there is an inverse relationship between laser power and welding speed. So that for a given laser power a decreasing welding speed will result in a both a wider HAZ and melt width. Or for a given welding speed a greater laser power will have a similar effect. The effects of thermal cycling upon weld microstructure will be examined later.

(b) Weld Surface Topography Analysis.

A smooth flat surface finish is desirable when welding in order to increase the fatigue life of a welded component. The loss of fatigue properties can be attributed to striations or ripples acting as stress raisers on the surface of the weld which on loading can initiate fatigue cracks. In laser welding a relatively smooth weld bead is produced as compared to arc welding. However, careful examination of the surface topography is still important.

Laser welds were produced in 1.7 mm thick Ti-6Al-4V, using a laser power of 1500 W, at welding speeds of between 20 and 70 mms\(^{-1}\), using a 100 mm focal length lens. The weld surfaces were then examined using an SEM.
The Relationship Between Weld Striation Frequency And The Laser Power Ripple.

Fig. 6.13 shows the surface topography of laser welds produced using welding speeds of 70, 50, 40 and 20 mms\(^{-1}\). Using these and higher magnification photographs the spacings between the weld striations could be measured. Table 6.2 shows the distance between the principal (coarsest) weld striations, and their frequencies. These frequencies were calculated from the spatial wavelength and the cutting speed.

The general pattern of the striations was similar in each of the welds. Basically there was a regularly spaced coarse striation, with finer striations interspaced between them. From Table 6.2 it can be seen that the frequency of the coarse striations was almost constant at about 300Hz. This suggests that these striations were caused by a power ripple in the 3 phase supply to the HT arcs, which is the fundamental power supply to the lasing medium. The frequency of this ripple was measured using the laser beam analyser. A typical oscilloscope trace is shown in Fig. 4.13. The frequency of the power ripple (300 Hz) was found to correspond approximately with that of the coarse weld striations.

Interspaced between these coarse power ripples are a series of fine striations which are probably caused by fluid dynamic effects. A mechanism has been proposed (67) in which wave resonance in the weld pool could be caused by the interaction of gas pressure and surface tension, as shown in 6.14. In this case a depression is formed beneath the beam which generates a bow and a stern wave. If the bow wave reaches a critical size and flows around the depression, it spills over and solidifies at the back of the pool.
Oxygen contamination, and subsequent deleterious effects upon the mechanical properties, have been reported as a major problem during the welding of titanium alloys (68), and so special precautions were taken to prevent this problem during the laser welding program. For welds produced using the Control laser a special shielding nozzle and an argon filled welding jig, as described in Chapter 3, page 61 were constructed. For the welds produced using the Coherent Everlase argon was simply introduced coaxially with the laser beam from above the weld, and the dross jet was positioned underneath the weld zone to blow argon onto the underside of the weld. Auger analysis was used to analyse the extent of oxygen contamination within weld beads produced using a variety of argon gas pressures, laser powers and welding speeds.

The Variation In Oxygen Contamination With Welding Speed

Fig. 6.15 shows the variation in oxygen contamination as a function of depth and welding speed, in 1.7 mm Ti-6Al-4V, produced using the Control laser, with a power of 1500 W, and a relatively low argon gas pressure of 100 KPa. It can be seen from the graph that the overall level of oxygen contamination falls with increasing welding speed. The extent to which titanium, and its alloys, absorb oxygen is determined by three factors which are as follows.

(a) The dwell time the metal is at elevated temperatures.

(b) The peak temperature that is reached during the thermal cycle.

(c) The area of the weld bead exposed to the atmosphere.
Therefore if fast welding speeds are used in conjunction with high laser powers (in this case 1800W) then narrow "keyhole" welds can be produced. Such welds have low melt volumes which result in a rapid cooling rate and as a result the weld surface is at elevated temperatures for only a short time. In addition "keyhole" welds also have a narrow weld bead so that the molten metal is not exposed to the atmosphere outside the trailing argon shroud.

However if laser welds are produced using a low welding speed (using the same laser power of 1800W) then a large melt volume is produced with a correspondingly low cooling rate. This results in a long dwell time at elevated temperatures during which oxygen can be absorbed, and also a wide weld bead which may be exposed to the atmosphere outside the effect of the trailing argon shroud. It is apparent, from Fig. 6.15, that the base level of oxygen (0.2 wt%) can be reached within a depth of only 300 nm at a welding speed of 90 mms⁻¹.

Fig. 6.16 shows both the oxygen and nitrogen gradients with a weld produced in 1.7 mm Ti-6Al-4V under the same conditions as in Fig. 6.15 with a welding speed of 30 mms⁻¹. Fig. 6.16 shows that there is no nitrogen present at the surface of the weld and that there is then a sharp increase to a maximum value (10.5 Wt%) causing the oxygen level to fall slightly. One possible explanation for this effect is that there are two distinct layers present on the surface with a nitride layer immediately below it created as a result of the manner in which nitrogen is absorbed during welding. Nitrogen is absorbed as atomic N much more rapidly than Sievert's law would predict. However it must be mentioned that nitrogen absorption was only encountered at very low welding speeds exhibiting high levels of oxygen contamination and so is not a problem in well shielded weld beads.
Oxygen Contamination In Electron Beam Welds

Electron beam welding is the established process for the welding of Ti-6Al-4V as a result of the process being carried out under vacuum so that oxygen contamination is not a problem. A series of electron beam welds were prepared and analysed using Auger analysis to determine the level of oxygen contamination and compare them with the laser welds. The parameters of the electron beam welds produced are given in Table 6.3.

The oxygen gradient within a full penetration electron beam weld in 1.7 mm Ti-6Al-4V is shown in Fig. 6.17. If this is compared with a laser weld produced using the maximum welding speed and maximum argon gas pressure it can be seen that the level of oxygen contamination is greater in the laser weld than in the electron beam weld. However the levels of oxygen contamination in both cases are very small with the base level of 0.2% being reached at a depth of only 150 nm, even in the laser weld.

Thus it can be concluded that the level of oxygen contamination in laser welds is of the same order as that in electron beam welds. With greater care and more sophisticated argon shielding comparable welds could be produced by either process.

Summary.

The aim of sections 6.2.1(a-c) was to establish the effect of variation of the laser parameters upon bead profile, weld penetration and HAZ width. These three aspects are critical to any welding process and thus the laser welding parameters were optimized with respect to them. Detailed
Metallographic analyses of laser welded Ti-6Al-4V showed that increasing energy input resulted in both deeper weld penetration and a wider weld bead. The HAZ width was also found to increase with increasing energy input.

The surface topography of any weld bead is also important since the weld surface can act as a fatigue crack initiation site. Therefore extensive surface analyses were carried out in order to assess the effect of speed variation upon weld surface finish. It was concluded that one of the causes of weld surface striations was the power ripple which is characteristic of the laser output.

One factor which is of special importance when joining titanium alloys using any welding process is oxygen contamination. As a result particular attention was paid in this section to assess the variation in oxygen contamination with welding speed, so that the process could be optimized. It was concluded that an increased welding speed resulted in a reduced level of oxygen contamination.

Having established the optimum laser parameters with respect to oxygen contamination, weld surface topography, weld penetration and HAZ width it is important to examine certain other factors such as weld microstructure, grain growth, microhardness and porosity. It is the aim of the following sections is to examine these factors in detail.
(d) Microstructur al Analyses.

In order to analyse the microstructures produced as a result of laser welding Ti-6Al-4V alloy the weld thermal cycle must be considered. When the laser impinges upon the workpiece the metal immediately below the beam is rapidly melted and reaches a temperature approaching the boiling point at its centre. Once the laser melted zone has been established the beam moves away and the melt begins to cool. Heat is then lost by conduction to the adjacent metal which as a result melts. Due to convective currents within the weld pool a less severe temperature gradient is soon established.

The molten weld pool cools as heat is conducted away to the surrounding metal. Initially the weld cooling rate is very rapid (approximately $10^4 \text{Cs}^{-1}$) since the adjacent metal is cool and acts as a very efficient heat sink. The weld solidifies and transforms to beta phase (cf. phase diagram in Fig. 5.28). Further rapid cooling results in complete diffusion being unable to take place so that the martensitic transformation occurs. This process results in a totally martensitic weld structure in the weld on cooling, which is shown in Fig. 6.18.

The metal immediately adjacent to the fusion zone (the heat affected zone) is heated by conduction from the weld, during the cooling cycle, to a temperature just below the melting point, but well above the alpha-beta-beta transition temperature. The duration for which the metal is above this transition temperature (dwell time) is not sufficient for a totally martensitic microstructure to result on cooling and a small amount of primary alpha is retained. The mixed primary alpha + martensitic alpha HAZ structure is shown in Fig. 6.19. As the distance from the fusion line is increased both the dwell time and the peak temperature reached decrease, so that progressively more primary alpha is retained. Thus the proportion of primary alpha increases with distance from the fusion line until the peak
temperature attained in the HAZ falls below the martensite start temperature (cf. Fig. 5.28) at which point martensite formation ceases. The parent metal structure, as shown in Fig. 6.20 is unaffected by the thermal cycle.

The microstructures resulting from the laser welding of Ti-6Al-4V can be summarized as follows:

<table>
<thead>
<tr>
<th>Zone</th>
<th>Microstructure</th>
<th>Reference</th>
</tr>
</thead>
<tbody>
<tr>
<td>Fusion zone</td>
<td>Martensitic alpha</td>
<td>Fig 6.18</td>
</tr>
<tr>
<td>HAZ</td>
<td>Martensitic + Primary alpha</td>
<td>Fig 6.19</td>
</tr>
<tr>
<td>Parent metal</td>
<td>Primary alpha + Primary beta</td>
<td>Fig 6.20</td>
</tr>
</tbody>
</table>

These results are similar to those attained by Mazumder in his laser welding study (22). In addition the structure of the weld and the HAZ were found to be similar to those encountered during the electron beam welding of Ti-6Al-4V (69) showing that both laser and electron beam welding have similar cooling rates. However when submerged arc welding this alloy an accicular structure is formed in the weld zone (70), which indicates that the cooling rate in this process is lower than in either laser or electron beam welding.

Excessive grain growth has been reported as a serious drawback during TIG welding (13) and so careful attention was paid to this problem throughout the laser welding programme. During the welding process the parent alpha+beta alloy is melted and heated well above the melting point and on cooling solidifies to form b.c.c beta phase. It is during this stage of the weld thermal cycle that grain growth occurs. The size of the beta grains determines the final room temperature grain size since on cooling below the beta transus martensite laths are formed within the prior beta grain boundaries. If the cooling rate is sufficiently rapid then the material is not in the beta phase
field for sufficient time for grain growth to occur, as is the case when producing small melt volume "keyhole" welds in 1 and 1.7 mm thick Ti-6Al-4V.

When welding 2.7 mm thick material some grain growth did occur as a result of the much slower cooling rate caused by the large melt volume. Fig. 6.21 shows the variation in average grain size (measured by the linear intercept method) with welding speed for laser welds produced using welding speeds of between 6 and 50 mms⁻¹, with a laser power of 1800 W. Here the effect of grain growth can clearly be seen, with welds produced at lower speeds showing grain sizes of up to 1250 μm.

The size of the grains within a weld has a serious effect upon its mechanical properties (71), and the effect of grain size on tensile strength is given by,

\[
\text{Tensile Strength} \propto \frac{1}{(\text{Average Grain Size})^{1/2}}
\]

Thus it is desirable to have a very fine grain size within the weld which can be achieved by laser welding.

(e) Microhardness Examination.

Microhardness traverses were carried out on welds produced at various speeds using a Reichert Me F3 microhardness tester with 40 g load. A small load was used in order that a great many indentations could be fitted into a small area. The microhardness results were all slightly greater than would be expected due to relaxation effects on the indent at the small loads used.
Fig. 6.22 shows the microhardness traverse of a partial penetration weld produced using a welding speed of 32.5 mms$^{-1}$. The traverse was carried out 0.5 mm from the top surface of the weld. Initially there is a microhardness of about 355 Kgmm$^{-2}$, which corresponds to the parent metal hardness. This then increases to between 360 and 380 Kgmm$^{-2}$ which corresponds to the HAZ, which is a mixture of primary alpha and martensitic alpha. In the centre of the traverse the highest hardness values were attained, these were about 410 Kgmm$^{-2}$. This was a result of the martensitic nature of the weld bead, and could not have been attributed to oxygen contamination since Auger analysis indicated that oxygen only diffused into the weld bead to a depth of around 0.3 μm (see Fig. 6.17).

Fig. 5.23 shows a microhardness traverse of a weld produced using a welding speed of 17.5 mms$^{-1}$. If this is compared with Fig. 6.22 it can be seen that the width of the hardened zone is increased. This was due to both the increased weld bead and HAZ widths. Although the width of the hardened zone was increased the microhardness values were very similar. A microhardness traverse of a weld produced using a welding speed of 10 mms$^{-1}$ is shown in Fig. 6.24, this again shows a wider hardened zone, as would be expected, due to the lower welding speed.

Similar microhardness traverses were carried out on laser welds produced using the Control laser. Fig. 6.25 shows 3 microhardness traverses carried out 0.5, 1 and 1.5 mm from the top of a full penetration keyhole weld, produced in 1.7 mm Ti-6Al-4V, using a welding speed of 60 mms$^{-1}$, and a laser power of 1800 W. It can be seen that the microhardness values in the centre of the traverses (the weld zone) are similar to those shown in Fig. 6.22. The microstructures were also identical.
(f) Porosity

Porosity has been reported as a persistent problem during the electron beam welding (72), and TIG welding (73) of titanium and its alloys and so all the welds produced in this experimental programme were examined radiographically. The X-ray and film parameters are given in Table 6.4. Kodak MX ultra fine grain film was used in order that printed enlargements of the weld bead could be produced.

Previous investigations (82) had found no significant porosity in laser welded Ti-6Al-4V. However porosity did initially appear to be a problem in this series of welds. Fig. 6 shows a series of weld radiographs of 2.7 mm Ti-6Al-4V, produced using a laser power of 1800 W, at welding speeds of between 50 and 4 mms⁻¹. It is clear that at welding speeds of 4-8 mms⁻¹ where full penetration welds were obtained there was no porosity. Under all other conditions the welds were not full penetration welds and porosity was observed. This suggests that the porosity was caused by boiling of the titanium metal, in the root of the weld, producing a vapour which was trapped during solidification forming porosity. However, when full penetration welds were produced the vapour could escape from the root of the weld. All full penetration welds were found to be free of porosity down to the limit of resolution of the system, which was approximately 200 μm.

This argument also explains why there is porosity present in electron beam welded Ti-6Al-4V. EB welding uses a vacuum chamber which suppresses the boiling point of the molten metal so that a larger proportion of the weld is boiling producing metal vapour. This results in increased weld porosity on solidification.
During the welding of Ti-6Al-4V a very fine deep blue powder was deposited on one side, or both, of the weld bead. This had been reported previously by (74) as a fine deposit of titanium nitride and titanium oxide. Detailed ESCA analysis of the deposit showed it to be a mixture of 12 % TiN, 47 % titanium oxides (including mainly TiO₂ but with some non-stoichiometric oxides) and 41 % Al₂O₃.

The deposit was probably formed by the titanium and aluminium in the weld pool being vapourised resulting in metal particles being ejected outside the argon shield, and then forming oxides and nitrides in the atmosphere. There are no vanadium oxides present because both aluminium and titanium are more volatile than vanadium and so can easily be ejected outside the argon shield. The deposit is shown clearly in Fig. 6.27 which is a bead on plate weld produced using a welding speed of 20 mms⁻¹ with a laser power of 1500 W.
General Aspects And Overview.

The previous section has examined a number of aspects which are of importance when laser welding titanium alloys. However laser welding is only one of several welding processes that can be used to join titanium alloys. In order to compare the results of the present work with various welding processes it is possible to calculate both the melting efficiency and the joining efficiency of laser welds produced during the present study.

The melting efficiency of a welding process is given by the energy required to melt the weld bead divided by the total energy input. This approach of assessing the efficiency of laser welding has been tried previously by Mazumder (1). Therefore, so that results from the present study could be compared directly with Mazumder's, melting efficiency values were calculated for a number of welds. Although melting efficiency is a good guide to the actual efficiency of conduction limited welds, produced by TIG, MIG, MAG etc, in laser welding large conduction limited welds can have higher melting efficiencies than narrow, deep keyhole welds. A better measure of the efficiency of a welding process is the joining efficiency, which is given by

\[(\text{welding speed}) \times (\text{material thickness}) / \text{laser power}\]

This much simpler expression can be used to compare results from the present study not only with Mazumder's work, but also from other welding processes.
Melting Efficiency.

The melting efficiency \((Z_m)\) of a welding process can be expressed as the energy required to melt the weld bead divided by the total energy input over the same distance. This can be described as follows.

\[
Z_m = \frac{(\text{mass of weld bead}) \times [C_p(T_m - T_a) + L_f + 0.5L_t]}{\text{Total Power Input}}
\]

\[
Z_m = \frac{N_a P V [C_p(T_m - T_a) + L_f + 0.5L_t]}{P}
\]

where

- \(P\) = Laser Power (W)
- \(N_a\) = Weld Bead Area \((m^2)\) \((\text{transverse})\)
- \(V\) = Welding Speed \((\text{ms}^{-1})\)
- \(C_p\) = Specific Heat Capacity = 599 J/Kg K \((58)\)
- \(T_m\) = Melting Temperature = 1950K \((58)\)
- \(T_a\) = Ambient Temperature = 293K
- \(L_f\) = Latent Heat Of Fusion = 436000 J/Kg \((8)\)
- \(L_t\) = Latent Heat Of Transformation = 95000J/Kg \((8)\)
- \(\rho\) = Density = 4420 Kgm\(^{-3}\) \((58)\)

The melting efficiency for the bead on plate welds produced could be calculated as all these values were known. However a value of half the latent heat of transformation was used because as the Ti-6Al-4V was 50% alpha and 50% beta phase the only half the material transformed on heating into the beta phase field. The weld bead areas were measured accurately by using a Quantimet 800 image analyser.

Fig. 6.28 shows the variation in melting efficiency with welding speed, for welds produced using the Control laser. It can be seen that the melting efficiency rises with increasing
welding speed. This effect can be explained by examining the two basic laser welding mechanisms, which are related to the welding regimes.

(a) The use of high incident energy densities results in a very narrow weld bead ("keyhole" weld) being produced. The weld bead width approximates to that of the focussed laser beam (as shown in Fig. 6.29), which suggests that the weld bead is melted by direct laser irradiation. This direct melting is an efficient process and as a result high melting efficiency values are attained at rapid welding speeds.

(b) However as the energy density is reduced the weld bead width increases so that the bead is much wider than the incident beam. In this case melting occurs in two stages. A small volume of metal (approximately the same width as the focussed beam) is directly melted by laser irradiation. The remainder of the weld bead grows by conduction from the directly irradiated region (as shown in Fig. 6.30). This two stage melting mechanism is a less efficient process than the direct melting mechanism outlined in section (a). Thus the use of low welding speeds results in low melting efficiency values.

Another factor that reduces the melting efficiency values at low welding speeds is re-radiation from the weld bead surface. A wider weld results in more heat being lost from its surface reducing the melting efficiency.

Previous work (1) carried out on the laser welding of Ti-6Al-4V alloy also indicated that the melting efficiency increased with processing speed. However Mazumder carried out most of his work at low welding speeds where full penetration welds were produced and did not consider partial penetration welds. In addition the majority of Mazumder's welding programme was carried out using different thicknesses of material to the present study. Therefore direct comparisons between the two
sets of results are difficult. However the observed experimental trends were similar, although in a different range.

Joining Efficiency.

The melting efficiency, as described in the previous section, was not an accurate guide to the actual efficiency of the welding process in that it favoured wide weld beads, whereas the desired weld bead shape is narrow and keyholed. Thus previous work (f) has used joining rate efficiency as a measure of the process efficiency. The joining efficiency is given by the area joined per unit energy and can be expressed as,

\[ \text{Joining Efficiency} = \frac{(\text{Welding Speed}) \times (\text{Material Thickness})}{\text{Laser Power}} \]

Table 6.5 shows the calculated joining efficiencies for full penetration welds produced using the maximum possible welding speed, for both the Control and the Coherent lasers. The data in Table 6.5 shows two important features.

(a) The maximum joining rate efficiencies are very similar, for both the Control and the Coherent laser, for 1 mm thick Ti-6Al-4V alloy sheet. This is because a six fold increase in laser power, in this case, results in a six fold increase in maximum welding speed.

(b) Increasing material thickness results in a reduction in maximum joining efficiency. This effect can be attributed to similar factors which explain the reduction in melting efficiency with welding speed described in the previous section. Basically in 1 and 1.7 mm thick Ti-6Al-4V alloy sheets narrow full penetration "keyhole" welds can be produced, using rapid processing speeds. Such welds are produced by the efficient
direct melting process outlined in the previous section (Fig 6.29). However laser welds produced in the 2.7mm thick material exhibited bead widths well in excess of the focussed beam diameter which resulted in the majority of the weld bead being melted by conduction from a small directly melted zone underneath the beam, as shown in Fig. 6.30.

If the results of this work (cf. Table 6.5) are compared to joining rate efficiency values obtained in previous studies (Table 6.6) using electron beam, plasma arc and GTA it can be seen that laser welding compares very favourably. The joining rate efficiency values are greater in the 1 and 1.7 mm material than those attainable with electron beam welding, which are about 30 mm²/KJ (16). It must be mentioned however that it is misleading to compare joining rate efficiencies for different welding process unless the same power and material thickness are used.

(h) Mechanical Testing.

In order to assess the performance of welded components in service conditions careful mechanical testing is required. The mechanical testing programme involved both impact testing and tensile testing. In each of these the properties of the parent metal were compared to those of the welded specimens, in order to show the effect of laser welding upon the mechanical properties of Ti-6Al-4V alloy.

Impact Testing.

The impact testing programme was carried out using a Losthausen Charpy impact tester, on sub-size standard Charpy specimens, the dimensions of which are given in Fig. 3.14.
For reasonable impact test results it is important to utilize as thick a specimen as possible. For this reason bead on plate welds were produced in 3.2 mm thick Ti-6Al-4V sheet. However with the laser power limitation of 1800W full penetration could not be achieved. As a result the notch was then machined out from the underside of the weld specimen so that weld metal would be tested during the impact test. Impact tests were carried out at a number of temperatures varying between 77K and 373K for both laser welded and parent metal specimens. The laser welded specimens were produced using a welding speed of 6 mms⁻¹, and a power of 2KW.

Fig. 6.31 shows the variation in absorbed impact energy with temperature, for both the welded and parent metal specimens. It can be seen that energy absorbed by the parent metal specimens decreased with test temperature from 19J (at 373K) to 14J (at 77K). Previous work carried out by IMI Titanium (58) has found a similar trend. Figure 6.32 is an SEM of the fracture surface of a parent metal specimen fractured at a test temperature of 293K, where the ductile nature of the fracture surface is clearly visible.

Impact testing of the weld metal revealed a similar drop in impact energy from 17J (at 373K) to 13J (at 77K). However the overall absorbed energy values were approximately 12% lower than those of the parent metal. This effect can be attributed to three factors.

(a) The weld metal microstructure was martensitic so that there were greater internal stresses within the weld metal than the parent equiaxed structure.

(b) The grain size of the weld metal was also larger than that of the parent equiaxed structure reducing the impact properties.
SEM analyses of weld impact test fracture surfaces revealed some porosity to be present in the root of the weld, as shown in Fig. 6.33. The porosity in the welded specimens occurred because the impact specimens were machined out of non full penetration bead on plate welds, so that any gases evolved during welding could not escape from the toe of the weld.

**Tensile Testing.**

The aim of this series of tests was to compare the yield strength and the UTS of the parent metal and laser welded 1 and 1.7 mm thick Ti-6Al-4V. Butt welds were produced between two 120x1100 mm blanks, from which the test pieces were machined out. The dimensions of the test pieces are given in Fig. 3.13. The parameters for the laser welded specimens are given below in Table 6.7.

These parameters were used in order to produce full penetration welds using the minimum power and maximum speed. This was because previous work (7) had found that increasing laser power and decreasing welding speed had little effect upon mechanical properties.

All tensile testing was carried out using a 10T Instron, in accordance with BS 18 (75). The cross head speed was 0.1 cms⁻¹. Load-extension curves were produced for both welded and parent test pieces, from which the yield strength, tensile strength and % elongation could be determined. Four tensile tests were carried out on each thickness, of both the welded and parent specimens. A typical load-extension curve is shown in Fig. 6.34, and Table 6.8 shows the results of the tensile testing programme, with the values being the mean of 4 tests.
In all the tensile tests carried out on the laser welded specimens failure was found to occur in the parent metal. Therefore it can be concluded that the weld metal has a higher UTS than the parent. This correlates with the microhardness results in section (e), page 163 which showed that the weld metal was substantially harder than the parent. SEM analysis of a typical fracture surface, shown in Fig. 6.35 and 6.36, showed that the fracture was totally ductile and occurred in a "cup and cone" manner.

It is apparent from Table 6.8 that there is little difference between the UTS and yield strength of the parent and laser welded samples. However the % elongation, which is a measure of the ductility of the material, was found to drop slightly from about 12% for the parent metal, to about 10% for the laser welded samples, which is due to the martensitic microstructure of the weld. This slight loss of ductility has been found in previous work(21).

6.2.2 Pulsed Welding.

At present the most common mode of CO₂ laser operation is the continuous wave mode, in which the lasing gas is constantly supplied with electrons from the HT arcs to produce a continuous output. However many gas lasers have the facility to pulse the laser output. Previous work has found that pulsed laser welding has led to increased processing speeds with stainless steels (37). However there has been very little published work in this field and so the aim of this series of experiments was to assess the feasibility of pulsed laser welding Ti-6Al-4V alloy sheet. Particular attention was paid to the pulse parameters and their relationship to the resulting weld bead morphology.
Experimental Procedure.

Both the Control and the Coherent lasers were fitted with facilities for pulsed outputs. However since the Coherent laser produced the best Gaussian beam distribution this laser was chosen to carry out the pulsed welding programme. Bead on plate welds were produced in 1 mm thick Ti-6Al-4V alloy sheet using a wide variety of pulse lengths, frequencies and welding speeds. Variations in these upon the weld bead morphology were then studied. Throughout the pulsed welding programme the mark:space ratio was kept constant at 1. The mark:space ratio is the ratio of the pulse duration / dead time, which is shown in the schematic in Fig. 6.37. The average power output was monitored using a power meter with a response time of approximately one second.

Production Of Pulsed Laser Outputs.

Pulsing is achieved in CO₂ lasers by simple electrical switching of the HT arcs. The frequency of this electrical switching determines the repetition rate of the pulses. The proximity of the individual pulses to each other is limited by two factors which are as follows.

(a) The response times of the electrical and mechanical systems which vary according to the particular laser.

(b) The lasing response time and the time lag between individual pulses. These two factors are discussed in detail later.
Monitoring Of Pulsed Outputs.

Until recently it has been almost impossible to accurately monitor pulsed laser outputs. However a relatively new innovation, the laser beam analyser (LBA), can now be used to accurately show the pulse shapes. From this display the pulse duration and frequency can also be measured. During the pulsed welding programme the pulsed output was displayed on a storage oscilloscope and copies of the output were made using a chart recorder. The mode of operation of the LBA is described in detail in Chapter 3, page 62.

Pulse Shape Analysis.

The output pulse profile (monitored by the LBA) is directly related to the electrical characteristics of the HT arcs. Therefore in order to explain the pulse shape the electrical properties of the arcs must be considered.

A typical low frequency (60 Hz) pulsed output is shown in 6.38. The power curve can be divided into three distinct regions which are as follows.

(a) Plasma formation zone
Initially when a voltage is applied across the electrodes there is no effect upon the lasing gas since it is non conductive. However as the voltage is increased the potential reaches a point at which breakdown occurs and ionization of the gas takes place. At this stage lasing commences.

(b) Steady state zone.
Once the lasing gas has become ionized the voltage required to sustain the plasma drops, and thus after a short time lag the intensity of the laser output also falls. The power intensity
then remains constant at this level until the electrodes are switched off.

(c) Extinction of the plasma.
When the electrodes are switched off the lasing action does not cease immediately. Some plasma remains and decays to a potential less than the breakdown voltage over a period of time (the decay time). Only when all the plasma has decayed does the lasing action cease.

As the pulse frequency is increased the dead time decreases (ie the time during which no lasing takes place). This occurs because there comes a point where the decay time is longer than the dead time. If the arcs are then reignited (with the gas already at a potential greater than zero) a smaller applied voltage is required to initiate the glow discharge needed for lasing. Therefore at higher frequencies the pulse shape is fundamentally different to that at lower frequencies. The change in pulse shape with frequency can be seen clearly in 6.39.

Estimation Of Peak Power Output During Pulsed Welding.

It is apparent from Fig. 6.39 that there is a power surge associated with the striking of the electrodes in pulsed welding which is especially prominent at lower pulse frequencies. Using the power curves it is possible to calculate the peak power attained during the striking surge.

Initially the individual pulse areas (which correspond to the pulse energy) were measured. From these areas it was possible to calibrate the power graphs since

\[
\text{Pulse area (energy) \times Frequency} = \text{Average power output.}
\]
By equating the areas required to produce 300W (pulsed) and 300W (continuous wave) it was possible to display the CW output on the same graph, as shown in Fig. 6.39.

Since the CW power output is known to be 300W the peak power output can be calculated as follows.

\[
\frac{\text{Pulse Peak Height}}{\text{CW Output Height}} = \frac{\text{Peak Pulse Power}}{\text{CW Output Power}}
\]

Table 6.9 shows the variation in peak power with frequency and it can be seen that the peak power decreases with increasing pulse frequency. This effect can be explained by examining the decay characteristics of each pulse. At low repetition rates (\(~300\text{Hz}\)) each pulse has sufficient time to decay so that the gas returns to its ground state and the lasing action ceases (ie. there is some dead time). However if the decay time is equal to or longer than the 'dead time' the gas never fully returns to the ground state. In this case a smaller applied voltage is required to produce gas breakdown and less of a power surge occurs.

Analysis Of Pulsed Welds

When laser welds are produced using the continuous wave mode the point at which transverse sections, for metallography, are taken is not important since the weld bead will have a continuous penetration depth. However with pulsed welding sectioning at different points along the weld bead can reveal almost random and very misleading bead profiles. This can occur because during pulsed welding individual packets of energy are produced resulting in discrete "spikes" of weld metal. The
pulse frequency and the welding speed determine whether or not each "spike" is joined to the one preceding it, producing a continuous weld.

Fig. 6.40 is a schematic representation of a pulsed laser weld produced using a low frequency and a high welding speed. It can be seen that if sectioning is carried out at point X then the weld bead will appear to penetrate through the metal, whereas sectioning at point Y will reveal little weld penetration.

In order to correctly analyse pulsed welds micrographs must be produced by polishing back from the underside of the weld bead to reveal any discontinuity. Fig. 6.41 shows two pulsed welds polished from underneath produced using the same welding speed. The pulse frequency in (a) was 200Hz and in (b) 1000 Hz. The continuous weld penetration can be clearly seen in (b) whereas in (a) individual spikes of weld metal can be seen.

The separation of each "spike" of weld metal in 6.41(a) was found to correspond very closely to the theoretical pulse separation of the laser output which is given by,

\[
\text{Welding Speed / Pulse frequency} = \frac{\text{Experimental Pulse Separation}}{\text{Theoretical Pulse Separation}}
\]

Experimental Pulse Separation = 0.181 mm
Theoretical Pulse Separation = 0.175 mm

From these results it can be postulated that each individual output pulse produces one "spike" of weld metal.
Solidification Of Pulsed Welds.

When pulsed welds are produced using high frequencies each pulse of energy does not have time to completely decay before the next one commences. Thus the lasing action never totally ceases so that the welds approximate to continuous wave welds. In this case solidification is continuous not only because of the constant lasing action but also because each pulse produces a weld metal "spike" before the previous one has solidified.

However when using low frequencies and high welding speeds each laser pulse is separate and produces a single spike of weld metal. These spikes solidify independently so that discontinuity in the weld metal solidification structure can be seen. This discontinuity is shown clearly in Fig. 6.42 which is a pulsed weld polished back from the top surface.

The Effect Of Pulsing On Weld Bead Morphology.

When laser welds were produced using the Coherent laser in the continuous wave mode full penetration keyhole welds could not be produced. Only conduction limited welds were formed in 1mm thick Ti-6Al-4V alloy sheet. However laser welds produced using the pulsed mode were continuous full penetration keyhole welds. This change in weld bead morphology can be explained by examining the power output of the laser.

Using the continuous wave mode a constant power of 300W was supplied. However by pulsing the laser with a mark:space ratio of 1:1 and keeping the average power constant at 300W it can be calculated that the peak power can be as much as 4x the CW power (as shown in Table 6.9). This increase in the peak power results in vapourization of the workpiece which leads to production of a "keyhole". This reduces the metal's
reflectivity so that the remaining pulse energy is absorbed readily by the molten puddle. A cross section of a typical pulsed keyhole weld is shown in Fig. 6.43.

Joining Rate Efficiency Of Pulsed Welds.

The joining rate efficiency of a welding process is given by

\[
\text{Material Thickness} \times \text{Welding Speed} \over \text{Laser Power} \quad \text{mm}^2 \text{J}^{-1}
\]

Using this relationship it is possible to calculate the maximum joining rate efficiency values for welds produced using the laser in both the continuous wave and the pulsed mode.

(a) Continuous wave mode.

Maximum welding speed (1mm thick Ti-6Al-4V) = 17.5 mm s^{-1}
Power output = 300W
Therefore maximum joining rate efficiency = 58 mm^2/KJ

(b) Pulsed mode.

Maximum welding speed (1mm thick Ti-6Al-4V) = 35 mm s^{-1}
Frequency = 100 Hz
Pulse length = 1 ms.
Average power output = 300W
Therefore maximum joining rate efficiency = 116 mm^2/KJ.

These results show that pulsed welding has much greater joining rate efficiencies than continuous wave welding. This can
be attributed to the onset of keyholing when using the laser in the pulsed mode. If the joining efficiencies for the pulsed laser at 300W are compared to those attained using the Control laser at 1500-1800W (Table 6.10), it can be seen that pulsed welding at 300W has a greater efficiency than even when using a continuous wave power output of 1800W. In addition the weld profiles produced using the Control laser and the Coherent laser (pulsed) were both keyholed.

Thus it can be concluded that similar weld profiles can be produced in 1 mm thick Ti-6Al-4V using either the Control laser in the continuous wave mode or the Coherent laser in the pulsed mode. These results offer the possibility of using much lower power lasers to produce welds of a similar penetration depth as much more expensive high power lasers.
6.2.3 Heat Flow Analysis.

In any welding process it is extremely useful to have a model capable of predicting experimental results from which it is possible to estimate the correct welding parameters when using new materials. Ideally such a heat transfer model should be able to predict the following.

(a) The minimum welding speed and power.
(b) The depth of penetration.
(c) The extent of the heat affected zone.
(d) The cooling rate within the weld and the HAZ.

A variety of different heat transfer models for welding have been developed using both analytical and numerical techniques. This section aims to review these models and assess their applicability to the laser welding process. In addition the latest available laser welding model developed by M.Sharp (76) at Imperial College (from Mazumder's (7) original model) is used to compare theoretical results with experimental data from the present welding programme. It must be stressed that no attempt has been made to modify the existing programme.

Review Of Previous Heat Transfer Models.

Analytical Techniques.

The first heat flow model designed for welding was developed by Rosenthal (77) in 1941. He specified two distinct welding situations.

(i) A point source of heat moving over the surface of an infinitly wide thick plate where the heat flow can be regarded as 3D.
(ii) A line source of heat moving through an infinitely wide thin plate, where the heat flow can be regarded as 2D.

The equations which Rosenthal developed are given below.

For the 3D case

\[(T-T_0) = \frac{Q \cdot \exp(-VX/2a) \cdot \exp(-VR/2a)}{2\pi K R}\]

For the 2D case

\[(T-T_0) = \frac{Q \cdot \exp(-VX/2a) \cdot K_0 (VR/2a)}{2\pi K g}\]

From these the centre line cooling rates were then derived, which are as follows.

For the 2D case

\[\frac{dT}{dt} = \frac{2\pi K C (Vg/Q)^2 (T-T_0)^3}{dt}\]

For the 3D case

\[\frac{dT}{dt} = \frac{2\pi K (V/Q) (T-T_0)^2}{dt}\]

In order to decide whether an actual situation is either 2D or 3D, the relative thickness parameter (RTP), was developed \((78)\) which is given by,

\[RTP = \frac{Cg^2p (T-T_0) V}{Q}\]
Where

\[ Q = \text{Energy input} \]
\[ v = \text{Welding speed} \]
\[ p = \text{Density} \]
\[ g = \text{Material thickness} \]
\[ C = \text{Specific heat capacity} \]
\[ T = \text{Weld temperature} \]
\[ T_0 = \text{Ambient temperature} \]
\[ K_0 = \text{Bessel function 2\textsuperscript{nd} kind, 3\textsuperscript{rd} order.} \]
\[ X = \text{Weld isotherm for a given } R. \]
\[ R = \text{Pool radius.} \]
\[ \alpha = \text{Thermal Diffusivity.} \]

For RTP less than 0.3 the cooling rate at the top and bottom of the plate were assumed to be equal and thus the conditions were 2D.

For RTP greater than 1 the cooling rate was assumed to be 3D, since the bottom of the plate has a zero cooling rate.

For values of RTP between 0.3 and 1 a graphical solution had to be used illustrating one of the drawbacks of Rosenthal's technique.

The main problem associated with the application of Rosenthal's technique to laser welding was that no account was taken of the keyholing effect. As a result of this predictions of experimental laser welding results can have errors of over an order of magnitude (7).

The first technique which took account of the keyholing effect was that of Swift-Hook and Gick (79). They assumed that the molten metal around the keyhole approximated to a cylinder, as shown in Fig. 6.44, and that the heat flow in the material was 2D. They then used the line source model developed by Carslaw and Jaeger (80), and deduced the following equation which gave the weld isotherms.
\[ S = q \exp(Ur \cos \phi)K_0(Ur) \]

where

- \( S = KT \)
- \( q = W/a \)
- \( U = v/2D \)
- \( a = \text{depth of penetration} \)
- \( W = \text{laser power} \)
- \( v = \text{welding velocity} \)
- \( K = \text{Thermal conductivity} \)
- \( K_0 = \text{Bessel function of the second order} \)
- \( y = \text{pool half width} = r \sin \phi. \)

By substituting the power per unit depth (X) into (6.1) it becomes,

\[ X = 2\pi \exp \left(-Ur \cos \phi \right)/K_0(Ur) \]

where \( X = W/a\phi \)

Swift-Hook and Gick(79) showed that at the widest point of the weld bead the following condition was satisfied,

\[ \frac{dX}{d\phi} = \frac{dy}{d\phi} \]
\[ \frac{dX}{dr} \quad \frac{dy}{dr} \]

where \( y = \text{half width of the weld pool} \)
\[ = r \sin \phi. \]

Using (6.2) and (6.3) they deduced the following relationships,

\[ \cos \phi = -K_0(Ur) / K_0'(Ur) \]
defining the normalized melt width as,

\[ Y = \frac{vb}{D} = 2Ub \]

where \( b = 2y \) = the full melt width.

Using (6.4) and (6.5) it follows that

\[ Y = 4Ur \frac{(1-K_0^2(Ur))}{K_0^2(Ur)}^{1/2} \tag{6.6} \]

And similarly using equations 4.5 and 4.10 the normalized power input \( (X) \) is given by,

\[ X = 2\pi \exp(UrK_0(Ur)) / K_0(Ur) \]

The relationship between the power \( (W) \) and melt width \( (b) \) can be found by eliminating \( Ur \) in (6.6) and (6.7) for two limiting cases, which are,

(i) High Speed Limit.

At high speeds (large \( Ur \)) the asymptotic expansions of the modified Bessel function can be used thus

\[ K_0(Ur) = (\pi/2Ur)^{0.5} \exp(-Ur) \]

\[ -K_0/K_0' = 1 - (1/2Ur). \]

So \[ X = (8\pi e Ur)^{0.5} \]

\[ Y = 4(Ur)^{0.5} \]

Thus \[ Y = (2/e)^{0.5} = 0.484X \]
(ii) low speed limit.

At the low speed limit (small Ur) approximations can be made to the modified Bessel function thus,

$$K_0 (Ur) = \ln (2e^{-h/Ur})$$

$$-K_0 = 1/Ur$$

Where $h = 0.577 = \text{Euler's Constant}.$

So $K_0/K_0'$ vanishes and

$$Y = 4U$$

While $$X = 2\pi / \ln (2e^{-h} / Ur)$$

Hence $$6.3/X = \ln (4.5/\gamma)$$

The relationship between $X$ and $Y$ is shown graphically in Fig. 6.45 with an interpolation in the intervening region where neither solution is strictly valid. Although Swift-Hook and Gick's model gave results closer to those attained experimentally, than Rosenthal's model, there was still a substantial error (1).

**Numerical Techniques**

Analytical techniques suffer from the drawback that non linear terms, such as surface re-radiation and keyholing cannot be included. This problem can be overcome by using numerical techniques. The first of these was developed by Mazumder and Steen (1,82).

In this model an axisymmetric laser beam with a defined power distribution strikes the surface of an opaque substrate
with finite width and depth, and of infinite length moving with uniform velocity along the length of the material. Part of the incident radiation is absorbed whilst the rest is reflected. The reflectivity was considered to be zero when the material had reached the boiling point, since the keyhole produced by vaporization acts as a black body absorber. The model also takes into account heat losses by radiation and convection from the surface, and conduction into the substrate. The main assumptions of the model are as follows.

(a) The laser beam is stationary relative to the earth, on an axis of symmetry, whilst the workpiece moves in a positive x direction with uniform velocity.

(b) The workpiece is of infinite length but of finite width and depth.

(c) Quasi-steady-state is assumed.

(d) The power distribution within the laser beam is Gaussian, and is given by,

\[ P_{x,y} = P_{tot} \frac{1}{\pi r_b^2} \exp \left(-\frac{2r}{r_b}\right) \]

where \( r = (x^2+y^2)^{1/2} \) is the radial distance from the centre, and \( P_{tot} \) is the total incident power. \( r_b \) is the radial distance at which the power density falls to 1/e² of the central value.

(e) Thermal conductivity, specific heat capacity and density are assumed to be independent of temperature.

(f) The latent heat of fusion is compensated by the latent heat of solidification.

(g) Reflectivity is zero above the boiling point.
(h) Heat is lost by radiation from both the upper and lower surfaces according to the Stefan-Boltzmann equation

\[ q_{\text{loss}} = A\sigma(T^4 - T_a^4) \]

where
- \( q_{\text{loss}} \) = radiative heat loss
- \( A \) = radiating area
- \( \sigma \) = Stefan-Boltzmann constant
- \( T \) = surface temperature
- \( T_a \) = ambient temperature

(i) Heat is lost by convection due to shielding gas flow, and the convective heat transfer coefficient is calculated from Gordon and Cobonque (60)

\[ h_C = 13Re^{0.5}Pr^{0.33}Kg / B \]

where
- \( h_C \) = convective heat transfer coefficient
- \( Re \) = Reynold's number
- \( Pr \) = Prandtl number
- \( Kg \) = thermal conductivity of the gas
- \( B \) = jet-plate distance

(j) When any location exceeds the boiling point it is considered to have evaporated.

(k) Radiation penetrating the workpiece is absorbed according to Bier Lambert's law,

\[ P = P_s \exp(-bL) \]

where
- \( b \) = absorption coefficient.
- \( P \) = radiant power at depth \( L \), which has a value \( P_s \) at the surface
Matrix points which have evaporated remain at fictitiously high temperatures to simulate the high reflection and radiation transfer effects from the fast moving plasma within the keyhole.

The overall matrix size was dependant upon the available computer core storage, and the grid spacing was decided by stability and computing time considerations. It was decided that for computing stability five matrix points had to lie within the incident beam diameter.

Basically a heat balance was performed at each matrix point within the substrate, using Fourier's conduction law.

\[ Q_x = -K \cdot A \cdot (dT/dx) \]

where

- \( Q_x \) = Heat flux in the x direction
- \( K \) = Thermal conductivity
- \( A \) = Area normal to the direction of the heat flow
- \( T \) = Temperature
- \( x \) = Direction of heat flow

For each matrix point the heat balance was given by

Heat in - Heat out = Zero.

If heat flow is considered positive in the x, y and z directions, from Fig. 6.46, the heat in is given by:

\[-Kdy.dx(dT/dx)p_\text{W} \quad -Kdy.dx(dT/dy)p_\text{S} \quad -Kdy.dx(dT/dz)p_\text{L} \]
\[ + p \cdot C_p \cdot dy \cdot dz ((T_\text{W} + T_\text{p})/2) \]
and the heat out is given by

\[-K \frac{dy}{dx}(dT/dx)_{PE} -K \frac{dy}{dx}(dT/dy)_{PS} -K \frac{dy}{dx}(dT/dz)_{PL} + p \cdot C_p \frac{dy}{dz} ((T_p + T_E)/2)\]

Thus the heat balance for each matrix point becomes

\[-K \frac{dy}{dx}(dT/dx)_{PW} -K \frac{dy}{dx}(dT/dy)_{PS} -K \frac{dy}{dx}(dT/dz)_{PL} + p \cdot C_p \frac{dy}{dz} ((T_W + T_p)/2) + K \frac{dy}{dx}(dT/dx)_{PE}\]

\[+ K \frac{dy}{dx}(dT/dy)_{PS} + K \frac{dy}{dx}(dT/dz)_{PL} - p \cdot C_p \frac{dy}{dz} ((T_p + T_E)/2) = 0\]

This can be simplified by using \(a\), the thermal diffusivity = \(K/pC_p\)

Therefore it follows that :-

\[-a \frac{dy}{dx}(dT/dx)_{PW} -a \frac{dy}{dx}(dT/dy)_{PS} -a \frac{dy}{dx}(dT/dz)_{PL} + \frac{dy}{dz} ((T_W + T_p)/2) + a \frac{dy}{dx}(dT/dx)_{PE}\]

\[+ a \frac{dy}{dx}(dT/dy)_{PS} + a \frac{dy}{dx}(dT/dz)_{PL} - dy.dz ((T_p + T_E)/2) = 0.\]

This equation was then transformed into finite difference terms for solution by the computer and the results displayed graphically by the dimensionless groups,

Dimensionless temperature

\[T^* = \frac{T_K D_b}{P_{TOT} (1-r_f)}\]

Dimensionless length

\[X^* = \frac{X}{D_b}\]

Dimensionless width

\[Y^* = \frac{Y}{D_b}\]
Dimensionless depth \[ Z^* = \frac{Z}{D_b} \]

Dimensionless speed \[ U^* = \frac{UD_b}{a} \]

Plots of \( T^* \) against \( Z^* \) are shown in Fig. 6.47.

Assuming that the beam diameter is known it is possible to calculate \( T^* \) for a given temperature. In this case the melt isotherm (1950K) was used with \( r_f \) taken as zero. Having calculated \( T^* \) Fig. 6.47 can be used to read off the corresponding \( Z^* \) for a given dimensionless speed, \( U^* \), from which the weld penetration depth \( Z \) can be calculated.

Mazumder's model can also be used to predict weld and HAZ widths, and other important factors such as cooling rate.

Problems Associated With Mazumder's Model.

The main problem associated with the early use of Mazumder's model is that it only predicted penetration depths, melt widths and HAZ widths at 3 speeds, due to computing considerations. This can be seen in Fig. 6.47. Also the assumption that the specific heat capacity and thermal conductivity remain constant with temperature is inaccurate which is shown in Fig. 6.48.

The main problem associated with Mazumder's original model is the Keyholing function. The model assumes that irradiated matrix points stay at ficticiously high temperatures (well above the boiling point) instead of vapourizing. This results in heat being conducted to adjacent matrix points whereas in practice these points are perhaps directly irradiated.
Modifications To Mazumder's Model.

Mazumder's original model has been extensively changed in order to overcome some of its drawbacks. One major alteration to the model was carried out by P. Henry who changed the fundamental basis of the model from quasi steady state to transient state conditions. In addition an allowance was made for both the latent heat of fusion and vaporization by incorporating melting and vapourization bands into the programme. Further modification has been carried out by M. Sharp (76). There is now a programme available for laser welding simulation, known as "Glaze 7". A listing of this programme is given in Appendix 1.

Operation Of The Model.

In order to use the model to simulate the experimental results achieved during the laser welding programme a number of parameters had first to be calculated.

(i) Reflectivity

The reflectivity $R(T)$ of a metal surface can be calculated from the emissivity $e(T)$:

$$e(T) = 1 - R(T) \quad (81)$$

For long wavelength radiation Bramson (81) developed a relationship from which the emissivity could be derived which is given below,

$$e(T) = 0.365 R(T)^{0.5} - 0.667 R(T) + 0.006 R(T)^{1.5}$$
Where \( R(T) \) = Electrical resistivity at a temperature \( T \).
\[ \lambda = \text{Wavelength} \]

From these two equations the room temperature reflectivity of Ti-6Al-4V can thus be calculated. Substituting in \( R = 170 \times 10^{-6} \text{cm} \) and \( \lambda = 10.6 \text{ \mu m} \)

\[ e(300K) = 0.041 \]

Therefore \( R(300K) = 1 - 0.041 \]

\[ = 0.95 \]

Bramson's equation is only valid for samples irradiated in a vacuum, as any oxide film that is present will decrease the reflectivity. All titanium alloys have an oxide film present on their surface so that the calculated reflectivity will be greater than the actual value. However since the laser welding of Ti-6Al-4V is carried out in the presence of an inert gas shield only a small degree of oxidation takes place during welding and therefore Bramson's equation can be regarded as a reasonable approximation of the initial reflectivity.

(ii) Calculation of the Heat Transfer Coefficient \( (h_c) \).

In order to calculate the heat transfer coefficient \( (h_c) \) the velocity of the gas jet shielding the surface of the weld must be known. In this case an 8mm diameter \( (d) \) aperture was used with a slow argon flow rate \( (F) \) of \( 1.3 \text{m}^3\text{s}^{-1} \). If lamellar flow is assumed then the gas velocity \( (V) \) can be expressed as follows \( (61) \)

\[ V = \frac{F}{2n(0.5d)^2} \]

\[ V = 1.28 \text{ ms}^{-1} \]

Gordon and Cobonque \( (60) \) reported that at the stagnation point of a vertically impinging jet
\[ \text{Nu}_0 = 13 \times \text{Re}^{0.5} \frac{D_j}{B} \]

using the correlations developed by Polhausen (35), Steen (42) and Mazumder (1) derived that

\[ \text{Nu}_0 = 13 \times \text{Re}^{0.5} \frac{D_j}{B} \times \text{Pr}^{0.33} = h_C \frac{D_j}{K_{gas}} \]

or

\[ h_C = 13 \times \text{Re}^{0.5} \times \text{Pr}^{0.33} \times \frac{K_{gas}}{B} \]

where

- \( \text{Nu}_0 \) = Nusselt Number
- \( \text{Re} \) = Reynold Number
- \( \text{Pr} \) = Prantl Number
- \( K_{gas} \) = Thermal conductivity.
- \( B \) = Jet plate distance

Since

\[ \text{Re} = \frac{p g V d}{M} \]

and

\[ \text{Pr} = \frac{\zeta M}{K_{gas}} \]

It is possible to calculate \( h_C \) by substituting in the following values

- \( p g \) = Gas density = 1.784 Kgm\(^{-3}\) (62)
- \( K_{gas} \) = Thermal conductivity = 162 x 10\(^{-4}\) Wm\(^{-1}\) (62)
- \( M \) = Gas viscosity = 0.000021 Nsm\(^{-2}\) (62)
- \( D \) = Nozzle diameter = 8mm.

Therefore \( h_C = 110 \text{ WmK}^{-1} \)

This value of \( h_C \) is lower than that used in previous calculations (1) which reflects the different shielding arrangements and gas flow rates utilized. Previous work (1) has been carried out using a nozzle whereas the present study has used a wide
diameter aperture with a slow gas flow rate, which results in much lower convective heat losses.

(iii) Choice of the convergency limit "BIG".

The choice of the convergency limit "BIG" proved to be a problem. If too low a value was used the programme never converged and if too high a value was used the results were inaccurate. Eventually a value of BIG = 100 was used to give an acceptable level of accuracy even although 150 iterations were required to satisfy the convergency criterion which took up a great deal of computing time.

Results And Discussion.

Once the programme has been run the results are presented in a format shown in Appendix 2, which shows typical results for a laser weld produced using a welding speed of 50 mms⁻¹. Temperature distributions are given in three planes which are,

(a) x-z Cross section through the weld (weld profile)
(b) x-y Weld surface temperature distribution.
(c) y-z Longitudinal section.

From these temperature distributions it is possible to plot weld bead profiles by defining the melting isotherm, which in this case was 1950K. In addition the HAZ width could be measured by defining the lower martensitic transformation temperature which was taken to be 1150K.
Weld Penetration Analysis.

The model can be used to predict weld profiles at various speeds and powers. Fig. 6.49 illustrates the theoretical weld bead profiles produced using a laser power of 1500W at speeds of 30, 50, 70 and 90 mms\(^{-1}\). It is apparent that the weld penetration increases with decreasing welding speed as would be expected.

Fig. 6.50 shows the variation in both theoretical and experimental weld penetration with welding speed. It is apparent that the model predicts results which are approximately 25% greater than the experimental results at 90 mms\(^{-1}\) and approximately 50% higher at 30 mms\(^{-1}\). It appears that the model over estimates weld penetration, and most closely correlates with experimental data at higher welding speeds. A number of factors might explain these effects.

(i) Welds produced in 2.7mm thick Ti-6Al-4V sheet at speeds below 30 mms\(^{-1}\) were not fully keyholed and so would have a lower level of weld penetration than would be expected in a fully keyholed weld.

(ii) The assumptions made in the model regarding reflectivity. In this case a reflectivity value of 0.95 was used (as calculated using Bramson's equation shown on page 214). M.Sharp's model assumes that the reflectivity remains constant (in this case 0.95) up to 500°C, and then falls linearly to zero at the boiling point, to simulate "keyholing". It is unclear however why 500°C was chosen as the temperature at which the reflectivity should begin to decrease. It may have been better to assume a linear decrease in reflectivity up to the melting point (1950K) at which point vapour is produced and a "keyhole" is established and the reflectivity can be regarded as zero. Such a modification of the model would decrease the calculated weld penetration depths bringing them more into line with experimental results. It is
clear that the whole problem of reflectivity requires a great deal of further investigation, so that the model may be modified.

(iii) M. Sharp's model takes no account of variation in the thermal constants with temperature which may lead to some inaccuracy in the theoretical results. As a result the model may require adaptation to take account of the increase in both thermal conductivity and specific heat capacity with temperature. This increase is shown in Fig. 6.48.

(iv) If it is accepted that vigorous stirring in the melt assists its lateral bead growth, as mentioned earlier, then this must be a method of redirecting energy which might otherwise be utilized in increasing the depth of weld penetration.

Weld Bead Width Analysis.

Fig. 6.51 shows the variation in both theoretical and experimental weld bead width with welding speed. It is apparent that both values increase with decreasing welding speed, and that the model underestimates the experimentally measured bead widths. However whereas weld penetration results in some cases only differed by 25% the weld bead width results showed differences of up to 70% at low speeds (approximately 30 mms⁻¹) although at higher speeds the difference was only approximately 30%.

Several factors may explain this discrepancy. One of the most important is that the experimental results were generated from only partially keyholed welds which as a result exhibited a wider bead than would have been present if the welds were fully keyholed.

Another possible explanation for the model's underestimation of the weld bead width is that the model takes no
account of stirring forces within the weld pool. These forces are driven by the surface tension gradient between the weld pool and the adjacent solid metal, as shown in Fig. 6.52. The surface tension gradient and the resulting stirring effect are set up as a result of the temperature difference between the molten metal beneath the beam and the solid metal at the weld interface. Therefore liquid moves from the centre to the side of the weld pool causing erosion of the adjacent solid metal. In this way more lateral melting is caused than would be predicted by this purely thermodynamic model.

Another possible explanation for the underestimation of lateral weld bead spread is that the model takes no account of the geometry of the keyhole. In practice once a keyhole has been established the liquid melt is moved sideways in order to move from the front to the back of the keyhole, as illustrated in Fig. 6.53.

The model is also capable of producing temperature distributions in the x-y plane (i.e. looking down onto the weld zone). From these temperature distributions melting temperature isotherms were produced at four different speeds and are shown in Fig. 6.54. This figure shows two important features.

(i) The extent to which melting occurs ahead of the centre of the incident beam is dependent upon welding speed. Using a speed of 90 mms\(^{-1}\) the melt front-beam centre distance is approximately the same as that of the focussed beam radius. This suggests that melting takes place in a direct manner, as described on page 189. However as the welding speed is decreased to 30 mms\(^{-1}\) the melt front-beam centre distance becomes greater than the beam radius. This result shows that melting ahead of the beam centre must be indirect with heat being transferred by conduction from the directly melted zone underneath the beam.
(ii) Fig. 6.54 also shows the variation in bead width with welding speed which correlates with the experimental trend. However it is also apparent from Fig. 6.54 that the model's estimation of weld bead width is much less than the experimentally measured values, for the reasons discussed on page 220.

Prediction Of HAZ Widths.

The results from the model can also be used to predict the HAZ width. This is achieved by plotting the lowest temperature isotherm at which transformation takes place. In the case of the Ti-6Al-4V alloy a value of 850°C was chosen since below this temperature the martensitic transformation cannot occur and so the metal remains unaffected on cooling. Fig. 6.55 shows a typical theoretical weld bead profile with its adjacent HAZ, produced using a welding speed of 30 mms⁻¹ and a power of 1500W. It is impractical to compare the corresponding experimental result since, as a result of the reasons discussed earlier, the weld bead width is underestimated by the model. As a result Fig. 6.55 compares an experimental weld and HAZ profile with a similar bead width to the theoretical bead profile. It is apparent that the HAZ widths at the top of the bead show a close correlation, showing that for a given weld bead width the model can provide an accurate reflection of the HAZ width. However as the weld depth increases the experimental and theoretical HAZ widths diverge, which is simply a result of the diverging bead widths.

It can be concluded that providing the experimental weld bead volume is similar to the theoretical weld bead volume then a good agreement can be obtained from the model. Any discrepancies may be a result of the model not taking into
account variations in thermal conductivity and specific heat capacity with temperature.

Summary.

It can be concluded that the "Glaze 7" model developed at Imperial College is a useful tool for predicting the effect of laser welding on Ti-6Al-4V. The most important parameter that a welding model should be able to predict is the weld penetration depth. It is apparent from the results in this section that the Imperial College model has had most success in this area. However, the model is less accurate in predicting weld bead widths, which are of less interest. If this aspect of the model is to be improved then modifying the model to take into account problems with reflectivity, variation of thermal constants and heat sink effects due to material thickness the model will be required. Particular emphasis should be placed on the effect of fluid dynamics on the distribution of heat during laser welding as the vigorous stirring experienced by the weld pool generates a broader shallower melt zone than purely thermodynamic models predict.

6.3 Overall Conclusions.

The following conclusions can be drawn from this series of experiments.

Continuous Wave Laser Welding

(a) Full penetration keyhole welds can be produced in both 1mm and 1.7 mm Ti-6Al-4V, and full penetration conduction limited welds in 2.7 mm Ti-6Al-4V using the Control laser (using the maximum power of 1800W). The Coherent Everlase (maximum power 350W) however was only capable of producing full penetration
conduction limited welds in 1 mm thick material.

(b) Porosity is not a problem during laser welding Ti-6Al-4V when full penetration welds are produced. However when full penetration is not achieved then porosity does occur in bead on plate welds.

(c) Joining efficiencies of up to 63 mm$^2$/KJ can be achieved in 1 mm thick material (with the Control laser) compared to only about 30 mm$^2$/KJ in electron beam welding.

(d) The black powder deposited during the laser welding of Ti-6Al-4V is a mixture of $\text{Al}_2\text{O}_3$, $\text{TiO}_2$ and TiN.

(e) Both the melt and HAZ width increase with laser power and decrease with welding speed.

(f) The depth of penetration increases with laser power and decreases with welding speed.

(g) Some grain growth occurs when welding thick (2.7mm) sections of Ti-6Al-4V at low welding speeds.

(h) The striations on the surface of laser welds are a result of a combination of both the ripple in the power supply and of fluid dynamic effects.

(i) Laser butt welds in 1 and 1.7mm thick Ti-6Al-4V alloy sheet were found to exhibit yield strengths greater than those of the parent metal, with slightly lower ductilities.

(j) Careful shielding of the weld zone resulted in optimized laser welds having similar levels of oxygen contamination to electron beam welds.
Pulsed Welding Conclusions.

(a) Joining efficiencies of up to 116 mm$^2$/KJ can be achieved in 1 mm thick Ti-6Al-4V using the Coherent laser in the pulsed mode compared to the continuous wave maximum of 63 mm$^2$/KJ.

(b) It is possible to produce full penetration "keyhole" welds with aspect ratios of less than one in 1 mm thick Ti-6Al-4V using the Coherent laser in the pulsed mode. Whereas the minimum aspect ratio that could be achieved using the continuous wave mode was 1:0.83.

(c) The use of large pulse frequencies results in continuous weld beads. However if low repetition rates are used then discrete spikes of weld metal can be seen.

(d) The pulse frequency determines the pulse shape and peak power. At low frequencies (60Hz) up to 4x the average power output level can be reached during the pulse. Whereas at high frequencies (1000Hz) a peak of only 1.75x average power output is reached.

Heat Flow Conclusions.

(a) The Imperial College model "Glaze 7" is a useful tool with the ability to predict penetration depths within 25% of those achieved in practice.

(b) The model works less well when predicting weld bead widths as it is purely thermodynamic and makes no allowance for fluid flow effects.

(c) The model can also predict HAZ widths very accurately, providing that the theoretical bead volume is similar to that obtained experimentally.
Table 6.1. The variation in depth of focus (Z) with lens focal length (f).

<table>
<thead>
<tr>
<th>Laser</th>
<th>f/mm</th>
<th>D/mm</th>
<th>Z/mm</th>
<th>d (focused spot size measured by the LBA)/mm</th>
</tr>
</thead>
<tbody>
<tr>
<td>Control</td>
<td>100</td>
<td>24</td>
<td>0.65</td>
<td>0.42</td>
</tr>
<tr>
<td>Control</td>
<td>150</td>
<td>24</td>
<td>1.46</td>
<td></td>
</tr>
<tr>
<td>Coherent</td>
<td>63</td>
<td>11</td>
<td>1.23</td>
<td>0.26</td>
</tr>
</tbody>
</table>

Table 6.2. Weld striation frequencies and wavelengths.

<table>
<thead>
<tr>
<th>Welding Speed mms⁻¹</th>
<th>Wavelength μm</th>
<th>Frequency Hz</th>
</tr>
</thead>
<tbody>
<tr>
<td>20</td>
<td>70.1</td>
<td>284</td>
</tr>
<tr>
<td>30</td>
<td>105</td>
<td>286</td>
</tr>
<tr>
<td>40</td>
<td>140</td>
<td>253</td>
</tr>
<tr>
<td>50</td>
<td>161</td>
<td>319</td>
</tr>
<tr>
<td>70</td>
<td>215</td>
<td>307</td>
</tr>
</tbody>
</table>

Table 6.3. Electron beam weld parameters for full penetration welds.

<table>
<thead>
<tr>
<th>Material Thickness mm</th>
<th>Beam Power KW</th>
<th>Welding Speed mms⁻¹</th>
</tr>
</thead>
<tbody>
<tr>
<td>0.95</td>
<td>1.1 (22mA, 50KV)</td>
<td>33</td>
</tr>
<tr>
<td>1.6</td>
<td>2 (33mA, 60KV)</td>
<td>30</td>
</tr>
<tr>
<td>2.7</td>
<td>2 (33mA, 60KV)</td>
<td>22</td>
</tr>
</tbody>
</table>

(Chamber vacuum = 5 x 10⁻⁴ Torr)
Table 6.4. X-Ray conditions used during radiography.

<table>
<thead>
<tr>
<th>Material Thickness/ mm</th>
<th>Voltage/KV</th>
<th>Current/mA</th>
<th>Time/Secs.</th>
</tr>
</thead>
<tbody>
<tr>
<td>0.95</td>
<td>100</td>
<td>10</td>
<td>85</td>
</tr>
<tr>
<td>1.6</td>
<td>100</td>
<td>10</td>
<td>115</td>
</tr>
<tr>
<td>2.7</td>
<td>100</td>
<td>10</td>
<td>150</td>
</tr>
</tbody>
</table>

(Film Focal Distance = 1m)

Table 6.5. Maximum joining rate efficiency values.

<table>
<thead>
<tr>
<th>Material Thickness mm</th>
<th>Welding Speed mms⁻¹</th>
<th>Laser Power W</th>
<th>Joining Efficiency mm²/KJ</th>
</tr>
</thead>
<tbody>
<tr>
<td>1.0</td>
<td>17.5</td>
<td>300</td>
<td>58.00 (Coherent)</td>
</tr>
<tr>
<td>1.0</td>
<td>100</td>
<td>1500</td>
<td>63.33</td>
</tr>
<tr>
<td>1.0</td>
<td>120</td>
<td>1800</td>
<td>66.66</td>
</tr>
<tr>
<td>1.7</td>
<td>30</td>
<td>1500</td>
<td>32.00</td>
</tr>
<tr>
<td>1.7</td>
<td>50</td>
<td>1800</td>
<td>44.00</td>
</tr>
<tr>
<td>2.7</td>
<td>8</td>
<td>1500</td>
<td>14.40</td>
</tr>
<tr>
<td>2.7</td>
<td>12</td>
<td>1800</td>
<td>18.00</td>
</tr>
</tbody>
</table>
Table 6.6. Joining rate efficiency comparison for welding Ti-6Al-4V. (76)

<table>
<thead>
<tr>
<th>Process</th>
<th>Thickness</th>
<th>Power</th>
<th>Welding Speed</th>
<th>Joining Efficiency</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>mm</td>
<td>KW</td>
<td>mms</td>
<td>mm²/KJ</td>
</tr>
<tr>
<td>Plasma arc</td>
<td>2</td>
<td>0.5</td>
<td>1.2</td>
<td>4.8</td>
</tr>
<tr>
<td>Plasma arc</td>
<td>2</td>
<td>0.76</td>
<td>2.3</td>
<td>6.0</td>
</tr>
<tr>
<td>EBW</td>
<td>2</td>
<td>1.17</td>
<td>16.9</td>
<td>28.8</td>
</tr>
<tr>
<td>EBW</td>
<td>1</td>
<td>0.78</td>
<td>16.9</td>
<td>21.7</td>
</tr>
<tr>
<td>GTA</td>
<td>1</td>
<td>0.76</td>
<td>1.5</td>
<td>1.9</td>
</tr>
</tbody>
</table>

Table 6.7. Welding parameters for the tensile test pieces.

<table>
<thead>
<tr>
<th>Material Thickness</th>
<th>Laser Power</th>
<th>Welding Speed</th>
</tr>
</thead>
<tbody>
<tr>
<td>mm</td>
<td>W</td>
<td>mms⁻¹</td>
</tr>
<tr>
<td>1</td>
<td>1800</td>
<td>100</td>
</tr>
<tr>
<td>1.7</td>
<td>2000</td>
<td>60</td>
</tr>
</tbody>
</table>

Table 6.8. Tensile testing results.

<table>
<thead>
<tr>
<th>Material mm</th>
<th>Parent Metal</th>
<th>%Elong</th>
<th>σYield MPa</th>
<th>UTS MPa</th>
<th>Laser Welded</th>
<th>%Elong</th>
<th>σYield MPa</th>
<th>UTS MPa</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>12.0</td>
<td>942</td>
<td>998</td>
<td>10.2</td>
<td>938</td>
<td>1001</td>
<td></td>
<td></td>
</tr>
<tr>
<td>1.7</td>
<td>11.7</td>
<td>934</td>
<td>963</td>
<td>10.0</td>
<td>913</td>
<td>962</td>
<td></td>
<td></td>
</tr>
</tbody>
</table>
Table 6.9. The variation in peak power with pulse frequency (average power = 300W).

<table>
<thead>
<tr>
<th>Pulse Frequency</th>
<th>Peak Output Power</th>
<th>Peak Output Power</th>
</tr>
</thead>
<tbody>
<tr>
<td>Hz</td>
<td>W</td>
<td>Average Power</td>
</tr>
<tr>
<td>100</td>
<td>1312</td>
<td>4.37</td>
</tr>
<tr>
<td>200</td>
<td>674</td>
<td>2.02</td>
</tr>
<tr>
<td>400</td>
<td>562</td>
<td>1.87</td>
</tr>
<tr>
<td>600</td>
<td>544</td>
<td>1.81</td>
</tr>
<tr>
<td>1000</td>
<td>527</td>
<td>1.75</td>
</tr>
</tbody>
</table>

Table 6.10. Maximum joining efficiency values for both pulsed and continuous wave laser welds.

<table>
<thead>
<tr>
<th>Laser</th>
<th>Mode</th>
<th>Power</th>
<th>Material Thickness</th>
<th>Maximum Joining Efficiency</th>
</tr>
</thead>
<tbody>
<tr>
<td>Coherent</td>
<td>Pulsed</td>
<td>300</td>
<td>1</td>
<td>116</td>
</tr>
<tr>
<td>Coherent</td>
<td>CW</td>
<td>300</td>
<td>1</td>
<td>58</td>
</tr>
<tr>
<td>Control</td>
<td>CW</td>
<td>1800</td>
<td>1</td>
<td>66</td>
</tr>
<tr>
<td>Control</td>
<td>CW</td>
<td>1500</td>
<td>1</td>
<td>63</td>
</tr>
</tbody>
</table>
Fig. 6.1. Schematic representation of the laser welding arrangement used for the Control laser.

Fig. 6.2. Schematic representation of keyhole formation.
Fig. 6.3. The variation in weld bead shape and penetration depth for welds produced using the Coherent laser in 1 mm thick Ti-6Al-4V. (x30)
Fig. 6.4. Keyhole welds in 1 and 1.7 mm Ti-6Al-4V. (x30)
Fig. 6.5. The variation in depth of penetration with welding speed in 2.7mm thick Ti-6Al-4V.

Fig. 6.7. The variation in depth of penetration with welding speed for 1.7 and 2.7 mm Ti-6Al-4V.
Fig. 6.6. The variation in weld bead shape with welding speed for welds produced in 2.7 mm thick Ti-6Al-4V using a laser power of 1800W.
Fig. 6.8. Welds produced using welding speeds of 50 mms$^{-1}$ and laser powers of 1500 W, in 1 and 2.7 mm Ti-6Al-4V.
Fig. 6.9. The variation in depth of penetration with laser power in 1mm Ti-6Al-4V, using a welding speed of 30mm/s.

Fig. 6.10. The variation in depth of penetration with welding speed for 100 and 150 mm focal length lenses.
Fig. 6.11. The variation in HAZ and melt width with welding speed in 1.7 mm Ti-6Al-4V, using a laser power of 1500W.

Fig. 6.12. The variation in HAZ width (including melt width) with laser power in 2.7 mm Ti-6Al-4V.
Fig. 6.13. SEM photographs of the surface topography of laser welds. (x1K)
Fig. 6.14. Schematic representation of striation formation in laser welding.

Fig. 6.15. The variation in oxygen contamination in laser welded 1.7 mm thick Ti-6Al-4V, produced using a welding speeds of 30, 50, 70 and 90 mms$^{-1}$ and a power of 1500W. (argon pressure = 100KPa).
Fig. 6.16. Oxygen and nitrogen gradients in a laser weld produced using a welding speed of 30 mm s⁻¹ and a power of 1500 W. (Argon pressure = 278 kPa).

![Graph showing oxygen and nitrogen gradients in a laser weld.]

Fig. 6.17. The oxygen gradients in a typical electron beam weld compared to a laser weld produced at the maximum welding speed (45 mm s⁻¹, 1800 W).
Fig. 6.18. The weld zone structure. (x700)

Fig. 6.19. The HAZ structure. (x700)
Fig. 6.20. The parent metal structure.

Fig. 6.21. The variation in average grain size in fusion zone with welding speed.
Fig. 6.22. Microhardness traverse carried out on a laser weld produced using the Coherent laser (welding speed 32.5 mms⁻¹).

Fig. 6.23. Microhardness traverse carried out on a laser weld produced using the Coherent laser (welding speed 17.5 mms⁻¹).
Fig. 6.24. Microhardness traverse carried out on a laser weld produced using the Coherent laser (welding speed 10 mms⁻¹).

Fig. 6.25. Microhardness traverses carried out on a laser weld produced using the Control laser (welding speed 80 mms⁻¹).
Fig. 6.26. Radiographs of welds produced using laser powers of 1800 W and speeds of between 50 and 6 mms$^{-1}$.
Fig. 6.27. Black deposit produced during laser welding.

Fig. 6.28. The variation in melting efficiency with welding speed for welds produced using the Control laser.
**Fig. 6.29.** Schematic representation of direct melting mechanism.

**Fig. 6.30.** Schematic representation of indirect melting mechanism.
Fig. 6.31. Impact test results for parent metal and laser welded Ti-6Al-4V.

Fig. 6.32. SEM of the fracture surface of a typical unwelded Ti-6Al-4V impact test sample (test temperature = 0°C)
Fig. 6.33 SEM of the fracture surface of a typical laser welded Ti-6Al-4V impact test sample (test temperature = 0°C) showing root porosity.

Fig. 6.34. SEM of the fracture surface of a typical laser welded Ti-6Al-4V tensile test sample showing ductile failure.
Fig. 6.35. SEM of the fracture surface of a typical laser welded 1.7 mm thick Ti-6Al-4V tensile test sample showing cup and cone fracture.

Fig. 6.36. Load extension curve for a typical laser weldment in 1.7 mm thick Ti-6Al-4V sheet.
Fig. 6.37. Schematic representation of a pulsed laser output.

Fig. 6.38. Pulsed laser output produced using a frequency of 60 Hz, with a pulse length of approximately 7 ms.
Fig. 6.39. Pulsed laser outputs produced using a nominal mark:space ratio of 1:1 with an average laser power of 300W.
Fig. 6.40. Schematic representation of the effect of sectioning position upon weld bead morphology.

Fig. 6.41. Pulsed laser welds polished from the underside produced using frequencies of (a) 200Hz and (b) 1000Hz.
Fig. 6.42. Discontinuous solidification in a pulsed laser weld (produced using a frequency of 100Hz and a welding speed of 30 mms$^{-1}$).

Fig. 6.43. Typical pulsed keyhole weld produced in 1mm thick Ti-6Al-4V. (100Hz, 1mS, 35 mms$^{-1}$).
Fig. 6.44. Penetration weld geometry. (29)

Fig. 6.45. Theoretical curve for penetration welds. (79)
Fig. 6.46. Control volume for Mazumder's model (82).

P is the lattice point of the control volume drawn and represents the calculation centre for that elemental volume.

Fig. 6.47. \( T^* \) versus \( Z^* \) through the point of interaction (7).
Fig. 6.48. Variation in thermal conductivity and specific heat capacity with temperature.
Fig. 6.49. Theoretical laser weld bead profiles produced using a laser power of 1500W and welding speeds of 30, 50, 70 and 90 mms⁻¹.
Fig. 6.50. The variation in theoretical and experimental weld penetration with welding speed.

Fig. 6.51. The variation in theoretical and experimental weld bead width with welding speed.
Fig. 6.52. Schematic representation of the effect of surface tension forces within the weld pool.
Fig. 6.53. Schematic representation of the effect of keyholing upon the weld pool.
Fig. 6.54. Theoretical weld bead profile in the longitudinal direction.
Fig. 6.55. Theoretical and experimental (HAZ and weld bead) profiles.
CHAPTER 7. CONCLUSIONS AND RECOMMENDATIONS FOR FURTHER WORK.

The aim of this section is to discuss the results of both the laser cutting and welding chapters highlighting their major conclusions and the implications for the laser processing of other alloys. In addition recommendations for further investigation are also made.

One of the major objectives of the laser cutting programme was to postulate a mechanism by which inert gas assisted laser cutting takes place. Previous work (56) has put forward an oxygen assisted mechanism. However the present study has shown that the two laser cutting mechanisms are fundamentally different. SEM analysis of inert gas assisted laser cut edges has shown there to be two distinct regions present which are, (a) a directly melted zone, and (b) a fluid flow zone.

The striation patterns observed on the directly melted zone surface were created as a result of the power ripple within the laser which was monitored using a laser beam analyser.

A new device developed at Loughborough University, known as the "dross jet", was used during the cutting programme to overcome the problem of dross adhesion to the underside of laser cuts which has been reported by several workers (10). This problem was tackled by Arata, using his tandem nozzle method, with only limited success. It was concluded from the present study that the dross jet was a successful device for producing dross free laser cut edges. However the problem of dross adhesion to the underside of laser cuts is not confined to titanium alloys. Already the dross jet has been used during the oxygen assisted laser cutting of stainless steel (65) and the results of the present cutting programme suggest that the dross jet might be useful when inert gas assisted laser cutting other metals such as zirconium or tantalum. In the case of these metals as well as
titanium alloys the dross jet has two functions. It manipulates the dross away from the component side of the cut, and also provides inert gas protection from deleterious oxidation.

Oxygen contamination is one of the major problems associated with the laser cutting of titanium alloys as a result of the deleterious effects of oxygen upon mechanical properties. Careful attention was paid throughout the laser cutting programme to shielding the cut zone from oxygen contamination. It was concluded that laser cut edges could be produced in thin section (up to 2.7mm thick) Ti-6Al-4V alloy with surface levels of oxygen contamination no greater than those of a conventionally guillotined edge, providing that the shielding gas and laser parameters were optimised. The use of the dross jet, primarily designed to remove dross, was found to provide an efficient argon shield on the underside of the cut zone.

The surface roughness of the laser cuts in Ti-6Al-4V sheet were also investigated in detail. Previous investigations of laser cut edge roughness has concentrated on mild steels (70) which are cut with oxygen as the assist gas. The current cutting programme succeeded in producing laser cut edges in 1 mm thick Ti-6Al-4V alloy sheet with $R_a$ (roughness average) values better than those of a conventionally guillotined edge.

The major drawback of using the Coherent Everlase laser during the cutting programme was that as a result of the maximum power output being only 300W the maximum thickness of Ti-6Al-4V that could be cut was 2.7 mm. In order to cut thicker section material a higher power laser would have been required. The use of high power lasers for laser cutting thick sections has previously been difficult to investigate in detail due to the poor quality of surface finish attainable, as a result of the poor laser mode often associated with such lasers in the past. However more recently developed lasers now produce near Gaussian
modes at powers of up to 2500W (75). The use of such lasers to cut thick section titanium alloys would seem a natural extension of the present study and a potentially interesting area for future research.

The work undertaken during the laser cutting programme concentrated on Ti-6Al-4V alloy but has implications for a wide variety of titanium alloys and gas sensitive metals such as zirconium, hafnium, beryllium, molybdenum and tungsten. These all suffer to varying degrees from embrittlement due to oxygen or nitrogen contamination. As a result similar experimental arrangements could be used to produce low oxygen, dross free, smooth laser cut edges. Potentially the most interesting area for further investigation would be the laser cutting of zirconium alloys which are now finding increasing use in the nuclear industry. However the absorptivity of a metal is also important when deciding whether or not inert gas assisted laser cutting is a feasible option. Titanium has an absorptivity of 0.08 to 10.6μm radiation, whereas molybdenum for example has an absorptivity of 0.027. Thus for such low absorptivity metals inert gas assisted laser cutting may only be feasible with higher power lasers.

Only a small amount of research into the laser welding of Ti-6Al-4V alloy has been undertaken prior to the present work. Although the major problem associated with the welding of titanium alloys is oxygen contamination little attention has been paid in previous work to this problem. One of the major aims of the present study was to examine the extent of oxygen contamination during the laser welding Ti-6Al-4V alloy sheet using Auger analysis. It was concluded that provided proper care was taken with the shielding arrangements laser welds could be produced in Ti-6Al-4V alloy with oxygen levels similar to those present in electron beam welds.
The present laser welding programme has also shown the limitations of using 2KW CO₂ lasers. The major problem is that full penetration welds can only be achieved in Ti-6Al-4V sections up to approximately 3mm thick. Indeed in order to produce "keyhole" welds with aspect ratios less than 1.0 the maximum thickness was approximately 2mm. These results show that whilst 2KW lasers may be used to join Ti-6Al-4V alloy sheet in sections thinner than 2mm, thicker material would require either higher power laser welding or electron beam welding.

Porosity has been reported as a persistent problem during the electron beam welding of Ti-6Al-4V (68). However the present study has shown that this may not be a problem providing that full penetration is achieved.

This investigation has also examined the mechanical properties of laser welded Ti-6Al-4V alloy sheet. Whereas previous work (6) has concentrated on tensile testing, in the present programme tensile testing was augmented with impact testing. It was concluded that there was a 15% drop in absorbed impact energy in laser weldments as compared with the parent metal samples. The tensile results revealed that the laser weldments showed no significant loss of ductility and that all the welded samples fractured in the parent metal.

The problem of oxygen contamination is not unique to titanium and its alloys. Other alloys especially those of zirconium also suffer embrittlement. Thus laser welding with similar shielding arrangements as those employed during the present welding programme might be used to weld zirconium alloys. This would seem an interesting new topic for further investigation especially in view of the low thermal conductivity of zirconium, which would add to the efficiency of the process as a result of the lower conductive losses.
Most CO₂ lasers can not only be operated in the continuous wave mode, but also have a facility for pulsing the laser output. There has been little or no previous investigation into the feasibility of pulsed laser welding Ti-6Al-4V (or any other metals). As a result one of the major objectives of the present study was to investigate the effect of variation in the major pulse parameters upon the weld bead morphology of Ti-6Al-4V alloy. It was concluded that for a given material thickness pulsed welding offered the possibility of producing full penetration "keyhole" welds using a lower average power input than continuous wave welding. These results suggest that with the optimization of the pulse parameters lower power, and less costly, lasers may be utilized in certain welding situations where presently a high power laser is being used.

Precise monitoring of pulsed laser outputs has been difficult in the past, but has been overcome by the use of the laser beam analyser. It is clear from the present study that the laser beam analyser could be used successfully to monitor pulsed laser outputs.

The present study has shown pulsed laser welding to be a feasible alternative to the continuous wave laser welding of Ti-6Al-4V alloy. It is also possible that pulsed welding may be an attractive process for the joining of a wide variety of other metals. In particular the high peak powers attained during pulsed welding might be useful for welding highly reflective and high thermal conductivity materials such as aluminium alloys, where at present very high continuous wave powers are required to produce any weld penetration.

It is clear from this intensive study of laser processing of the Ti-6Al-4V alloy that high power CO₂ lasers are ideal tools for various applications at present carried out by more traditional methods eg mechanical profiling and TIG welding. As confidence in the process spreads throughout the various areas
of industry which use titanium alloys and other materials which are similarly problematic (Zr, Ta, Mo etc.) it is certain that laser cutting and welding will become commonplace.
References

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PROGRAM GLAZET (PARAM,OUTPUT,TAPE10,TAPE5,PARAM,TAPE5,OUTPUT,TAPE11)

VERSION 1.2 34/39/19. M.C.SHARP

THermal ConDUCTION MODEL FOR LASER PROcessING

Based on Welding Model Developed by J. MazUMDER 1977
AND REFINED BY P.S.HENRY 1983

This version is designed to use a Cubic Mesh

written in FORTRAN 77 on the CDC FinF Compiler

common blocks: variable type initialisation: array initialisation

common: X, Y, Z, DELTA

the temperature is the initial temperature: also the ambient temperature

COMMON/CLIMIT/X, Y, Z, DELTA

the number of grid points in each dimension

COMMON/NX, NY, NZ

the index point of the beam centre

COMMON/NP, NPX, NPY, NPZ

surface power input array

COMMON/CSP, CSF, CSB, CSB2, CSB3, CSB4

set the temperatures at which the reflectivity function

COMMON/REFL/X, Y, Z, DELTA

changes from the constant value reflect

COMMON/OMEL/RMT, VMEL

temperature change/length at which the reflectivity function

COMMON/CRNS/UCF, HCF, VCF

time is the total interaction time after n iterations

COMMON/CORNG/OL, X, Y, Z, DIF, DELTA

max time at which the limits of the temperature array

COMMON/CTEM, CTEM+, CTEM-

on which the calculations are made

COMMON/CBEAM, CBEAM, PIOTAL

COMMON/CTHERM, STPT, CONDENS

COMMON/CUPER

the main arrays are dimensioned here: the array dimensions are

DO 10 COMMON/TEMP(1000), PSOF(2000), POER(3000)

dimensionally redefined by each subroutine: mostly as a three

dimensionally array: this feature is only available in the

FORTRAN 77 standard

character: FNAME, INFORM, B30

program data

data: SPT, CONDENS, S30, S33, S37, S38, S39, S40, S41, S42

multi thermal properties: sp., heat, thermal conductivity, density

DATA: RPT, VMEL, T775, 0.5, 0.5, 269000.0

melting data: melt, pt, (m), melt, temp, band, latent heat

DATA: RPT, V3, QVAP, 3300.0, 330.0, 5.279630.0

bulking data

DATA: 35, 75, 0.0

plasma data: 3ier, laMber, cjeFF

DATA: TEMP, HCNV, S35, S36, S37, S38, S39, S40, S41, S42

heat transfer data: melt, temp, convective heat transfer

DATA: RPT, VMEL, T775, 0.5, 0.5, 269000.0

data: nGIO, C11, L11, 0.05, 0.038

initial dimensions of region of interest

DATA: X111, XI12, XI13, XI14

DATA: X30, X33, X36, X39, X42

set jk pays < jk> dp array bounds

call inpary(j30, j31)

inference version 1.2 34/39/15.

WRITE(6, 171) INFORM(128)

calculate thermal diffusivity: alpha

alpha = cond/(density * sp.)

heat process and iteration control parameters

read(15, j1, PIOTAL, <=c1, reflectivity limit, reset

read(15, j2, PIOTAL, <=c1, reflectivity limit, reset

999 continue
READ ?? FMT(1L,5F5.3) INFORM
GOTO 398
999 CONTINUE

STORE REFLECTIVITY VALUE READY FOR MAIN CALCULATION

RFSAVE=REFLEC
REFLEC=0.0

CALCULATE DELTA(I+1,1)/KI

CALL RANGE(OLENX,CXYY,CTHIZ,REAM)
KI=KI(IY+1)/3+1
CALL SEFTEM(ITEMP)

SET SURFACE POWER DISTRIBUTION

CALL PDISP(POWER)

CALCULATE A VALUE FOR DELTIME - A 'WAITING' FACTOR

CALL DELTIME

INITIALISE GRID CALCULATION LIMITS AND ITERATION COUNTERS

BEGIN

IF=IX
IYL=IY
IZL=IZ
ITER=3
TIME=0.0
NLT=0
IPRINT=1
IDDAT=1

START POINT OF MAIN CALCULATION LOOP

BEGIN

200 CONTINUE

CHECK FOR RESET

IF (ITER.EQ.IRESET) THEN
CALL RESET(TIME,TEMP,PROF,POWER)
REFLEC=RFSAVE
IDDAT=1

VIRTUAL

INCREMENT ITERATION COUNTER

ITER=ITER+1

CALCULATE GRID CALCULATION LIMITS

IF (ITER.GE.IRESET+3).AND.(ITER.GT.3.AND.ITER.LT.IRESET)) THEN
ENDN

MAIN FUE CALCS. IF PR=1 PRINTS SUMMARY

IF (IPRINT.EQ.IPRINT) THEN
IPRINT=IPRINT+1
IPRINT=IPRINT
ENDF
CALCULAT SIGMA(0,0,TEMP,PROF,IDDAT)

TIME=TIME+DELTIME

CHECK FOR CONVERGENCE LIMIT IS THE MAX. TEMP CHANGE IN ITERATION

IF (SIGMA.GE.0.01).AND.(SIGMA.LT.LIMIT) GOTO 20

PRINT OUT CONVERGENCE DETAILS

IF (ITER.LT.LIMIT) THEN
WRITE(*,99) ITER
ELSE WRITE(*,100) ITER

ENDIF

CALCULATE WLD PROFL IF JOU1=1

IF (JOUT1.EQ.1) THEN
CALL PROFILE(1,TEMP,PROF)

99 CONTINUE

100 CONTINUE

99 CONTINUE
ENDIF

OUTPUT SUMMARY INFORMATION IF IOUT3=1

IF (IOUT3.EQ.1) THEN
CALL SUMMARY(IOUT1,TEMP,PROF,POWER)
ENDIF

OUTPUT DATA TO TAPE IF IOUT2=1

IF (IOUT2.EQ.1) THEN
CALL OUTPUT(FNAME,TEMP,POWER,PROF)
ENDIF

WRITE(6,'(1X,A1)')GLAZES FINISHED
STOP

END

SUBROUTINE RANGE

SUBROUTINE RANGE(X,Y,Z,X BEAM)

COMM:CLIMT/I,X,Y,Z,DX,DELTA

DELTA=X BEAM/5.0

IF (ITOT.LE.3250) THEN
10 DELTA=DELTA*1.75
20 IF (ITJ+LT.500) GOTO 110
30 IF (ITOT.GT.3250) THEN
40 DELTA=DELTA*1.35
50 IF (ITJ+LT.900) GOTO 203
60 IF (ITJ+LT.900) GOTO 203
70 IF (ITJ+LT.900) GOTO 203
80 IF (ITJ+LT.900) GOTO 203
90 IF (ITJ+LT.900) GOTO 203
100 RETURN

100 FORMAT(6,1000)I,X,Y,Z,DELTA*,1000.

SUBROUTINE PEDIST

SUBROUTINE PEDIST(P,X=Z)

COMM:CLIMT/I,X,Y,Z,DX,DELTA

COMM:BEAM*/RE4,PFIJAL

DIMENSION PD=(IX,Y)

DELTA2=DELTA2*2

PD=PD2

DO 100 K=1,IX
20 DO 100 I=1,IV
30 PD(I)=PD(I)+(K-1)*32+(I-1)*2+DELTA2
40 IF (K.LT.200) THEN
50 X1=(FLOAT(K-1)-5)*DELTA
60 Y2=X1*DELTA
70 X2=FLOAT(I-1)-5)*DELTA
80 Y2=Y1*DELTA
90 PD(I)=PD(I)+Y1*X2
100 IF (I.LT.1) THEN
110 PD=PD+34
120 ELSE
130 PD=PD+(PD2/2.0)
140 ENDIF
150 CONTINUE
160 WRITE(6,'(1X,I10)')PD*2.0
170 RETURN
SUBROUTINE DELTIM

COMMON/CSPEED, DTIME, DELTME, ALPHA
COMMON/SLIM, I, KI, DELTA

DELTME = (ALPHA / (DELTME * K)) + (U / (G * DELTA))

RETURN

END

SUBROUTINE GRLIM

COMMON/GROLIM, TEMP

COMMON/TMP, K, L, N, M, I

DIMENSION TEMP(N, M, I)

TEMP(2) = 2.0

RETURN

END

SUBROUTINE GRLIM() TEMP

COMMON/GROLIM, K, L, N, M, I

COMMON/TMP, K, L, N, M, I

DIMENSION TEMP(N, M, I)

TEMP(2) = 2.0

RETURN

END
00 DELTA42=1.0/(DELTA**2)
20 IF (K.GT.1.AND.I.LT.IX) THEN
30 DZEF=DELTA42*(TEMP(KP1,L)-TEMP)
40 DZF=DELTA42*(TEMP(KP1,L)-TEMP)
50 ELSE
60 ABSORP=1.0-2*F(TEMP)
70 IF (I.EQ.1) THEN
80 DZEF=DELTA42*(TEMP(KP1,L)-TEMP)
90 DZF=DELTA42*(TEMP(KM1,L)-TEMP)
100 ELSEIF (K.EQ.1) THEN
110 DZEF=DELTA42*(TEMP(KP1,L)-TEMP)
120 DZF=DELTA42*(TEMP(KM1,L)-TEMP)
130 ELSE
140 DZEF=DELTA42*(TEMPA-TEMP)
150 DZF=DELTA42*(TEMPA-TEMP)
160 ELSEIF (I.EQ.1) THEN
170 DZEF=DELTA42*(TEMP(KP1,L)-TEMP)
180 DZF=DELTA42*(TEMP(KM1,L)-TEMP)
190 ELSEIF (K.EQ.1) THEN
200 DZEF=DELTA42*(TEMPA-TEMP)
210 DZF=DELTA42*(TEMPA-TEMP)
220 ELSE
230 DZEF=DELTA42*(TEMPA-TEMP)
240 DZF=DELTA42*(TEMPA-TEMP)
250 ENDIF
260 IF (K.GT.1.AND.I.LT.IY) THEN
270 DYSF=DELTA42*(TEMPA-TEMP)
280 DYNF=DELTA42*(TEMPA-TEMP)
290 ELSEIF (I.EQ.1) THEN
300 DYSF=DELTA42*(TEMPA-TEMP)
310 DYNF=DELTA42*(TEMPA-TEMP)
320 ELSEIF (K.EQ.1) THEN
330 DELCOND=U*(TEMPA-TEMPA)/DELTA
340 ELSE
350 DELCOND=U*(TEMPA-TEMPA)/DELTA
360 ENDIF
370 IF (I.EQ.1) THEN
380 FLUX=FLUX+ABSORP/(COND+DELTA)
390 ELSE
400 FLUX=F+FLUX
410 T=T+DELTA
420 ENDIF
430 IF (NLT.EQ.1) THEN
440 IF (L.NE.1) THEN
450 IF (TEMP(KP1,L).*GT.*BPTK.AND.*TEMP(KP1,L).*LT.*BPTK+V) THEN
460 DZUF=DZUF+1
470 ELSEIF
480 IF (BPTK+V.GE.*BPTK+V) THEN
490 DZUF=BPTK+V(FLAT(L)-BPTK+V)*DELTA)
500 ENDIF
510 IF (TH+30.*PTK/3) THEN
520 IF (TH+L*DELTA))
530 ENDIF
540 ENDIF
550 ENDIF
560 ENDIF
570 FLUX=DZEF+F*L+DYSF+DZUF+CZEF
580 RESOU=FLUX+FLUX+ALPHA4*DELCOND
590 ENDIF
600 RETURN
610 ENDIF
620 RETURN
630 ENDIF
640 RETURN
650 RETURN
660 RETURN
670 RETURN
680 RETURN
690 RETURN
700 RETURN
710 RETURN
720 RETURN
730 RETURN
740 RETURN
750 RETURN
760 RETURN
770 RETURN
780 RETURN
790 RETURN
800 RETURN
810 RETURN
820 RETURN
830 RETURN
840 RETURN
850 RETURN
860 RETURN
870 RETURN
880 RETURN
890 RETURN
900 RETURN
910 RETURN
920 RETURN
930 RETURN
940 RETURN
950 RETURN
960 RETURN
970 RETURN
980 RETURN
990 RETURN
SUBROUTINE RESET

COMMON/CLIMIT/I,L,Y,IZ,IX,KI,DELTA
COMMON/CTEMP/TEMP
COMMON/CHANGE/CHANGE
COMMON/TEMP/TEMP
COMMON/CREAM/CREAM
COMMON/CGRD/KS,KF,ILY,IYL
DIMENSION TEMPL(IX,IY,IZ),PROF(IY,IZ),POWER(IY,IY)

CALL PROFILE(IY,IY,PROF)
WRITE(*,*(IX,1))"CENTERLINE TEMPERATURES"

KI=KI+5
WRITE(5,1000)(TEMP(J,J,1),J=KI,KI+5)
TSET=TEMPA+T
IF (POWER(IY,IY,LT,TSET)) GOTO 999
DO 100 I=IY+1
   IF (POWER(N+1,N+1,LT,TSET)) GOTO 115
   DO 200 CONTINUE
   V=IY*DELTA
   IF (I-EQ.0) THEN
      DO 300 L=1,LZ-1
         V=L
      ENDIF
      DO 400 J=1,IX+1
         N=I+1
         IF (POWER(N,N,N+1,LT,TSET)) GOTO 350
         DO 300 CONTINUE
         N=N+1
         IF (N.EQ.1) THEN
            XLI=KI*DELTA
            Z=N-1
            Z=Z*DELTA+(TEMP(IX-1,J)+((TEMP(N,N,N)-TSET)/
            *(TEMP(N,N,N)-TSET)))
         ENDIF
         DO 400 J=1,IX+1
         N=I+1
         IF (N.EQ.1) THEN
            TLI=KI*DELTA
            Z=N-1
            Z=Z*DELTA+(TEMP(IX-1,J)+((TEMP(N,N,N)-TSET)/
            *(TEMP(N,N,N)-TSET)))
         ENDIF
         XT=XL+XLI
         CALL RANGE(IY,IX,IK,IZ,XT,IT,IT)
         CALL JQIST(II)
         CALL DELTM
         CALL SETTEMP(TEMP)
   ENDIF
WRITE(*,*(IX,1))"RESET COMPLETED--RUN RESTARTING"
WRITE(5,1000)IY,ILY,ILY,*"MESH SIZE IS(4,1),DELTA=100".
WRITE(1,1000)IX,IX,IX,*"MESH SIZE IS(4,1),DELTA=100".

FUNCTION POWERIN

COMMON/BEAM/R3EAN,PTOTAL
DIMENSION F(i1,i1)
FUNCTION (R) = EXP(-2.0*(X/R3EAN)**2)
A = 2.0*PTOTAL/(3.142*(R3EAN**2))
SUM = 0.0
G = (X**2-X1)/10.0
DO 100 I1 = 1,i1
DO 500 J1 = 1,i1
S = FLOAT(I1-1)*G+X1
Y = FLOAT(J1-1)*H+Y1
SUM = SUM + F(I1,J1) + SUM + SUM
END

POWERIN = A*SUM**3.0
RETURN
END

SUBROUTINE LATTRAP

FUNCTION T2

CONVERT TEMP TERMS ALL ALONG FOR LATENT HEAT TO THE ENERGY
EQUIVALENT NOT ALL ALONG FOR LATENT HEAT

SUBROUTINE OUTPUT
FUNCTION RF(T)

COMMON/REFL,RFLEC,T1,T2

IF(T.LE.T1) THEN
    RF=REFLEC

ELSE
    RF=RFLEC*(T2-T)/(T2-T1)

END IF

RETURN

END

FUNCTION RF(T)

COMMON/REFL,RFLEC,T1,T2

IF(T.LE.T1) THEN
    RF=REFLEC

ELSE
    RF=RFLEC*(T2-T)/(T2-T1)

END IF

RETURN

END

FUNCTION RF(T)

COMMON/REFL,RFLEC,T1,T2

IF(T.LE.T1) THEN
    RF=REFLEC

ELSE
    RF=RFLEC*(T2-T)/(T2-T1)

END IF

RETURN

END

FUNCTION RF(T)

COMMON/REFL,RFLEC,T1,T2

IF(T.LE.T1) THEN
    RF=REFLEC

ELSE
    RF=RFLEC*(T2-T)/(T2-T1)

END IF

RETURN

END

FUNCTION RF(T)

COMMON/REFL,RFLEC,T1,T2

IF(T.LE.T1) THEN
    RF=REFLEC

ELSE
    RF=RFLEC*(T2-T)/(T2-T1)

END IF

RETURN

END
<table>
<thead>
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<th>DIMENSIONS OF REGION OF INTEREST (MM)</th>
<th>8.0254</th>
<th>1.1606</th>
<th>2.7000</th>
</tr>
</thead>
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<td>640.95</td>
<td>2049.4</td>
<td>3515.1</td>
</tr>
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<td>2977.4</td>
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<tr>
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<tr>
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<td>1659.4</td>
<td>3902.0</td>
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</table>
### Convergence Criterion Satisfied

1. Model Run Summary Output

### Sheet Parameters

- **Temperature**: 1500 (8302.5)
- **Latent Heat of Melt**: 602.5
- **Specific Heat**: 0.5
- **Emissivity**: 0.2
- **Temperature Profile**: 1.2
- **Grid Data**: 1500

### Thermal Conductivity

- **Conductivity**: 602.5
- **Diffusivity**: 0.2

### Material Properties

- **Density**: 0.5
- **Thermal Expansion**: 0.5

### Energy Content

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### Surface Temperature Distribution

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<th>Y (mm)</th>
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### Energy Distribution

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</tbody>
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### Additional Information

- **Latent Heat**: 602.5
- **Specific Heat**: 0.5
- **Emissivity**: 0.2
- **Temperature Profile**: 1.2
- **Grid Data**: 1500
- **Material Properties**: Density 0.5, Thermal Expansion 0.5
- **Energy Content**: 0.5
- **Surface Temperature Distribution**: X (mm) Y (mm) Temperature
- **Energy Distribution**: X (mm) Y (mm) Energy

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### Notes

- All values are approximate and subject to experimental error.
- Further analysis is required for precise calculations.
- The data presented is for educational purposes only.
