

LASER SURFACE TREATMENT OF
ALUMINIUM ALLOY SUBSTRATES

by

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In the Name of Allah(C.C.) (God), the most Gracious,
the most Merciful.

ABSTRACT

Laser surface treatment is the general name of surface modification processes such as surface alloying, cladding, particle injection, transformation hardening etc. by laser which alter the physical and/or chemical properties of surfaces.

In this work, laser surface alloying, cladding and particle injection of aluminium alloy (LM25) substrates were carried out using silicon, aluminium, mild steel, silicon carbide and aluminium oxide powders and mixtures of these powders.

Laser surface alloying is the process of modifying surface composition of a substrate by adding alloying elements into the laser generated melt pool on the substrate. Addition of alloying elements was carried out by a pneumatic screw feeder in form of powder. A 2kW CW CO₂ laser was used and various operation parameters such as scan speed, powder flow rate, beam diameter were investigated. Surface topographies of laser treated samples were examined. Microstructural examinations of both polished and fractured surfaces were undertaken by optical and scanning electron microscopy. X-ray diffractometry was used to identify phases. Compositional analyses were carried out on an SEM fitted with an EPMA instrument. Microhardness of many samples was measured. The microstructural features of hypereutectic Al-Si alloys, proeutectic silicon, proeutectic aluminium refinement, eutectic nucleation and growth are

reported.

Laser surface cladding using mild steel+aluminium and mild steel powders was investigated. Microstructural and phase analysis showed that different intermetallic phases with cellular and dendritic structures occurred. Cracking was the main problem making results unsatisfactory.

An investigation of laser particle injection of aluminium alloy (LM25) substrates with SiC powder was undertaken. No deep injection was obtained; however, shallow injection of carbide particles was successful. The microstructure of injected layers consists of undissolved SiC particles, dissolved and resolidified SiC particles and some other decomposition products.

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CHAPTER 1

INTRODUCTION

The first demonstration of laser action was in 1960. The laser was built by T H Maiman at Hughes Aircraft using synthetic ruby which is formed by the addition of chromium impurity to aluminium oxide. Since that time vigorous research and development have led to a rapid sustained growth in the number of laser types, in the output power produced, and in the scope of their applications. Laser technology had the reputation of being a tool looking for applications. A wide variety of solid, liquid and gaseous materials have been made to lase, with emissions that have ranged from ultraviolet to the infrared region of the spectrum, for isolated pulsed flashes or continuous wave beams, and with average powers from microwatts to many kilowatts.

Flexible control of the laser as a thermal energy source broadened the processing applications from welding, cutting, and drilling using a high intensity focused spot, to hardening, tempering, glazing, alloying, and cladding using a lower intensity beam.

The use of aluminium and its alloys during the past thirty years has been extending faster than many other metals including iron and copper. There are numerous reasons for this, the most important ones being: high strength to weight ratio, excellent corrosion resistance, ease of fabrication, high electrical and thermal conductivities, low cost and high scrap value. These properties together with a

virtually inexhaustable supply for many years in the future have been encouraging industry to research new metallurgical processing techniques to allow the full potential of this versatile material to be realised.

Aluminium alloys are used in manufacturing several components in the automotive industry; however, they are not used in places where they need good wear resistance combined with hardness.

Until recently research efforts into the improvement of wear resistance properties have been restricted to developing new aluminium alloys. The most common alloying addition to aluminium to improve bearing properties has been silicon. Silicon improves castability, increases strength to weight ratio, improves wear resistance and reduces the coefficient of thermal expansion. Hypereutectic aluminium silicon alloys have been used in internal combustion engines as cylinder blocks, cylinder heads and pistons. However, aluminium-silicon alloys are prone to scuffing under conditions of poor lubrication such as those that exist during starting and warm up of an engine. For this reason aluminium alloy cylinder blocks usually have cast iron or steel liners and inserts for bearing surfaces which are expensive and reduce the significance of the benefits gained by using aluminium alloy. Therefore, automobile manufacturers are interested in possible alternative techniques to using cast iron and steel liners and inserts.

Due to interest in the automotive industry in

aluminium-silicon alloys, the development of wear resistant, hard, light weight, high silicon hypereutectic Al-Si alloys, using non-conventional methods, is important. Therefore the practical aim of the present work is to investigate the potential for producing hard, wear and/ or corrosion resistant surfaces for automotive applications and other applications.

Laser techniques provide potential for surface alloying, cladding and particle injection to be carried out. Rapid solidification can produce considerable refinement of structure in relation to dendrite arm spacing and interphase spacing in the eutectic. The alloying/cladding element can be preplaced on the substrate or can be continuously fed into ~~the~~ laser melt pool as a powder. The latter method has some advantages over the first one, as discussed in this thesis.

Requirements for the surfacing may differ according to the particular component e.g. shape, area of surface to be covered and depth of treatment. Therefore, the effect of processing parameters on track geometry were studied. In some applications, the laser processed surface finish may be suitable for use without machining. Generally, some machining is required to remove the part of the bead to produce a flat surface.

Several investigations on aluminium have been done with silicon and other powders by several investigators in recent years. (9,10)

The present research work is concerned with the laser

surface alloying/cladding, and particle injection of aluminium alloy (LM25) substrates, using Si, Al, SiC, Al_2O_3 , mild steel powders and the mixture of these powders. The influence of process parameters on alloying, cladding and particle injection characteristics are reported. Microstructural analysis of laser treated samples were carried out using both optical and electron microscopy facilities. Composition microanalysis and phase analysis were done utilising both EPMA and X-ray techniques. Micro hardness values of processed samples were measured.

Additionally, some development work which was carried out on the powder feeding system is also presented.

CHAPTER 2

LITERATURE SURVEY

2.1 Lasers in Materials Processing

Though many materials have now been made to lase and though there are now commercially available many types of lasers, most materials processing is done by only a few.(1) An interesting review of the development of commercial lasers has been given by Klauminzer in Ref.(2)

Lasers can be classified as ultraviolet, visible and infrared lasers according to the wavelength at which they emit radiation. For materials processing purposes, infrared lasers occupy the most important place. Infrared lasers are available in two major groups.(3-6)

1) Near-infrared; solid state lasers e.g. YAG or Nd glass which emit light at the 1.06 μm wavelength, ruby at the 0.694 μm wavelength.

2) Far-infrared; CO_2 lasers which emit light at 10.6 μm wavelength.

Lasers are generally named after the "medium" from which the laser light is extracted. The near-infrared, solid state lasers of significance here are:

- 1) Nd: YAG (Neodymium doped yttrium-aluminium garnet).
- 2) Nd: Glass (Neodymium doped glass).
- 3) Ruby.

Three fundamental elements, "medium, excitation source,

and optical resonator" are common to all lasers.(1,3-5)

The former (Nd: YAG) employs yttrium aluminium garnet and can tolerate continuous operation for power outputs of several hundred watts. Average power outputs of 1 kW or more can be obtained with pulsed operation.

The use of glass rather than YAG allows economic fabrication of large laser rods and large energies per pulse. However, the low conductivity limits the repetition rate to about 1 pulse per second at maximum power. It is particularly suitable for drilling applications. For commercial solid state metal working lasers, single pulse output values are from a few millijoules to few hundreds of joules.(1,4)

Ruby lasers emitting at $0.694\text{ }\mu\text{m}$, contain triply-ionized chromium, Cr, in crystalline aluminium oxide as host and commercial products are restricted to low repetition rate pulsed operation of 1 pulse/sec, like Nd: glass lasers. The normal pulsed output is actually an envelope of random spikes and particularly suited, again like the glass laser, to drilling applications. Presently the role of the ruby laser in materials processing is being overshadowed by its application in diagnostics such as holography.(1,2,4-6)

The CO_2 laser can operate on CO_2 alone but normally employs a mixture of CO_2 , N_2 and He for vastly greater efficiency. The nitrogen greatly increases the excitation efficiency and is usually present with a partial pressure a few times greater than that of the CO_2 . Helium comprises about three-quarters of the total gas mixture and is provided to dissipate the heat from the discharge. Despite

the very high reflectivity of many metals at 10.6 μm wavelength the high powers available from the CO_2 laser and its reliability have led to its use throughout the whole spectrum of materials processing. Further the CO_2 laser can be operated in a pulsed mode by choice rather than necessity, in addition to the CW (continuous) mode. Higher peak powers can thereby be obtained, greatly facilitating some processing applications such as drilling.

CO_2 lasers are all electrically excited. They all circulate the lasing medium for the purpose of cooling. It is in the design of the gas circulating system that the most obvious differences occur.(1,3-5) In fact the type of the gas circulating is often used to characterize the various lasers such as:

- 1) Slow axial flow (Fig.2.1)
- 2) Fast axial flow (Fig.2.2)
- 3) Transverse flow (Fig.2.3)

These lasers range physically from desk-size to room size, and their power outputs may exceed 20 kW average power.(1-3)

Excimer lasers whose development began in 1975 have a high potential for industrial processing because, many materials have very high absorption at the excimer wavelengths. These are typically in the ultraviolet region, such as 0.193 μm for ArF or 0.308 μm for XeCl. The short wavelength allows even smaller focused spot sizes than with a ruby laser. These wavelengths will produce

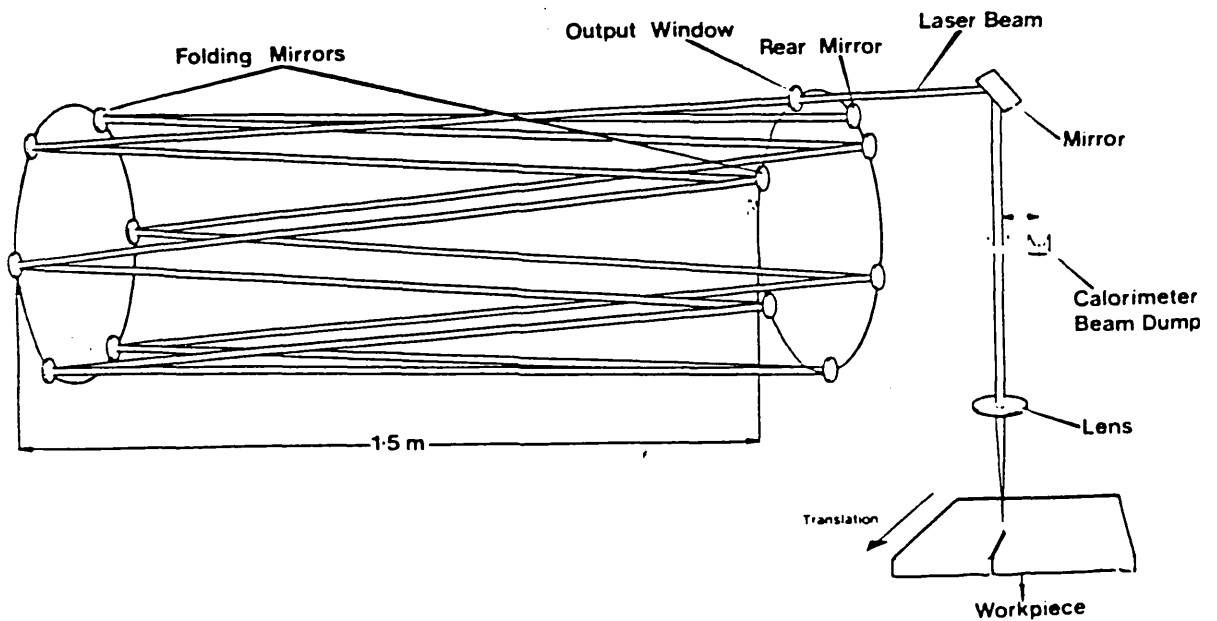


Fig 2.1 Slow Axial-Flow CO₂ Laser (courtesy Ferranti Ltd) (5)

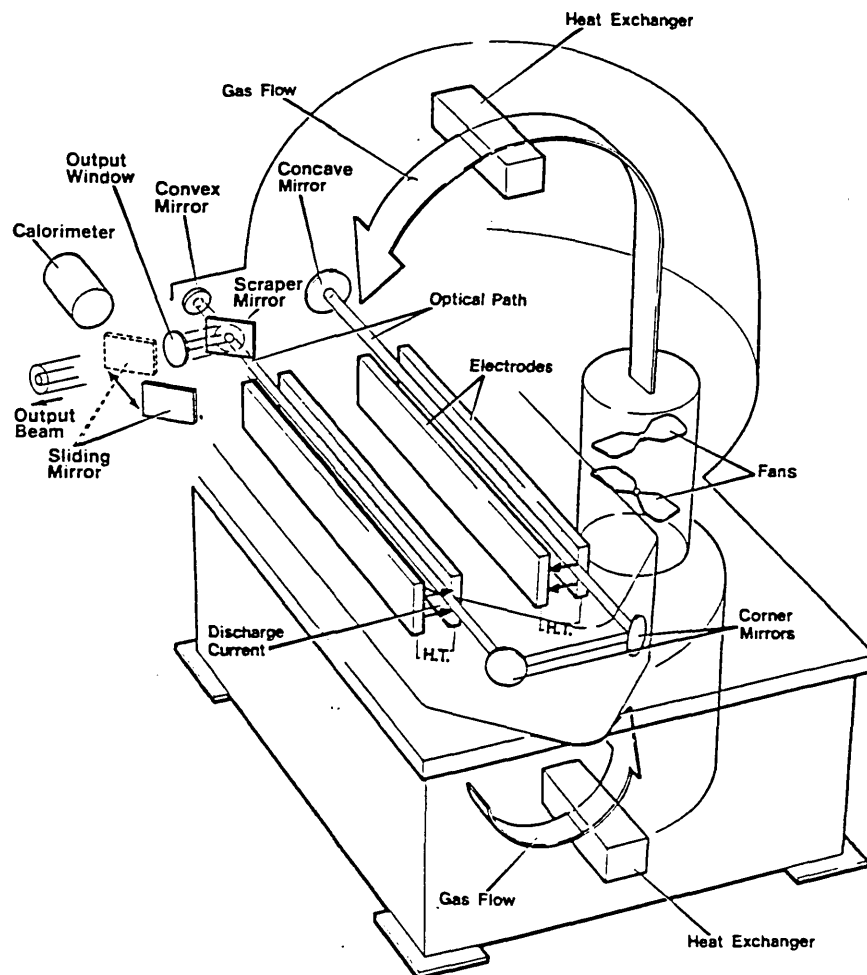
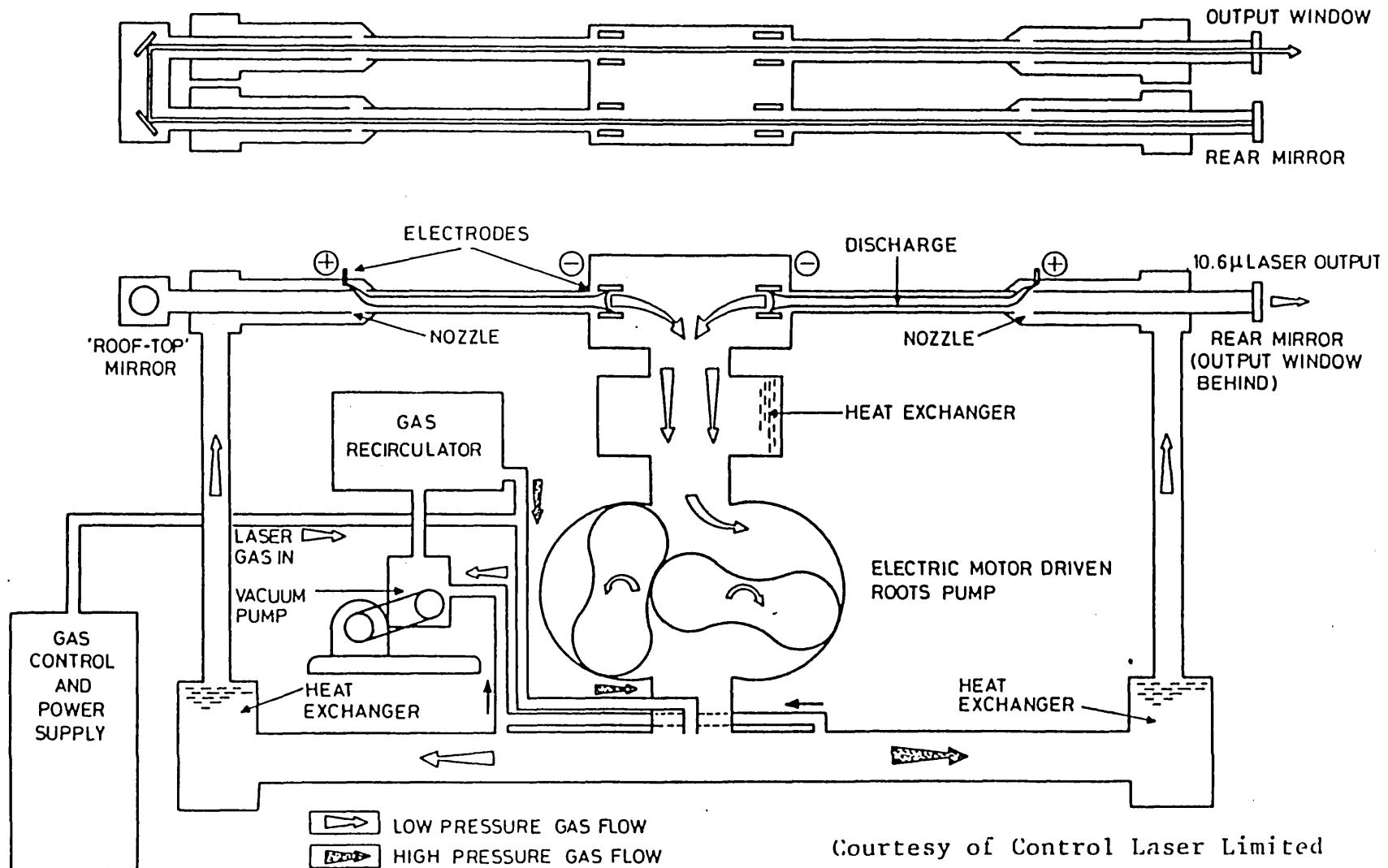


Fig.2.3 Fast Transverse-Flow 5kW CO₂ Laser (schematic) - Culham Laboratory (5)

OVERHEAD VIEW OF LASER PATH



Courtesy of Control Laser Limited

Fig.2.2 Fast axial-flow, (5)

photoassociation in some materials. Thus, very selected areas of the surface can be treated with little heating of adjacent areas. This attracts considerable interest in marking applications.(1)

The useful output power from the laser is usually no greater than 10% (for CO₂) and often less than 1% (for Ruby) of the total power input so that a major consideration in the designing and operation of lasers is the dissipation of unwanted heat.(1,3-5)

Important requirements for lasers which are used for industrial machining include (5):

- a) Sufficient power at the work (cw, or pulsed).
- b) Controlled focal intensity profile.
- c) Reproducibility (of power, modes, polarisation etc.).
- d) Reliability (long interval between services).
- e) Capital and running costs which are economic for the application.(5)

TABLE 2.1 Lasers Commonly Used for Material Processing

Laser Medium	Wavelength	Average Power or Pulse Energy
CO ₂	10.6 μm	< 20 kW; cw, pulsed
Nd:YAG	1.06 μm	< 1 kW; cw, pulsed
Nd:Glass	1.06 μm	< 100J; pulsed
Ruby	0.694 μm	< 100J; pulsed
Excimer (XeCl)	0.308 μm	< 1J; pulsed

The basic lasers which presently dominate in materials processing application are shown in Table (2.1)(1)

2.2 Laser Beam - Material Interaction

Assuming controlled and consistent delivery of some form of laser energy to a work surface, one may consider the mechanisms by which the laser light energy is absorbed by the materials surface. In most laser-induced materials processing the absorbed light generates heat and hence the temperature necessary for the modification of the work piece. In the majority of cases one or more phase changes occur and can consume much of the energy. The thermal conductivity, k , and the thermal diffusivity of the substrate, α , ($\alpha = \frac{k}{\rho \cdot C}$), determine how rapidly the heat is conducted away from the irradiated region and the consequent temperature distribution. Here, C , is the specific heat, ρ , is density. Heat transfer by radiation from the surface and by convection are often assumed negligible by comparison with conduction; this approximation, although convenient, becomes less acceptable at temperatures much above the melting point of the substrate.(1)

Comparison of radiant loss to conduction loss:

$$q_r = \sigma A_r \varepsilon T^4$$

$$q_c = k A_c \frac{dT}{dx} \cong \frac{k A_c T_z}{\sqrt{\pi \alpha \frac{D}{V}}}$$

$$A_c = \frac{2\pi D^2}{4} = \frac{\pi D^2}{2}$$

$$A_r = \frac{\pi D^2}{4}$$

$$\frac{q_r}{q_c} \cong \frac{\sigma \varepsilon T^4 \sqrt{\pi \alpha \frac{D}{V}}}{2 k T} \cong \frac{\sigma \varepsilon T^3}{2 k} \sqrt{\pi \alpha \frac{D}{V}}$$

$$\sigma = 5.67 \times 10^{-8} \text{ W/m}^2 \text{ K}^4$$

$$\varepsilon = 0.5$$

$$T = 2000 \text{ }^\circ\text{K}$$

$$k = 204 \text{ W/mK}$$

$$= 54 \text{ W/mK}$$

$$\alpha = 0.84 \times 10^{-4} \text{ m}^2/\text{s}$$

$$= 0.148 \times 10^{-4} \text{ m}^2/\text{s}$$

$$D = 0.4 \times 10^{-3} \text{ m}$$

$$V = 0.1 \text{ m/s}$$

$$\frac{q_r}{q_c} = \frac{5.67 \times 10^{-8} \times 0.5 \times 8.10^9}{2 \times 54} \sqrt{\frac{\pi \cdot 0.148 \times 10^{-4} \times 0.4 \times 10^{-3}}{0.1}} \cong 1 \times 10^3$$

q_r = Heat flow by radiation (W)

q_c = Heat flow by conduction (W)

σ = Boltzmann Constant ($5.67 \times 10^{-8} \text{ W/m}^2 \text{ K}^4$)

k = Thermal conductivity (W/mK)

ε = Emissivity of surface

T = Temperature ($^\circ\text{K}$)

A_r = Area of radiation (m^2)

A_c = Area of conduction (m^2)

D = Beam Diameter (m)

V = Scanning velocity (m/s)

α = Thermal diffusivity (m^2/s)

Some materials possess a combination of thermal and optical parameters rendering them difficult to work. For example, the high thermal conductivity of copper and aluminium removes heat very rapidly from the region of treatment. Further the very high reflectivity of these metals in the solid state gives very inefficient coupling of the incident light into the material.(1,4,7)

The surface modifications of widest industrial impact are likely to be practised on iron-based materials. For metals in general, it is at least safe to state that absorption of laser energy is poor. In some cases, the thermal and optical parameters may make laser processing too inefficient to be commercially viable.

2.3 Reflectivity of Materials

Laser beam coupling and absorption of energy is of major importance for materials processing. Clean, machined surfaces are poor absorbers of infrared light at temperatures below the melting point. 85-95% of the incident laser beam can be lost by reflection.

The reflectivity of a given surface depends on the following properties.(7-10)

- 1) Materials.
- 2) Surface finish.
- 3) Temperature.
- 4) Wavelength of beam

The effects of these properties on reflectivity have been investigated by several researchers.(7,8,11,12) A temperature dependence relationship between emissivity and electrical resistivity was derived by Bramson (4)

$$\epsilon_{\lambda_w}(T) = 0.365 \left[\frac{P_r(T)}{\lambda_w} \right]^{1/2} - 0.667 \frac{P_r(T)}{\lambda_w} + 0.006 \left[\frac{P_r(T)}{\lambda_w} \right]$$

where: P_r = Electrical resistivity at a temperature ($\Omega \cdot \text{cm}$)

ϵ_{λ_w} = Emissivity of substrate at T ($^{\circ}\text{C}$)

λ_w = Wavelength of radiation (cm)

for a model of photon absorption by electrons on an optically flat surface.

Roughened metals have a considerably higher absorption coefficient than polished metals because as a result of multiple reflections or higher surface dislocation density a larger part of the incident beam is trapped. An absorption coefficient of metals has been measured under several conditions as shown in Fig.(2.4) The absorption coefficient of oxides is considerably larger than that of metals.(4,7,8) Therefore, formation of an oxide layer on the substrate changes the counterface materials (from metal to oxide) of laser beam.

In order to increase the efficiency of beam coupling the following surface alteration methods can be applied.(7-13)

- 1) Covering the surface with materials such as graphite,

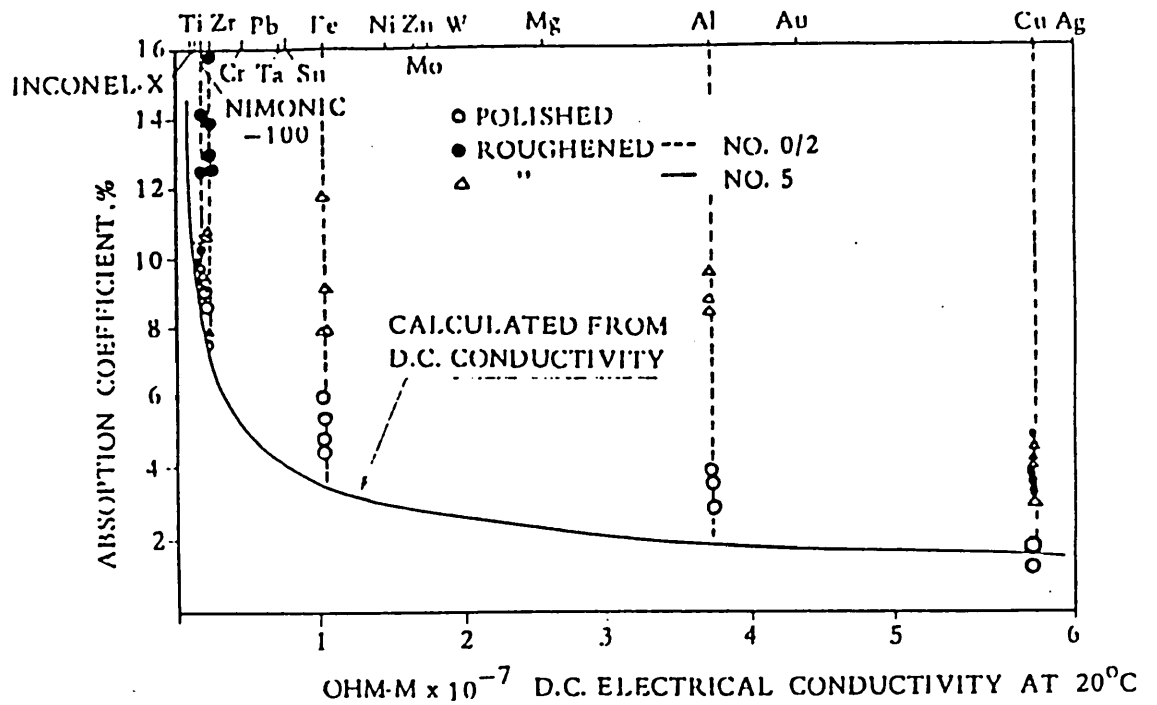


Fig.2.4 Absorption coefficient of roughened and polished metals. A solid line indicates the theoretical value calculated from the dc conductivity. (7)

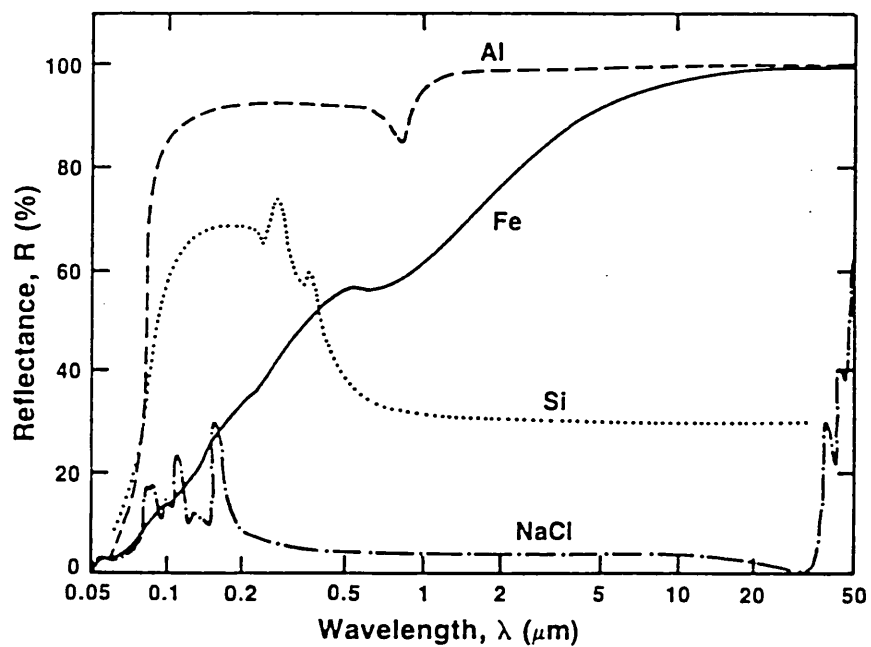


Figure 2.5 The reflectivity of some metals (Fe, Al), a semiconductor (Si), and rocksalt (NaCl) as a function of electromagnetic wavelength. The data correspond to normal incidence and 20° C. (1)

MnO , MoS , black paint or graphite/sodium silicate. If however, melting occurs these materials can affect the composition of the surface.

2) Shot blasting the surface (this will change the reflectance of the surface finish).

3) Using the reflective dome or nozzle (this makes possible reuse of the reflected energy).

4) Using a polarised beam incident at the Breuster angle.

Metals are more absorptive at shorter wavelengths than 1 μm . But at the wave length of the CO₂ laser (10.6 μm) the reflectance of metals such as aluminium, or even iron is very high so that the coefficient of coupling is only a few percent at the best.(1,11,12) Fig.(2.5) shows the normal incidence, room temperature reflectivity of several materials as a function of wave length.(1)

2.4 Materials Processing with Laser

The possible variations of the laser energy widen the processing applications. High energy levels achieved with a focused beam is used for welding, cutting, and drilling. Lower energy is achieved using a defocused or broadened beam and is used for hardening, annealing, glazing, alloying, and cladding using a lower intensity.(1,3,4,9,10,14-17) Table.2.2 shows the major classifications of laser processing.(1)

The growth of significant interest in the use of lasers for surface modification of metals began much later than it did for cutting, drilling, and welding of thin sheets. This reflects the development time necessary for the appearance of the high power lasers, the need to develop a market for this new technology and the need for the laser to prove it gave a high quality product. The surfacing possibilities of a laser were new. Laser surface treatments have got some advantages; these are: (9,10,14)

- 1) The process permits localised treatment.
- 2) Cleanliness of process.
- 3) Controlled heat input; therefore thermal distortion is low and heat affected region is minimised.
- 4) The process is easy to automate.
- 5) Non-contact processing, and inaccessible areas can be treated.
- 6) Complex shapes can be treated.

TABLE 2.2 Major Classification of Laser Processing (1)

Sub-Melting	Melting	Vaporization
Annealing	Cladding	Cutting
Hardening	Alloying	Scribing
	Glazing	Trimming
	Welding	Stripping
	Soldering	Drilling
	Grain Refining	Marking (Engraving)
	Cutting	Shock Hardening

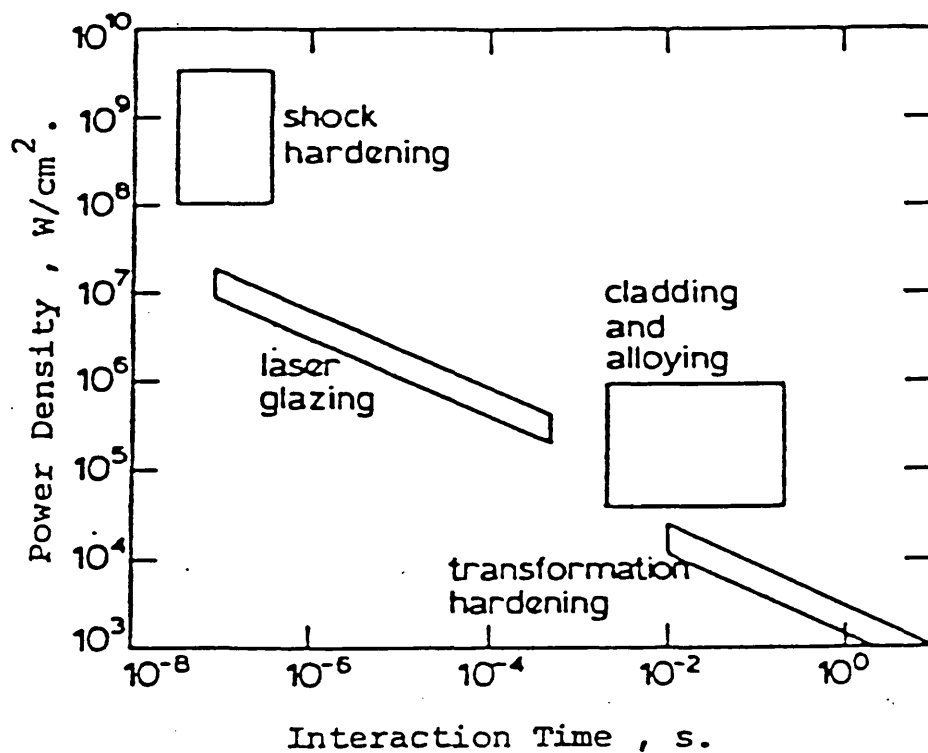


Fig.2.6 Operational regimes for laser surface treatment.(15)

7) It provides less after-machining.

8) It is easy to use between several work stations.

Operational regimes for laser surface treatment of materials are shown in Fig.(2.6) (15)

2.5 Laser Surface Alloying (LSA)

The process consists of melting a substrate surface to a controlled depth and width and adding alloying elements to the molten pool to change the composition of the treated area. The alloying element can be placed on the melt zone area in the form of preplaced powder, foil or by painting, electroplating, vacuum evaporation, wire feed, pneumatic powder feed, diffusion, ion implantation, and reactive gas. (1,3,9,10,17-21)

Most materials can be alloyed into different substrates. The high quench rate ensures that ^{the scale of} segregation is minimal.

The alloyed region shows fine micro-structure and it improves hardness, and mechanical properties.

The process makes possible a thickness of alloyed layer from a few microns to a few millimeters. Very thin layers can be obtained using pulsed lasers. (17-20)

Several investigations have been carried out on the alloying of one or more elements or alloys such as Cr, C, Mn, Al, Ni, Si, Mo, B, N, stainless steel, mild steel on both ferrous and the non-ferrous substrates. (9,10)

Several ferrous and non-ferrous materials have been

treated by LSA. Mild steel, stainless steel, and cast iron are the most important materials in the ferrous group. More detail about LSA of ferrous materials can be found in Ref.(21-27)

The non-ferrous group, include titanium, aluminium and their alloys, superalloys, copper, and nickel; of these copper and aluminium have been found difficult to laser process due to their high reflectivities and thermal conductivities. In Refs.(23,28-31) more information has been given about several non-ferrous materials such as aluminium, titanium and nickel which have been treated by LSA.

The results of laser surface alloying may produce a surface material layer which has never previously been studied. Thus, lack of fundamental knowledge may hinder the development of this very promising laser surface modification method.

Development of alternative processes, such as cladding, plating, carburizing, boronizing, and chromizing, can offer better results.

2.6 Laser Surface Cladding

Laser surface cladding is a technique similar to conventional hardfacing in which the laser beam melts either a preplaced or continuous fed material causing it to flow over the surface of the substrate with the minimum of dilution to the substrate. Then the clad layer freezes and forms a protective or wear resistant layer. By overlapping

single tracks large areas of cladding can be achieved.(14)

Process variables are similar to those for laser alloying. The main characteristics of the laser cladding process are:(9,10,14)

- a) Localized heating with minimum thermal distortion.
- b) Smooth surface finish.
- c) Changeable clad thickness.
- d) Good fusion bonding.
- e) Fine microstructure.
- f) High process flexibility and easy automation.
- g) Controllable levels of dilution and shape of clad.
- h) Non-contact method of application.
- i) Minimum surface preparation.

The powder can be fed not only as a single powder alloy but also either as a premixed powder or as separate streams from different hoppers. Optical feed back (re-use of reflected energy) and vibro laser cladding methods have been examined. Ref.(14,32-34)

As strategic material costs and availability continue to fluctuate in our uncertain world, all those processes which reduce the use of these materials become increasingly attractive. Laser cladding offers the opportunity to produce surface layers of rare and/ or expensive materials on substrates of more common materials. Even in the absence of strong cost driven material selection, the unique combinations of surface layers and substrates may themselves be adequate justification for use of the process. Hard,

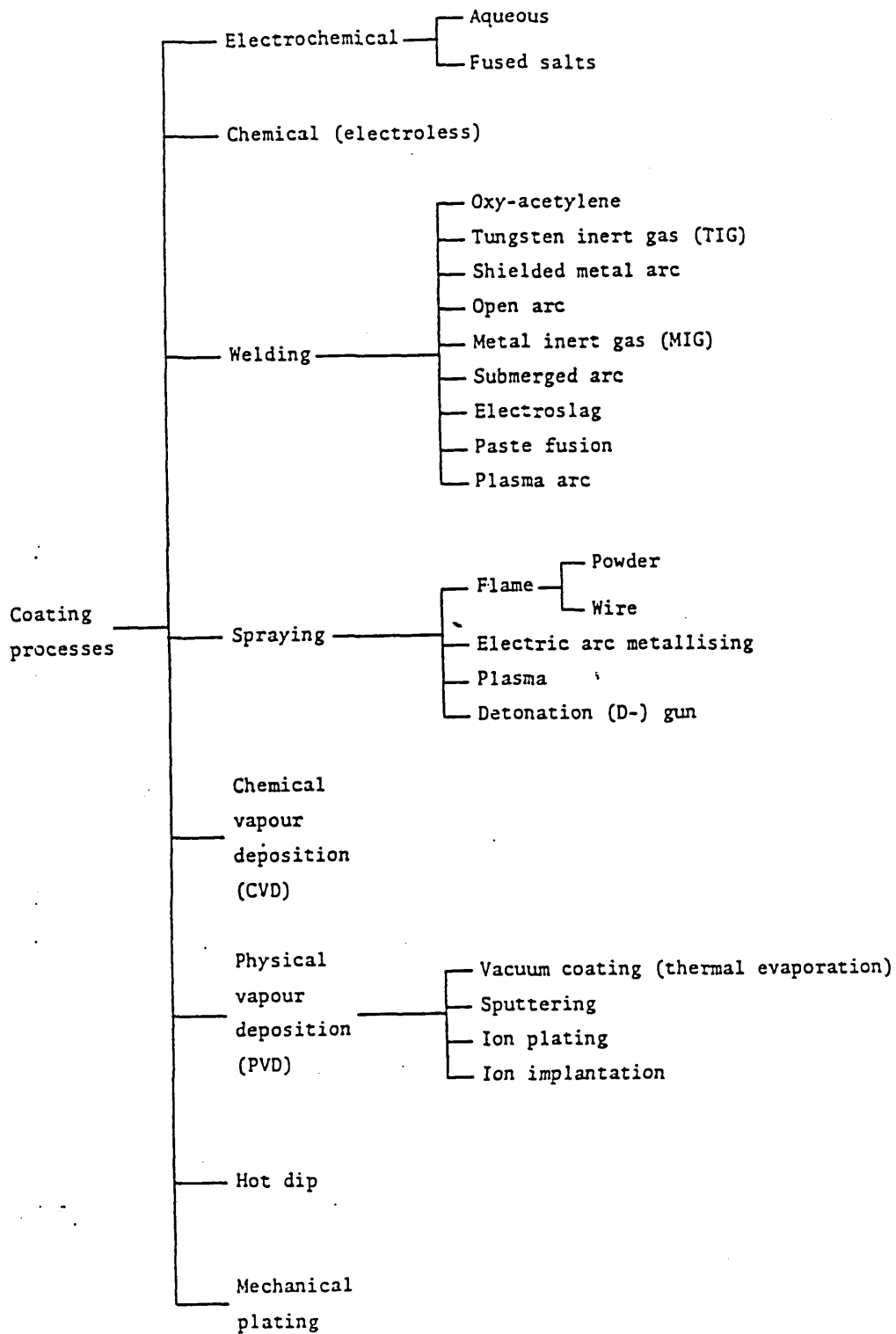


Fig.2.7 Types of surface coating techniques. (36)-

wear-resistant surface layers may be produced on tougher, fracture resistant substrates. Or, corrosion resistance may be imported to internal surfaces of vessels without the large heat inputs of conventional cladding techniques. Especially in the case of thin layers this large heat input results in melting and important distortion of materials. Laser cladding does not compete economically with conventional processing for the production of thick clad layers.(9,10,14,32-34)

There are several alternative surface coating techniques for both thick and thin coatings; these are shown in Fig.2.7 (35)

Today, industrial use of laser cladding in industry is carried out successfully by several companies.(9) However, the first industrial use of laser cladding was by Rolls Royce in 1981 to clad the nickel based superalloy turbine blade shroud interlocks on the RB 211 jet engine.(36)

2.7 Laser Melt Particle Injection

The laser-melt particle injection process is a technique that can be used to accomplish diverse modifications in the structure and the chemistry of metal surfaces. This process is similar to laser cladding and alloying by the blown powder route except that the particles blown into the laser melt pool do not completely melt. The desired result is to form a surface layer of the metal with a uniform mix with the required volume fraction of hard particles.

The laser-melt particle injection process was developed

to form hard, wear resistant surfaces on alloys. Extensive development work has been carried out especially on light weight high strength alloys such as 5052 Al, 6061 Al, Ti-6Al-4V, for which alternative surface hardening processes are not plentiful.(37-47) By this process, ceramic particles, such as carbides and oxides, are blown into the melt pool formed by a high energy laser beam on a moving substrate. Subsequent solidification results in a composite metal-carbide layer exhibiting continuity of properties across the substrate melt interface.(37-47)

With a single melt pass and a stationary beam, layers 2-3 mm wide could be processed. Wider areas were processed by overlapping several melt passes. There are five characteristics which are important when using particle injection technique:(37)

- 1) Uniform structure in the surface layer.
- 2) Minimum dissolution of particles.
- 3) Maximum surface wetting of particles.
- 4) Correct volume fractions.
- 5) Strong bonds between the particles and the metallic substrate.

Besides being a relatively slow process, the overlapped regions sometimes showed excessive thermal cracking and carbide dissolution. For example, in Ti-6Al-4V alloy injected with WC, complex resolidified carbides were found within the overlapped region (Fig.2.8). As a first step towards the commercialization of the process, a means to

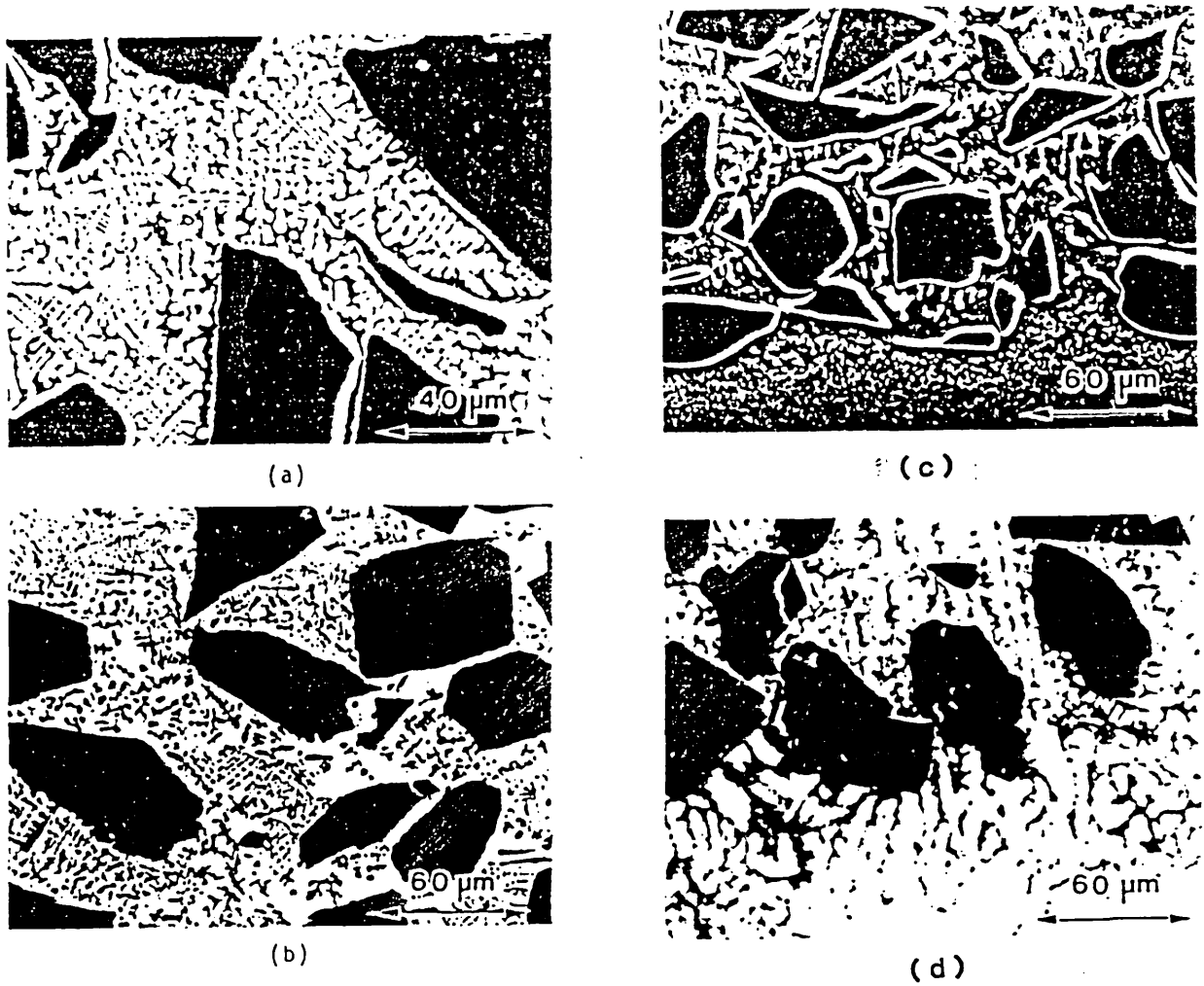
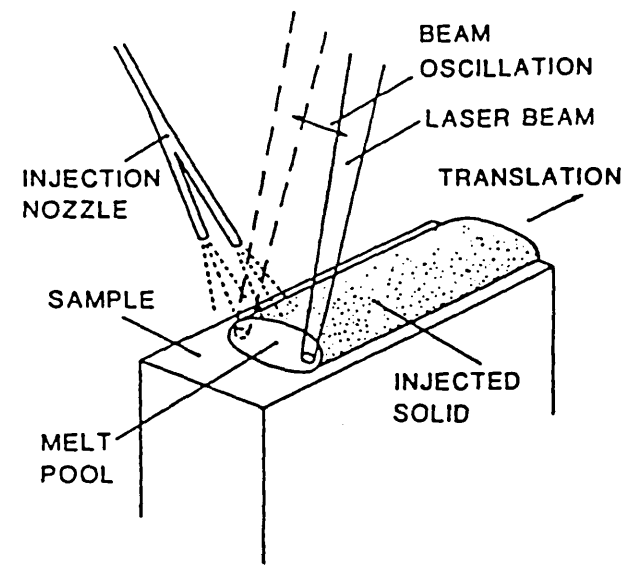


Fig.2.8 Network of resolidified carbides found in the upper regions (a) and (b) and microstructures in the lower portions (c) and (d) of TiC injected layers. (a) and (c) Ti-6Al-4V, (b) and (d) Inconel 625. (37)

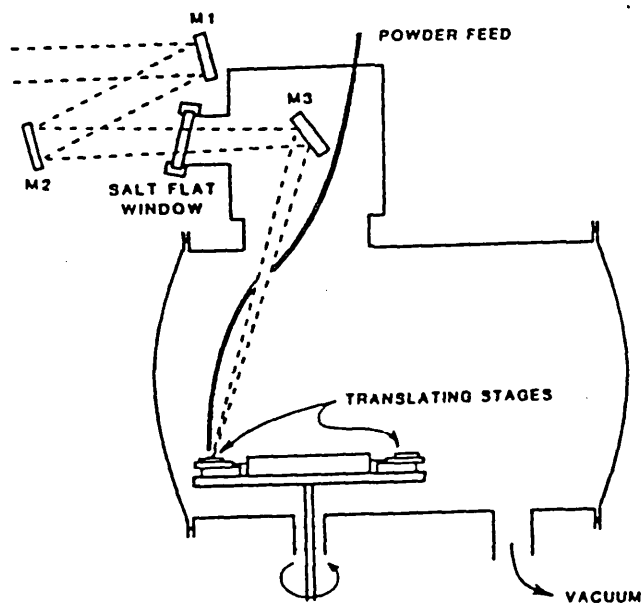
produce wide injected strips with a single melt pass has been developed.(37) For this purpose the process was modified to include the laser beam oscillations to form wider melt pools. With this technique, injected layers 5-10 mm wide were produced. Fig.2.9 shows a diagram of laser-melt particle injection processing.(37)

The process variations are mainly particle delivery system, particle delivery pressure, and gas shrouding system. Microstructure of laser-melt particle injected alloys is shown in Fig.2.8. (37)

Wear resistance of carbide injected aluminium and titanium alloys has been studied by several investigators. (42-44,46) Abrasion testing with a dry sand rubber wheel apparatus demonstrated that the resistance to abrasion of both aluminium alloys and Ti-6Al-4V can be improved by this processing.



(a)



(b)

Fig.2.9 (a) Schematic diagram of laser-melt particle injection processing with an oscillating beam.

(b) General arrangements of the vacuum laser-melt particle injection processing. (37)

2.8

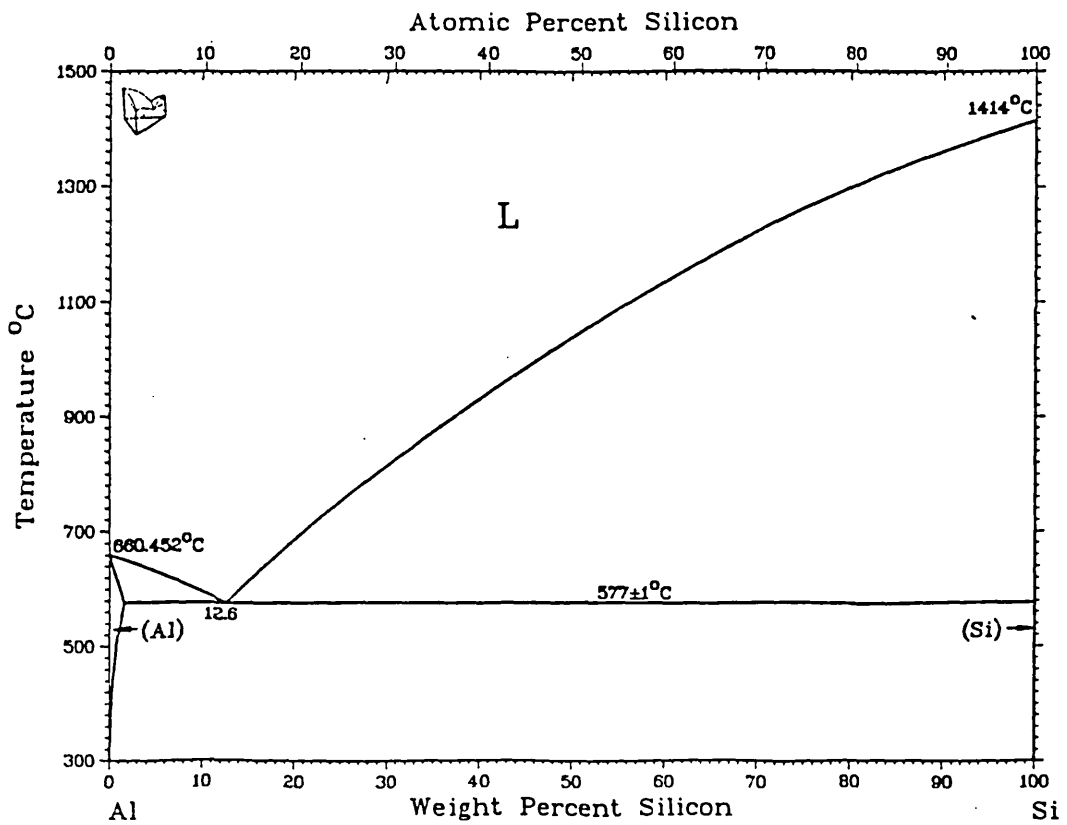
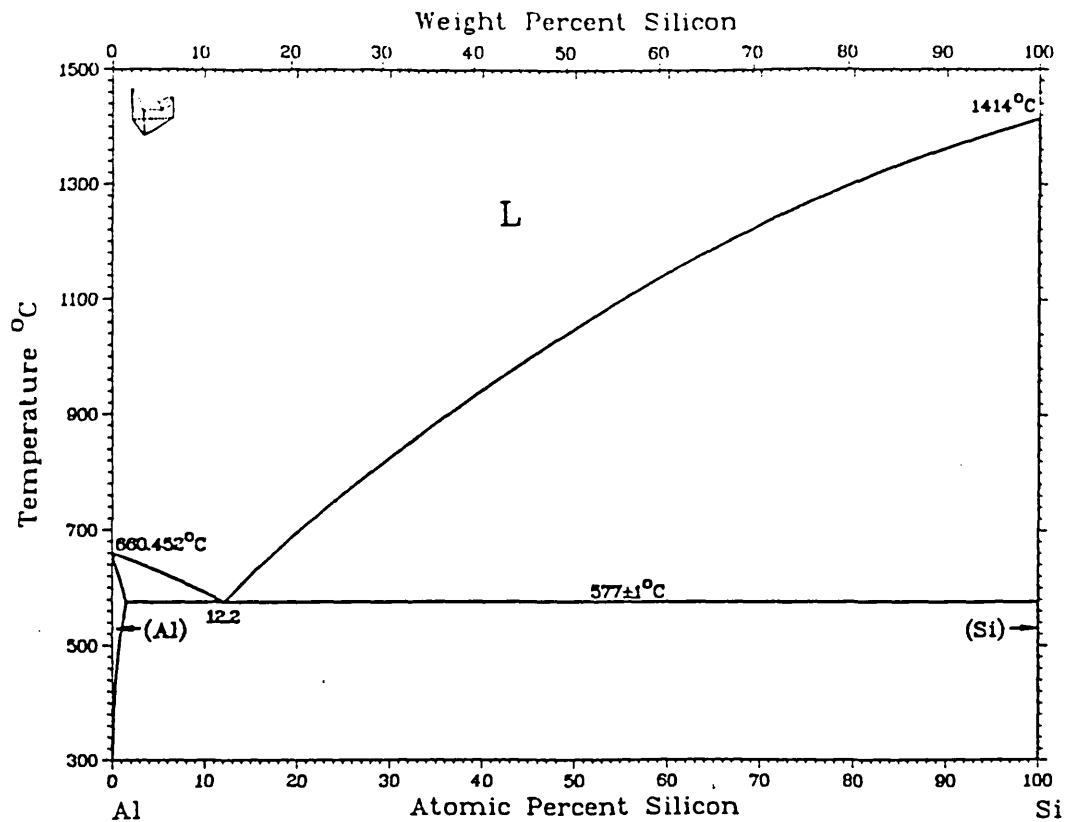
Aluminium-Silicon System

Silicon is one of the main alloying element in aluminium alloys, the others are Cu, Mg, Mn, Zn. The aluminium-silicon system is a simple eutectic, a combination of fcc metal and a non-metal diamond cubic structure. The solid solubility of silicon in aluminium is 1.65 wt% Si and aluminium in silicon about 0.5 wt% Al at 577 °C, at which temperature the eutectic occurs. The composition of the eutectic point has been reported as ranging from 11.7 wt% Si to 14.5 wt% Si, with the most probable equilibrium value as 12.6 wt% Si. (Fig.2.10).

Silicon imparts high fluidity and low shrinkage, which provides good castability and weldability. Because of the low thermal expansion coefficient it is used for pistons and the high hardness of silicon containing alloys enhances the wear resistance. The maximum amount of silicon in cast alloys is about 22-24 wt% Si. But powder metallurgy alloys may go as high as 40-50 wt% Si. Sodium and strontium are used for the structural modifications and phosphorus nucleates the silicon to permit of a fine distribution of the primary crystals in hypereutectic alloys.(48,49)

Iron is the main impurity and in most alloys efforts are made to keep it as low as economically possible, because it has a deleterious effect on both ductility and corrosion resistance. The upper limit is usually 0.6-0.7 wt% Fe in sand and permanent mould castings. Cobalt, chromium, manganese, molybdenum, and nickel are sometimes added as

Assessed Al-Si Phase Diagram



J.L. Murray and A.J. McAlister, 1984.

Fig.2.10 Aluminium Silicon Binary Phase Diagram. (50)

correctives for iron. Addition of Mn increases the high temperature strength and improves creep resistance through the formation of high melting point compounds. Copper increases strength and fatigue resistance; (it has not got a deleterious effect on castability). Magnesium increases strength; however, it decreases ductility. Zinc is permitted up to 1.5-2.0 wt% Zn, as it has got no substantial effect on properties at room temperature. Ti, and B additions are sometimes used for grain refinement. (48,49)

The lattice parameter is decreased slightly by silicon in solution and somewhat more by copper. A solution heat treatment and quench brings more silicon into solid solution. The lattice parameter of Al-Si alloy is between $a = 4.045 \times 10^{-10}$ mm and $a = 4.05 \times 10^{-10}$ mm depending on composition and treatment. Undissolved silicon and magnesium decrease the density. But most of ^{H₂} other additions increase it. Effects of alloying additions on the density of binary aluminium alloys are shown in Fig.2.11 (49)

Thermal expansion is reduced substantially by silicon, the reduction being reported to be linear up to 20 wt.% Si. Fig.2.12 shows the effect of alloying elements on the thermal expansion. (49)

Increasing silicon content increases strength, but reduces ductility. Sodium modification shows significant increases in ductility especially in sand castings. The effect of cell size and dendrite arm spacing on mechanical properties of alloys above 8 wt% Si is not very marked. Iron may slightly increase the strength, but importantly decreases the ductility, especially above 0.7 wt% Fe.

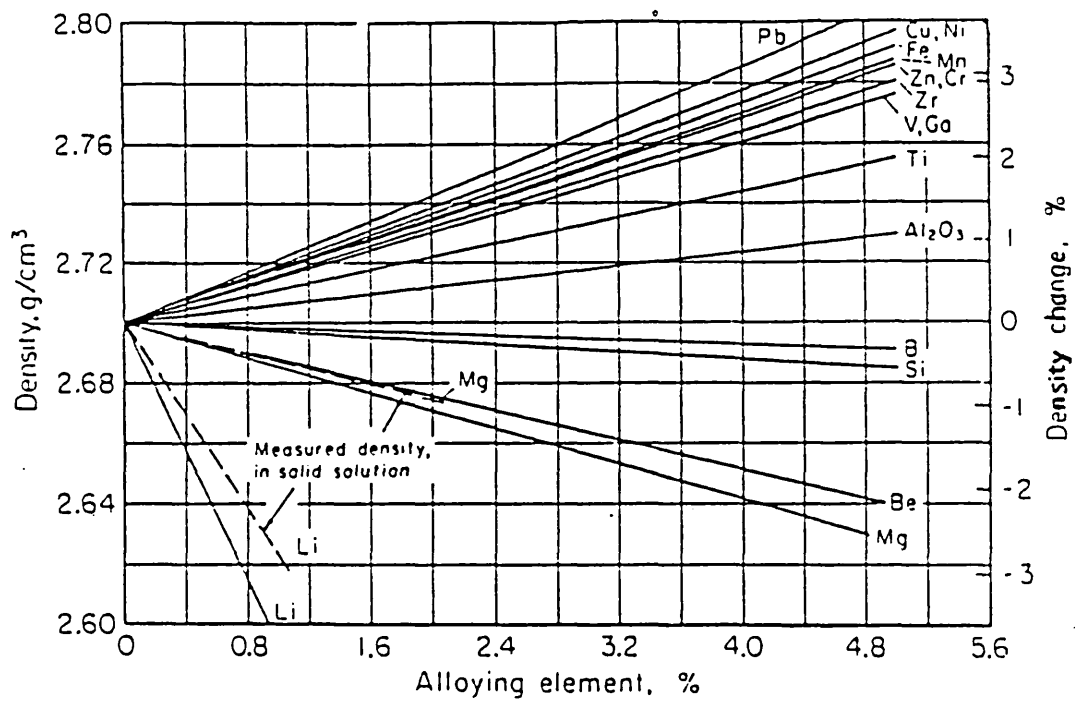


Fig.2.11 Density of binary aluminium alloys. (49)

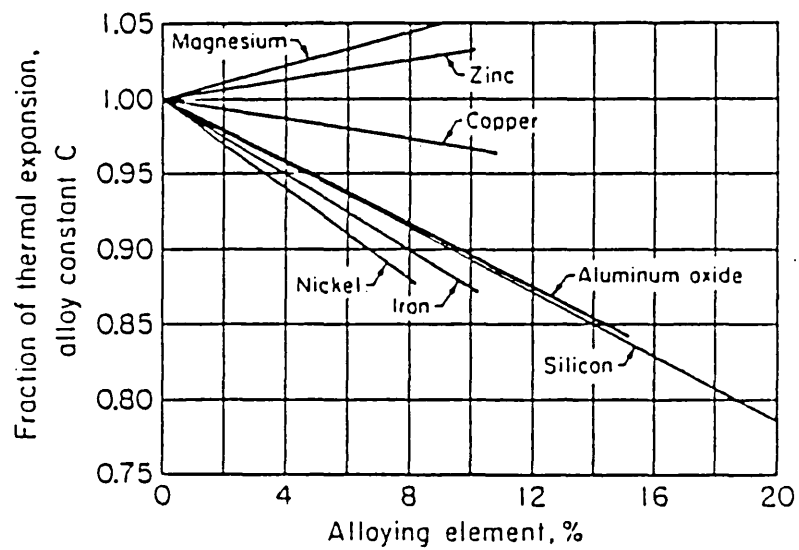


Fig.2.12 Effects of alloying elements on the thermal expansion of aluminium. (49)

The strength and ductility are inadequate unless modified and partial modification and overmodification causes a reduction in mechanical properties. Small additions of P refine rather than modify the microstructure. Both primary and eutectic silicon morphologies change from flake to a granular form with an increase in strength and ductility. The action of P is attributed to the heterogeneous nucleation of silicon by the compound AlP.(48,49,51)

At high temperatures the strength declines and the ductility increases. Compressive strength is higher than tensile strength by some 10-15%.

Creep resistance is not particularly good; silicon increases the creep resistance of aluminium much less than other alloying elements.(48,49)

The electrolytic potential of aluminium is made more positive by silicon in solution. Silicon has a less negative potential than aluminium, but, because of its positivity in most corrosive environments, there is little or no attack. When there is attack, it takes place in aluminium, only in alkaline solutions, which attack silicon as well as aluminium. Copper reduces the corrosion resistance appreciably. In contact corrosion aluminium silicon alloys are worse than aluminium 99.8 wt% Al.(48,49)

2.9 Industrially Important Aluminium-Silicon Alloys

The outstanding characteristics of alloys in the aluminium-silicon group are castability, ductility, and

AA No.	Former designation	Former ASTM designation	SAE designation	Processing(a)
305.0	...	...	...	S, P
A305.0	...	...	...	S, P
308.0	A108	...	...	S, P
319.0	319, as-cast	SC64D	326	S, P
A319.0	...	...	329	S, P
B319	...	...	329	S, P
324.0	324	...	...	P
328.0	ReJ N-8	SC82A	327	S
332.0	F332.0 (F132)	SC103A	332	P
333.0	333	SC94A	331	P
A333.0	...	...	...	P
336.0	A332.0 (A132)	SN122A	321	P
339.0	Z332.0 (Z132)	...	334	P
343.0	X443Z	...	...	P
354.0	354	SC92A	...	P
A355.0	...	...	...	S, P
355.0	355	SC51A	322	S, P
C355.0	C355	SC51B	335	S, P
356.0	356	SG70A	323	S, P
A356.0	A356	SG70B	336	S, P
B356	...	...	...	S, P
F356.0	...	...	...	S, P
357.0	357	...	...	S, P
A357.0	A357	SG71A	...	S, P
B357.0	...	...	...	S, P
C357.0	...	...	...	S, P
358.0	B358.0 (Tens-50)	...	...	S, P
359.0	359	SC91A	...	S, P
360.0	360	SC100B	...	D
A360.0	A360	SC100A	309	D
361.0	...	...	...	D
363.0	363	...	...	S
364.0	364	...	...	D
369.0	Special K-9	...	...	D
380.0	380	SC83B	308	P
A380.0	A380	SC83A	306	P
B380.0	A380	SC84A	306	P
383.0	...	SC102A	383	P
384.0	384	SC114A	303	D
A384.0	384	...	...	P
385.0	B384.0 (384)	...	...	P
390.0	390	...	...	S, P
A390.0	A390	...	...	P
B390.0	...	...	...	P
392.0	392	...	...	P
393.0	Nanasil	...	...	S, P, D
413.0	413	S12B	...	D
A413.0	A413	S12A	305	D
443.0	443	S5B	35	S, P
A443.0	443 (0.30 max Cu)	...	...	S
B443.0	443 (0.15 max Cu)	SSA	...	S, P
C443.0	A43	SSC	304	D
444.0	...	...	...	S, P
A444.0	A444	...	...	P
511.0	F511.0 (F214)	S7A	...	S

Table 2.3 Composition of Some Aluminium-Silicon Alloys.

			Chemical composition(b), %			
Silicon	Iron	Copper	Magnesium	Zinc	Tin	Other
5.0	<0.6	1.25	...	<0.35	...	...
5.0	<0.20	1.25	...	<0.10	...	...
5.5	<1.0	4.5	<0.10	<1.0	...	...
6.0	<1.0	3.5	<0.10	<1.0	...	...
6.0	<1.2	3.5	0.3	<1.0	...	...
6.0	<1.0	3.5	<0.10	<3.0	...	...
7.5	<1.2	0.50	0.55	<1.0	...	...
8.0	<1.0	1.5	0.40	<1.5	...	0.40 Mn
9.5	<1.2	3.0	1.0	1.0	...	...
9.0	<1.0	3.5	0.3	<1.0	...	...
9.0	<1.0	3.5	0.3	<3.0	...	...
12.0	<1.2	1.0	1.0	<0.35	...	2.5 Ni
12.0	<1.2	2.25	1.0	<1.0	...	1.0 Ni
7.2	<1.2	0.7	<0.10	1.6	...	...
9.0	<0.20	1.8	0.5	...	...	...
5.0	<0.09	1.25	0.5	<0.05	...	...
5.0	<0.6	1.25	0.5	<0.35	...	...
5.0	<0.20	1.25	0.5	<0.10	...	...
7.0	<0.6	<0.25	0.35	<0.35	...	...
7.0	<0.20	<0.20	0.35	<0.10	...	...
7.0	<0.09	<0.05	0.35	<0.05	...	...
7.0	<0.20	<0.20	0.21	<0.10	...	...
7.0	<0.15	<0.05	0.55	<0.05	...	...
7.0	<0.20	<0.20	0.55	<0.10	...	0.05 Be
7.0	<0.09	<0.05	0.50	<0.05	...	...
7.0	<0.09	<0.05	0.60	<0.05	...	0.05 Be
8.1	<0.30	<0.20	0.50	<0.20	...	0.2 Be
9.0	<0.20	<0.20	0.6	...	...	...
9.5	<0.20	<0.6	0.5	...	...	...
9.5	<1.5	<0.6	0.5	...	...	...
10.0	<1.1	<0.50	0.5	...	...	0.25 Cr, 0.25 Ni
5.25	<1.1	3.0	0.30	3.8	0.25	<0.25 Pb, <0.08 Mn +
8.5	<1.5	<0.20	0.30	<0.15	...	0.35 Cr, 0.03 Be
11.5	<1.3	<0.50	0.35	<1.0	...	0.35 Cr
8.5	<2.0	3.5	<0.10	<3.0	...	...
8.5	<1.3	3.5	<0.10	<3.0	...	...
8.5	<1.3	3.5	<0.10	<1.0	...	...
10.5	<1.3	2.5	<0.10	<3.0	...	...
11.3	<1.3	3.75	<0.10	<3.0	...	...
11.3	<1.3	3.75	<0.10	<1.0	...	...
12.0	<2.0	3.0	<0.30	<3.0	...	...
17.0	<1.3	4.5	0.55	<0.10	...	<0.1 Mn
17.0	<0.50	4.5	0.55	<0.10	...	<0.1 Mn
17.0	<1.3	4.5	0.55	<1.5	...	<0.5 Mn
19.0	<1.5	0.6	1.0	<0.50	...	0.4 Mn
22.0	<1.3	0.9	1.0	...	...	2.25 Ni, 0.12 V
12.0	<2.0	<1.0	<0.10	...	...	...
12.0	<1.3	<1.0	<0.10	...	...	...
5.25	<0.8	<0.6	<0.05	<0.50	...	...
5.25	<0.8	<0.30	<0.05	<0.50	...	...
5.25	<0.8	<0.15	<0.05	<0.35	...	...
5.25	<2.0	<0.6	<0.10	<0.50	...	...
7.0	<0.6	<0.25	<0.10	<0.35	...	...
7.0	<0.20	<0.10	<0.05	<0.10	...	...
0.5	<0.50	<0.15	4.0	<0.15	...	...

corrosion resistance. They are used extensively for sand and permanent mould castings for which strength is not a prime consideration, for example, domestic tools, pump castings and certain automobile castings. The alloy with a predominantly eutectic structure (LM 6) must be modified to ensure adequate mechanical strength and ductility.(49,51) Table 2.3 shows the composition of some aluminium-silicon alloys.(49)

The alloys of hypereutectic composition contain 15-25 wt% Si which are used in the manufacture of cylinders and pistons. These alloys display excellent wear resistance which, together with machinability, improves as the silicon particles become finer and more evenly distributed. Both of these effects are promoted by P refinement.(48,49)

Although binary aluminium-silicon alloys show some response to heat treatment, a much greater strengthening is achieved by making small additions of Cu (<3%), Mg, Ni. This gives rise to a family of alloys of hypoeutectic composition which contain a significant volume fraction of solid solution dendrites which are amenable to heat treatment.(48,49). LM 25 which contain ~7 wt.% Si is a very popular general purpose sand and permanent mould casting alloy in this group and illustrates the use that can be made of structure control to obtain a wide range of properties.

When cast, the alloy consists of aluminium dendrites with inter-dendritic, modified eutectic. In the full heat treatment condition, the alloy is solution treated to take all the magnesium into solution and partially spheroidize

the eutectic Si. The magnesium is retained in solution by quenching, and on ageing a uniform distribution of Mg_2Si is formed in dendrites to strengthen the alloy. The mechanical properties of cast alloys are influenced by porosity, inter-dendritic second phases, dendrite spacing and grain size.(49,51)

Particular uses are aircraft and automobile applications. It is used exclusively at present in Australia in the form of premodified Sr ingot, CP 601, for wheel castings.

2.10 Wear Properties of Aluminium-Silicon Alloys

Wear resistance of aluminium-silicon alloys is very good especially in hypereutectic alloys in which the hard primary silicon particles are well distributed, either by phosphorus nucleation or by powder metallurgy process.(48)

The alloys currently of most interest for the engine block in internal combustion engines are those based upon Al-Si whose casting characteristics and mechanical properties are already well established and some of these are detailed in Table 2.4. (52)

The tribological characteristics of these alloys were widely investigated.(52-59) Under dry sliding conditions these alloys exhibit wear with characteristics similar to other groups of alloys including steel and copper base alloys. When an aluminium-silicon alloy is rubbed against a steel counterface under dry sliding conditions two stages of wear behaviour are observed. Up to 120 N, load wear is

Table 2.4 COMPOSITION OF SOME TYPICAL ALLOYS FROM DIFFERENT COUNTRIES

Alloy Type	Country	Alloy Designation	Composition (WT %) - Remainder Aluminium *										
			Cu	Si	Mg	Ni	Fe	Mn	Ti	Zn	Sn	Pb	Others
Lo-Ex	Britain	LM13	.5-1.3	11.0-13.0	.8-1.5	.7-2.5	.8	.5	.2	.1	.1	.1	
		LM26 (designate)	2.0-4.0	8.5-10.5	.5-1.5	.5	1.2	.5	.25	1.0	-	-	
	USA	Al32 (SAE 321)	.5-1.5	11.0-13.0	.7-1.3	2-3	1.3	.35	.25	.35	.05	.05	
		Fl32 (SAE 332)	2.0-4.0	8.5-10.5	.5-1.5	.5	1.2	.5	.25	1.0	-	-	
	Germany	AlSi 12 Cu Ni Mg	.8-1.5	11 - 13	.8-1.3	.8-1.3	.7	.2	.2	.2	-	-	
		AlSi 12 Ni Mg	.2	11.5-12.5	1.3-1.7	.8-1.2	.7	.2-.3	.2	.2	-	-	
		AlSi 10 Cu Mg	2.5-3.5	9-10.5	.7-1.2	.5	.9	.5	.2	.8	-	-	
	France	A-S 12 UN	.5-1.5	11.5-13	.8-1.5	.5-1.5	.75	.3	.1	.2	.05	.1	
		A-S 12 U 4 N Zr	3.5-4.5	11.0-13.5	.9-1.6	1.1-1.7	.6	.6-1.0	-	-	-	-	Zr.1-2
Al-Cu	Britain	LM14	3.5-4.5	.6	1.2-1.7	1.8-2.3	.6	.6	.2	.1	.05	.05	
	USA	142 (SAE 39)	3.5-4.5	.7	1.2-1.8	1.7-2.3	1.0	.35	.25	.35	-	-	Cr.25
	Germany	Al Cu 4 Ni 2Mg	3.5-4.5	.5	1.2-1.8	1.7-2.3	.7	.2	.1-.2	.2	-	-	
Al-HIGH Si	Britain	LM28 (designate)	1.3-1.8	17-20	.8-1.5	.8-1.5	.7	.6	.2	.2	-	-	
		LM29 (designate)	.8-1.3	22-25	.8-1.3	.8-1.3	.7	.6	.2	.2	-	-	
	Germany	Al Si 18 Cu Ni Mg	.8-1.5	17-19	.8-1.3	.8-1.3	.7	.2	.2	.2	-	-	
		Al Si 21 Cu Ni Mg	1.4-1.8	20-22	.4-.6	1.4-1.6	.7	.4-.6	.2	.2	-	-	Co.7
		Al Si 25 Cu Ni Mg	.8-1.5	23-26	.8-1.3	.8-1.3	.7	.2	.2	.2	-	-	Co.5
	France	A-S 18 UN Zr	.8-1.2	17-19	.8-1.3	.8-1.2	.6	-	1	1	-	-	Zr.15-.35
		A-S 22 UN	1.0-2.2	20-22	.8-1.3	.8-1.6	.85	.3-.8	.2	.2	-	-	Co.8-1.5

low, adhesive wear does not occur and an oxide film is produced on both surfaces which adds further protection. A transition to a higher wear rate occurs when there is transfer of aluminium to disc surface and this is typical of adhesive wear. Because these transferred particles ^{become work} hardened they promote further wear by abrasion.

When the silicon content of alloy is varied there is an approximately linear relationship between the silicon content and the onset of undesirable severe and adhesive wear. Fig.2.13 shows the relationship between transition load and silicon content.(52)

Wear resistance of high silicon alloys (20-25% Si) is 10 times better than that of plain carbon steel and comparable with that of surface hardened steel. It is therefore necessary to use as high a silicon content as possible, but this must be balanced by the requirement for other mechanical properties and also machinability; in practice a 16 to 18%wt.Si alloy is widely used.(58) There is little effect of silicon particle size on wear behaviour even though this has a significant effect on both mechanical properties and machinability. Because of plastic deformation at the surface under contact conditions the silicon particles break up to a particle size of less than 10 μm , even if they are 100 μm in unworn alloy.(52)

If a grey cast iron is used instead of steel the transition load is considerably increased in comparison with steel. The presence of graphite at the friction surface inhibits the adhesion of aluminium to the counterface.(52)

There is little effect due to the hardness of

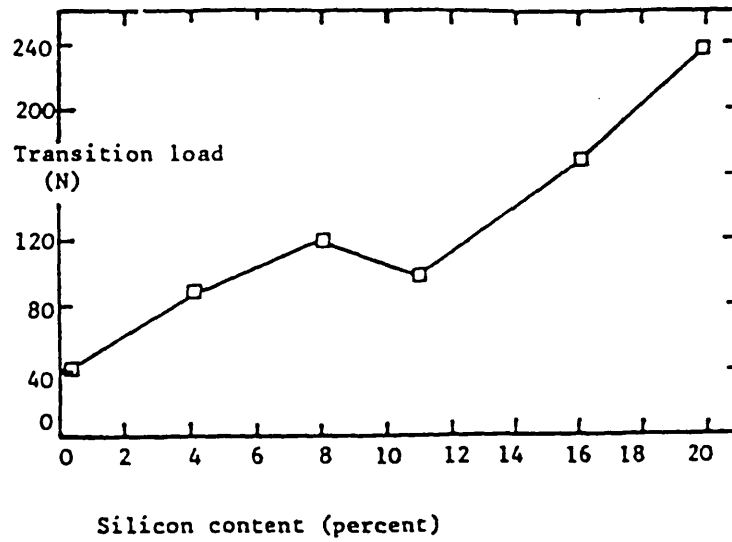


Fig.2.13 Effect of silicon content on the transition load for aluminium alloys. (52)

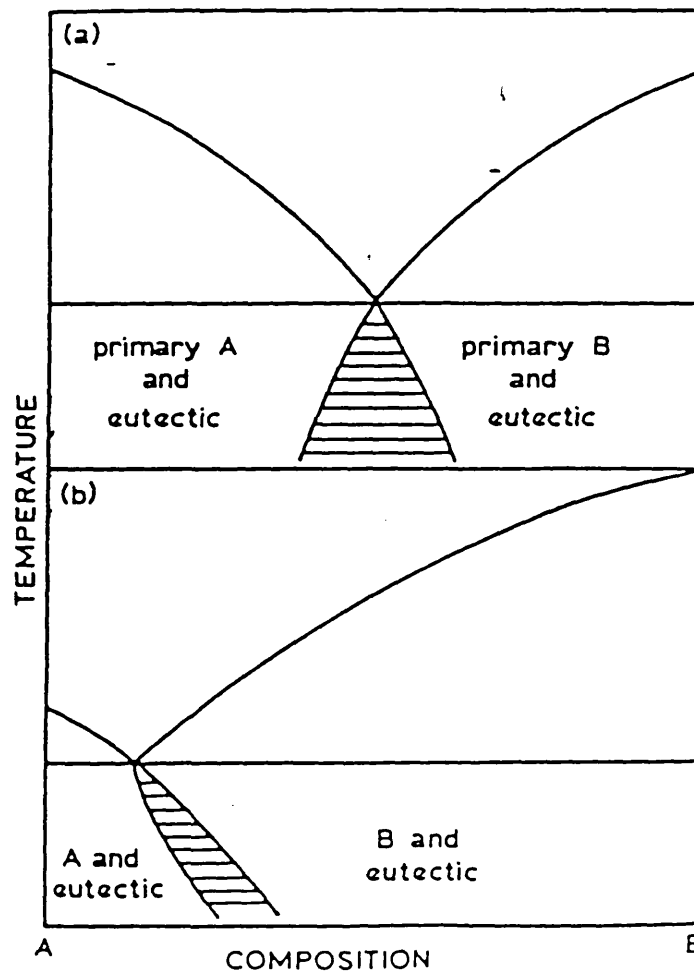


Fig.2.14 Phase diagram for normal and anomalous structures.

a) Symmetrical phase diagram.

b) asymmetrical phase diagram. (61)

the counterface although care must be taken to ensure that the surface is not rough. In the development of the Vega engine it was found that a problem arose from the contact of the aluminium piston skirt with the the aluminium alloy bore. It was overcome by plating a hard iron coating on the piston skirt.(52-54)

In recent years, there has been considerable development in aluminium-graphite alloys stimulated by the marked improvements brought about by graphite in reducing the incidence of adhesive wear. However, these alloys are still somewhat difficult in terms of their casting characteristics and also their comparatively low strength.(59)

Megaw et al. reported that the effect of silicon enhancement by laser alloying is to reduce the wear rate to approaching one quarter the level seen with the untreated substrate.(60) Walker et al. showed that there is a significant increase in wear resistance of laser alloyed Al alloy with increasing silicon content.(24)

2.11 Solidification Structure of Aluminium-Silicon Alloys

2.11.1 Eutectic Structure

The eutectic reaction is one of the oldest phenomena in solidification. Alloys at or close to, the eutectic composition are of the lowest melting point, are fluid and have a short freezing range. These characteristics promote castability. Casting of engineering components preceded understanding of the solidification process.(61-63)

Eutectic solidification shows two different structures, normal and anomalous eutectic microstructures. Normal structures display lamellar or rod morphologies and occur in systems with a symmetrical phase diagram, Fig.2.14a, whereas anomalous structures like Al-Si flake morphology have been associated with an asymmetrical phase diagram. Fig.2.14b.(61) The shaded region in Fig.2.14 shows the coupled eutectic zone in which eutectic formation occurs.

The microstructures displayed by Al-Si alloys have been described in several previous publications. (51,61-67) Controlled directional solidification studies in particular, have shown that the variety of microstructures displayed by this non-faceting/faceting system is a consequence of the different growth modes of the faceting phase, Si, and the dependence of their operation on the solidification conditions.

Si nucleates relatively easily in the Al-Si system and is sensitive to the impurities present in the liquid. Both eutectic and the primary phase form outside the coupled zone. The boundaries of the coupled zone can be determined if the rates of growth for primary and eutectic phases are known unless repeated nucleation of one of the phases occurs.(65)

The structures observed in directionally solidified eutectic alloys are represented in Fig.2.15 as a function of the temperature gradient and growth velocity and illustrate the range of structures that may form in one system.(51,62,65)

At high temperature gradient (G) and low growth velocity

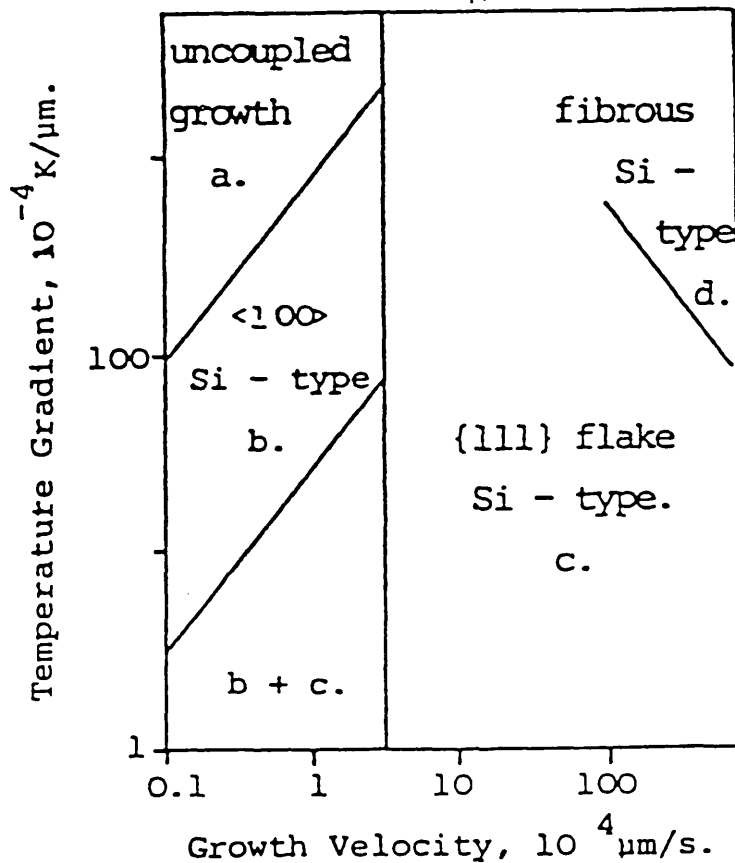


Fig.2.15 Range of microstructures observed in directionally solidified Al-Si eutectic alloys. (62)

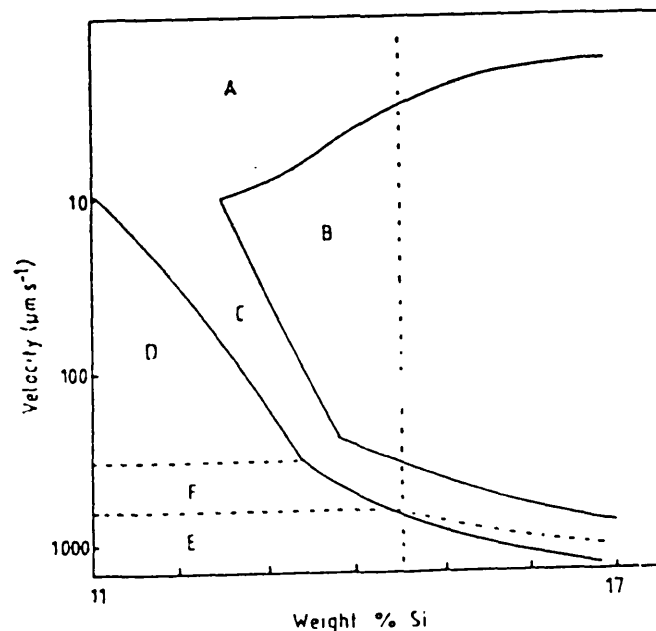


Fig.2.16 Microstructure as a function of growth velocity and silicon content in high purity Al-Si alloys grown at a temperature gradient of 12.5 K/mm. A, plate-like, angular and complex regular silicon; B, complex regular star-like polyhedral silicon (> 14.5 wt.% Si), angular and flake eutectic silicon; C, flake and angular silicon, zone of coupled eutectic growth; D, (aluminium) dendrites and flake eutectic silicon; E, (aluminium) dendrites and fibrous eutectic silicon; F, transition of flake and angular silicon to fibrous silicon. (65)

(V), region a, the two phases grow independently and silicon solidifies in a faceted manner. Decreasing G, region b, causes break down of the aluminium interface and the Si forms rod with a $\langle 100 \rangle$ texture. Region c, corresponds to typical casting conditions, Si grows as a twinned irregular interconnected flake network via the twin plane re-entrant edge mechanism (TPRE). Controlled rapid cooling produces steep G and high V necessary to form fibrous Si, region d.(65)

Frequent twinning allows a more flexible growth mechanism which can approach the diffusion-controlled growth of a non-faceted/non-faceted system. It is shown that twins are not essential for fibrous growth; the transition in morphology is attributed to change from layer to continuous growth.(65-67)

2.11.2 Proeutectic Silicon

Proeutectic silicon crystals with $\{111\}$ facets form from hypereutectic alloys. The growth rate anisotropy of Si is such that massive geometric primary silicon grows from the melt by layer growth and is bound by slow growing $\{111\}$ faces giving an octahedral growth habit. Atasoy et al.(65) remark that twins which form during growth do not necessarily contribute to the growth process. The changing morphology of silicon in terms of alloy content and growth velocity is shown in Fig.2.16. The silicon morphology is the dominant structural feature in these alloys and it is convenient to define the various structural types with

reference to the silicon morphology. The most easily defined types are primary phase comprising massive crystals of geometric, star-like, dendritic shape and eutectic phases of flake and fibrous form. Three other morphologies observed frequently are angular Si, star-like Si, and complex regular Si. (65-67)

Fredriksson et al. showed that increasing cooling rate causes nucleation to occur at edges and corners of proeutectic silicon. Corners and edges become preferred growth sites at higher growth velocities and extended layer growth was observed. (66)

CHAPTER 3

EXPERIMENTAL PROCEDURE

3.1

Apparatus and Devices

This surface modification work was carried out using a 2 kW continuous wave CO₂ laser (Control Laser Ltd.) of the fast axial type. The arrangement of the work station is shown in Fig.3.1

A mixture of CO₂, He, N₂ flows through the four discharge tubes. An electrical current flowing through the tubes in the counter direction to the gas flow, pumps the CO₂ by way of electron collisions and indirectly by way of vibrationally excited N₂. The He acts as a coolant.

Argon gas is passed at a minimum flow of ~1.5-2.0 l/min to prevent ingress of smoke/vapour and particles into the lens chamber. This gas flow rate is insufficient to disturb the powder flow into the molten pool. In addition, use of a long focal length lens (150 mm) and a 1.5 mm nozzle aperture reduces the gas pressure required. The argon gas also acts as a shrouding gas to prevent oxidation.

A gas shrouding device was fitted onto the laser head. It had a gas inlet with some small holes on it. Gas passes out through these small holes. For more effective protection it can be moved up and down to keep it near the surface when larger spot diameters are required. When the spot diameter is increased, the distance between^{the} nozzle and substrate will be increased and shrouding efficiency will not be good enough. The gas shrouding device is shown in Fig.3.2.

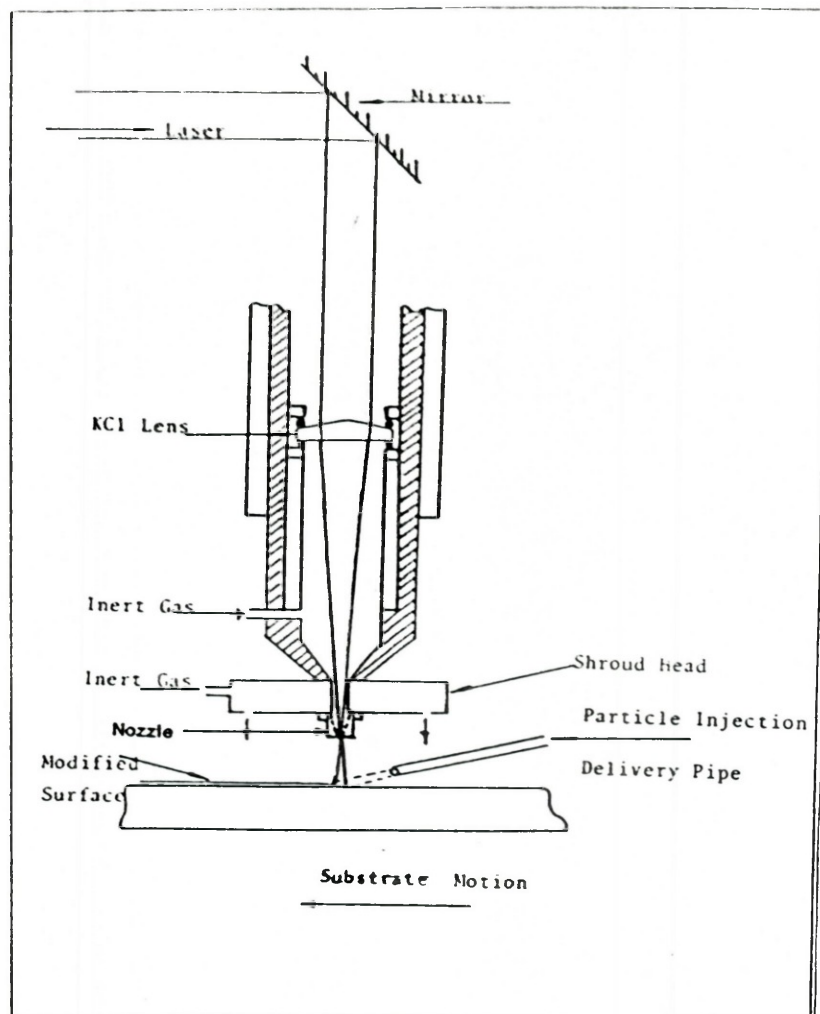


Fig.3.1 General arrangement of the work station.

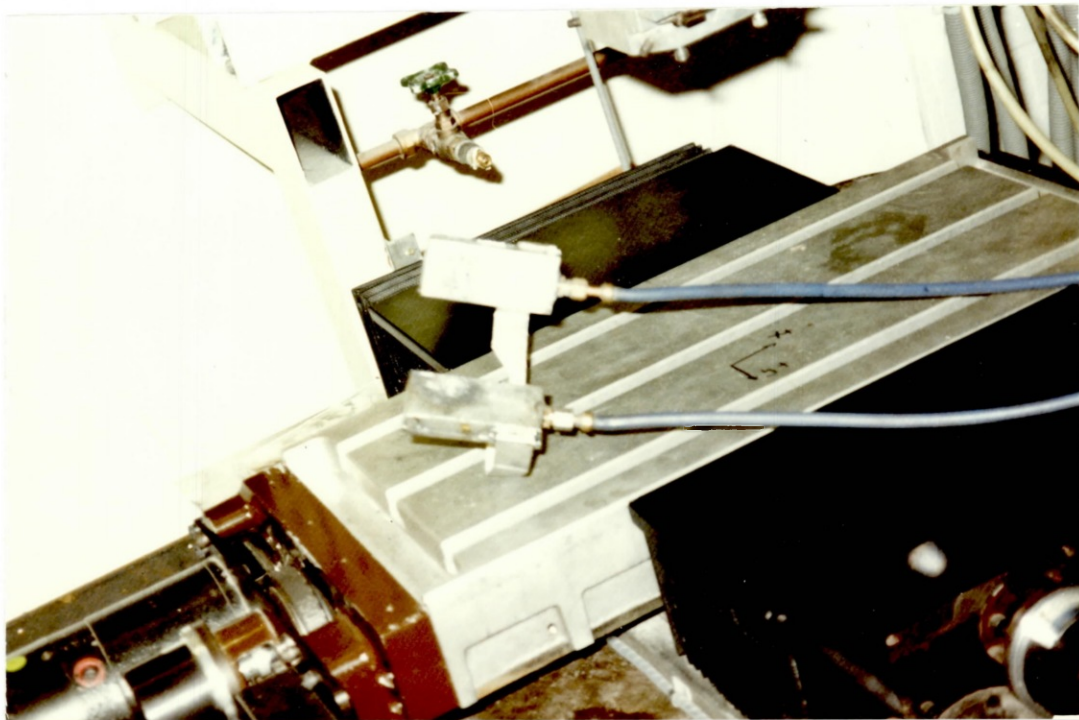


Fig.3.2 Gas shrouding device.

3.2 Powder Delivery System

Two categories of deposition technique have emerged for incorporating the added constituent to the melt pool.

- 1) Preplaced material before laser melting.
- 2) Application of material in the form of powder, wire, etc. during the laser melting process.

There are some advantages to adding material during laser processing. Firstly, the unit operation of preplacing the material is eliminated. This is an important factor in production systems. In the case of preplaced powder, powder usually comes off during processing and it stops the process. Use of adhesives can change the composition. The favoured powder delivery system in the present work is to blow powder directly into melt pool. Secondly, powder particles in or near to the surface appear to enhance the absorption of laser energy. Lastly, metallurgically, the use of powder provides more uniform mixing and alloying with the base material, and it reduces macro-segregations effect during solidification. There is an important point which presents a difficulty to the system. This is especially for powders finer than 44 μm , that is agglomeration of powder, and the uneven (non-uniform) feeding of powder as a result of agglomeration. Some work, which was carried out to overcome this problem, is mentioned below.

A gas blown screw powder system was used in these experiments and by other investigators also. Further information about the use of the system is given in Refs. (14, 32, 33). A diagram and general arrangements of screw

feeding system are shown in Fig.3.3 and Fig.3.4.

This powder feeder was developed at Imperial College. However, further development with a new mixing system, in order to feed finer and sticky powders, was carried out by the author. Principally, the powder feeder consists of the following parts: (5 and 6 represent the new features of the modified system)

- 1) Screw.
- 2) Powder stocking hopper.
- 3) Powder pick up and transportation line.
- 4) Pressure balancing line.
- 5) Vertical powder mixing and liberation system. (Fig.3.3)
- 6) Vibrator on transportation line. (not shown in Fig.3.3)

The powder is blown by a stream of an inert gas (argon). The stream is controlled directionally by the outlet tube. The screw speed controls the mass flow rate of the injected material.

The gas pressure is also important because of its effect on feeding rate, uniform feeding, transportation of powder and particle velocity. In systems where the particle density is less than the metallic substrate density e.g. with TiC particles (4.9 g/cm^3) on a steel substrate (7.9 g/cm^3) it has been found that the particles float on the melt pool surface unless they enter with sufficiently high velocity; any mixing which does occur will be due to particles being dragged by convection currents.

If the powder velocity is too high, powder will blow away

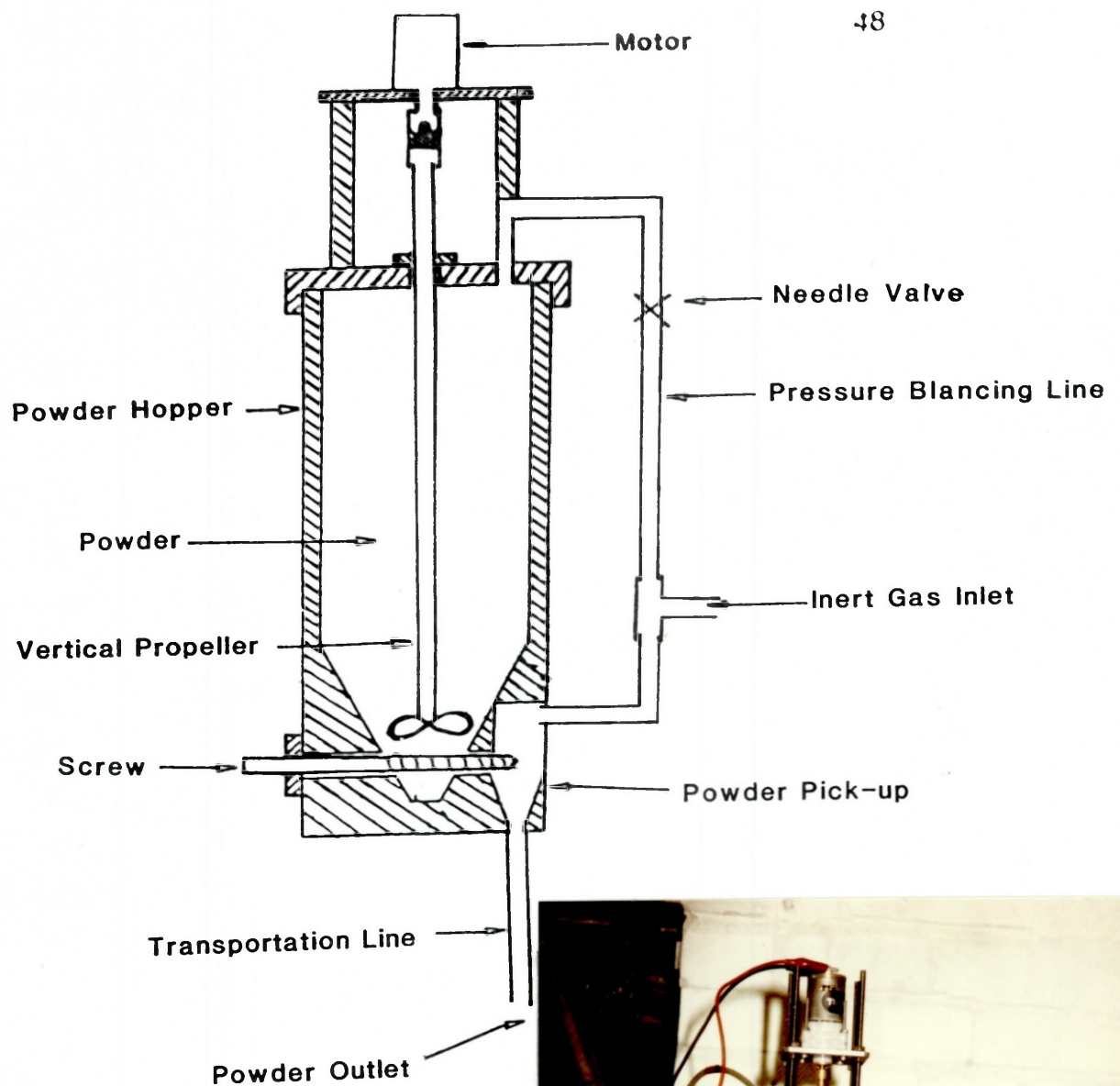
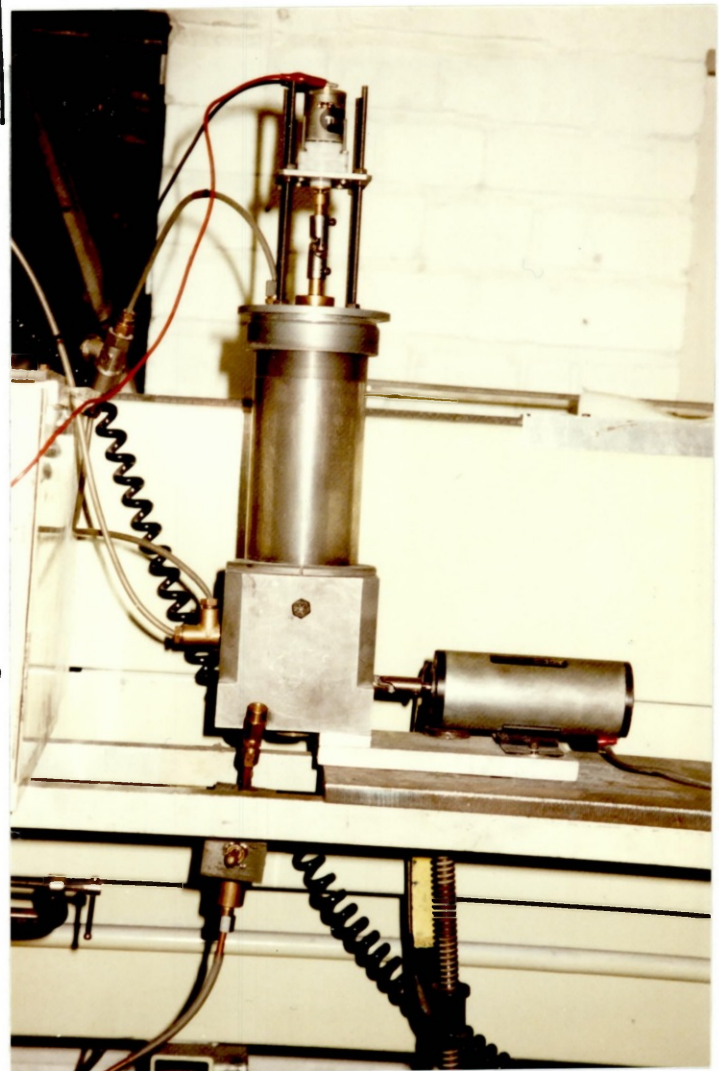


Fig.3.3 General arrangement of screw powder feeder with top driving system.

Fig.3.4 General view of screw powder feeder with top driving system.



and these powder particles will stick at the side of the melt pool as well. Spreading particles on the substrate will affect the further tracks by changing surface reflectivity and powder deposition.

The angle of the injection tube has a major effect on the powder flux (g/sec.mm^2) input into the melt pool. The effect of injection angle on powder deposition is illustrated in Fig.3.5; with decreasing angle the area of contact of powder increases. An adjustable injection tube angle system which was made during this work is shown in Fig.3.6. The angle of the injection tube was chosen between 35° - 40° .

3.3 Recycling The Reflected Energy

In previous investigations involving laser processing of aluminium alloys (23,31,68) problems have been encountered due to their high reflectivities at infrared wavelengths and their high thermal conductivities. These two properties have also made aluminium alloys the most difficult to laser weld.

Reduction of laser reflectivity by oxidised surfaces is well known. A similar effect is shown by roughened and shot blasted surfaces.

All ^{the present} experiments ^{were} carried out using a reflective dome and a reflective nozzle. The reflective dome has been described by Weerasinghe.(14,33)

A new reflective nozzle was designed and produced by the author. (Fig.3.5a)

This optical feed back system (dome or nozzle) reflects back the laser energy reflected by the material surface. A

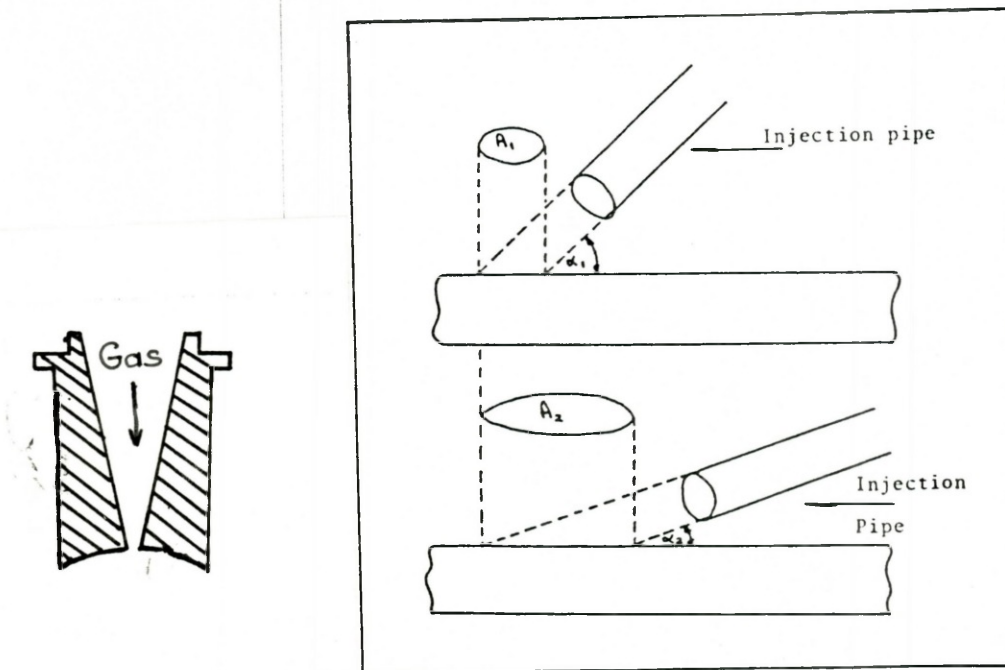


Fig.3.5 Effect of injection angle.

Fig.3.5a A view of the reflective nozzle

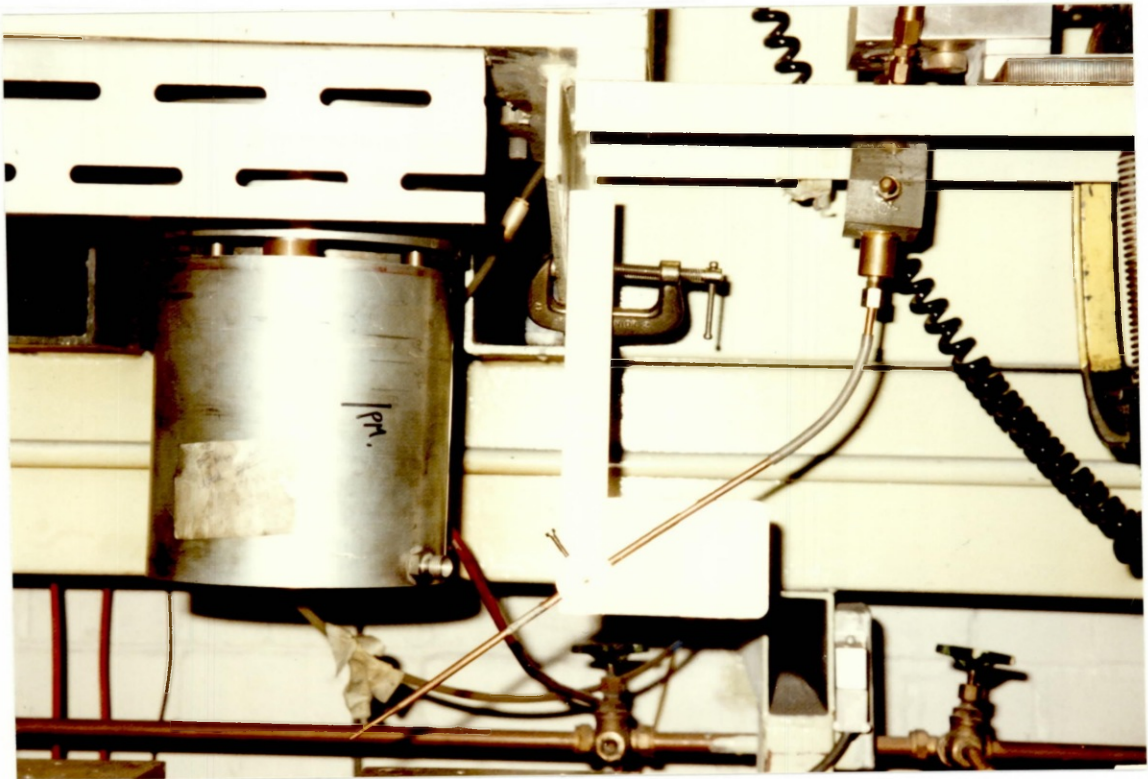


Fig.3.6 Injection tube angle adjusting system.

grit blasted aluminium alloy (LM25) surface was not affected at 5mm beam diameter and 1.95 kW without optical feed back; however, melting was obtained with the reflective dome under the same conditions.

3.4 Materials Used

The LM 25 Al-Si alloy was supplied by BL Cars Ltd., as cast, flat plates 150X75X15 mm; prior to shot blasting and degreasing, some of the plates had their surfaces machined. The chemical composition of this alloy is shown below:

wt %	Si	Fe	Cu	Mn	Mg	Ni	Zn	Pb	Sn	Al
	6.5-7.5	0.5	0.1	0.3	0.2	0.1	0.1	0.1	0.1	balance

The cladding, alloying and particle injection materials used in this work were in the form of powder. Powders used were:

Powders	Particle Size
Silicon (99 % Si)	44-125 μm
Silicon Carbide	~125 μm
Aluminium Oxide	very fine (not defined)
Aluminium (Commercially Pure)	44-125 μm
Mild Steel	~125 μm

3.5 Specimen Preparation/Processing

Specimens were clamped on an x-y hydraulic table with screws between two pieces of metal bar.

Firstly, a beam print was made on the substrate without any powder flow; then powder was blown onto the beam print by adjusting the powder tube for alignment. For this purpose usually a visible HeNe laser beam was used.

Processing parameters investigated were laser power, scanning speed, beam diameter, and powder flow rate. Laser outlet power was usually over 1.35 kW because of the high reflectivity of aluminium alloys, which causes energy loss; it was altered by changing the current into the discharge tubes.

The scanning speed of LM 25 samples was changed by adjusting the flow of liquid to a hydraulic ram to the X-Y table and the time read out from an electronic timer fixed to a sensor of hydraulic ram. In experiments a defocused beam was used; the KCl lens used for focusing the beam was moved inside the laser focusing head. The spot diameter of the beam is adjusted by moving the laser focusing head together with the lens up or down to the specimen.

Beam diameters were typically between 2 and 5 mm, and the scanning speeds between 3 and 35 mm/sec.

The laser power density, P/A , (P is power, A is spot area) could then be calculated.

3.6 Post-Laser Treatment Specimen Preparation

Specimens prepared by laser treatment were cut by a cutting off machine and mounted in Araldite. After mounting, specimens were ground on SiC abrasive to 1200 grit and then polished to 0.25 μm with diamond compound. The tracks produced were generally examined for size and shape using a Projectina Model 4011-4016/MMA macroscope.

3.7 Optical Microscopy

An optical microscopical examination was made of all specimens. Samples were sectioned in transverse, longitudinal and plan sections. Nikon and Reichert optical microscopes were used to obtain micrographs. Samples were etched in different etchants. Etchants used were 0.5% HF in H_2O ; Keller's Reagent and 1 vol. HCl 1 vol. $\text{C}_2\text{H}_5\text{OH}$, 1 vol. H_2O (for SEM 30 sec etching time), for aluminium-silicon system. 20% H_2SO_4 , 80% H_2O solution at 70 $^{\circ}\text{C}$ and Keller's Reagent were used for the aluminium-iron system. Mild steel clad layers were etched in 4% nital for 5 sec. Keller's Reagent and 1 vol. HCl, 1 vol. $\text{C}_2\text{H}_5\text{OH}$, 1 vol. H_2O for SiC injection.

3.8 Scanning Electron Microscopy and Composition Analysis

Scanning electron microscopy was used to examine both polished, deeply etched and fractured surfaces of samples. A JEOL T200, and a JEOL JSM 35 SEM were employed in these examinations. Samples were examined upto 6000X magnification. Compositional micro-analyses were carried out on a JEOL JSM 35 SEM fitted with an EPMA instrument. Samples for analysis were polished to 1 μm and coated with a thin carbon layer to obtain a good thermal and electrical conductivities before putting in the microscope. Accuracy of analyses was within $\sim 0.3\%$. Energy dispersive X-ray spectra and quantitative analysis were obtained from the computer fitted to the microscope.

3.9 X-Ray Diffractometry

X-ray diffraction examination of surfaces of various specimens was carried out using a Phillips PW 1050 diffractometer using $\text{Cu K}\alpha$ radiation source. Specimens were cut off from laser treated surfaces and then ground and polished to 1 μm . These specimens were mounted on the rotating stage of the X-ray diffractometer which was operated at 40 kV and 40 mA. The specimen size of single tracks was, about 2-3 mm wide, which was smaller than the desirable size due to limitation on track width. Therefore, some peaks given in the references were not always obtained.

Peaks were evaluated in two theta, θ , and d values calculated

from $\lambda = 2d \sin \theta$ (Bragg Law) and searched in the Hanawaltt index. The accuracy of measurements was within 0.5°/cm.

3.10 Hardness Testing

Microhardness measurements of polished samples were carried out on a Leitz Miniload 2 microhardness tester. The average track hardness was determined from a minimum of four impressions on uniform alloyed samples. Micro hardness of several tracks against depth below track surface were recorded as well.

CHAPTER 4

RESULTS

4.1 SURFACE ALLOYING

4.1.1 SILICON ALLOYING

4.1.1.1 Operation Parameters

4.1.1.1.1 Scanning Speed/Powder Flow Rate

Two parameters, scanning speed, and powder flow rate affect the mass deposition rate at a constant feeding angle. For a constant powder flow rate, mass deposition on the substrate increases with decreasing scan speed. Mass deposition rate is an important factor which affects the bead profile i.e. particularly height, and consequently cross sectional area of a track and level of dilution. The width of a track is not greatly affected by mass deposition rate, because, the width of a track is limited by beam diameter. Therefore, the change of track width with scanning speed is not clear. Fig. 4.1.2.

Track height shows a regular and meaningful change i.e. it decreases with increasing scan speed. This is shown in Fig. 4.1.1.

For slower scanning speeds, due to increase of height of track, width tends to increase while depth remains constant and cross sectional area of melted substrate become less. Photomacrographs of silicon alloyed samples for different scanning speeds are shown in Fig. 4.1.3b

Useful deposition of powder and alloying does not occur at velocities faster than 25 mm/sec. For slower scan speeds useful tracks have been produced. However, scan speeds of slower than 5 mm/sec produce beads on the substrate higher

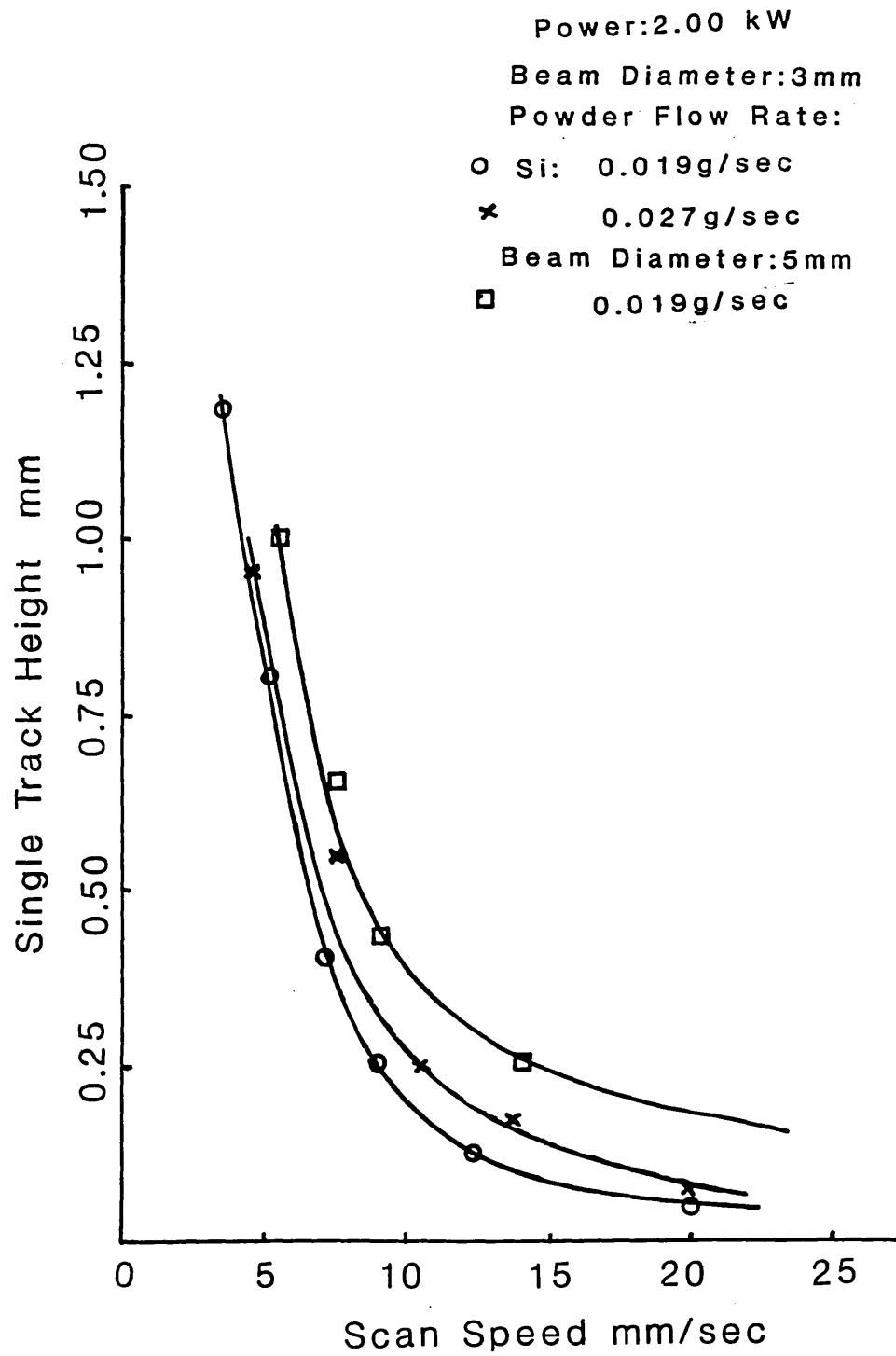


Fig.4.1.1. Single track height vs scan speed in silicon alloying.

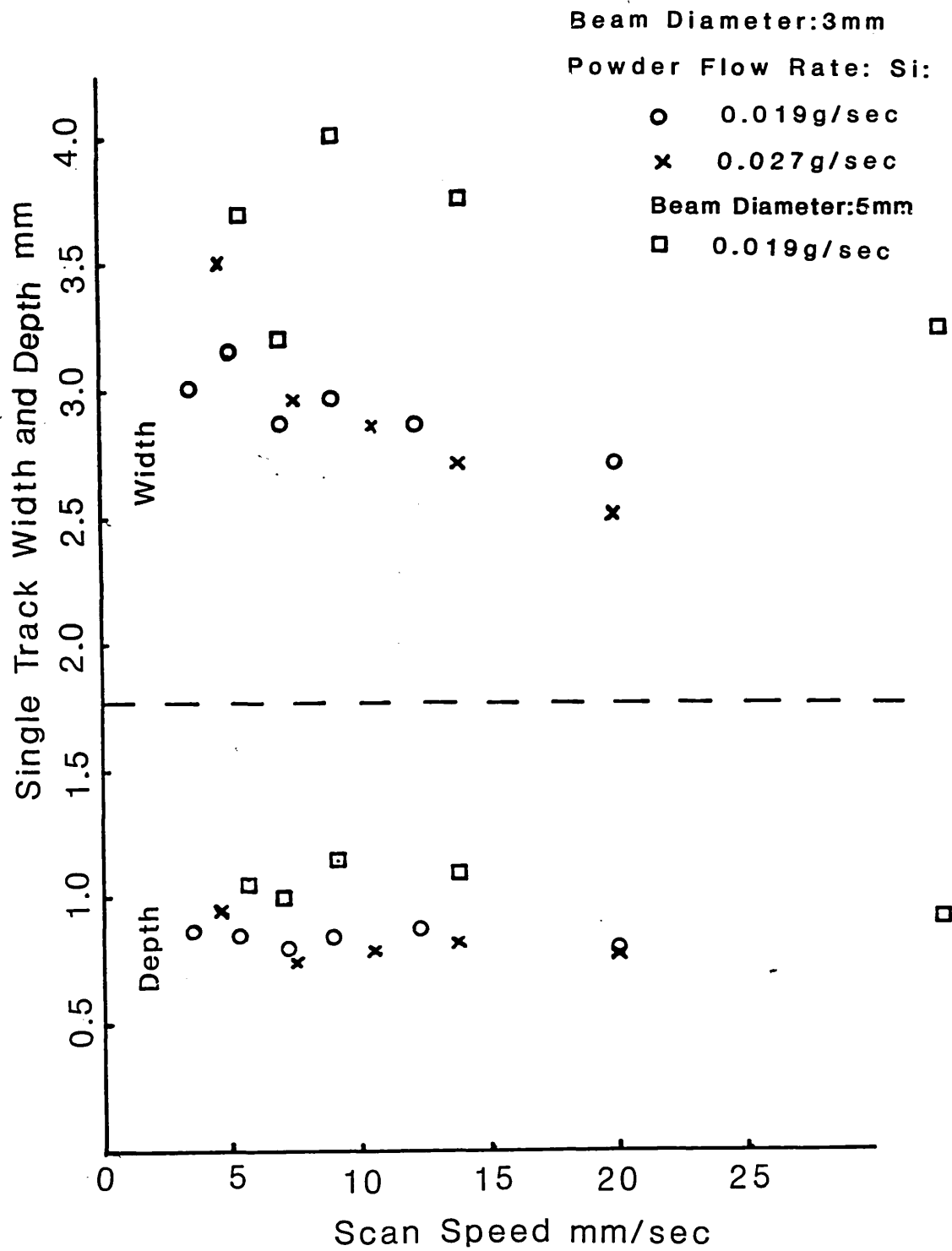
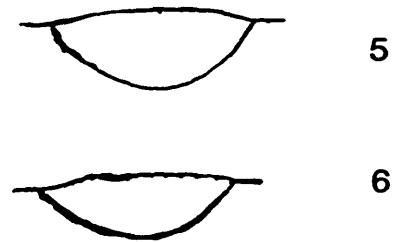
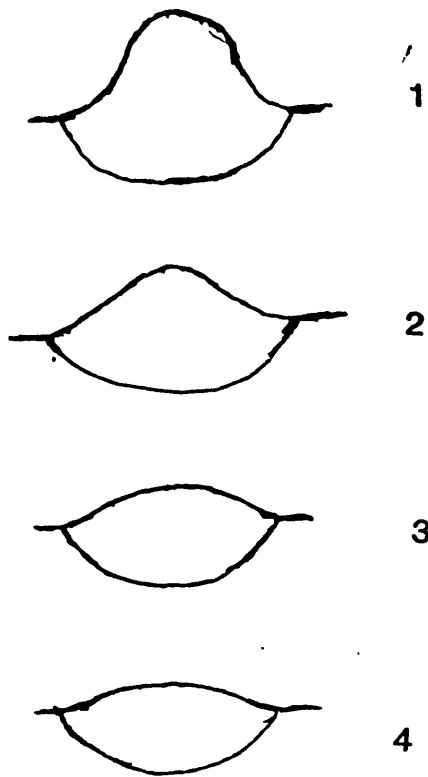
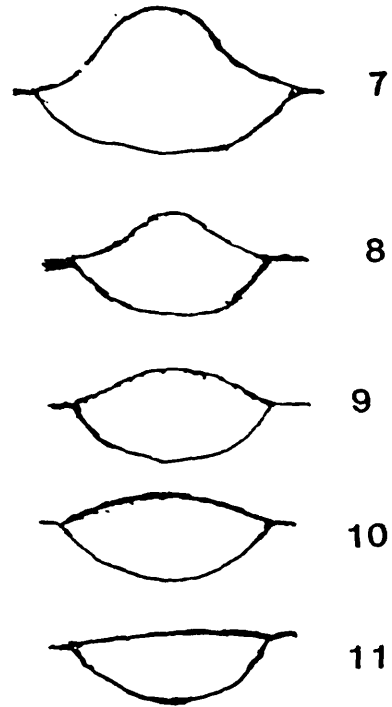


Fig.4.1.2 Single track width and depth vs scan speed in silicon alloying.

Beam diameter: 3 mm



Beam diameter: 3 mm



Beam diameter: 5 mm

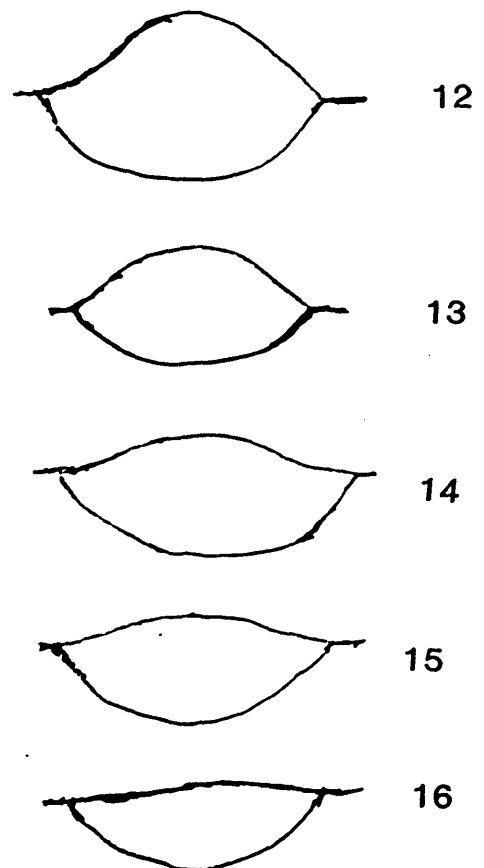


Fig.4.1.2a Bead profiles of silicon alloyed tracks.
(Operational parameters and bead sizes are given in table 1
and 1a in appendix.)

than required; this kind of bead high profile is not suitable for overlapping of areas. Fig. 4.1.3a illustrates this kind of bead.

However, bead profile is not only dependent on scanning speed but is also dependent on powder flow rate. Therefore, scan speeds faster than 5 mm/sec for a high flow rate may produce the high profile of the bead. The important effect of scan speed is on the distortion and extensive melting of substrates. Interaction time increases with decreasing scan speed. It means more energy input and it distorts the samples, as is quite clear on thinner samples, and it causes residual stress in the substrate. It also causes unnecessary melting of the substrate. The effect of lower scan speeds on sample distortion and melting is shown in Fig. 4.1.4a

4.1.1.2 Track Properties

4.1.1.2.1 Surface Appearance

Surface appearance of silicon alloyed layers is dependent on uniformity of powder flow, size of powder transportation tube at outlet point, powder flow velocity and tube angle. Fig.4.1.3b-c illustrates the surface morphology of some silicon alloyed tracks. Some tracks have comparatively flat surfaces (surface roughness is $\sim 50 \mu\text{m}$ and can be accepted for production). It is clear that fluctuation of powder flow is very low, therefore surface quality is better. On the other hand some tracks illustrate the effect of fluctuation of powder flow on bead morphology. In this case, the surface roughness is $\sim 400 \mu\text{m}$. Remelting of

Fig.4.1.3a Surface appearance of silicon alloy/clad overlap layers and single tracks. Single tracks next to the overlap layer from right show the effect of non-uniform and high powder feeding phenomena. This type of track is not suitable for overlapping process.

P=1.93 kW, BD=4.8 mm

Track 1: SS=8.80 mm/sec, PF=0.05 g/sec

Track 2: SS=3.00 mm/sec, PF=0.05 g/sec

1st overlap: SS= 5.50 mm/sec PF=0.027 g/sec, 2mm overlap

2nd overlap: same parameters remelted without powder flow.

Scale:1/1

Fig.4.1.3b Surface appearance of silicon alloy single tracks. Scale:1/1

P=1.94 kW, PF= g/min. SS= mm/sec Track No

BD=4 mm	1.05 (1,2,8)	5.40	1,4,12
	0.65 (3-7)	3.50	2,3
	1.20 (9-12)	13.75	7,8,9
		7.90	5,11
		6.25	6
		18.50	10

Fig.4.1.3c Surface appearance of another silicon alloy layers with more uniform and flat surface. Scale:1/1

P=2.00 kW, PF=g/min SS=mm/sec

BD=3 mm 1.00 (1,2,3,6,7) 9.45 (1,2,4,5,8,9,15,17)

1.25 (4,5,8,9,16) 4.80 (3,6,7,16)

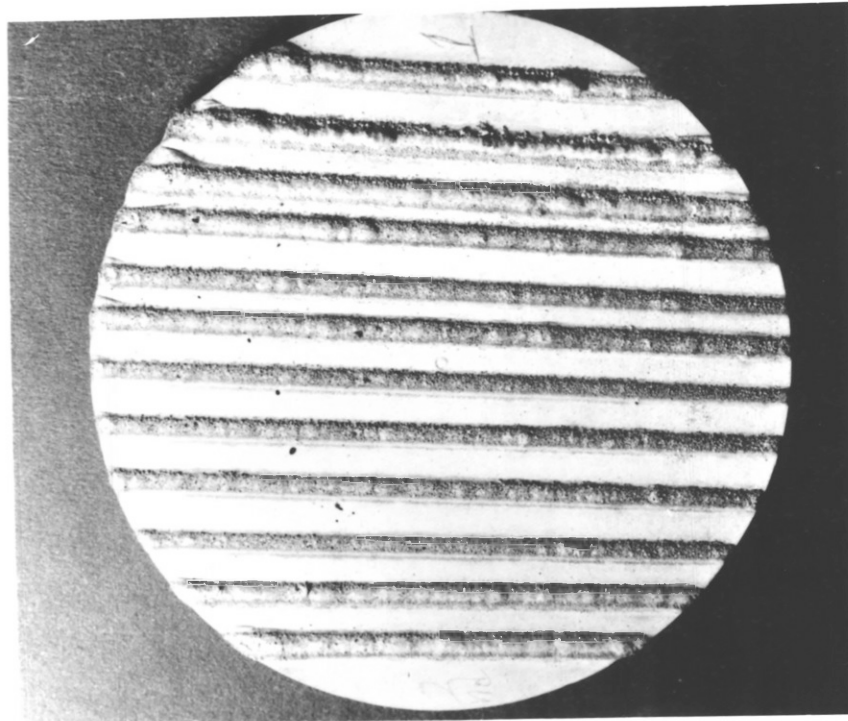
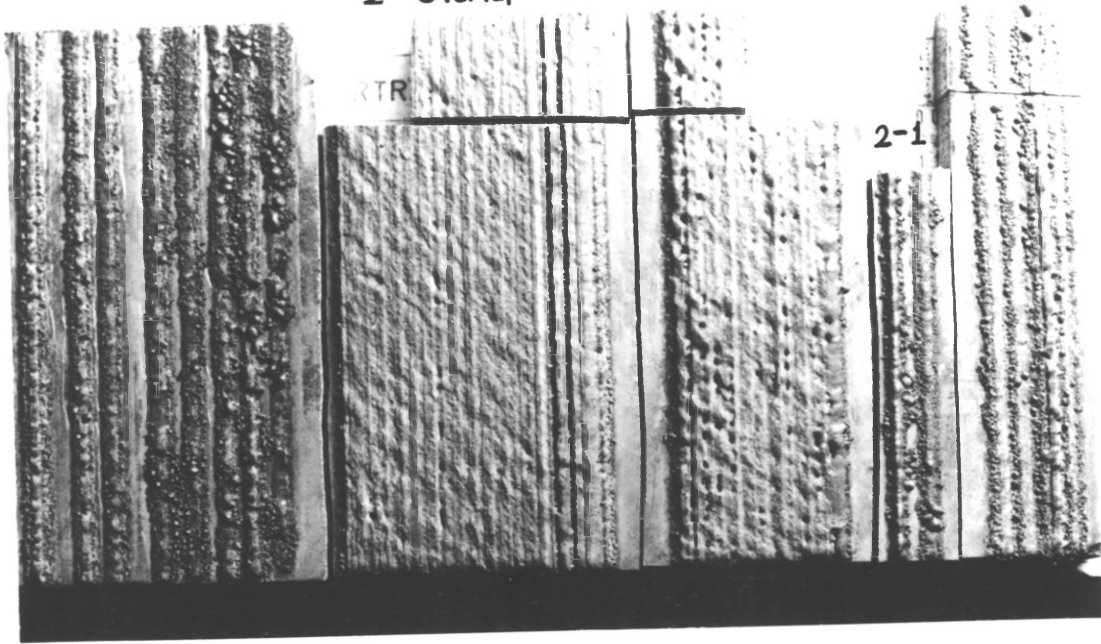
Parentheses show track number. Track 1 is the top one.

P:power, Bd:beam diameter, SS:scan speed, PF:powder flow

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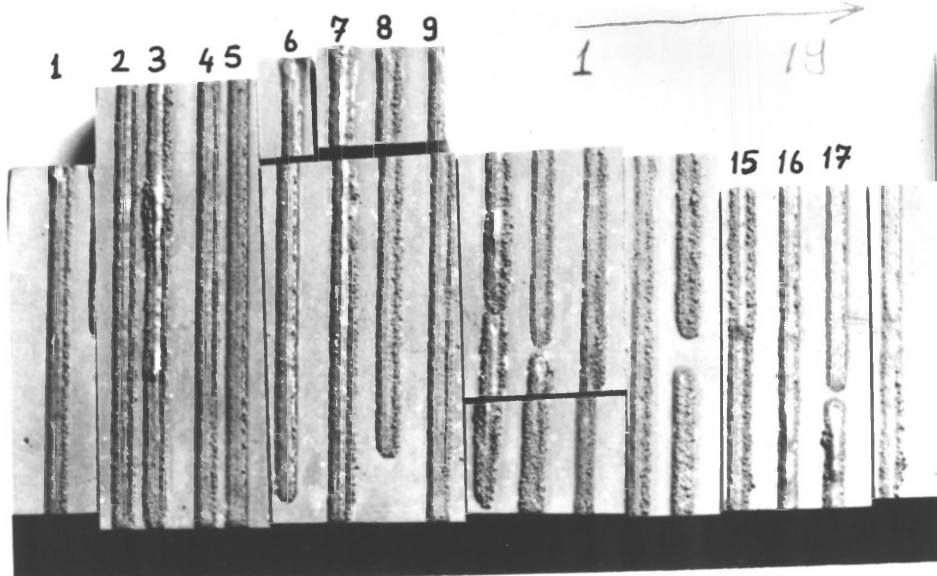
2nd Overlap

1st Overlap



1

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tracks and overlap areas without blowing powder may produce smoother flat surfaces. Powder particles around beads are the result of the following:

- a) Wider tube diameter.
- b) Scattering of powder from substrate around bead.
- c) Particles spread from previous track processing on the substrate partly melting around the bead in the low energy density area.

This effect can be minimized at optimum tube size and flow velocity, but it cannot be eliminated totally.

4.1.1.2.2

Cracking

The presence of cracks in a material is an important phenomenon affecting its physical properties and use. There are no cracks and fissures observed in silicon alloyed layers. The aluminium-silicon phase diagram is a simple eutectic. Constituents are proeutectic aluminium, eutectic and proeutectic silicon. It has no intermetallic compound which may cause a brittle structure. However, at high silicon ratios cutting off produced some mechanical cracking but it is not related to structural cracking. All photographs relating to silicon alloying illustrate the absence of cracking.

4.1.1.2.3

Porosity

The main problem to produce good quality alloy layers

apart from uniform feeding of powder is the presence of both large and small porosities. Fig.4.1.4b shows one of the tracks with porosity. The cause of porosity may be one or more of the following:

1) ^{Gas present in} Substrate material.

2) Moisture and other impurities and organic materials in the powder.

3) Gases during melting.

4) Entrapment of shrouding gas.

Experiments were carried out using two different silicon powders. When the samples of the two powders were examined metallographically two situations concerning porosity were encountered. Samples produced using one kind of powder, which was cleaner than the other powder were almost porosity free. Fig.4.1.4c. With the other powder there was extensive porosity. Fig.4.1.10c shows the situation of a fractured surface with much porosity. Therefore, materials used should be chosen carefully.

Micro analysis results of a few samples showed the presence of argon in the bead; this means the shrouding gas may cause porosity as well.

Lastly, the substrate has considerable porosity as a manufacturing fault.

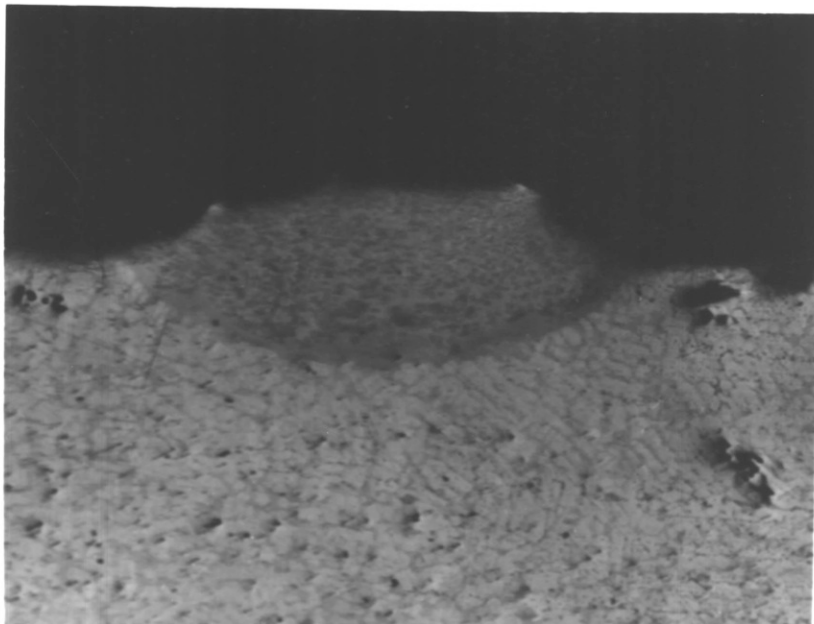
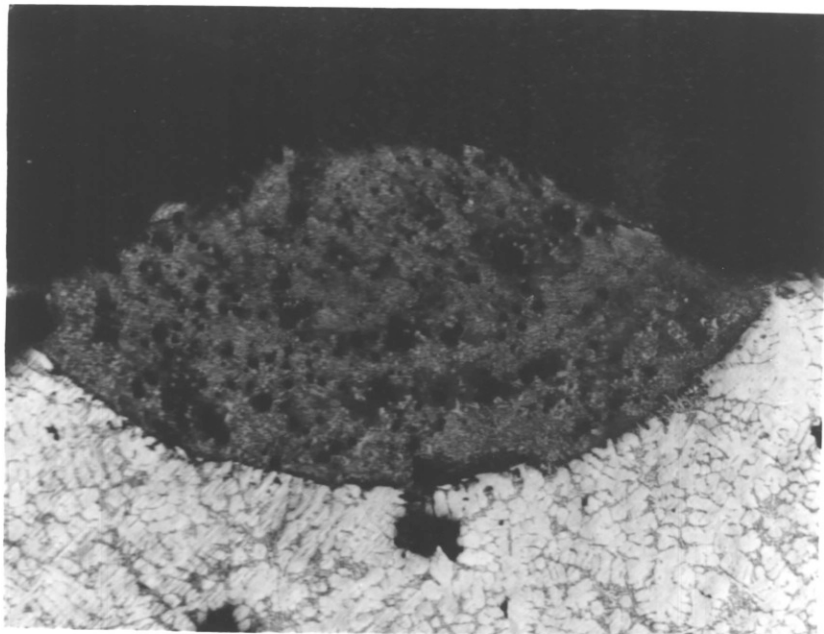
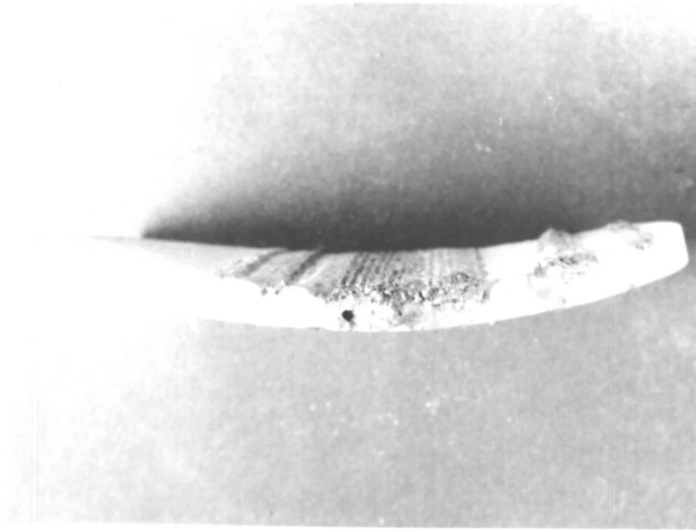
4.1.2 SILICON + ALUMINIUM ALLOYING

LM 25 Al-Si alloy substrates were alloyed with Si+Al powder mixture at different powder flow rates (table 2 in

Fig.4.1.4a An overlap silicon alloyed sample shows the distortion of sample. Scale:1/1

Fig.4.1.4b A silicon alloy layer showing considerable amount of porosity. Mag.22

Fig.4.1.4c Macrograph of a silicon alloyed bead which processed with different silicon powder is almost porosity free. Mag.22.



appendix) and scanning speeds with changing Si/Al ratio at different beam diameters (Si/Al ratio could not be changed easily due to difficulty of feeding silicon powder; therefore premixed powder was used). The effects of scan speed and powder flow rate on track geometry are illustrated in Fig. 4.1.5a

The surface appearance is similar to silicon alloying. There is a good bond between the substrate and track layer.

No cracking was observed, and the general structure is the same as to silicon alloying. However all of the beads contain both large and small porosity.

Increasing powder feeding rate increased the bead height and reduced the dilution and depth of the melted zone. Increase of aluminium ratio in the Si/Al mixture makes laser processing more difficult ($\text{Si/Al} = 1/3$). Tracks become discontinuous and the melt formed into large lumps, Fig. 4.1.5a, as a result of the high surface tension forces. Higher aluminium ratio would reduce the beam coupling efficiency; therefore, the energy density would decrease, and the melting would become less efficient and the extent of the melted zone in the substrate would be reduced.

A few experiments were carried out using two separate powder feeders. The ratio of Al/Si was changed by altering the screw speed of the powder feeder. In this case there is a limitation on the minimum amount of powder deposition. The mass is doubled due to coupling of two feeders and the control of powder flow becomes more difficult; therefore, the lower limit on the size of bead for a constant scan

Fig.4.1.5a (Top left) Appearance of Si+Al alloy layers at different processing parameters. Globular, discontinuous tracks illustrate the effect of high powder flow rate or slow scan speed. Scale:1/1 (See table 2 in appendix for the operating parameters; tracks 1-17)

Fig.4.1.5b Microstructure of a silicon alloyed track with plate like morphology. Mag. 200.

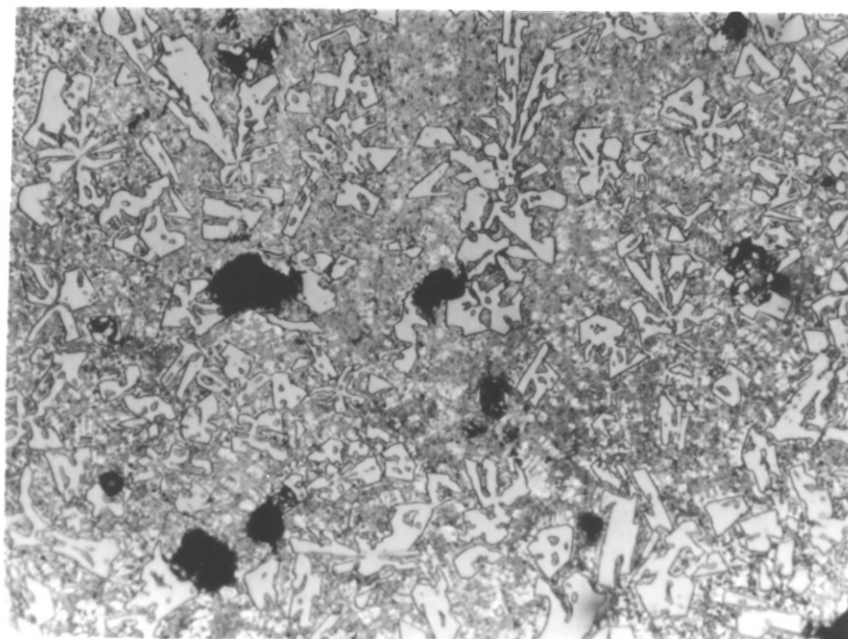
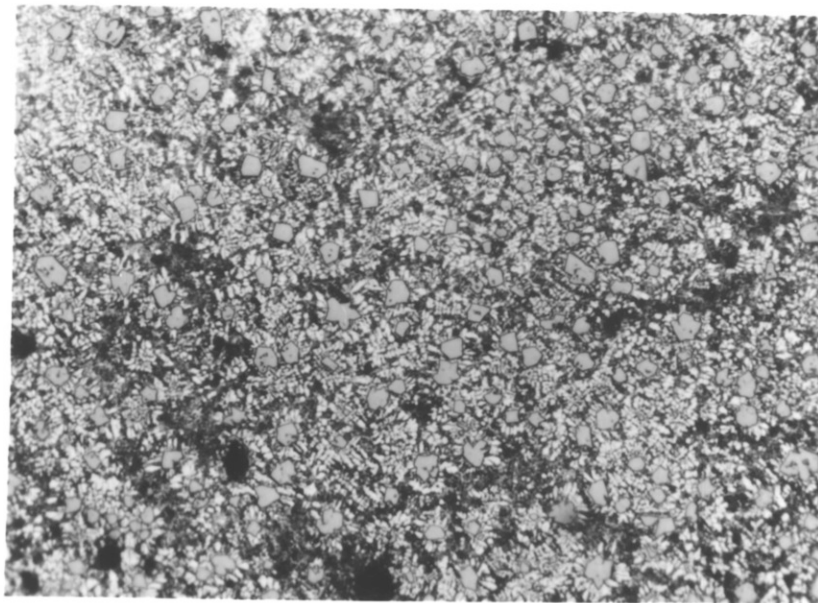
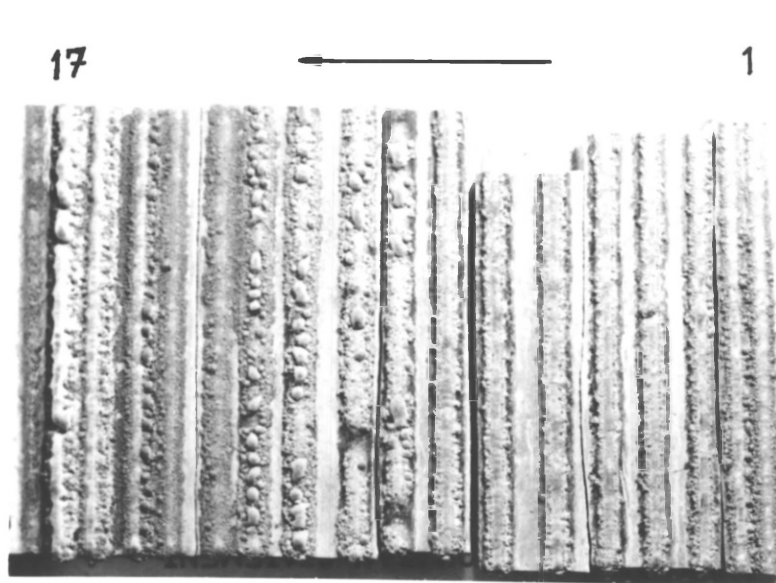
P=1.93 kW, BD=4.8 mm, SS=5.4 mm/sec, PF=0.05 g/sec

Fig.4.1.5c Optical micrograph of a silicon alloyed track with polyhedral silicon particles. Mag. 186.

P=1.90 kW, BD=4.8 mm, SS=7.75 mm/sec, PF=0.019 g/sec

Fig.4.1.5d Optical micrograph of dendritic and star like proeutectic silicon in a silicon alloyed track. Mag. 186.

P=1.93 kW, BD=4.8 mm, SS=8.80 mm/sec, PF=0.027 g/sec



speed becomes larger.

The effect of increasing aluminium ratio on composition is the reduction of silicon in the alloyed region.

Microstructural features of both silicon and Si+Al alloyed layers are the same; therefore these results are given together.

4.1.3

MICROSTRUCTURAL ANALYSIS

Highly alloyed tracks were produced by blown powder silicon alloying (over 50 wt% Si). The microstructures of these alloy beads are shown in Figs. 4.1.5b and 4.1.6a-b. These hypereutectic alloys have microstructures of proeutectic silicon, proeutectic aluminium and fibrous aluminium-silicon eutectic. Morphology of proeutectic silicon exists as plate-like, star-like, dendritic or polygonal shapes depending on Si content. Fig. 4.1.5b-d. A distinctive aluminium halo (shell) coats proeutectic silicon particles in general and separates them from eutectic. Fig. 4.1.7. Proeutectic aluminium was located around proeutectic silicon and in the eutectic area. However, this surrounding proeutectic aluminium does not form a continuous shell. Therefore, eutectic silicon nucleates from proeutectic silicon. In some tracks, the distribution of proeutectic silicon is uneven due to inhomogeneous mixing of alloyed silicon.

The microstructure of laser processed area is finer than the substrate and shows more ductile fracture. Fig.4.1.10a.

4.1.3.1 Proeutectic Silicon

Silicon alloyed tracks were produced of the hypereutectic composition. Therefore, proeutectic silicon is the main feature of this alloy. The morphology of proeutectic Si varies with the amount of silicon alloyed. Highly alloyed regions show plate-like silicon morphology. Figs. 4.1.5b and 4.1.6a. In these areas the silicon content was found to exceed 50 wt%. The plates are directionally oriented and initially formed plates behave as nucleation sites for new silicon plates.

Where the amount of silicon is less than about 50 wt% Si, proeutectic Si forms as star-like, dendritic or polygonal shape. Fig. 4.1.5c-d and 4.1.8c. These star-like and dendritic particles grow with faceted layer growth. Fig. 4.1.9a-b.

In one track containing ~85 wt.% Si massive dendritic silicon was observed in the region close to the substrate. Fig. 4.1.6b. Twinning occurs in some proeutectic silicon particles. Fig. 4.1.9c. The growth mode of dendritic silicon is illustrated in Fig. 4.1.7b. It appears that silicon polyhedra form and these polyhedra twin by producing star-like crystals which grow as faceted dendrites.

4.1.3.2 Eutectic Structure

Eutectic silicon nucleates from pro-eutectic silicon as

Fig.4.1.6a SEM view of a highly silicon alloyed track, taken from the cross section of a track. Proeutectic silicon grows as plates. (Comp. ~85 wt% Si)

Fig.4.1.6b A proeutectic large silicon dendrite with fibrous eutectic network (upper side of the dendrite) around it. (Lower portion of above sample.)

Fig.4.1.6c Very small geometric polyhedra of proeutectic silicon particles were found in the lower portion of a silicon alloyed track.

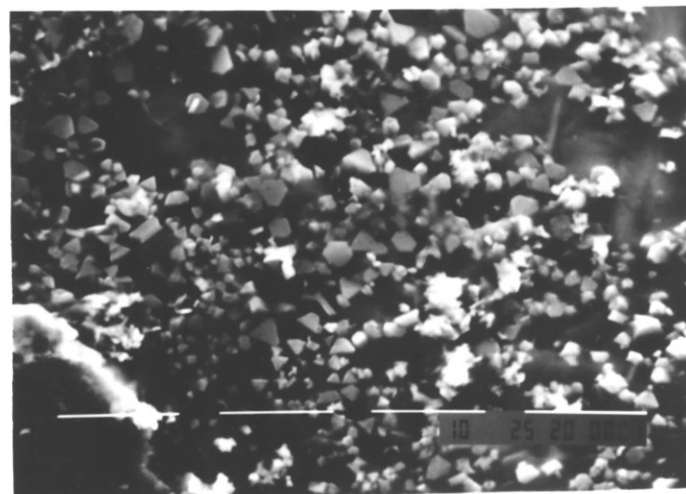
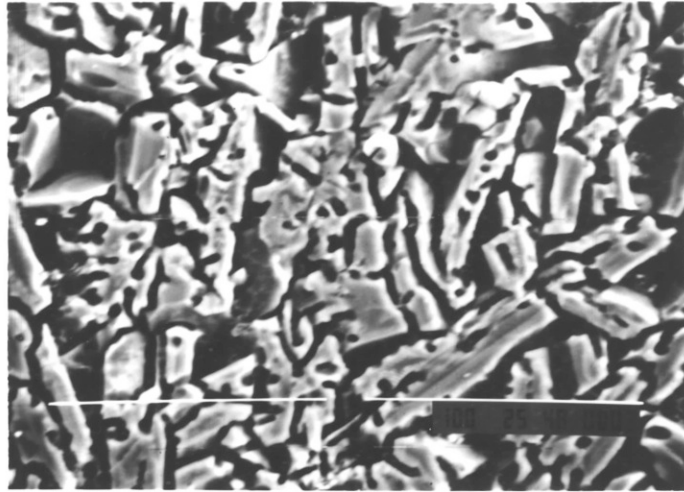
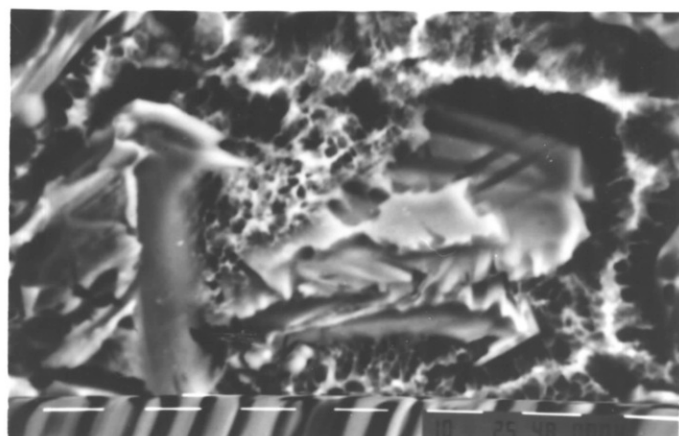
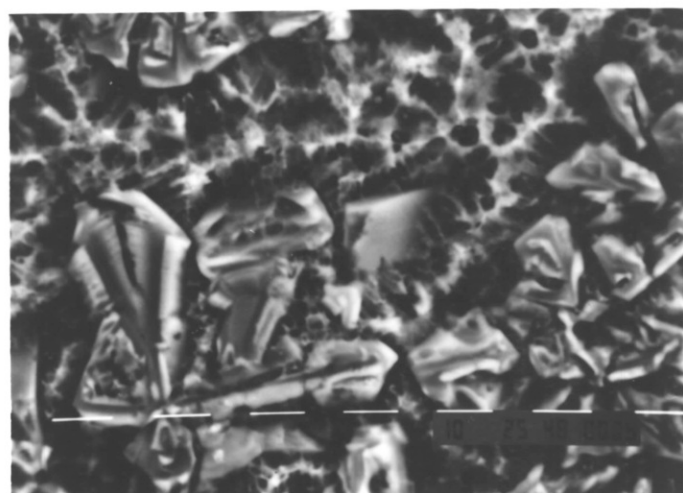
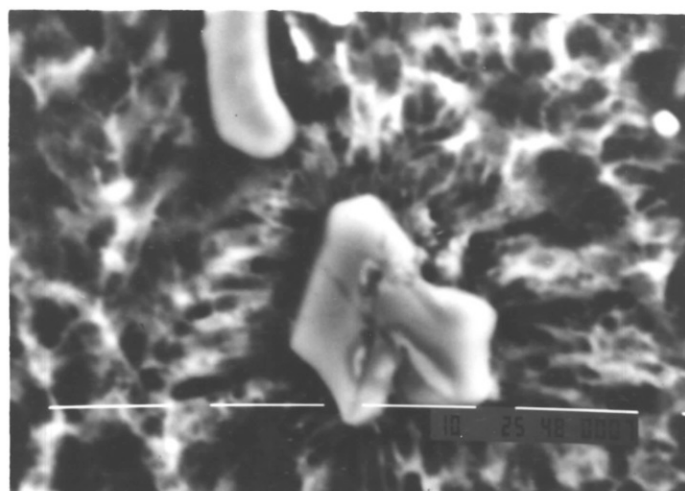


Fig.4.1.7a SEM view of a proeutectic silicon polyhedron surrounded by a fibrous eutectic silicon network.

Fig.4.1.7b SEM view of dendritic and star like proeutectic silicon particles with fibrous eutectic and proeutectic aluminium.

Fig.4.1.7c Another SEM micrograph illustrating fibrous fine eutectic surrounding proeutectic silicon particle (lower section) and the aluminium shell (upper section).



is clear in most figures. The eutectic shows short fibrous morphology growing as rods. Fig.4.1.7. Sometimes it grows through an aluminium halo and in some cases surrounds proeutectic silicon. Edges and corners act as a preferential growth sites. These eutectic silicon rods seem to penetrate the aluminium halo.

Eutectic silicon radiates like a branch of a tree from silicon particles and colonies expand by continuous branching. The eutectic network looks like a sponge and the eutectic spacing is a function of cooling rate. As a result of rapid cooling high microstructural refinement was observed, even though the eutectic interphase spacing was not measured. The eutectic composition was shifted from 12.6 wt.% Si to ~18 wt.% Si as determined by electron microprobe analysis of a track showing a predominantly eutectic structure.

4.1.3.3

Proeutectic Aluminium

When liquid cools down to the liquidus first proeutectic silicon nucleates and grows from the liquid. The liquid composition approaches the eutectic composition and as a consequence of rapid cooling undercooling occurs below the eutectic temperature. The Al-Si eutectic system shows an asymmetrical phase diagram and undercooling initiates proeutectic dendrites. Fig.2.14.

In low silicon hyper-eutectic alloys, proeutectic aluminium nucleates from substrate/melt interfaces

Fig.4.1.8a General SEM view of a hypereutectic Al-Si alloy with proeutectic silicon plates. (composition:~53 wt.% Si)
P=2.00 kW, BD=3 mm, SS=4.90 mm/sec, PF=0.02 g/sec

Fig.4.1.8b Higher magnification of the hypereutectic laser processed bead (4.1.8a) containing proeutectic plate like silicon, Al-Si eutectic and aluminium shell. Eutectic nucleates from proeutectic silicon plates.

Fig.4.1.8c SEM view of a proeutectic silicon dendrite growing by layer growth.

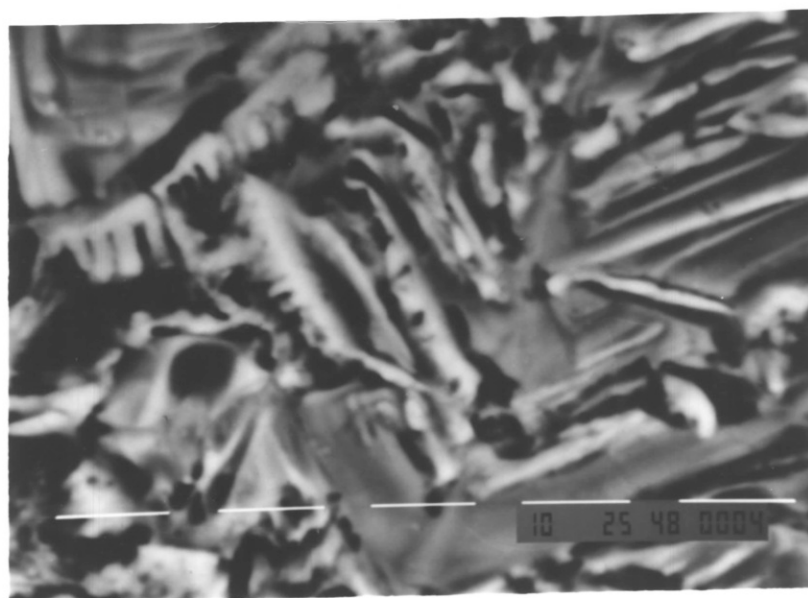
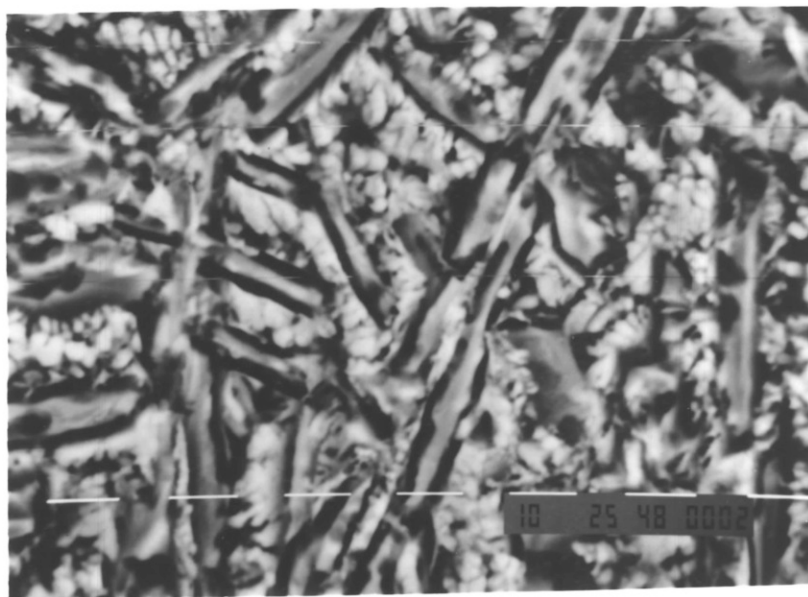
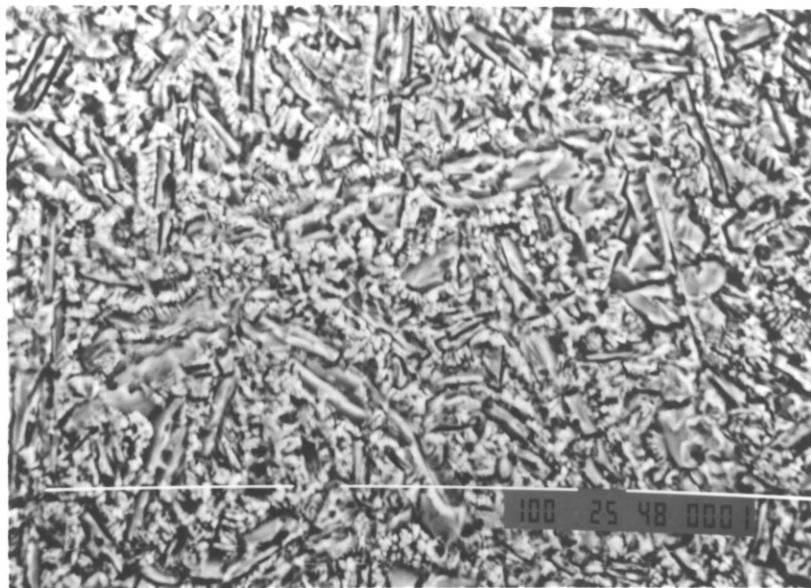
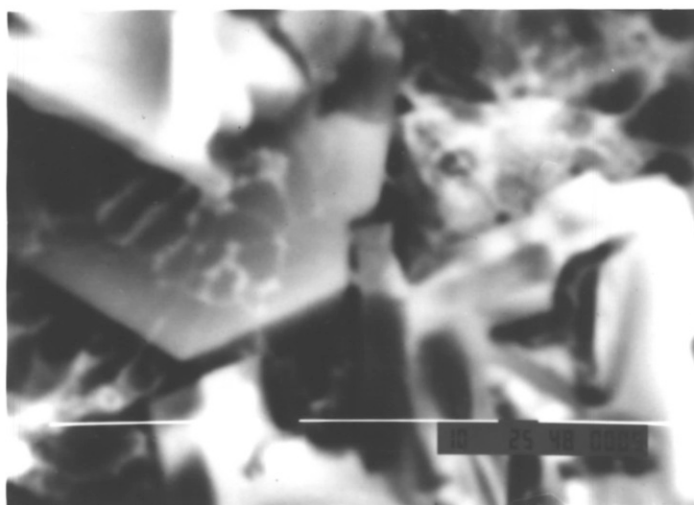
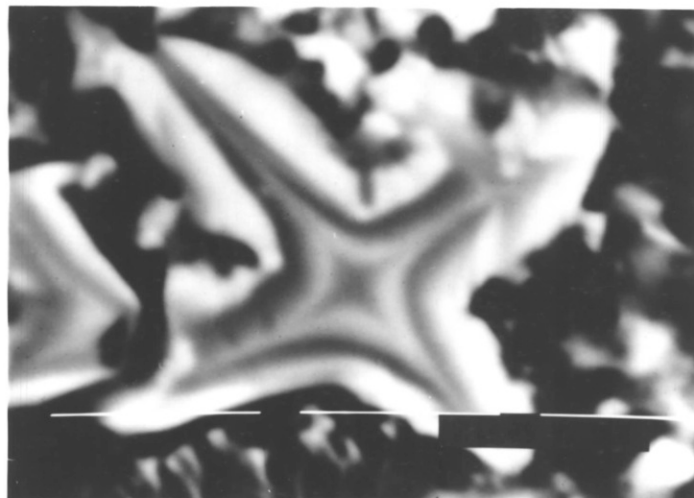
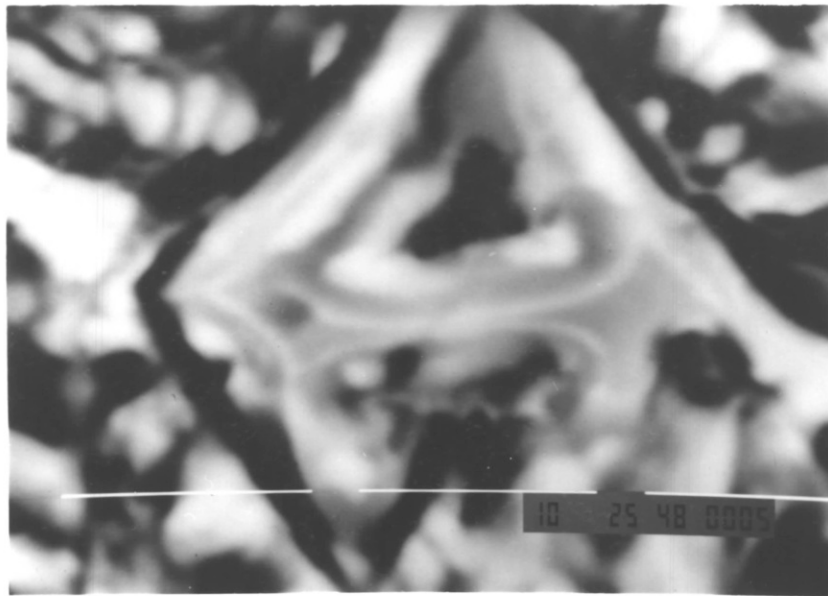


Fig.4.1.9a High magnification SEM view of a silicon particle in hypereutectic Al-Si alloy illustrates faceted layer growth of silicon particle and aluminium shell. (composition: ~40 wt% Si)

Fig.4.1.9b High magnification SEM micrograph of a proeutectic star like silicon with faceted layer growth.

Fig.4.1.9c SEM micrograph of a twinned proeutectic silicon polyhedron and proeutectic silicon /eutectic interface. Eutectic silicon nucleates preferentially on proeutectic silicon.



independently. Where proeutectic silicon particles intercept misgrowth occurs and the obstacle is bypassed; then growth continues either in the original direction or a new dendrite of Al nucleates from the aluminium around proeutectic silicon.

In high silicon hypereutectic alloys proeutectic aluminium seems to nucleate from the aluminium shell surrounding silicon particles. Fig. 4.1.5b. Localised dendritic network growth occurs in all directions instead of directionally growing dendrites.

The aluminium shell surrounding proeutectic silicon results from undercooling below eutectic temperature as the liquid around proeutectic silicon becomes richer in aluminium during solidification. Aluminium in this liquid shell nucleates and grows preferentially below the eutectic temperature. It behaves as nucleation sites for proeutectic aluminium. The thickness of this shell varies on different particles and some proeutectic silicon particles are coated directly by the eutectic silicon network without an aluminium shell. Fig. 4.1.7

4.1.4

X-Ray Diffractometry

A silicon alloyed overlapping track (~60 wt.% Si) was examined in a Phillips PW 1050 diffractometer. The phases present were found to be aluminium and silicon. Experimental and reference d spacings and hkl parameters are given below. (Reference is Hanawalt Index and card numbers are given)

4-0787 (Al), 27-1402 (Si)

Phase	System
Al	Cubic $a_f=4.0494 \text{ \AA}$
Si	Cubic $a_f=5.43088 \text{ \AA}$

d $\text{\AA}$ (Exp.)	d $\text{\AA}$ (Ref.)	hkl	Phase
3.1325	3.13552	111	Si
2.3350	2.338	111	Al
2.0191	2.024	200	Al
1.9151	1.92011	220	Si
1.6357	1.6374	311	Si
1.4284	1.431	220	Al
1.3565	1.35772	400	Si
1.2456	1.24593	331	Si
1.2189	1.221	311	Al
1.1675	1.1690	222	Al
1.1081	1.10857	422	Si
1.0446	1.04517	511	Si
1.0114	1.0124	400	Al

4.1.5 Compositional Analysis

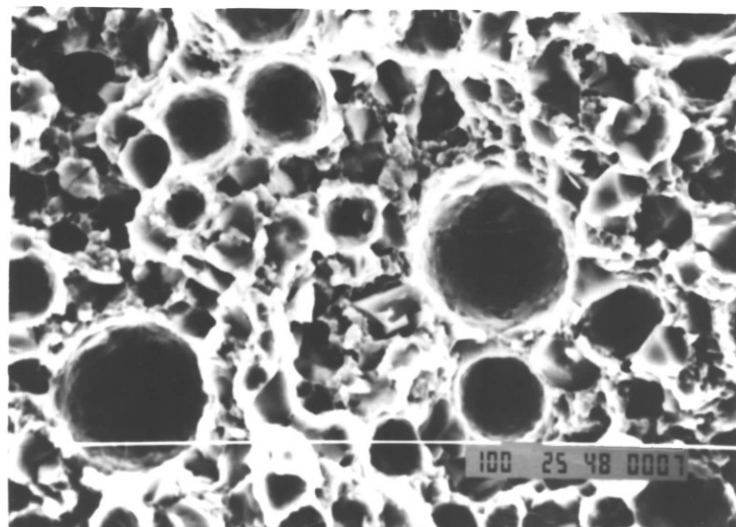
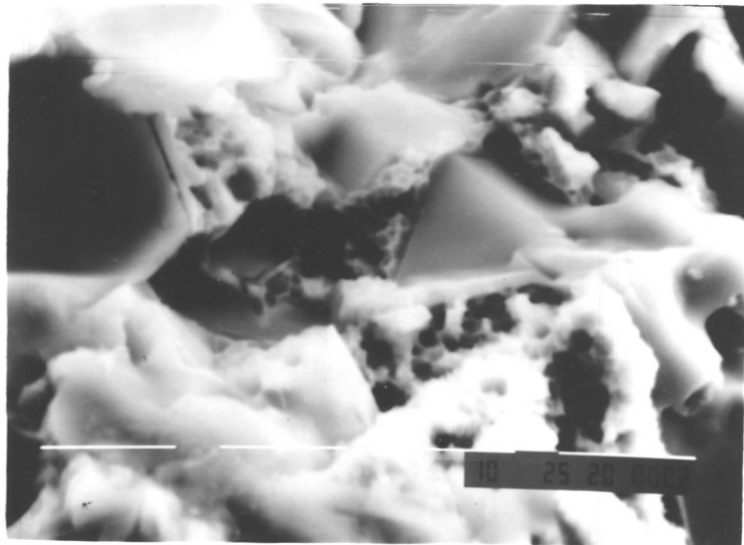
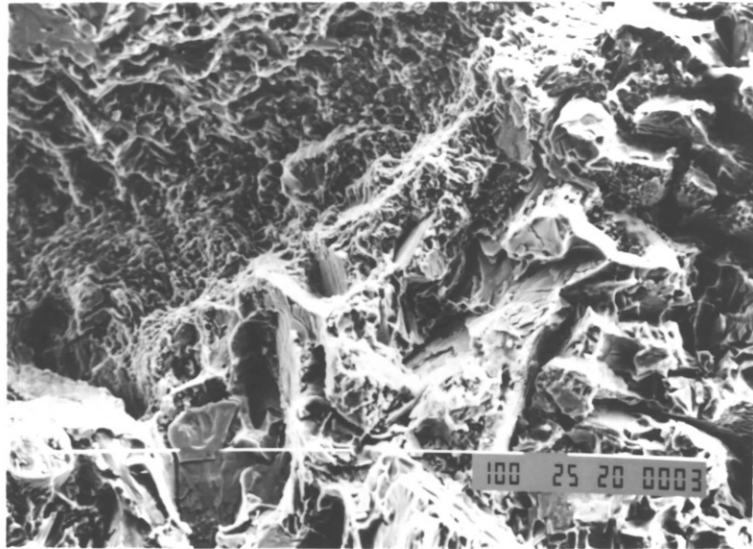
Energy dispersive X-ray spectra analysis was carried out using a JEOL JSM 35 SEM fitted with an EPMA instrument. The results are presented as graphs showing major elements in the alloy tracks together with quantitative analysis results. Figs. 4.1.11 to 4.1.14. Different samples at various operation parameters were analysed to find out the

Fig.4.1.10a Fractured surface of a laser processed sample. Top left section is laser melted/alloyed area and bottom right section is substrate.

P=1.93 kW, BD=4.8 mm, SS=8.80 mm/sec, PF=0.05 g/sec

Fig.4.1.10b Fractured and etched section of an alloyed track illustrating proeutectic silicon and eutectic.

Fig.4.1.10c SEM view of porosity in a fractured and etched sample.



36 CNT

16K FS: A

800 EV

20 EU/CHAN

Link Systems 860 Analyser

14-Oct-87

Si K α Al K α

0 - 0

CO 27

2 - 1

Fig.4.1.11 Energy dispersive X-ray spectrum of a silicon alloyed track.

COMPOSITION: 43.4 at% Al, 56.4 at% Si

124 CNT

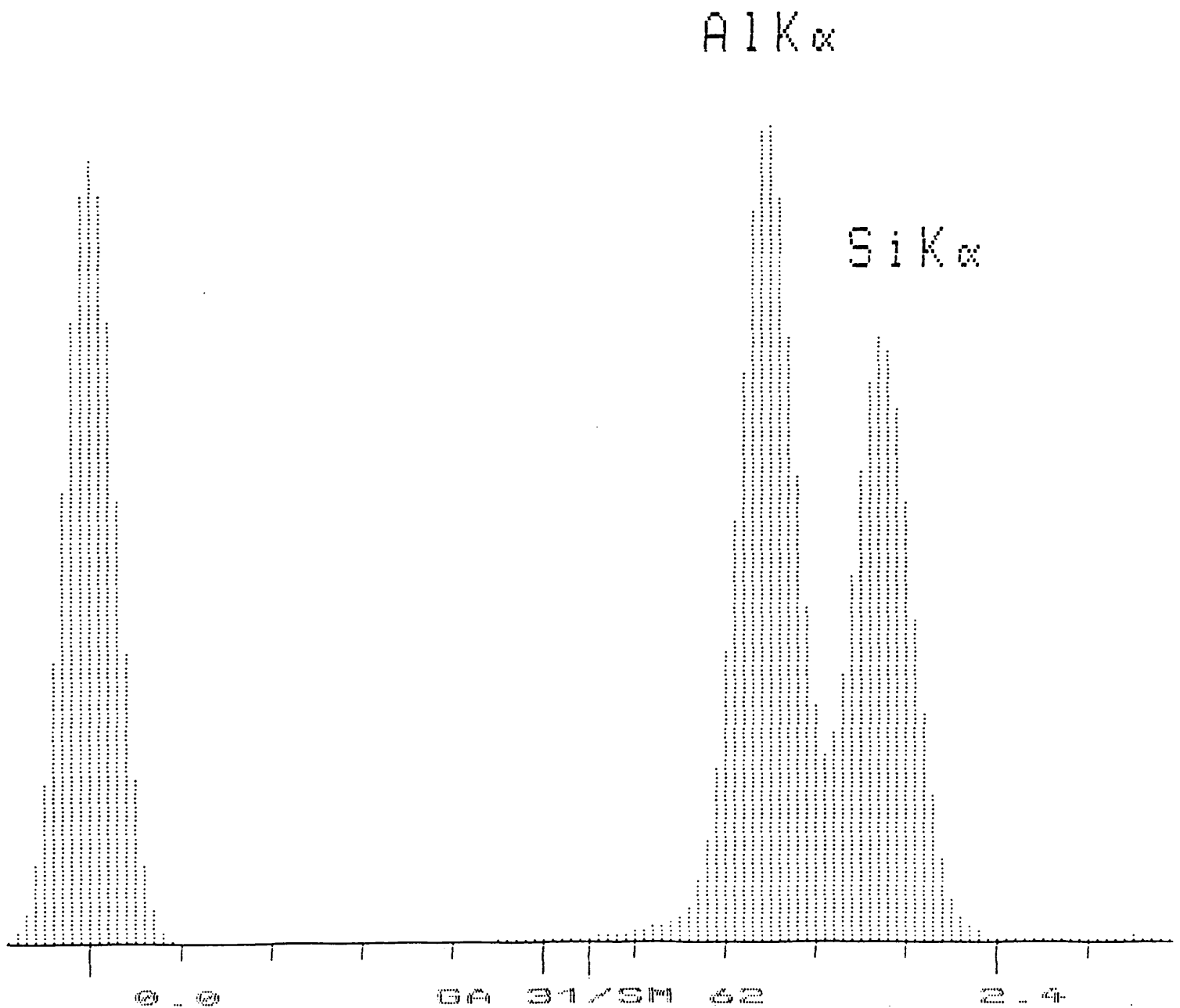
1100 EU
Link Systems 860 Analyser16K FS: A
20 EU/CHAN
14-Oct-87

Fig.4.1.12 Energy dispersive X-ray spectrum of a silicon alloyed track.

COMPOSITION: 56 at% Al, 43 at% Si

107 CNT

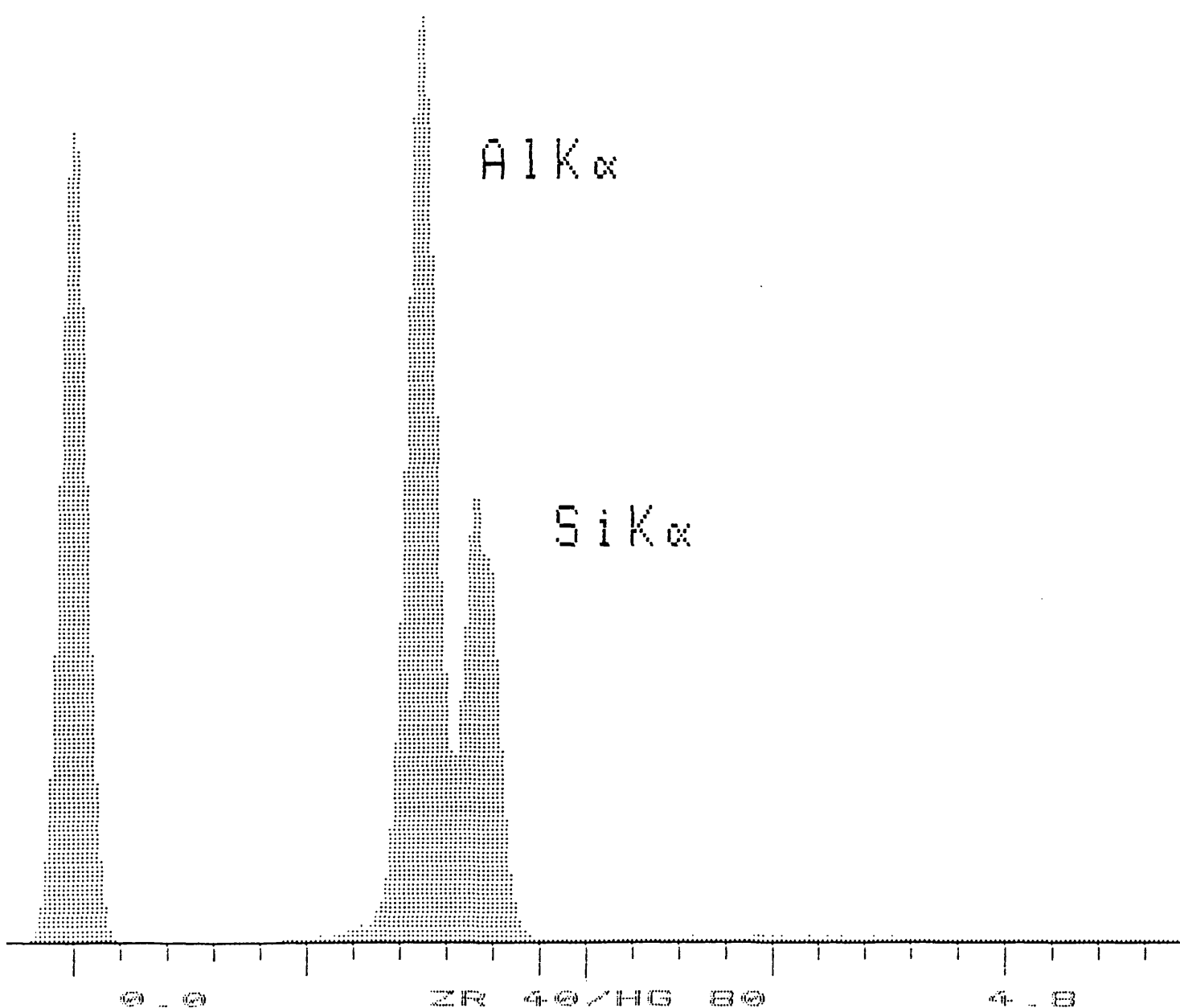
2200 EV
Link Systems 860 Analyser16K FS: A
20 EU/CHAN
14-Oct-87

Fig.4.1.12a Energy dispersive X-ray spectrum of a silicon alloyed track.

COMPOSITION: 70 at% Al, 29.5 at% Si

0 CNT 255 FS: B
1280 EV 20 EV/CHAN
Link Systems 860 Analyser 14-Oct-87

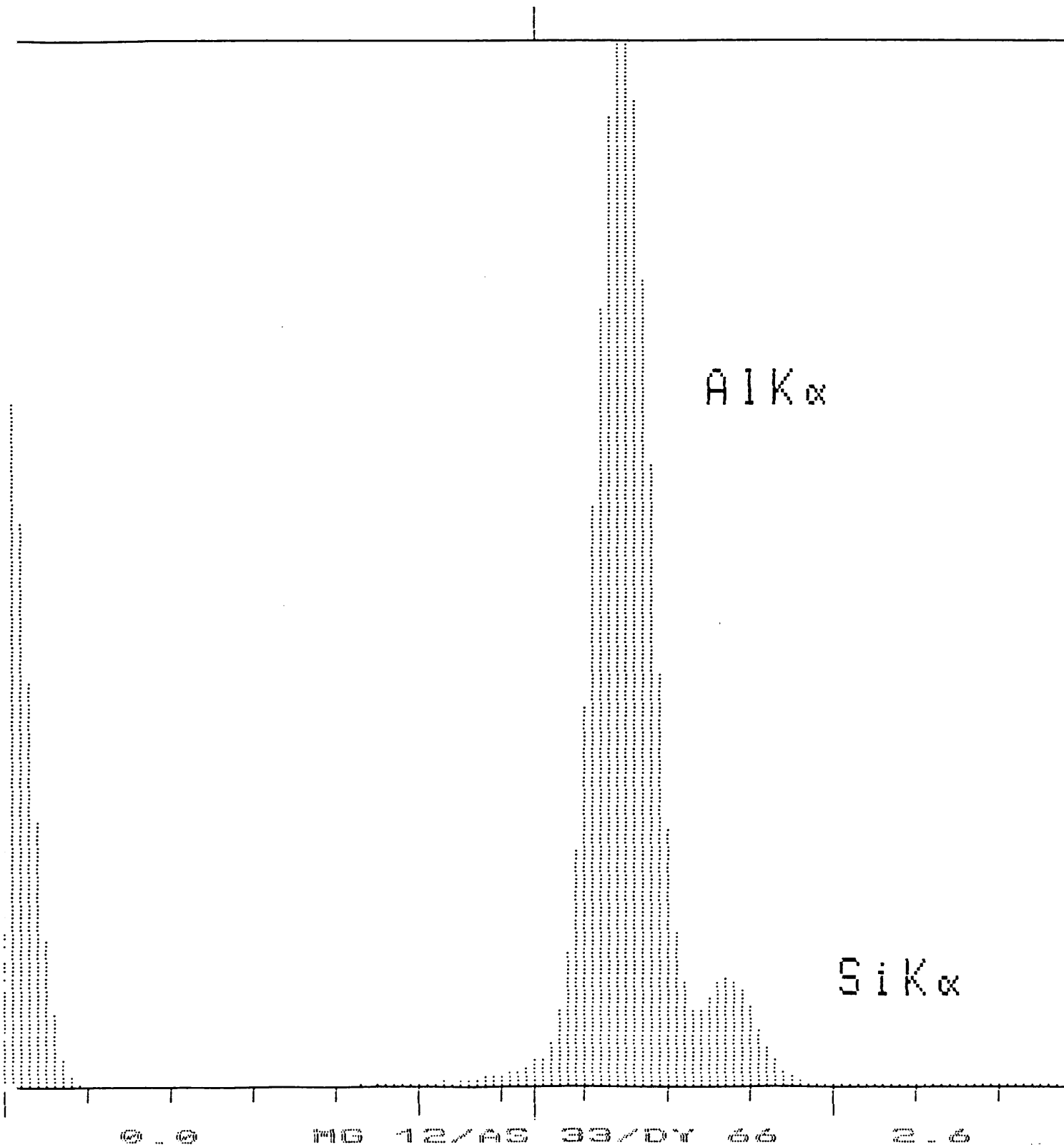


Fig.4.1.13 Energy dispersive X-ray spectrum of a silicon alloyed track. (Eutectic region)

COMPOSITION: 82 at% Al, 17.4 at% Si

0 CNT 16K FS: B
1100 EU 20 EU/CHAN
Link Systems 860 Analyser 14-Oct-87

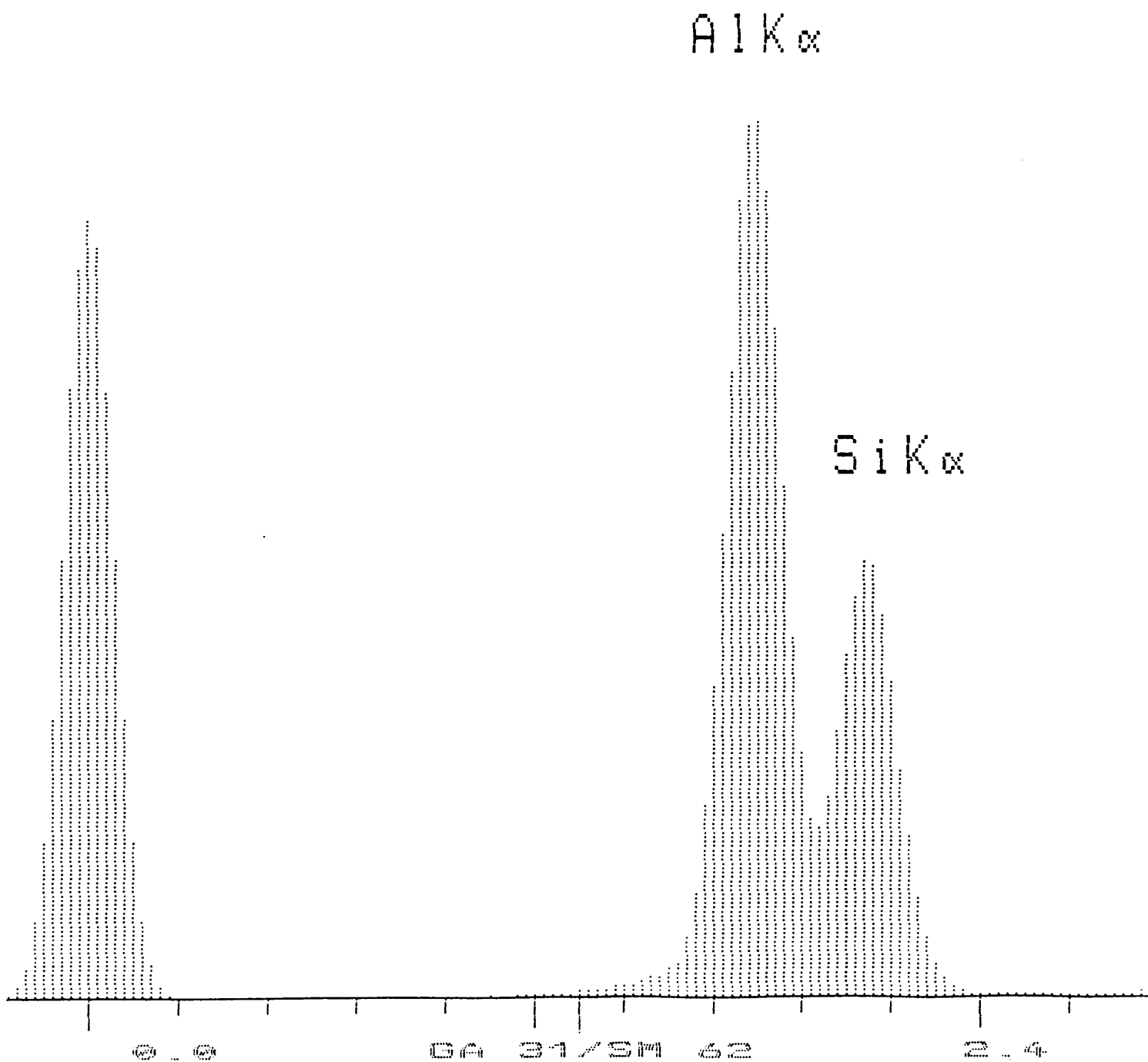


Fig.4.1.14 Energy dispersive X-ray spectrum of a silicon alloyed track.

COMPOSITION: 64 at% Al, 36 at% Si

relationship between composition and microstructures and hardness.

4.1.6 Hardness Testing

Micro-hardness values of samples were measured to find out the effect of silicon alloying on track hardness and the role of primary silicon on total hardness. There is a relationship between hardness and wear resistance. This however does not indicate that hard materials are necessarily more wear resistant. Wear resistance depends on some other factors such as particle size, particle shape, brittleness of particle, kind of friction and load etc. However, some previous work which was carried out by Walker (23) and Megaw (60) reports that abrasive wear decreases as hardness increases in hyper-eutectic aluminium silicon alloys.

From the combination of compositional analysis and hardness results it was found that hardness depends on volume fraction of proeutectic silicon in ^{the} alloyed layer. The hardness of a laser melted fine dendritic area is ~ 120 HV while substrate hardness is ~ 70 HV. Tracks were produced up to 90 wt% Si with hardness of ~ 800 HV. The hardness level obtained is 10-11 times higher than the hardness of the substrate material. See Fig.4.1.15.

Hardness levels of Al+Si alloyed tracks are lower than for pure silicon alloying as a result of the lower volume fraction of proeutectic silicon in alloyed layer (as expected). The plot of hardness against depth is illustrated

Fig.4.1.15 Microhardness against depth in a silicon alloyed track. (Composition: ~90 wt% Si, 9 wt% Al)

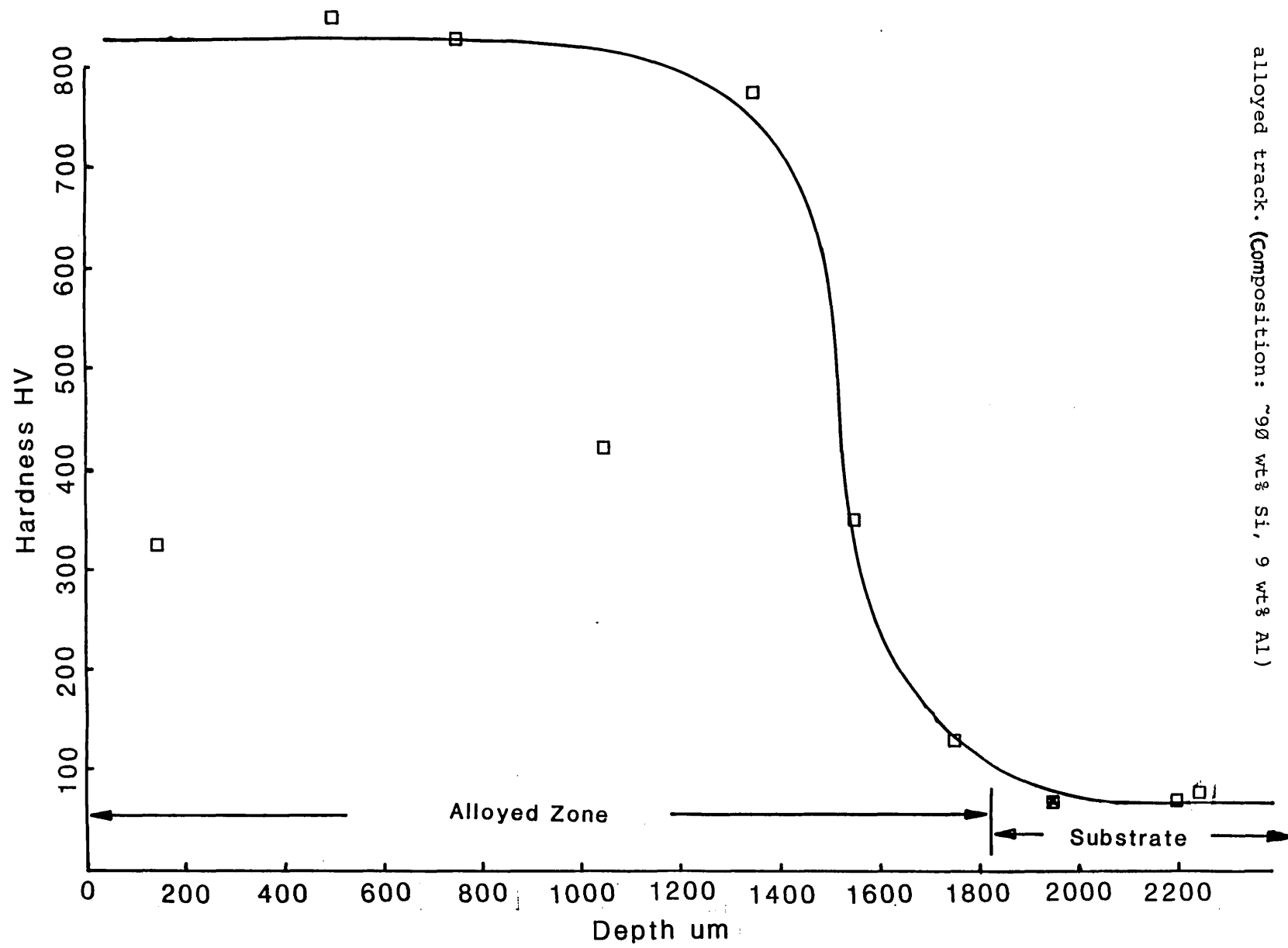


Fig.4.1.16 Microhardness against depth in a silicon alloyed track. (Composition: ~ 80 wt% Si, 20 wt% Al)

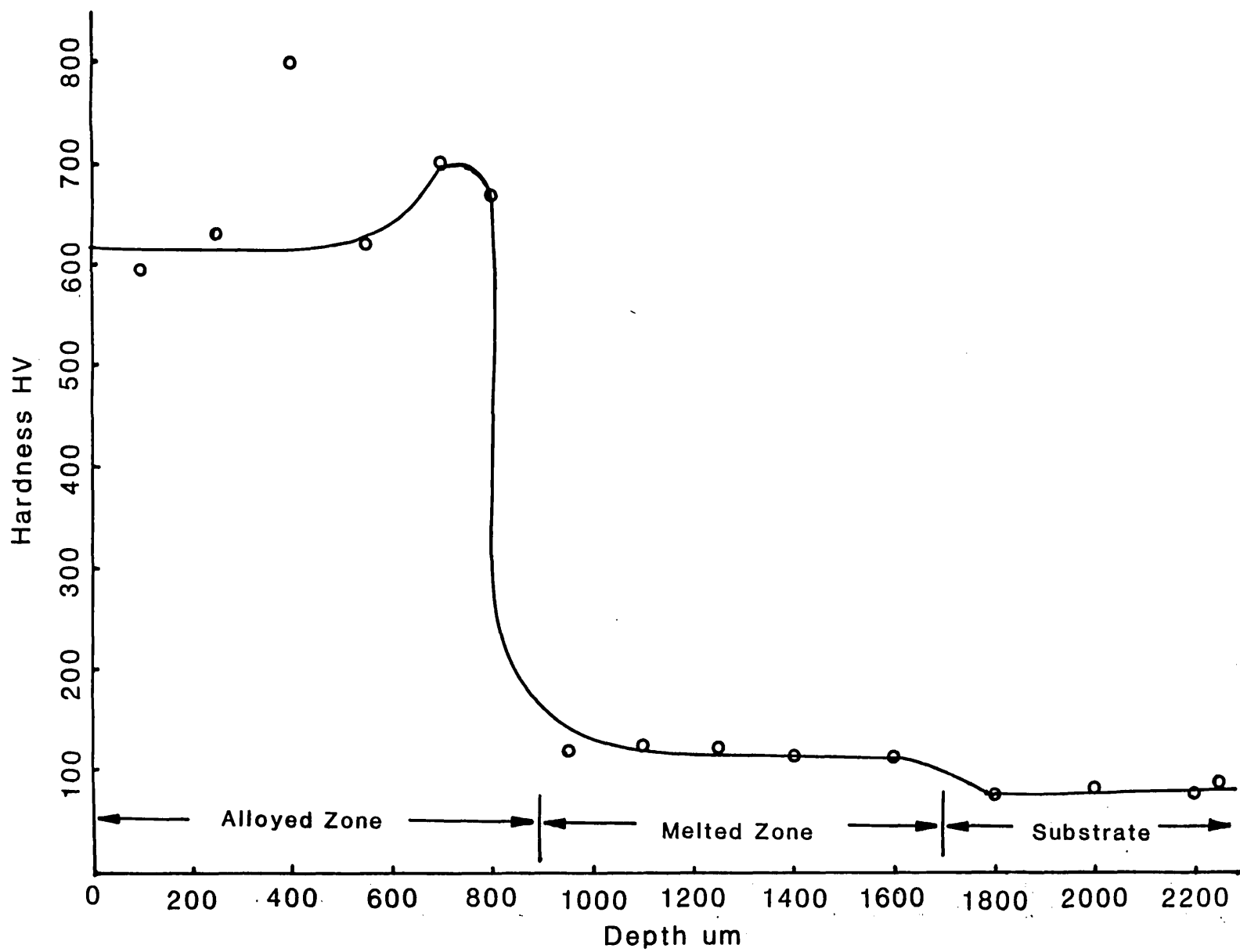


Fig.4.1.17 Microhardness against depth in a silicon alloyed track. (Composition: ~63 wt% Si, 37 wt% Al)

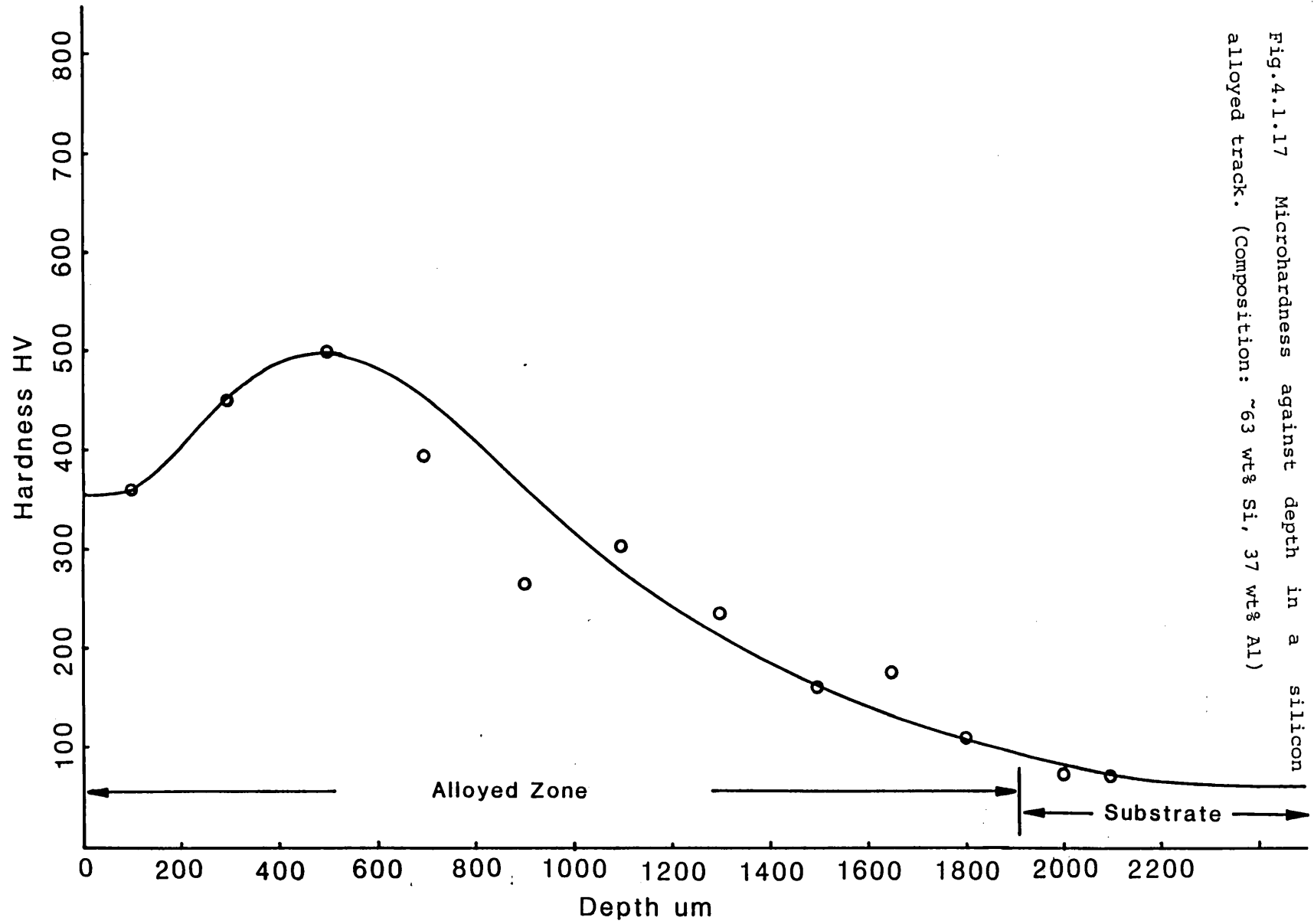
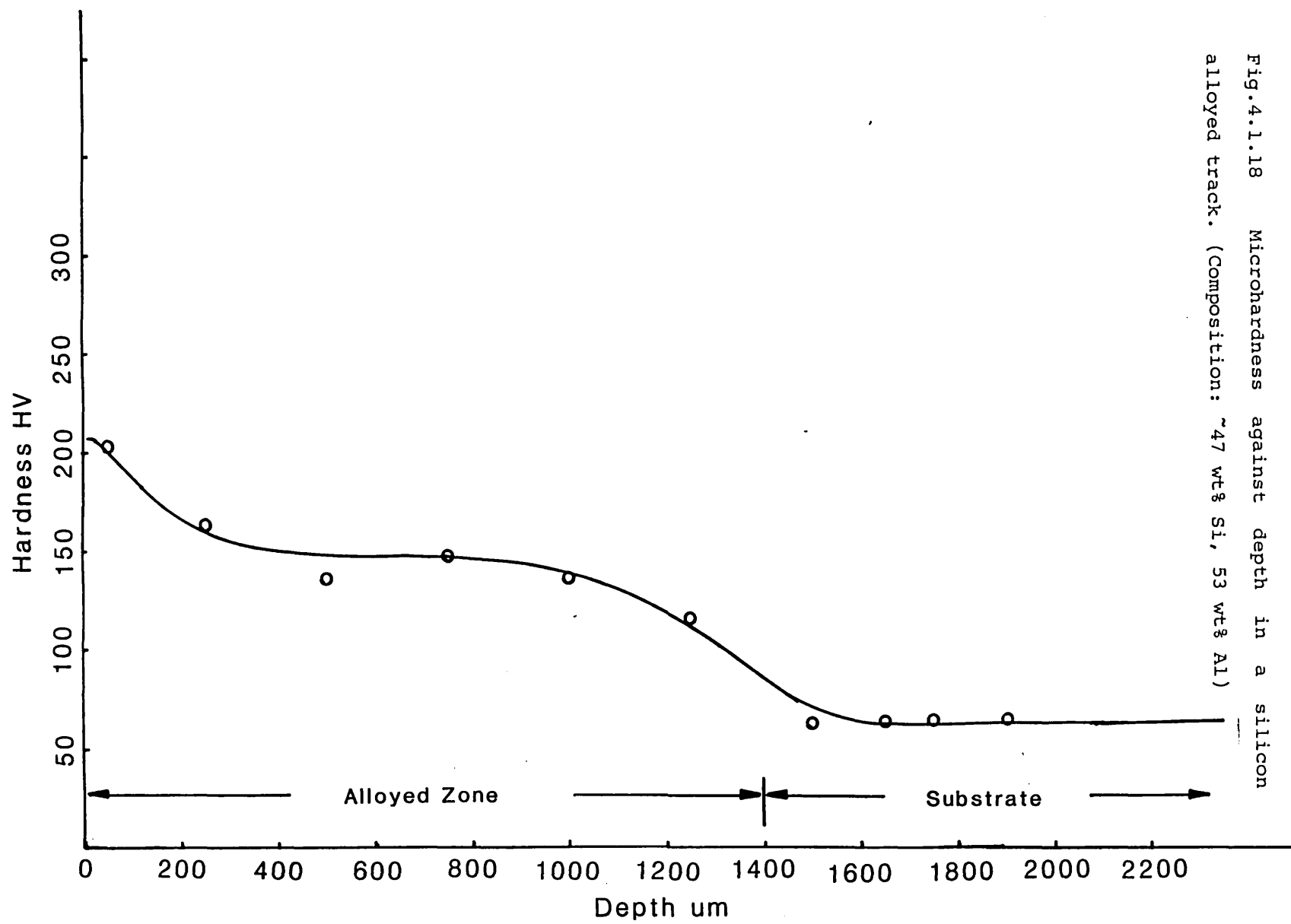


Fig.4.1.18 Microhardness against depth in a silicon alloyed track. (Composition: ~47 wt% Si, 53 wt% Al)



in Fig.4.1.14 to 4.1.18 for different tracks with various compositions.

Some graphs show irregular changes in hardness with distance. This is the result of uneven nonuniform distribution of pro-eutectic silicon particles in the bead.

4.2 SURFACE CLADDING

4.2.1 MILD STEEL + ALUMINIUM CLADDING

4.2.1.1 Operational Parameters

Surface cladding of Al-Si alloy substrates using ~ 2.0 kW power and 3 mm beam diameters at different scan speeds and powder flow rates was carried out. Laser processing parameters such as beam diameter, and power were not changed so as not to reduce the beam coupling efficiency. Different ratios of mild steel/aluminium was employed using two separate powder feeders. As a result, different aluminium-iron alloys were formed ranging from the Fe-rich region to the Al-rich region of the Al-Fe phase diagram.

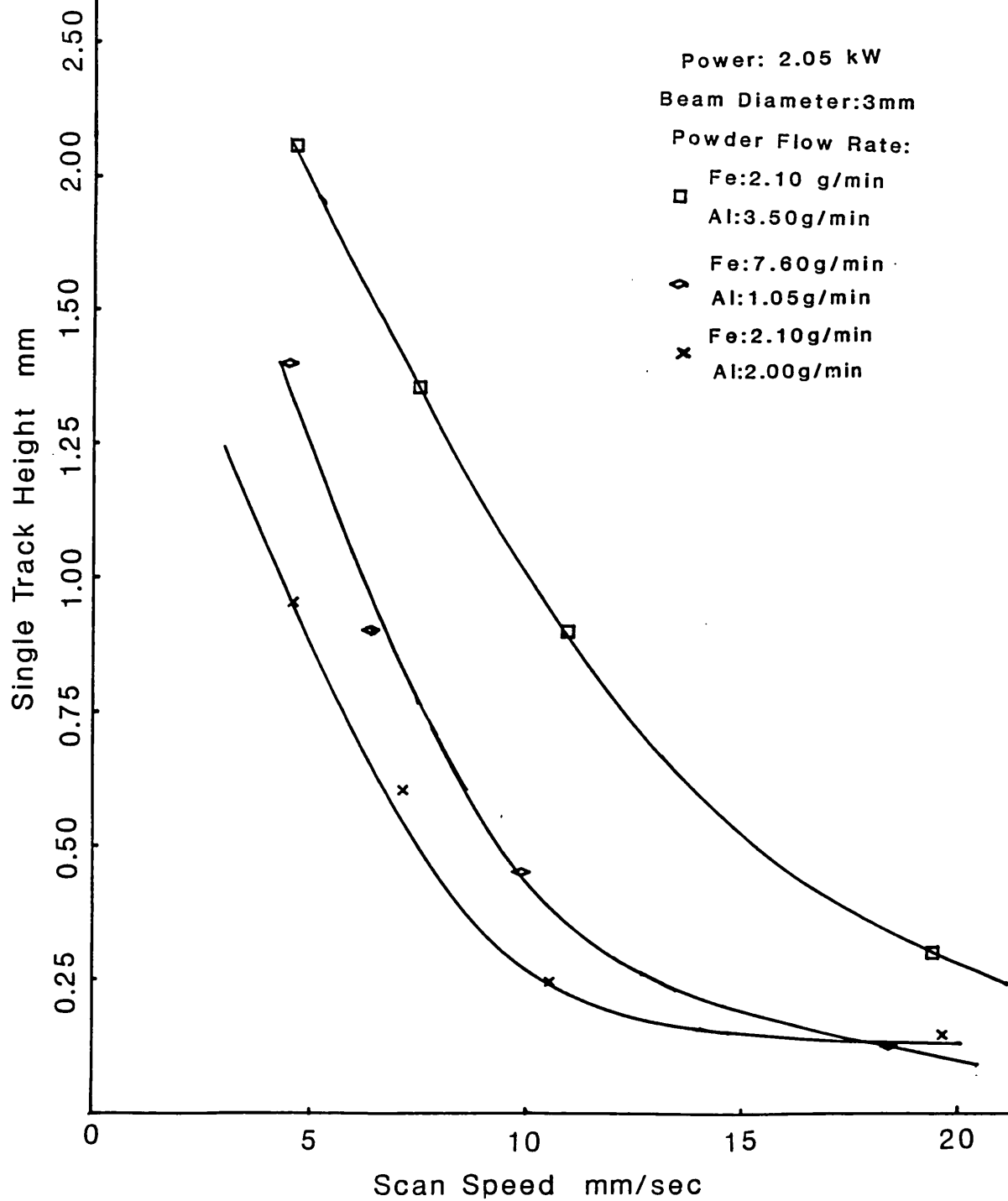
The operating parameters such as scan speed and powder flow rate on track geometry were studied. (For more detail see table 3 in appendix).

The effect of processing parameters on some engineering properties such as cracking, porosity, hardness and on structural features such as phases, microstructure and composition changes were studied.

4.2.1.1.1 Scanning Speed

The effect of scan speed on track geometry for constant and changing powder flow rates is illustrated in Fig.4.2.1 and Fig.4.2.2. Single bead height shows regular increase with decreasing scan speed, but, the bead width does not change consistently over the range. For faster scan speeds

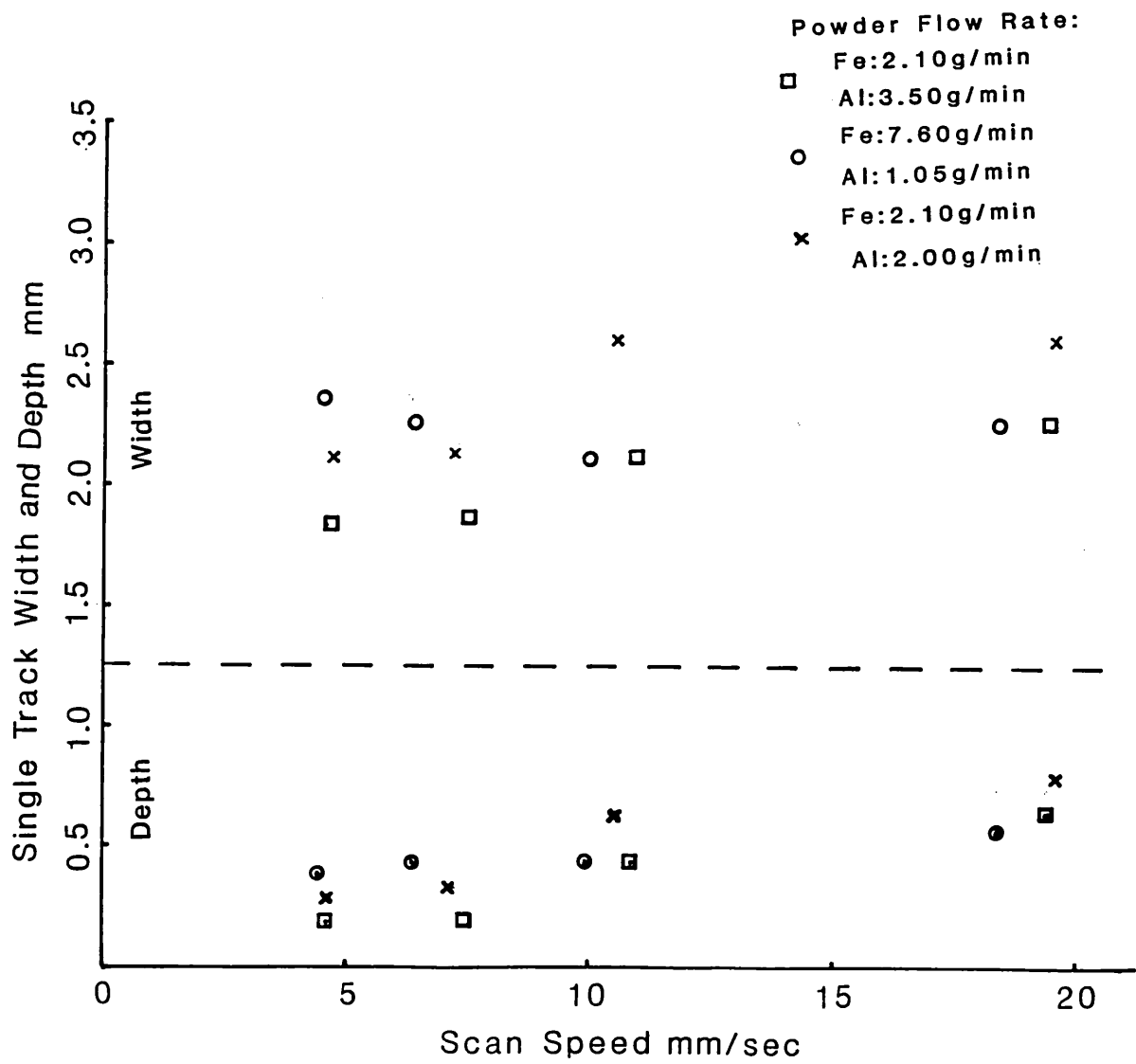
Fig.4.2.1 Single track height vs scan speed in Al+mild steel cladding at various powder flow rates.



Power:2.05 kW

Beam Diameter:3mm

Fig.4.2.2 Single track width and depth vs scan speed in Al+mild steel cladding at various powder flow rates.



(e.g. > 15 mm/sec) or lower powder flow rates width increases. At 9-10 mm/sec scan speed with a powder flow rate giving the width/height ratio of 1.5-2.5, the width is smaller; for a higher powder feed rate giving a width/height ratio of < 1.5 the width starts broadening with increasing bead height. This broadening is less than that of high scan speed and low feeding ratios. At scan speeds faster than 13 mm/sec, alloying occurs instead of cladding. This is shown in Fig.4.2.3d. Macrographs of different tracks at different scan speeds at a constant powder flow rate are illustrated in Fig.4.2.3.

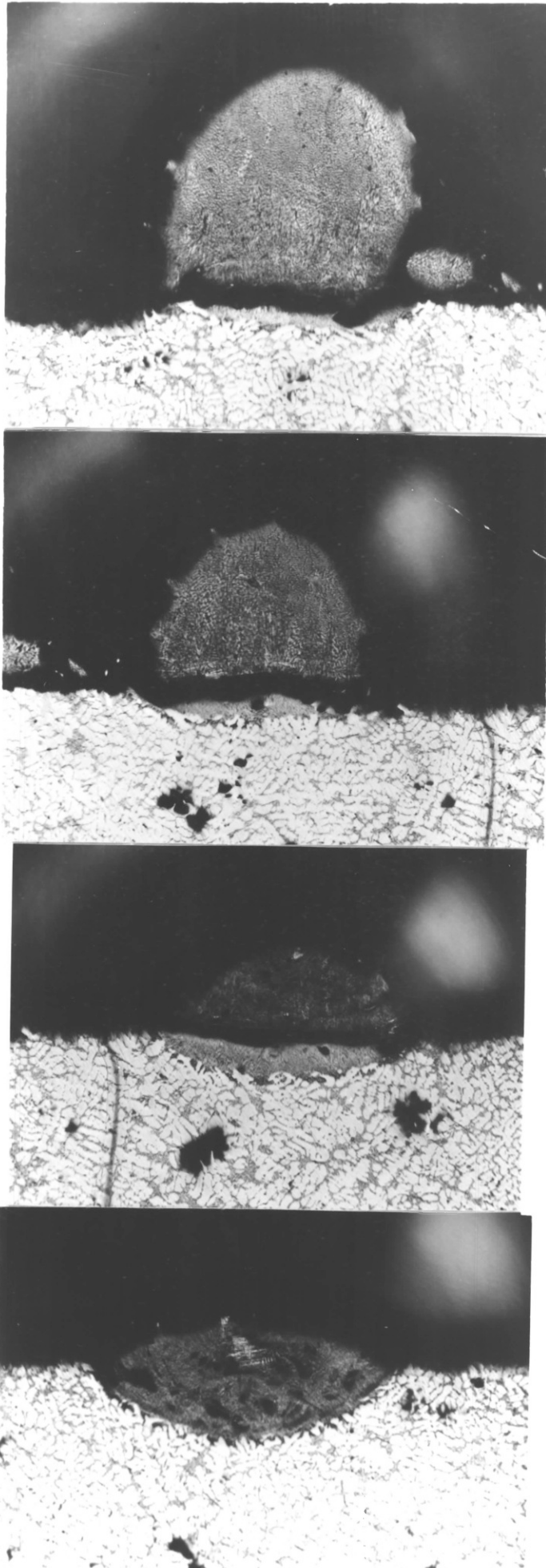
4.2.1.1.2

Powder Flow Rate

Changing powder flow rate has a similar effect on track geometry to scan speed, but these two parameters work directly proportional to each other i.e. for an increase in powder flow rate, to obtain the same track geometry an increase of scan speed is required, otherwise the height of the track increases. However, this is not applicable for the whole range of processing speed, because, there is an upper limit on powder feeding rate and a certain interaction time is necessary. Therefore, at speeds faster than 15 mm/sec no cladding was observed; only alloying occurs over this speed. The effect of powder flow rate on track geometry is shown in Fig.4.2.4. Operational parameters of mild steel/aluminium clad layers are given in Table 3 in the appendix.

Fig.4.2.3a-d A series of photomacrographs of mild steel/aluminium clad layers showing the effect of scan speed on track geometry. Mag. 21.

Scan speed from top to bottom in order 4.67mm/sec, 7.45mm/sec, 10.87mm/sec, 19.38mm/sec at powder flow rate of 5.60 g/min. (2.10g/min Fe and 3.50g/min Al). 4.2.3d shows alloying instead of cladding due to high scan speed. (composition of clad layer for (a-c): ~78 at.% Al, 21at.% Fe)



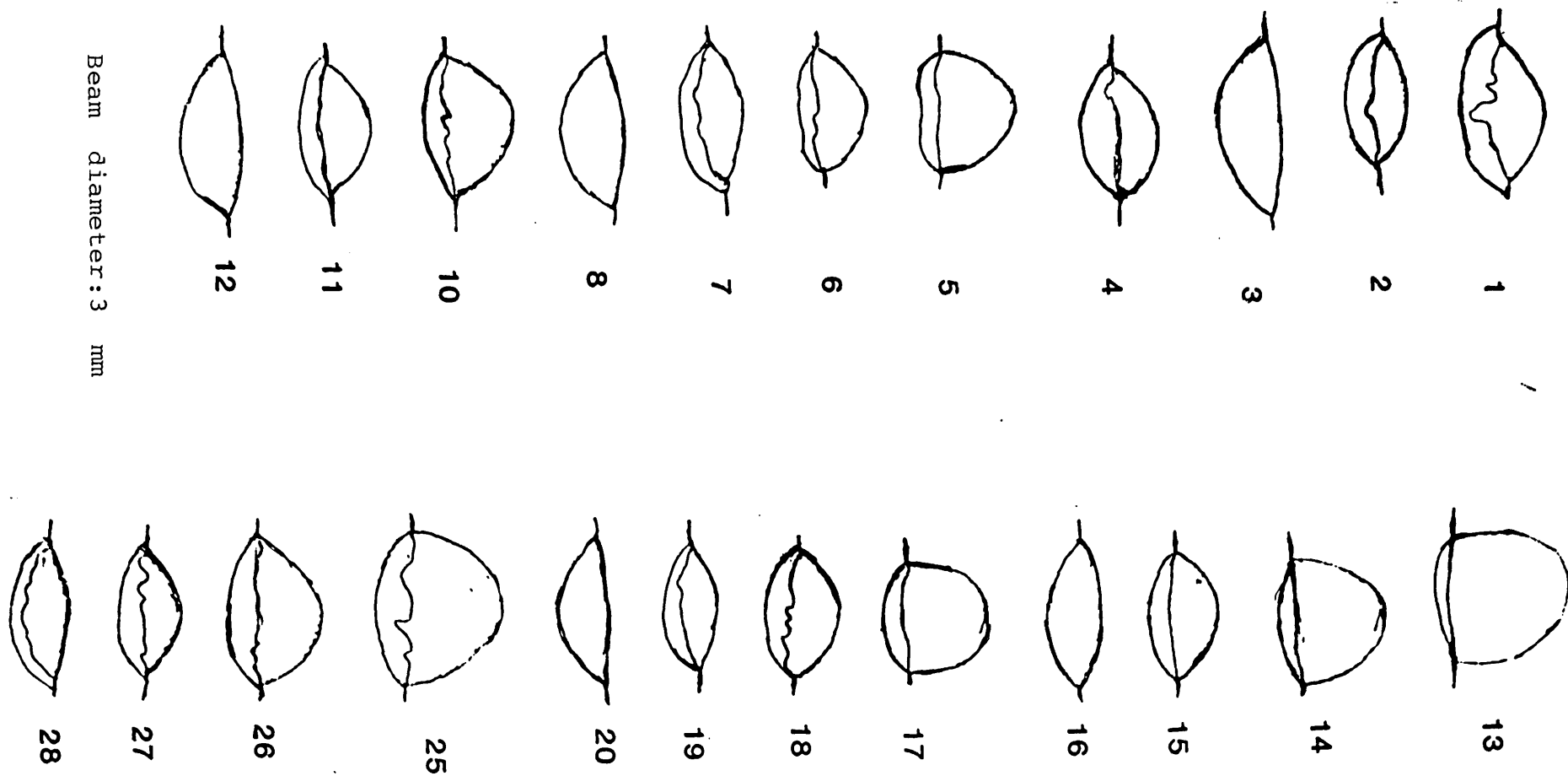


Fig.4.2.4 Bead profiles of Al+mild steel clad tracks with various operating parameters. (Operating parameters and bead sizes are given in table 3 and 3a in appendix.)

4.2.1.2 Track Properties

4.2.1.2.1 Influence of Mild Steel/Aluminium Ratio on Composition

Mild Steel and aluminium powders were fed from two separate powder feeder coupling into one outlet tube. The use of this system made it possible to change the ratio of powders during processing. The arrangement of the powder feeding system is shown in the experimental procedure. Compositional analysis measurements showed that the compositions of clad layers fit that of powder mixture fed within $\pm 2\%$. This can be further improved by more accurate control of the powder feeders. Compositional analysis and energy dispersive X-ray spectra of elements relating to mild steel/aluminium cladding are detailed in the compositional analysis and X-ray sections.

4.2.1.2.2 Porosity

In general, porosity is not a problem for mild steel/aluminium clad layers. Macrographs and micrographs in the related section give the evidence of this situation. (Figs.4.2.3a-d and 4.2.5a)

4.2.1.2.3 Cracking

One of the most important features which affects materials quality, producability, and service performance is the presence of micro or macro cracks, and fissures (cracks between the substrate and the clad layer).

Almost all beads have cracks and fissures between the

substrate and the clad layer. The cracks initiate from a narrow intermetallic zone between the clad layer and the substrate. The composition of this narrow zone is different from the substrate and most of the clad layers examined. Alloying with the substrate occurs in this zone and a brittle phase identified by microprobe analysis forms (mostly Al_3Fe). This is shown in Fig.4.2.5a. In most beads, when samples were etched in different etchants this zone shows different response to etching.

In most cases longitudinal cracks do not appear but, cracks parallel to the cross section are present. The causes of cracking will be discussed in chapter 5.

However, the layers produced at high speeds and low powder flow rates exhibit alloying rather than cladding, and there is no problem of cracking. The iron contents of those layers are less than 10 at% Fe and the volume of intermetallic particles is comparably low, being distributed in the aluminium matrix. Fig.4.2.6a.

4.2.1.3

Microstructural Analysis

Different microstructures were obtained showing dendritic, columnar, cellular and equiaxed features. The aluminium+mild steel clad layers consist of different Al-Fe alloys in the range of ~80 wt.% Fe, 20 wt.% Al to 80 wt.% Al, 20 wt.% Fe changing with the ratio of aluminium/mild steel powder mixture. Different etchants were used to identify microstructures such as Keller's reagent, 1% HF acid, 20% H_2SO_4 . Some tracks which did not show any response to Keller's reagent and 1% HF, were etched in 20% H_2SO_4 solution at 70 °C.

High Al/Fe ratios produced clad beads containing over 77 at.% Al and EPMA analysis and X-ray diffractometry of these beads showed the presence of Al_3Fe (or $\text{Al}_{13}\text{Fe}_4$ according to X-ray data). The microstructure of these beads appears to be essentially cellular; from the Al-Fe phase diagram the main phase is Al_3Fe and the intercellular regions are expected to be aluminium. Cellular and dendritic network of Al_3Fe phase is shown in Figs.4.2.5 and 4.2.7. This structure was found in both clad and alloy layers which were produced at faster scan speeds of cladding. The upper and middle sections of the clad beads have finer cellular grains, while the lower section contains coarser columnar grains oriented in different directions, generally in the direction of heat flow. Fig.4.2.3a. The effect of scanning speed i.e. interaction time on structure is also seen from these samples. Faster scan speeds, that is shorter interaction

Fig.4.2.5a Optical micrograph taken from transverse section of a mild steel/aluminium clad layer illustrates Al_3Fe and Al phases with cellular structure. (composition: 78 at% Al, 21 at% Fe) Mag. 126.

Processing parameters are the same as in Fig.4.2.3c.

Fig.4.2.5b SEM view taken from plan section of above sample illustrates the transgranular cracking and cellular structure of Al_3Fe phase.

Fig.4.2.5c Higher magnification view of 4.2.5b . Black cavity areas are aluminium phase which was removed by etching.

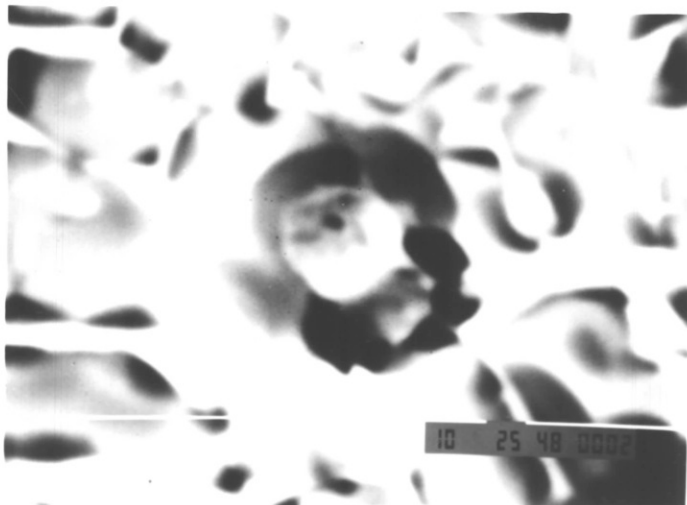
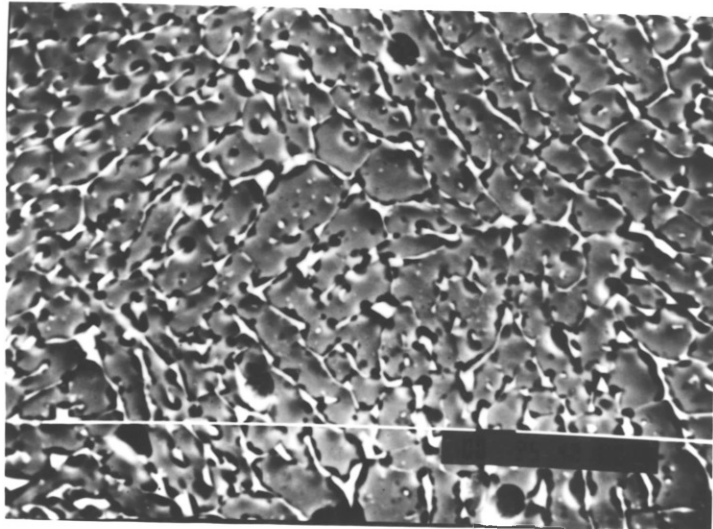
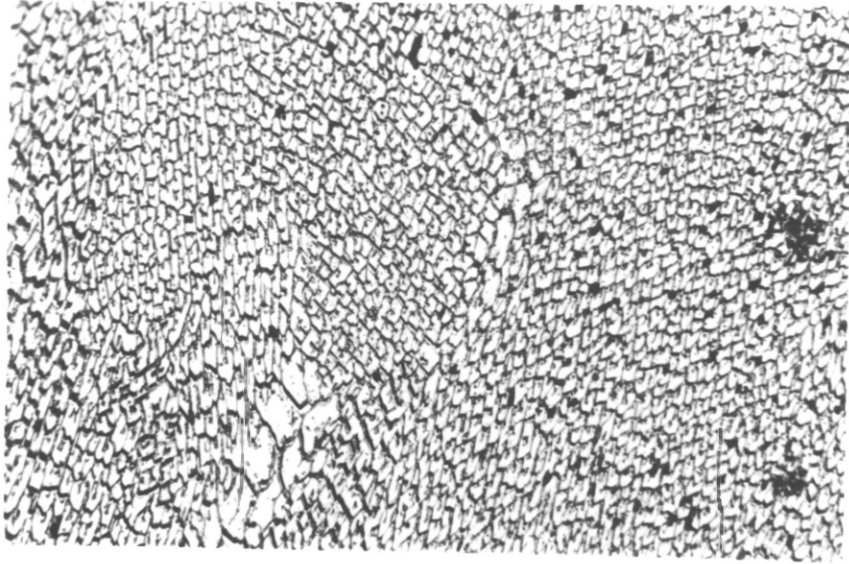
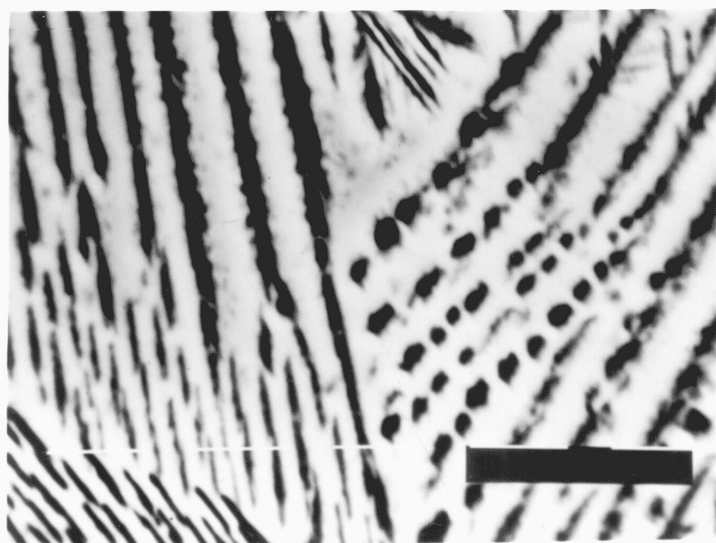
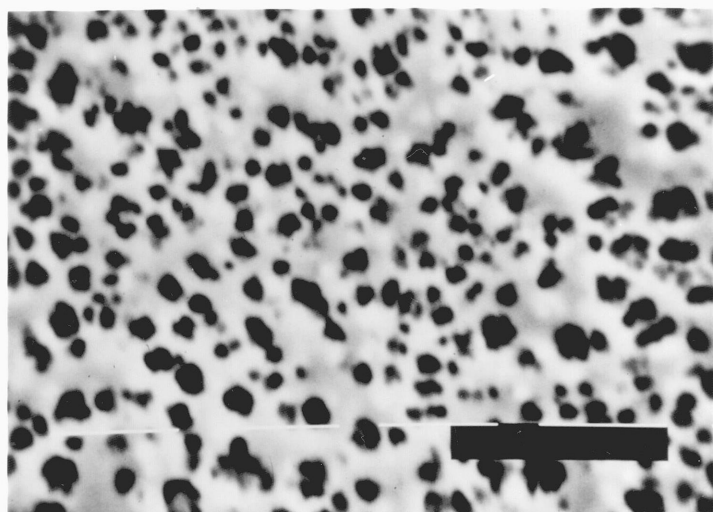
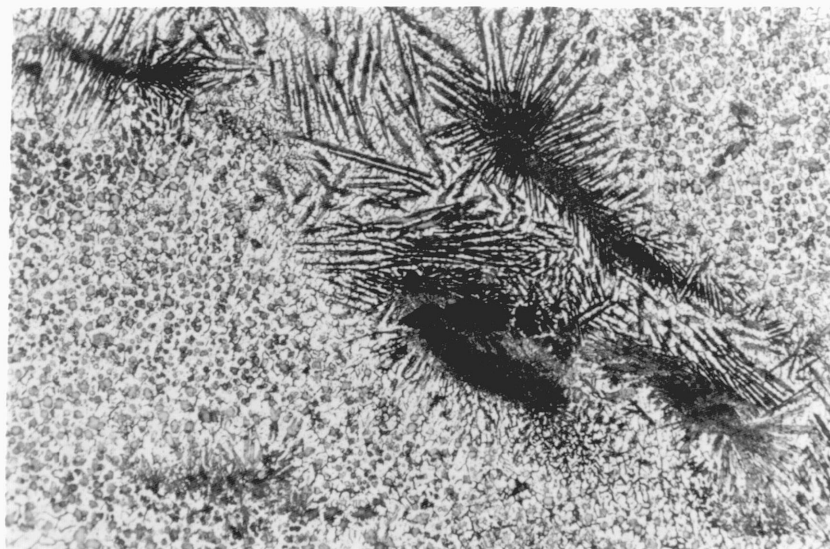


Fig.4.2.6a Optical micrograph of a mild steel/aluminium clad layer. This track was produced at high scan velocity therefore, alloying occurred instead of cladding. Fine cellular distribution of intermetallic Al_3Fe phase in aluminium occurred with some massive intermetallic. (Composition: ~ 90 at% Al, 5 at% Si, 5 at% Fe) Mag. 504.

Fig.4.2.6b SEM micrograph of fine distributed intermetallics in above sample.

Fig.4.2.6c SEM view of cellular section of the above sample consisting of two phases (Al and Al_3Fe).



times produce finer structures. (Fig.4.2.5a)

There are some small dark areas which look like porosity, but in fact they were found to be cavities caused by deep etching of the Al phase. Fig.4.2.7 shows an example of this situation.

In the case of scan speeds faster than 25 mm/sec (alloying occurs) some tracks have ~5at% Fe. The microstructure of these alloys consists of small equiaxed grains of intermetallic Al_3Fe phase in an aluminium matrix. The size of the grains is much finer than in clad layers and the distribution is uniform. On the other hand, some larger intermetallic compound particles of elongated shape are present in these tracks. Fig.4.2.6a-b

All of the clad layers have a narrow intermetallic zone between clad layer and the melted zone. The upper region of the melted zone contains an acicular distribution of the intermetallic Al_3Fe phase.

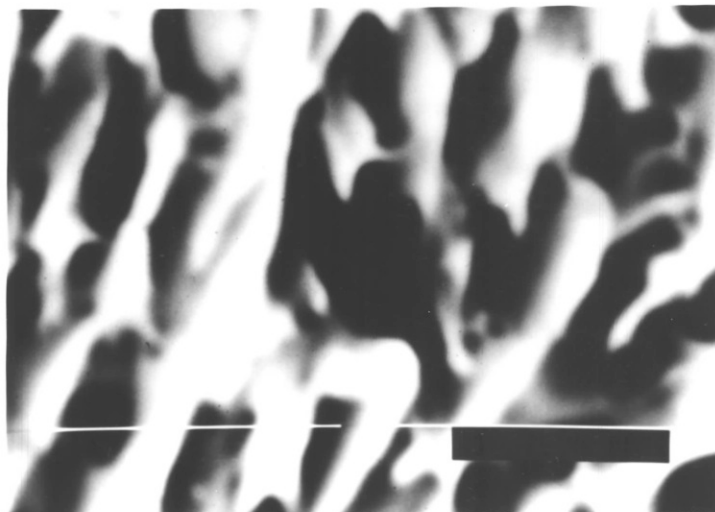
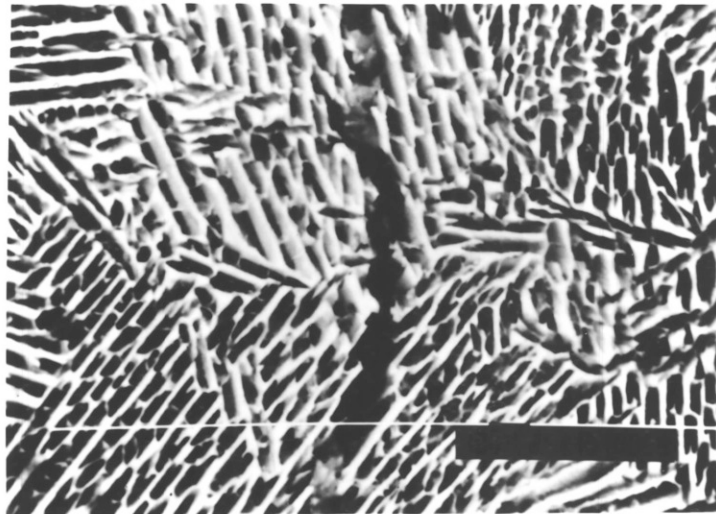
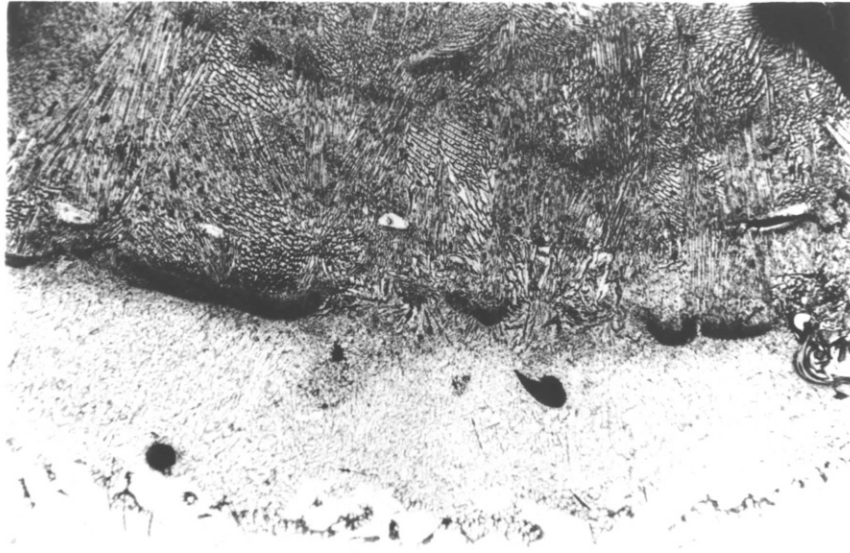
With a lower Al/Fe ratio beads with ~65 at.% Al were formed, and elongated grain structures were observed in which FeAl_2 intermetallic compound was identified by EPMA analysis; these tracks were found to be the most brittle tracks. Cracks are present in all directions and along the track. Fig.4.2.8a.

With higher Fe contents another phase produced was Fe_3Al which was identified by the microprobe analysis and X-ray diffractometry. The microstructure of these tracks consists of equiaxed grains. These grains (analysed as containing 77.5 at.% Fe) show oriented rod-like fibrous cellular and dendritic network as solidification structure. Fig.4.2.9.

Fig.4.2.7a Optical micrograph taken from transverse section of a mild steel/aluminium clad layer with cellular structure. (Composition: 77 at% Al, 22.5 at% Fe) Mag.252.

Fig.4.2.7b SEM micrograph taken from plan section of above sample shows intermetallic Al_3Fe matrix and Al phase between matrix.

Fig.4.2.7c High magnification view of 4.2.7b illustrates cellular aluminium phase in matrix. when these small particles are etched for a longer time this area looks like porosity.



The melted zone of the LM25 substrate consists of fine dendritic proeutectic aluminium phase with some eutectic between them; the dendrite spacing decreases with increasing scan speed. (A similar effect is shown in Fig.4.3.2a-c in the particle injection section). The hardness of this melted zone is typically ~120 HV (e.g. Fig.4.2.) as compared with the coarse substrate structure of ~70 HV.

4.2.1.4 Compositional Analysis

Compositional microanalysis of samples was done using a JEOL JSM35 scanning electron microscope fitted with an electron probe micro analysis instrument. Energy dispersive X-ray spectra taken from these alloys are given below together with quantitative results. Figs.4.2.12 to 4.2.16.

4.2.1.5 X-Ray Diffractometry

Results taken from PW 1050 Phillips diffractometer were evaluated and compared to standard values of phases which are given in the tables. d spacings, hkl parameters and systems of phases are given below. The experimental results fit with the standards but some peaks shown in the standard were not obtained because the samples were quite small in size. (The reference is Hanawalt Index). Card No: 29-42(Al_3Fe), 6-0695(Fe_3Al), 33-20(AlFe), 4-0787(Al).

Phase	System	A°
Al_3Fe	Monoclinic	$a_0=15.489$, $b_0=8.083$, $c_0=12.476$ $\beta=107.70^\circ$
Fe_3Al	Cubic	$a_0=5.78$
AlFe	Cubic	$a_0=2.3954$
Al	Cubic	$a_0=4.0494$

X-ray diffractometry results of an aluminium+mild steel clad layer. Composition: 76.1 at% Al, 20.5 at% Fe, 3.3 at% Si.

d Å (Exp.)	d Å (Ref.)	hkl	Phase
4.0312	4.040	203,020	Al ₃ Fe
3.9343	3.962	003	"
3.6688	3.674	402	"
3.5230	3.545	220,221	"
3.3265	3.342	022	"
3.2481	3.234	403	"
3.1319	3.118	203	"
2.5232	2.525	423	"
2.4552	2.461	224,603	"
2.3325	2.338	111	Al
2.1478	2.161	115,622	Al ₃ Fe
2,0937	2.095	623,205	"
2.0736	2.078	206,713	"
2.0425	2.041	423,710	"
2.0273	2.024	200	Al
1.9345	1.9361	802	Al ₃ Fe
1.4317	1.431	220	Al
1.2206	1.221	331	"
1.1693	1.1690	222	"
0.9294	0.9289	331	"
0.9060	0.9055	420	"

X-ray diffractometry results of an aluminium+mild steel clad layer. Composition: 23 at% Al, 76.3 at% Fe.

d Å (Exp.)	d Å (Ref.)	hkl	Phase
2.8884	2.89	200	Fe ₃ Al
2.0491	2.04	220	"
2.0469			
1.4447	1.45	400	Fe ₃ Al
1.1822	1.18	422	"
1.1793			
0.9206			
0.9189			
0.9165	0.9157	310	AlFe
0.9149			

X-ray diffractometry results of an aluminium+mild steel clad layer. Composition: ~40 at% Al, 60 at% Fe.

d Å (Exp.)	d Å (Ref.)	hkl	Phase
2.8902	2.899	100	AlFe
2.0460	2.048	110	"
1.6729	1.6722	111	"
1.4445	1.4472	200	"
1.2944	1.2949	210	"
1.1823	1.1820	211	"
1.0237	1.0238	120	"
0.9153	0.9157	310	"
0.9360	0.9358	222	"
0.8342			

Fig.4.2.8a Optical micrograph of brittle Al-Fe inter-metallic phase showing grain structure and cracks in clad layer. (Composition: 65 at% Al, 34.5 at% Fe) FeAl_2 Mag.252.

Fig.4.2.8b SEM view of an Al/mild steel clad layer illustrates cellular and dendritic structure of Al_3Fe particles. (Composition: 80 at% Al, 19 at% Fe)

Fig.4.2.8c Higher magnification of above sample illustrates dendritic and cellular structure.

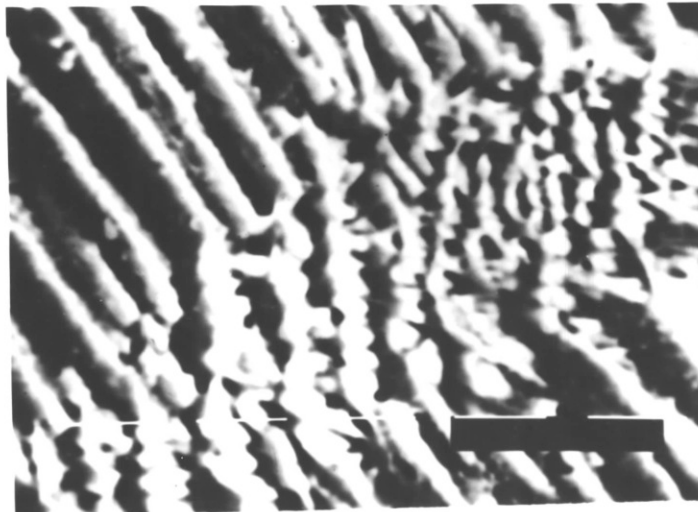
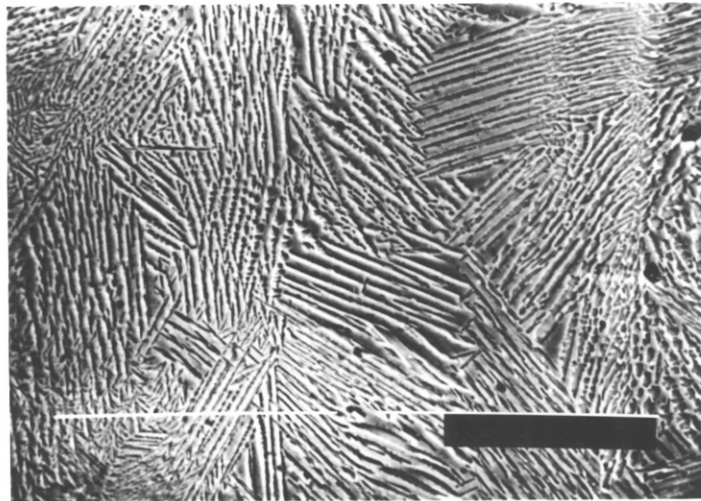
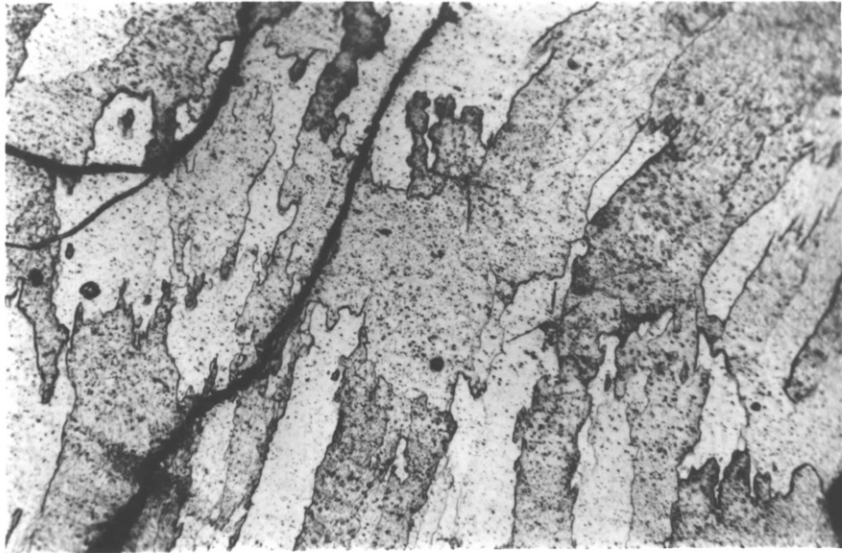
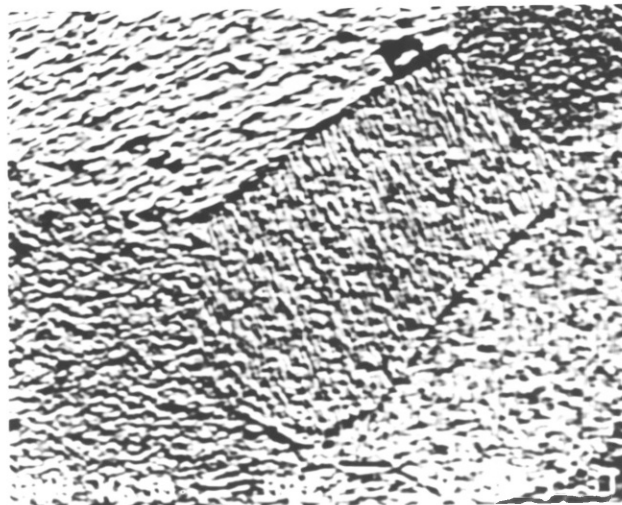
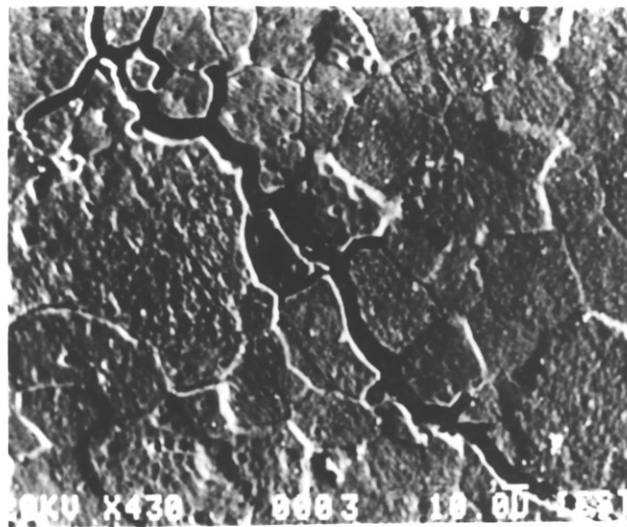
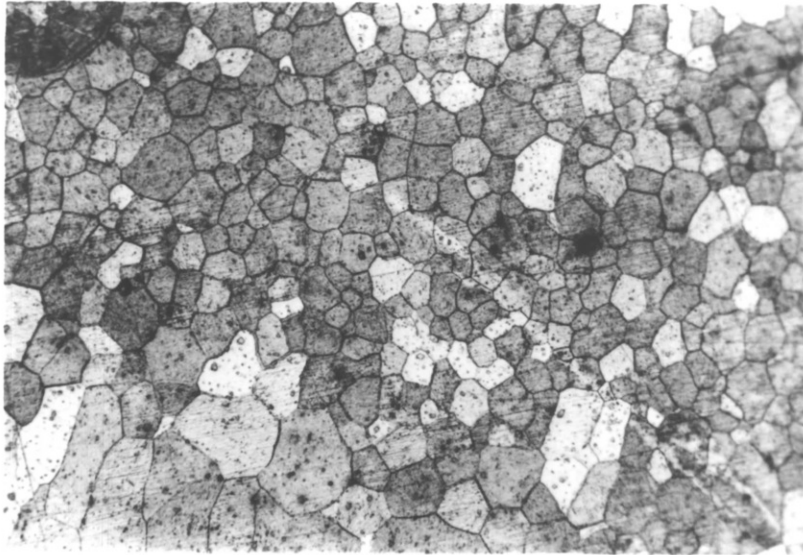


Fig.4.2.9a Optical micrograph of a Al/mild steel clad layer shows equiaxed grains of Fe_3Al . Picture was taken from transverse section. (Composition: 77.5 at% Fe, 22 at% Al) Mag.495.

Fig.4.2.9b SEM micrograph of same sample taken from plan section. Intergranular cracking is particularly noticeable.

Fig.4.2.9c SEM view of Fe_3Al grains shows grain boundaries and orientation of fibrous cellular and dendritic network in different grains.



4.2.1.6 Hardness Testing

Microhardness testing of Mildsteel+aluminium clad layers was carried out. The hardness of the tracks depends on the type of phase present and its distribution in the track. Hardness of an $\text{Al}_3\text{Fe}+\text{Al}$ track containing (78 at.% Al, 21.5 at.% Fe) is ~ 300 HV, ^{whereas} for Al_2Fe (65 at.% Al, 34.5 at.% Fe) it is ~ 650 HV. Hardness of the Fe_3Al (77.5 at.% Fe, 22 at.% Al) is 350-400 HV. An alloyed track with the composition of 90 at.% Al, 5 at.% Si, 5 at.% Fe has a hardness of ~ 125 HV.

4.2.2 Mild Steel Cladding

Cladding of the Al-Si alloy (LM25) substrates with mild steel powder was carried out producing a few tracks. The effect of operating parameters on the track geometry was similar that of mild steel+Al cladding.

Again cracking occurs in the substrate/clad layer interface and the tracks are almost porosity free.

The structure of the clad layers consists of ferrite+pearlite grains as is seen in Fig.4.2.10 a and b. White areas are the ferrite grains; among the ferrite grains small pearlite regions are present. Tempered martensite structure was observed in one track. Optical micrograph of this track is shown in Fig.4.2.10c.

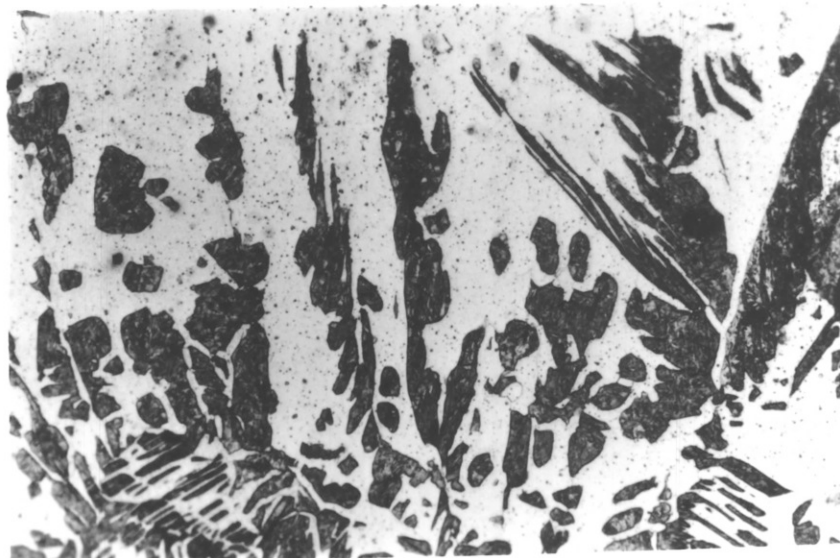
There is always a narrow inter-metallic zone between the clad layer and the substrate. As a result of the substrate melting, alloying occurs in this region and produces an inter-metallic phase.

Hardness measurements of these tracks were undertaken.

Fig.4.2.10a General optical microscope view of a mild steel clad layer. Lower section shows intermetallic zone (marked with arrows) between the clad layer and the melted zone. Mag.128.

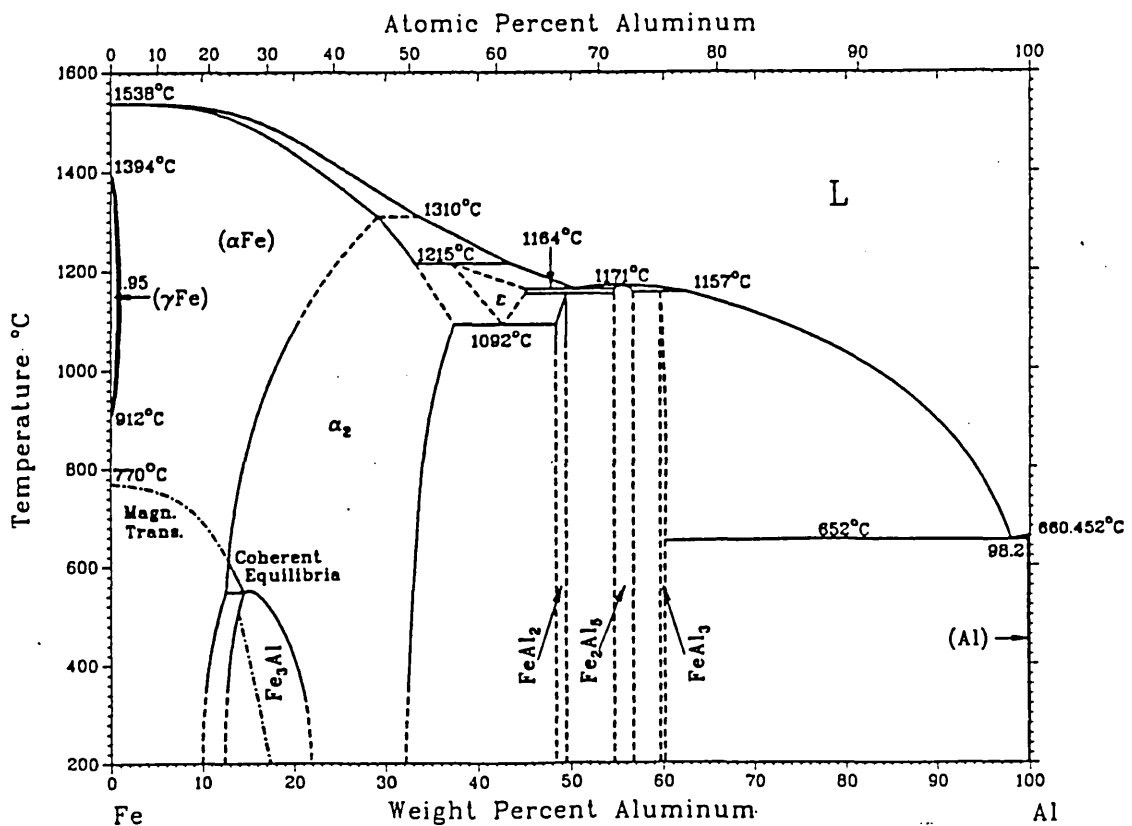
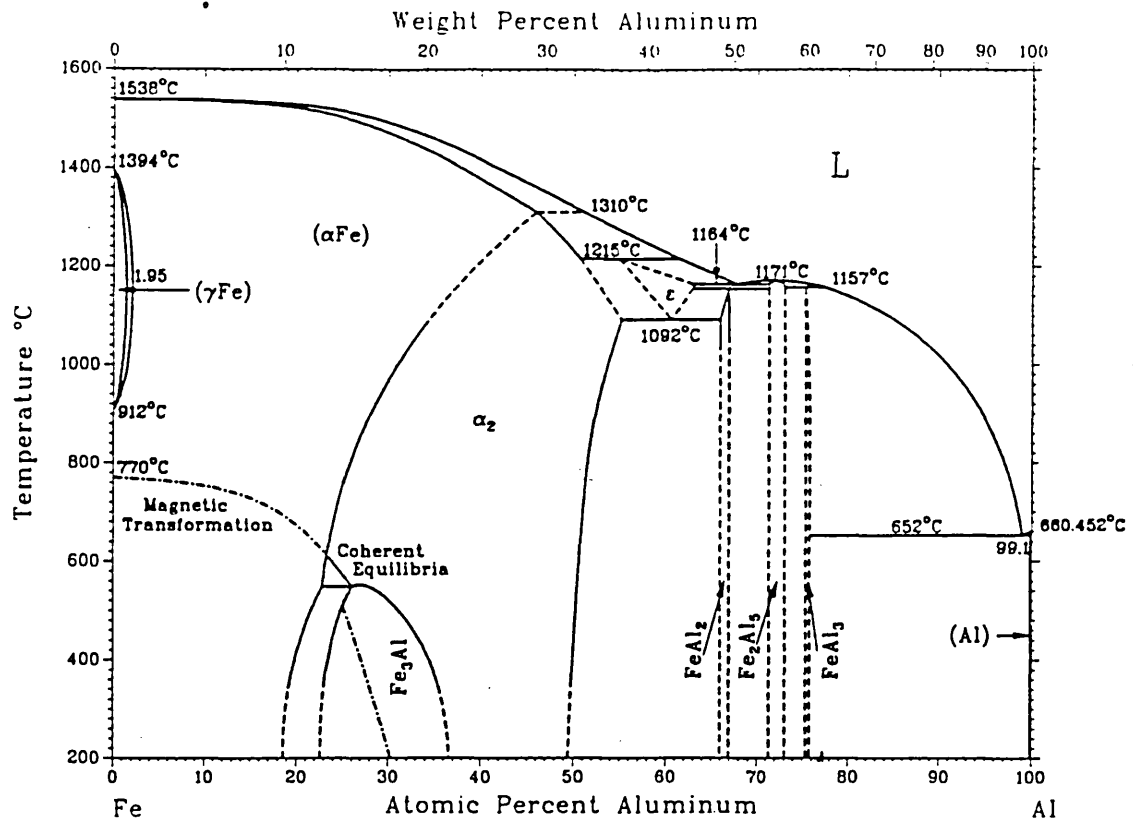
Fig.4.2.10b Higher magnification of above sample with ferrite and pearlite structure. Mag.510.

Fig.4.2.10c Tempered martensitic structure of a mild steel clad layer. Mag.255.



Hardness values are ~250 HV for ferrite+pearlite tracks and ~450 HV for tempered martensite structure as expected.

Fe-Al Phase Diagram



From [Kubaschewski].

Fig.4.2.11 Iron Aluminium Binary Phase Diagram.(50)

FA I 5-6 SECOND 2) / % / 4K FS: B
20 EV/CHAN, LIVE TIME = 100 SECS
SPECTRUM LENGTH = 1024 CHAN

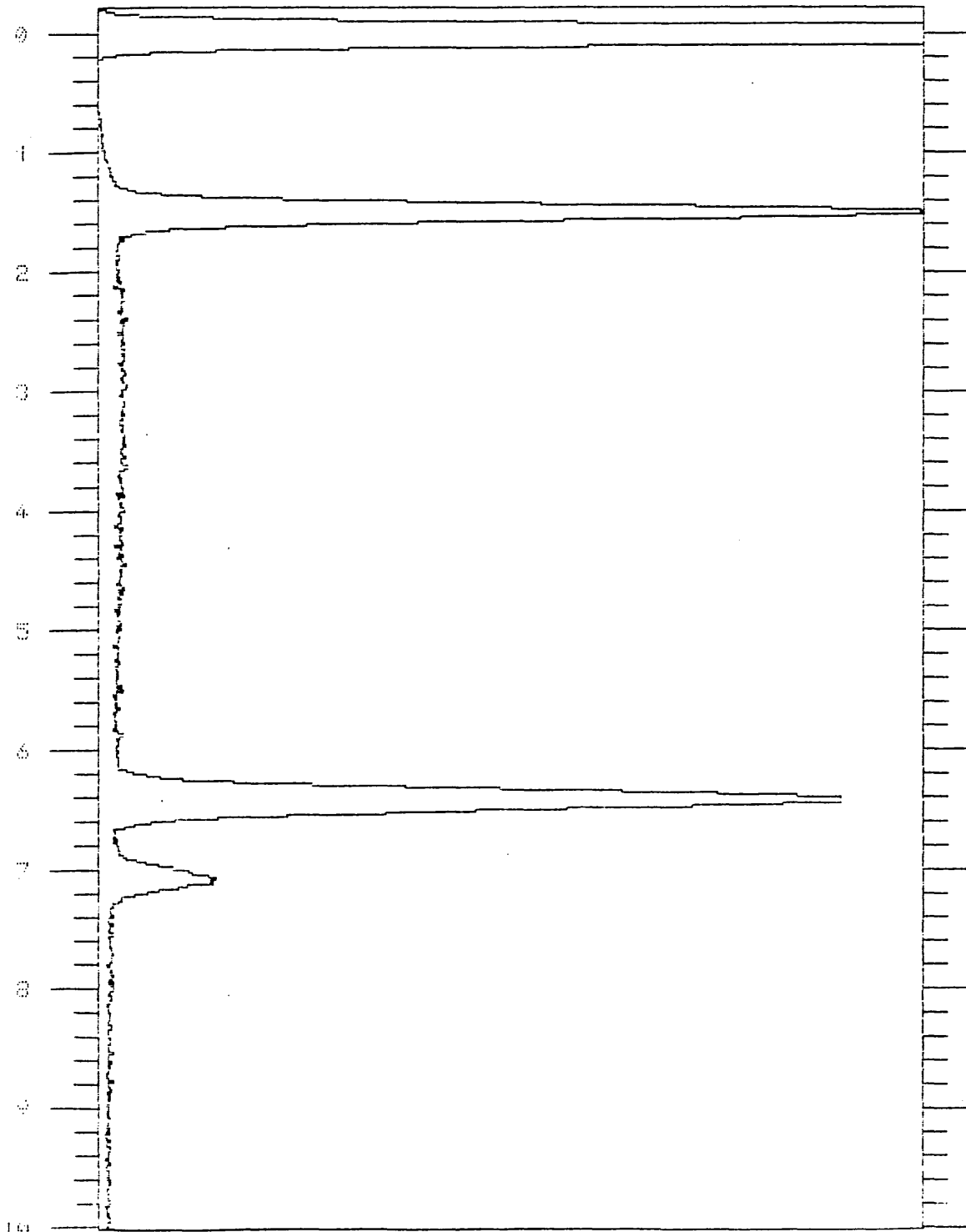


Fig.4.2.12 Energy dispersive X-ray spectrum of an Al+mild steel clad layer.

COMPOSITION: 63 at% Al, 36.5 at% Fe

FAI 5-8 THIRD (7/c)T/X< 8K FS: B
20 EV/CHAN, LIVE TIME = 100 SECS
SPECTRUM LENGTH = 1024 CHAN

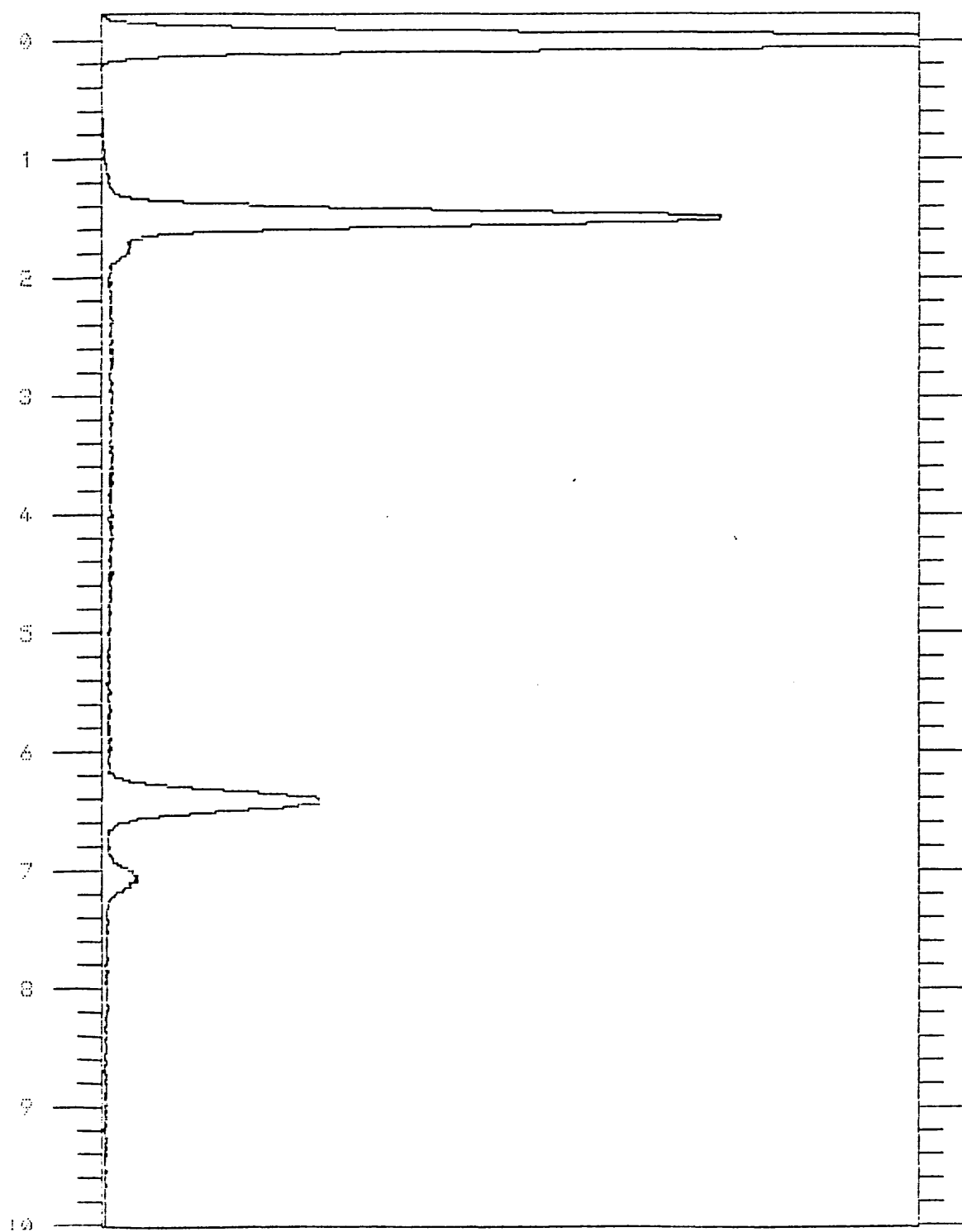


Fig.4.2.13 Energy dispersive X-ray spectrum of an Al+mild steel clad layer.

COMPOSITION: 76.1 at% Al, 20.5 at% Fe, 3.3 at% Si

FILE 1 0000000000 4K FSI 6
20 EV/CHAN. LIVE TIME = 100 SECS
SPECTRUM LENGTH = 1024 CHAN

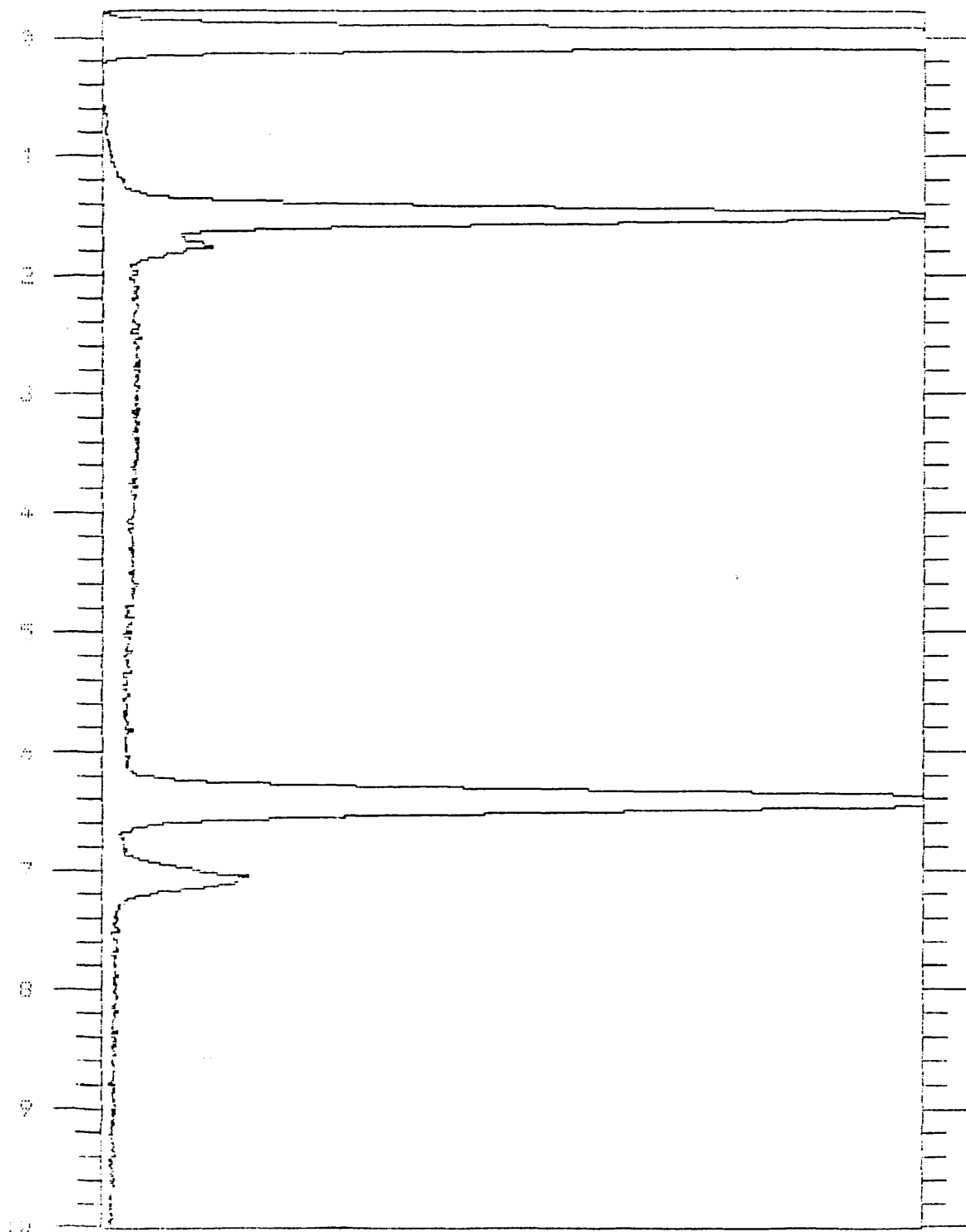


Fig.4.2.14 Energy dispersive X-ray spectrum of an Al+mild steel clad layer.

COMPOSITION: 55.7 at% Al, 39.5 at% Fe, 4.6 at% Si

FA-III 2 000000/17% 8K F31 B
20 EU/CHAN, LIVE TIME = 100 SECS
SPECTRUM LENGTH = 1024 CHAN

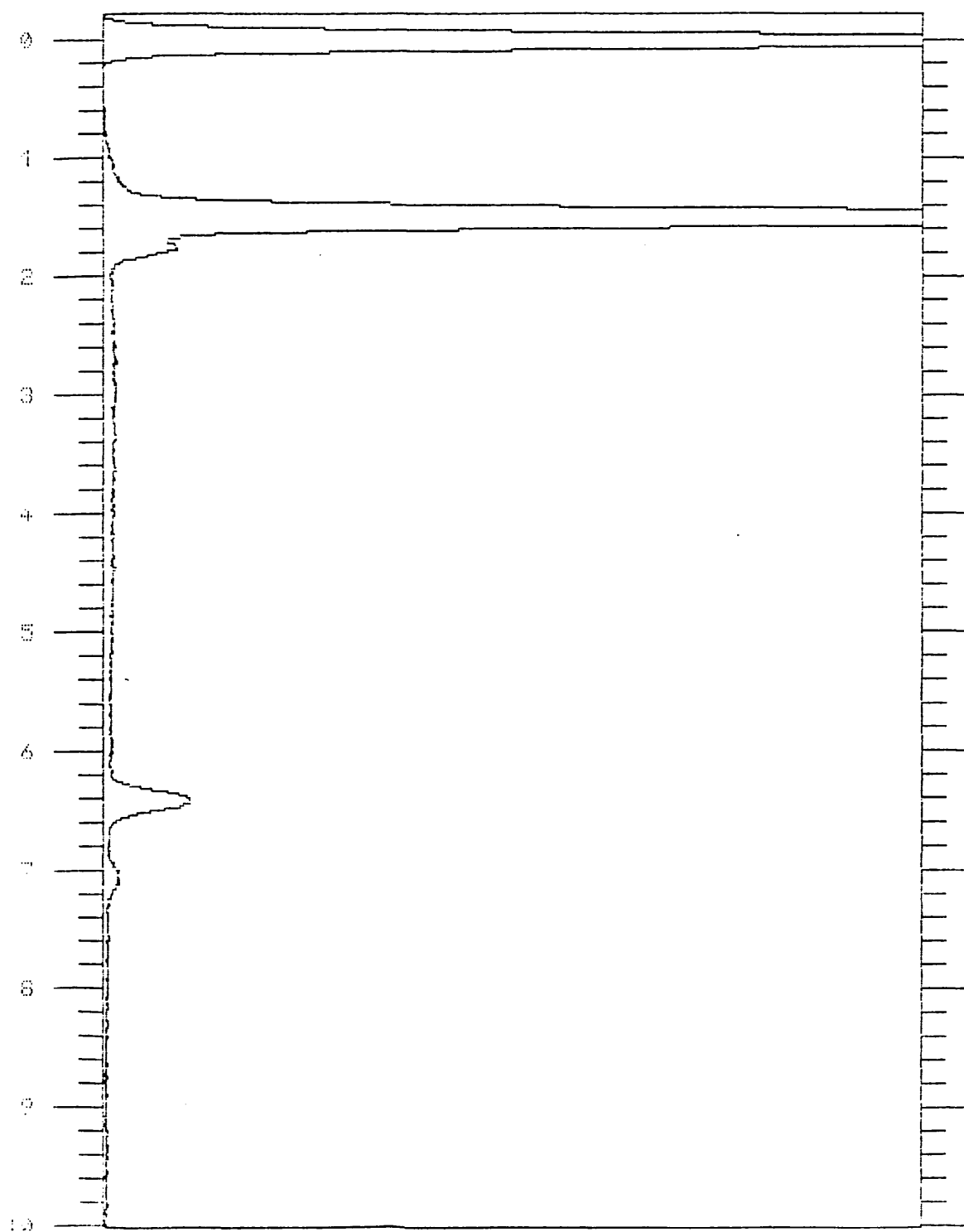


Fig.4.2.15 Energy dispersive X-ray spectrum of an Al+mild steel clad layer.

COMPOSITION: 86.2 at% Al, 7.5 at% Si, 5.5 at% Fe

IRON2A1)\)/)/e)T/X< BK FS: B
20 EV/CHAN, LIVE TIME = 100 SECS
SPECTRUM LENGTH = 1024 CHAN

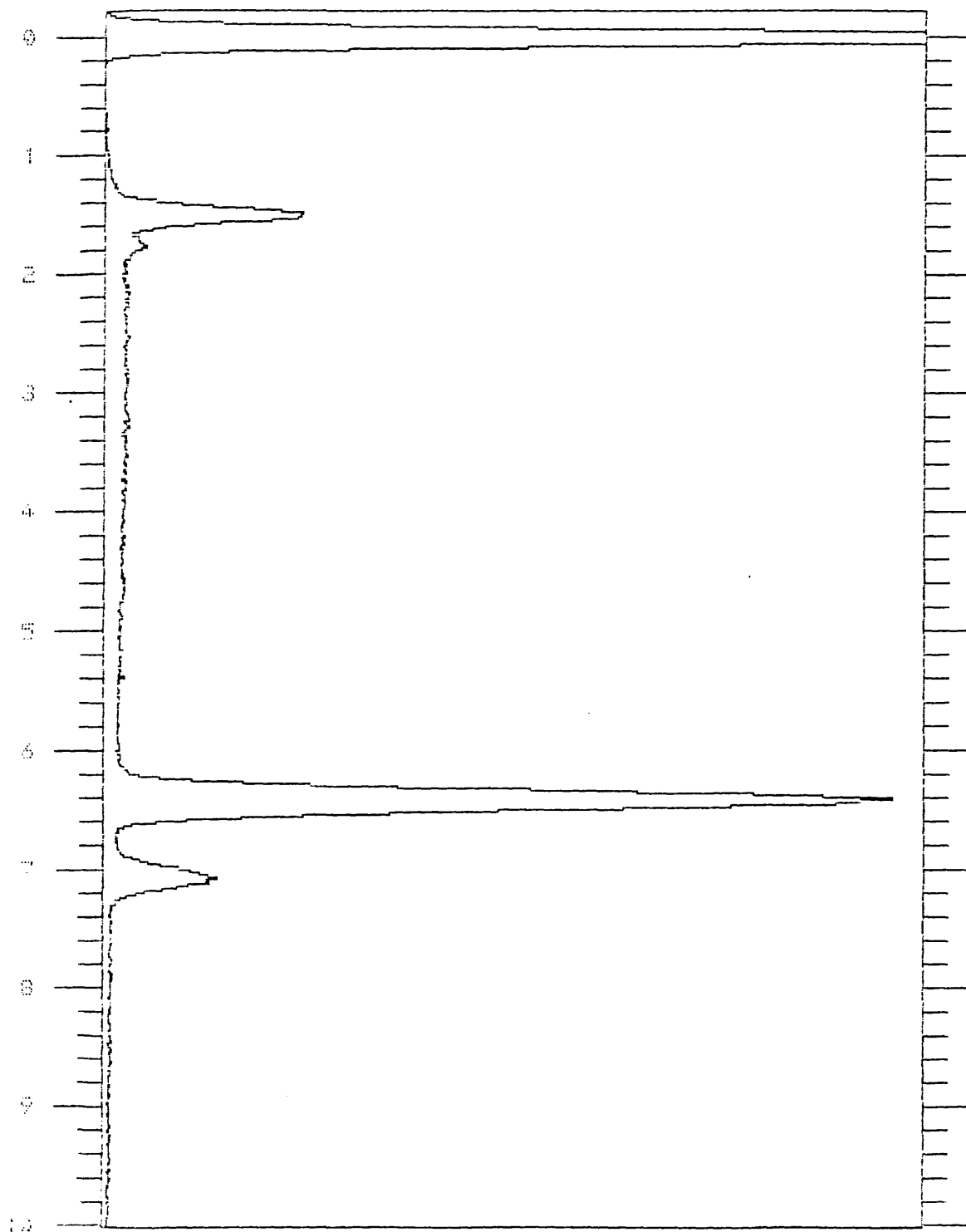


Fig.4.2.16 Energy dispersive X-ray spectrum of an Al+mild steel clad layer.

COMPOSITION: 31 at% Al, 65.5 at% Fe, 2.5 at% Si

Fig.4.2.17 Microhardness vs depth curves of an Al+mild steel clad layers.

COMPOSITION: 77.4 at% Al, 22 at% Fe

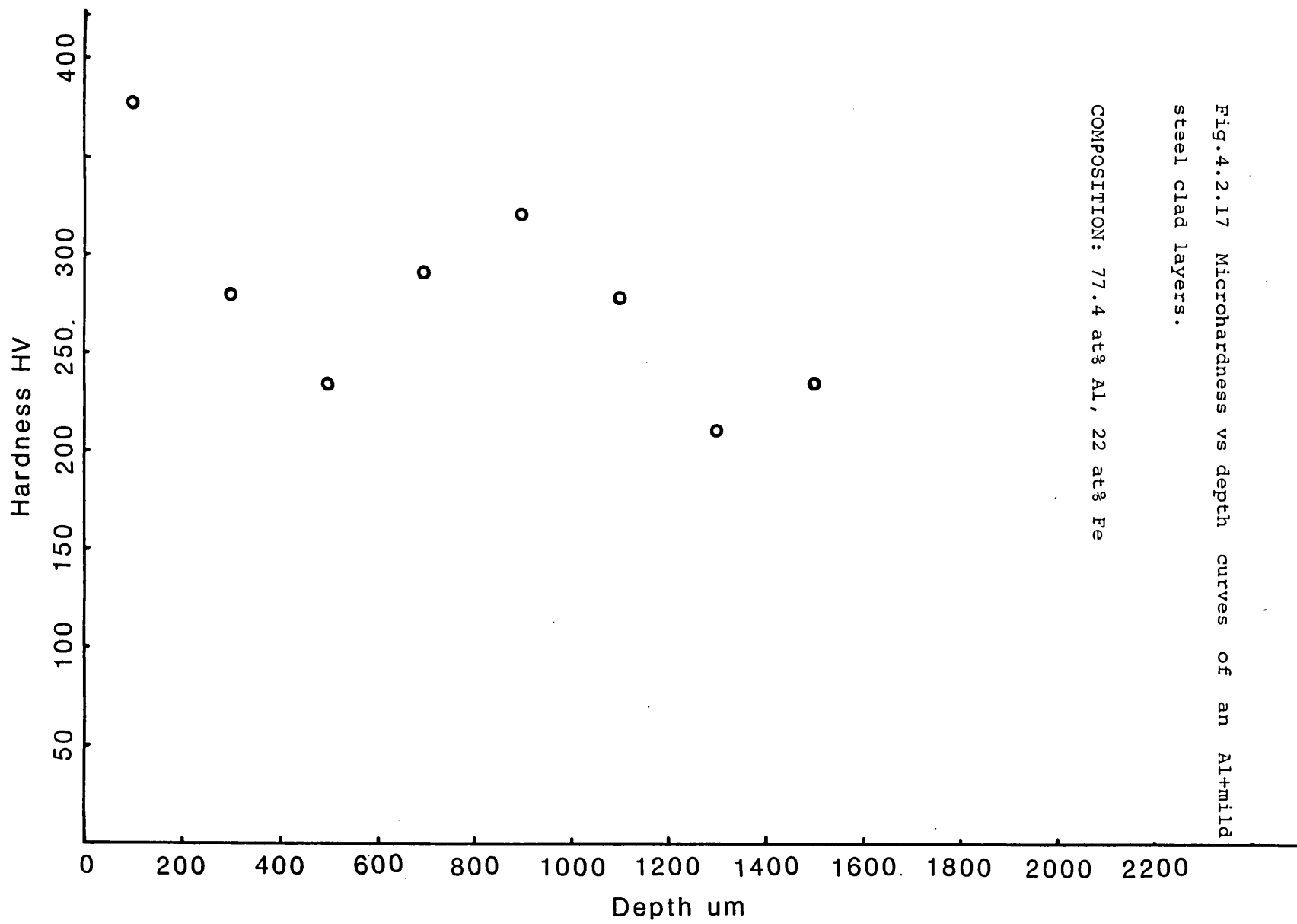
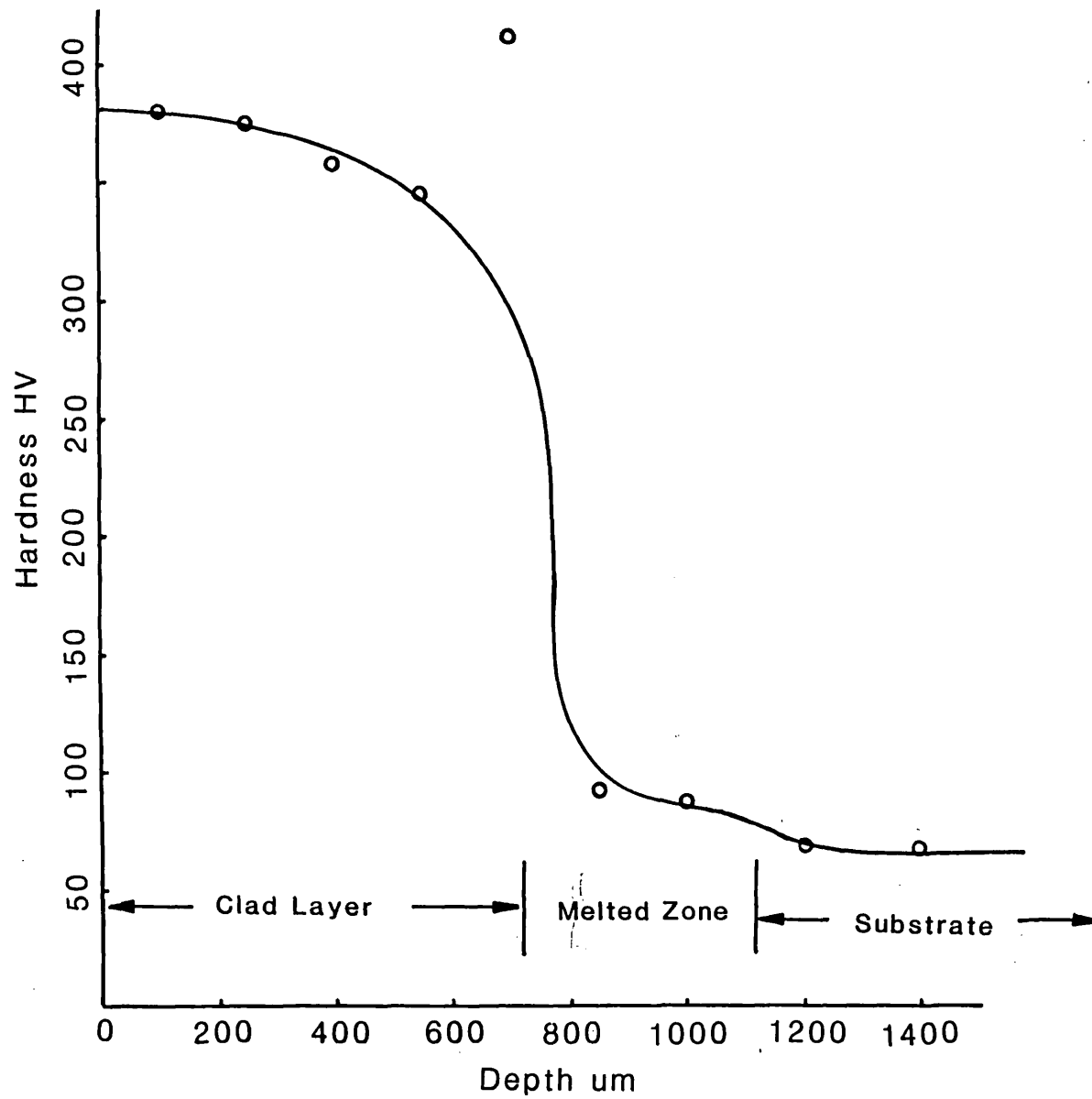


Fig.4.2.18 Microhardness vs depth curves of an Al+mild steel clad layer.

COMPOSITION: 30 at% Al, 69.5 at% Fe



4.3 PARTICLE INJECTION

4.3.1 Operational Parameters

Laser melt particle injection of aluminium alloy (LM 25) was carried out using SiC and Al_2O_3 particles of 125 μm for SiC average size. Grit blasted substrates were processed at various scan speeds and powder flow rates using a power of 2.0 kW with spot diameters of 2, 3, and 5 mm. The process is similar in experimental arrangement to the previous cladding and alloying. Powder particles were blown from a fixed tube into the melt pool, and incorporated into the surface as the trailing edge of the pool solidifies.

The experimental parameters used in the tests involving the forementioned techniques are outlined in table 4 in appendix.

With 5 mm spot diameter, it was not possible to produce any tracks successfully bonded to the substrate. When scan speed is high enough and/or the powder flow rate is low, injection does not occur. At lower scan speeds and/or high powder flow rates, a cavity appears in the middle of the track, and on each side of the track, SiC particles deposit with a very weak adhesion to the substrate. This cavity occurs as a result of vaporising the Al alloy substrate. On the other hand, some deposition with a strong bond to the substrate was achieved with 2 and 3 mm spot diameters. On laser melt tracks SiC particles stick to the substrate. However no carbide particles were seen in the depth of the laser melt zone; so that, no deep laser melt particle

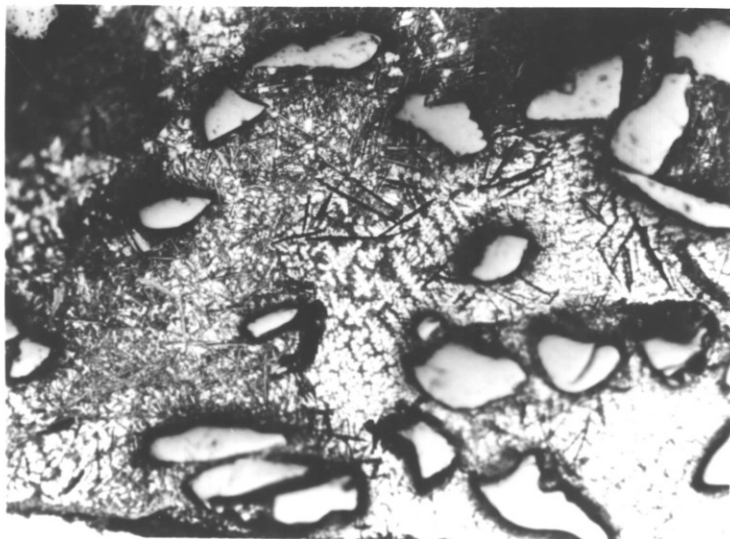
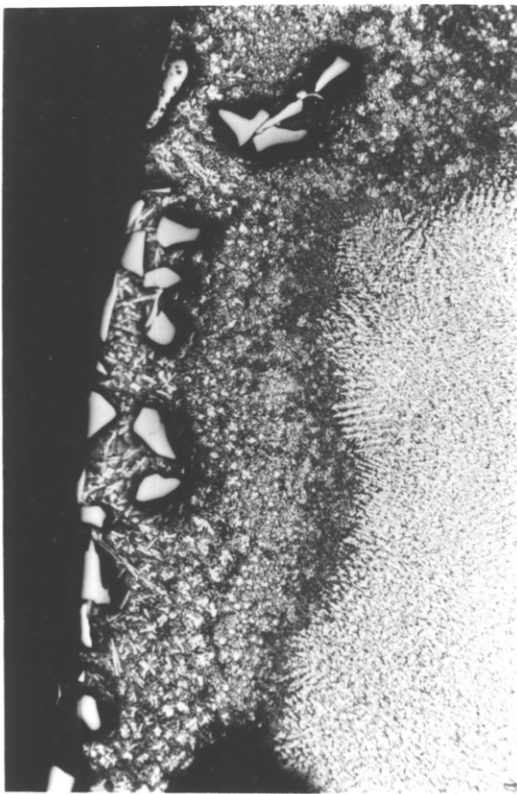
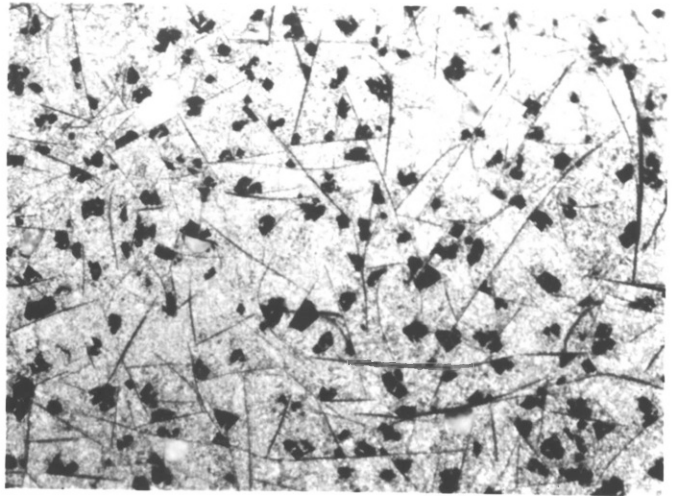
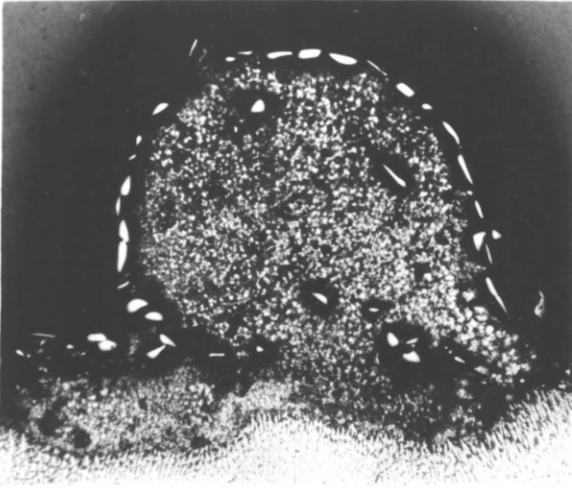
Fig.4.3.1a (Top left) Transverse section of an Al+SiC+Si injected layer. Mag.56. (Si/SiC=3/1)/Al=2/1
P=2.00 kW, BD=3 mm, SS=4.59 mm/sec, PF=3.15 g/min.

Fig.4.3.1.b (Top right) Micrograph taken from above sample (higher magnification) illustrates proeutectic silicon (black particles), acicular resolidified SiC particles, and eutectic matrix with some proeutectic aluminium. Mag.450 (Polarized light)

Fig.4.3.1c (Centre left) General appearance of SiC+Si injected layer. SiC particles stay in upper region of the track. Mag.113.

Fig.4.3.1.d (Centre right) Higher magnification optical micrograph of a SiC+Si injected layer illustrates undissolved SiC particles; plate like proeutectic silicon and eutectic among them. Mag.252.

Fig.4.1.3e (Bottom) Optical micrograph of a SiC injected track. Structure consists of undissolved SiC, dissolved and resolidified SiC, proeutectic Al, eutectic and small proeutectic silicon particles. Mag.252.



injection was achieved.

The SiC particles are very hard (~2000 HV). These particles partly dissolve in the aluminium melt and resolidified particles show a needle-like structure. The hardness of this dissolved resolidified region is about 135-140 HV although some very fine SiC particles were present.

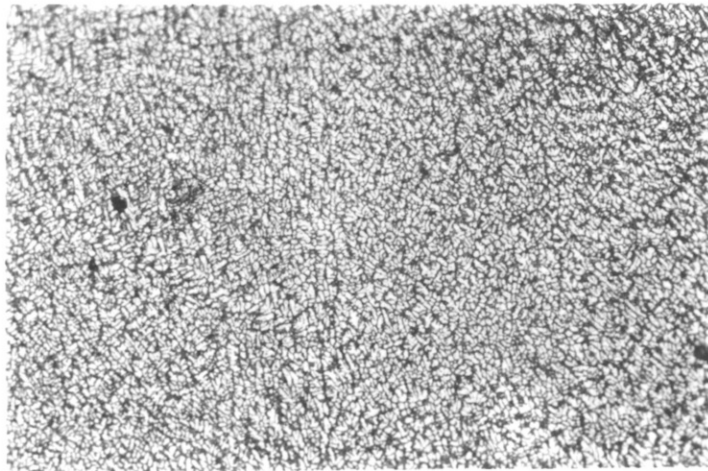
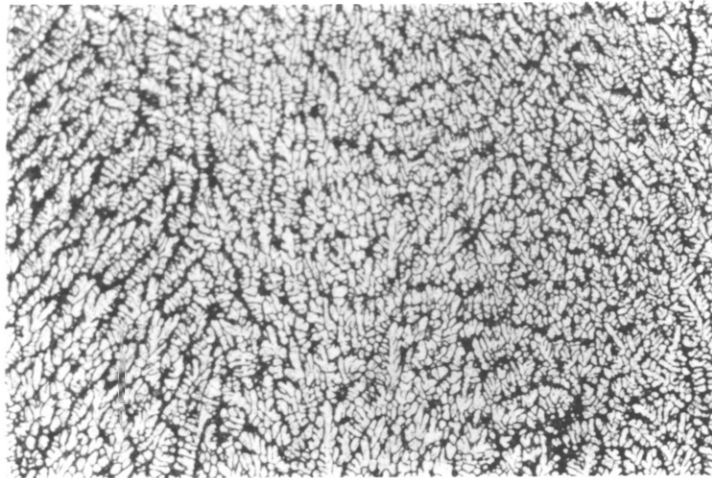
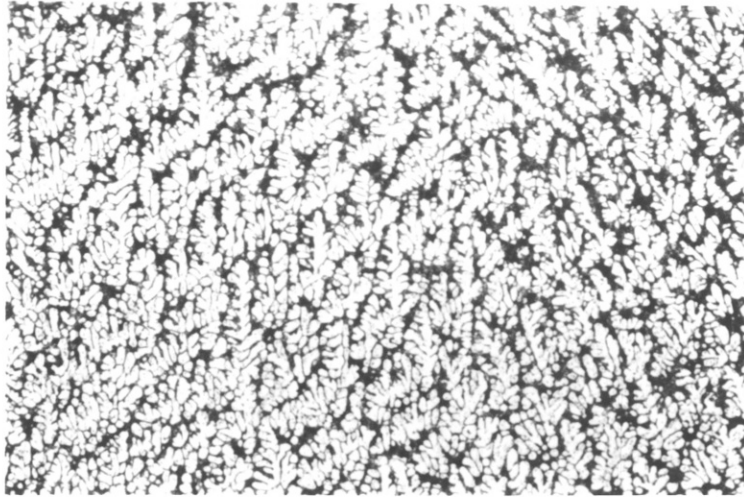
4.3.2 Microstructural Analysis

Surface properties such as wear and corrosion will depend upon the degree of micro structural homogeneity and modification within the solidified melt pool. Particle injected surfaces contain several regions of interest. These are ; the melted zone, the injected particles, the particle matrix interfaces, the matrix and the dissolution products.

An extensive melted zone was found in the laser particle injected aluminium alloy (LM 25) substrates. Particle injection is only found in the upper region of the melted zone Fig.4.3.1c. The structure of this region consists of very fine dendritic+eutectic structure of laser melted substrate material. Size of dendrites changes with scanning speed as a function of cooling rate. Slower scanning speed produces coarser dendrite structure. This is illustrated in Fig.4.3.2 as a function of different scanning speeds. Average dendrite arm spacings^(DAS) measured were:

D A S (λ) (μm)	Scan speed (mm/sec)
2.5 $\pm$ 0.1	4.49
1.7 $\pm$ 0.1	10.49

Fig.4.3.2a-c Dendritic microstructure of the laser melted zone of an Al-Si substrate ⁱⁿ which particle injection ~~was~~ not done . Micrographs were taken from different samples processed at various scan speeds. From top to bottom scan speed increases. These micrographs show the effect of scan speed i.e. interaction time and rapid cooling on microstructural refinement. Micrographs were taken at same magnification. Mag.432. Scan speeds from top to bottom a: 4.49mm/sec, b: 10.49mm/sec, c: 19.60 mm/sec.



The relationship between dendrite arm spacing (λ) and cooling rate ($\dot{E}$) for Al-Si and Al-Cu alloys is given in (Ref.68) as

$$\lambda = \beta \cdot \dot{E}^{-1/3} \quad \beta = 50 \mu\text{m} \cdot \text{K/s}^{1/3}$$

From this relationship, cooling rates are estimated between 8×10^3 and 2.5×10^4 .

The upper region consists of undissolved blocky silicon carbide particles, dissolved and resolidified needle-like elongated silicon carbide, proeutectic silicon, eutectic, proeutectic aluminium and other unidentified (probably Al_4C_3) particles.

Undissolved silicon carbide particles of 2000-2500 HV were only found in the upper region of tracks as is seen in Fig.4.3.1c-e. To make deductions about the distribution of these particles in the track is difficult; because, there is limited injection it would not be realistic. However, in general, particle distribution fluctuates. This is a subject related to laser operation parameters which was mentioned previously.

Melt-particle interaction is expected to result in some dissolution of the injected phase. A consequence of the dissolution process is the appearance of resolidified carbides in the inter-particle matrix. The solidification product is dissolved and resolidified silicon carbide particles. Figs.4.3.3 and 4.3.4. The shape of these particles was found to be long curvilinear and acicular needles; some of them appears in small blocky particles. Resolidified carbides do not appear in dendritic form. Fig.4.3.3 and 4.3.4.

The carbide injected samples also contained eutectic and proeutectic silicon particles. Fig.4.3.3a. This result relates to decomposition of silicon carbide particles according to the Si-C phase diagram Fig.4.3.5, because there is no other source to provide silicon enrichment in the bead. The eutectic morphology appears in fine fibrous form as in the silicon alloyed tracks, and proeutectic silicon appears as small polyhedral particles.

Proeutectic aluminium is an expected phase in the form of fine dendrites, because, the substrate material is a hypoeutectic alloy.

Lastly, there is another solidification product which has not been identified exactly. It appears in the form of fine acicular needles which nucleate from undissolved and resolidified silicon carbide particles. Figs.4.3.3b and 4.3.4d. This phase was found in silicon rich area and is expected to be Al_4C_3 .

A silicon+ silicon carbide+aluminium powder mixture (Si/SiC=3/1)/Al=2/1 was also used in a few experiments. The microstructures of these samples are illustrated in Fig.4.3.1a-d. The structure consists of silicon carbide, proeutectic silicon, proeutectic aluminium and resolidified SiC.

However, in this case, resolidified SiC particles were found extensively only in one track Fig.4.3.1a-b. Most of the tracks appear without resolidified carbide with undissolved particles. The morphology of silicon particles changes with volume fraction of proeutectic silicon as a result of alloyed silicon content which was explained in the

Fig.4.3.3a Optical micrograph of SiC injected Al-Si alloy shows undissolved SiC (big particles), dissolved and resolidified SiC (long curvilinear), proeutectic Si (small white polyhedra), eutectic (dark area), and proeutectic Al (white dendritic area). Mag.347.

Fig.4.3.3b SEM micrograph of a SiC injected track with dissolved SiC, proeutectic silicon and acicular needle-like probably Al_4C_3 phase nucleates from resolidified SiC particles.

Fig.4.3.3c (Bottom left) SEM view of deeply etched SiC injected sample.

Fig.4.3.3d (Bottom right) An optical micrograph illustrates acicular dissolved SiC particles, undissolved SiC particles and proeutectic Si particles (small white particles) Mag.347.

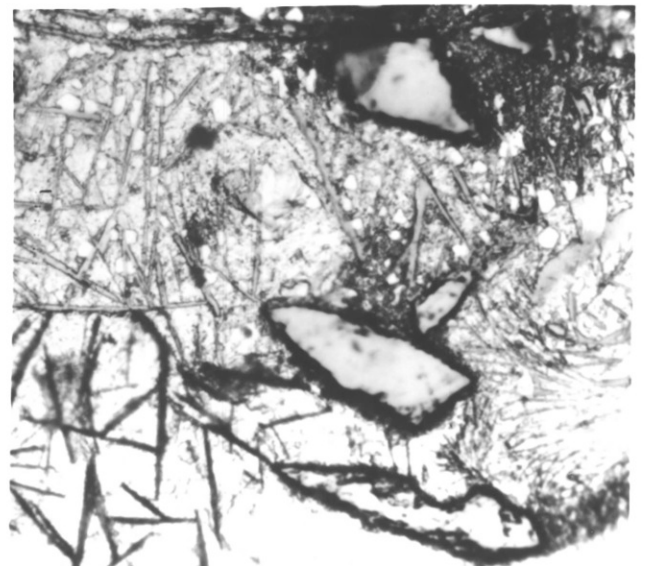
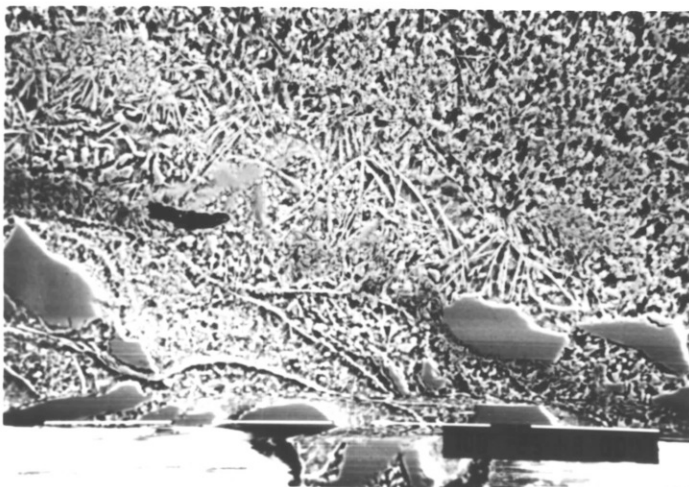
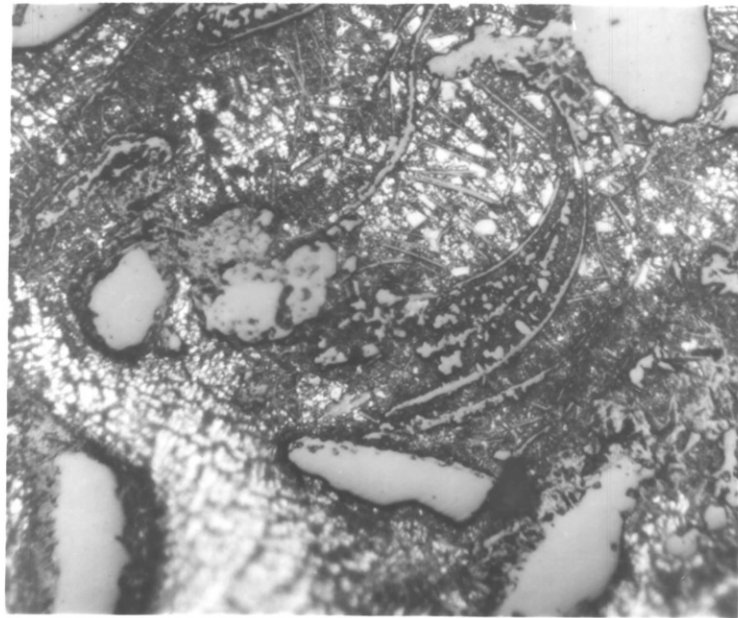
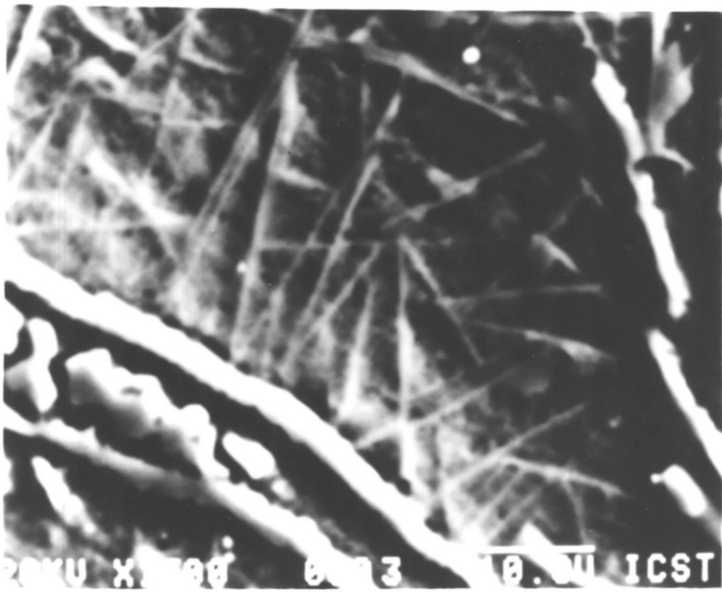
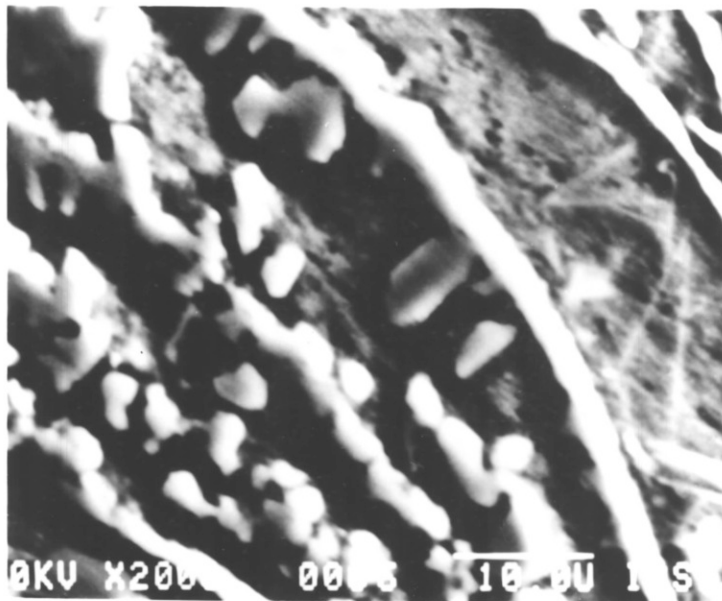
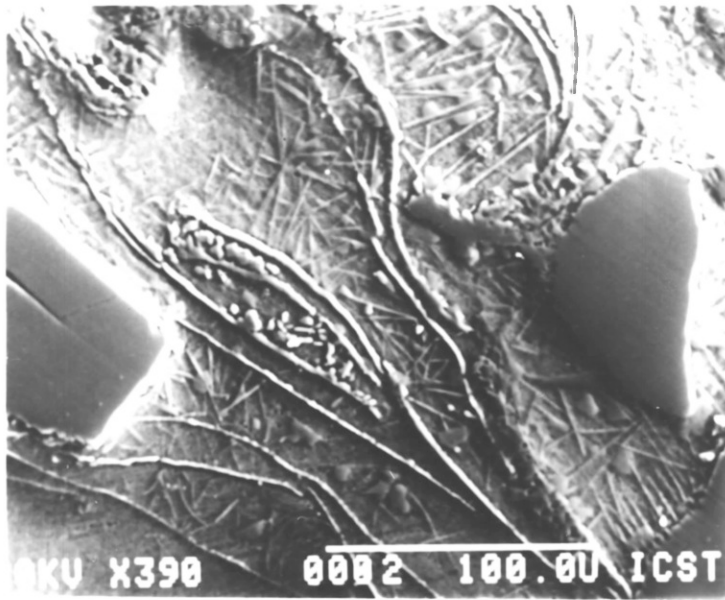


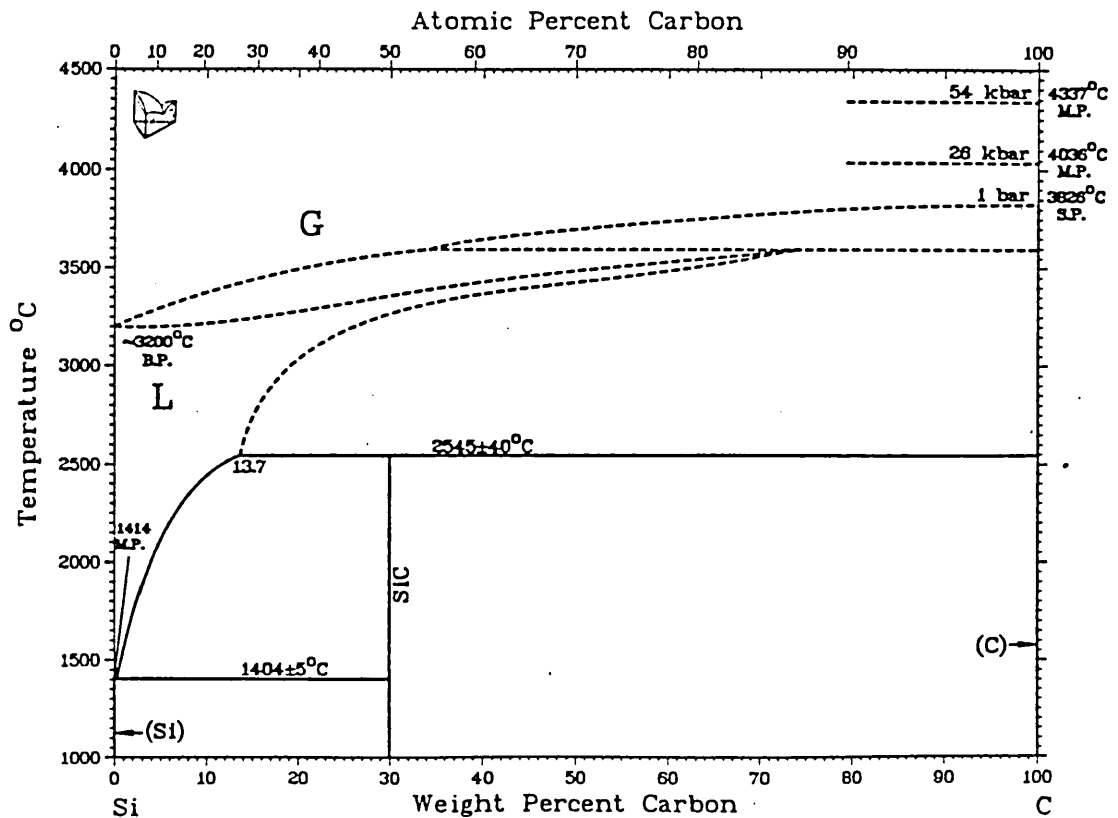
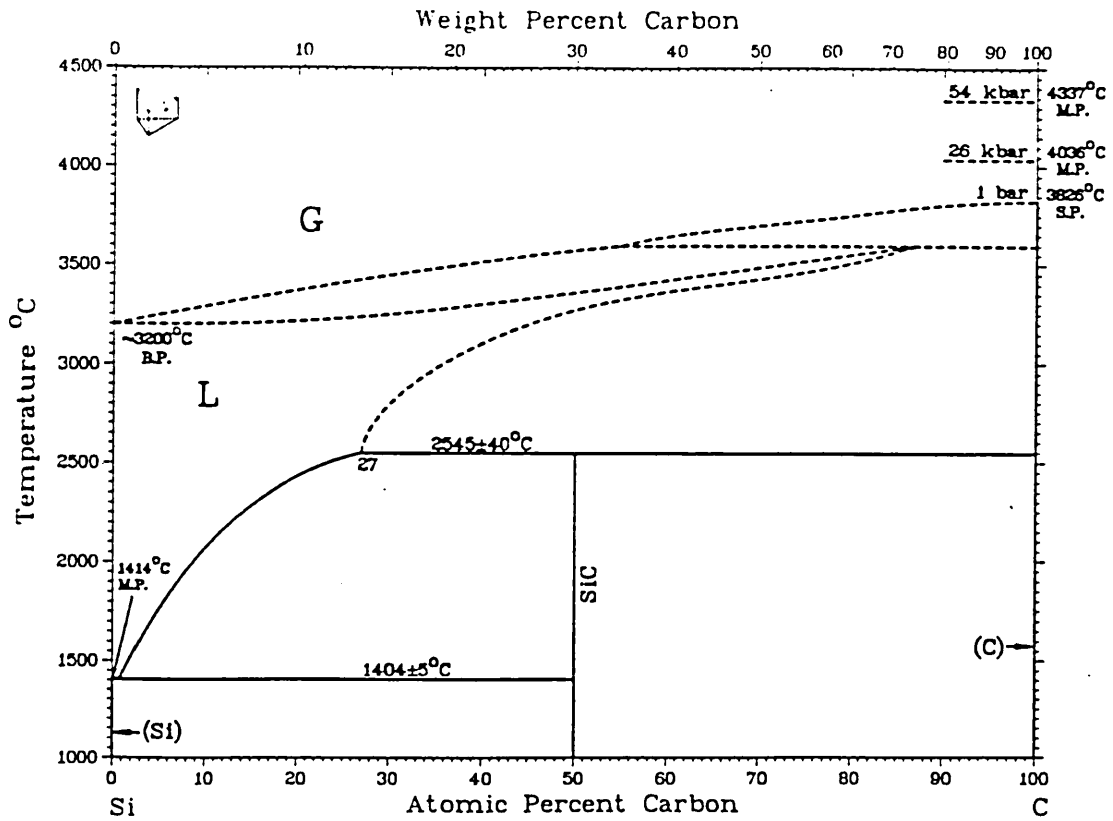
Fig.4.3.4a SEM micrograph of a SiC injected track with long curvilinear dissolved and resolidified SiC.

Fig.4.3.4b Higher magnification SEM view of dissolved and resolidified particles.

Fig.4.3.4c Fine acicular needle like (probably Al_4C_3) phase nucleates from resolidified SiC curves.



Assessed Si-C Phase Diagram



All temperatures have been adjusted to IPTS-68.

R.W. Olesinski and G.J. Abbaschian, 1984.

Fig.4.3.5 Silicon Carbon Binary Phase Diagram. (50)

the silicon alloying section. Identification of these phases was carried out according to appearance and shape of phase particles and EPMA analysis.

4.3.3 X-Ray Diffractometry

The X-ray diffractometry results of a silicon + silicon carbide injected track are given below. (Reference is Hanawalt Index. Card Numbers are (Al):4-0787, (Si):27-1402, (SiC)(H):29-1128, (SiC)(C):29-1129. (H):Hexagonal, (C):Cubic system.

Phase	System	
Al	Cubic	$a_0 = 4.0494 \text{ \AA}$
Si	Cubic	$a_0 = 5.43088 \text{ \AA}$
SiC(H)	Hexagonal	$a_0 = 3.081 \text{ \AA}$ $c_0 = 15.092 \text{ \AA}$
SiC(C)	Cubic	$a_0 = 4.3589 \text{ \AA}$

d Å (Exp.)	d Å (Ref.)	hkl	Phase
3.1329	3.13552	111	Si
2.62	2.628	101	SiC (H)
2.5751			
2.5150	2.516	102,006	SiC (H)
2.3354	2.338	111	Al
2.0239	2.024	200	Al
1.9180	1.92011	220	Si
1.6373	1.63747	311	Si
1.5389	1.5407	110,108	SiC (H)
1.4315	1.431	220	Al
1.3579	1.35772	400	Si
1.2982			
1.2899	1.2897	203	SiC (H)
1.2459	1.24593	331	Si
1.2208	1.221	311	Al
1.1693	1.1690	222	Al
1.1085	1.10857	422	Si
1.0895	1.0893	400	SiC (C)
1.0454	1.04517	511	Si
1.0130	1.0124	400	Al
0.9290	0.9289	331	Al
0.9170	0.91799	531	Si
0.9057	0.9055	420	Al
0.8266	0.8266	422	Al

4.3.4 Aluminium Oxide Injection

Laser melt particle injection of aluminium oxide particles was also examined. The results obtained were largely unsatisfactory. The results showed that no actual particle injection occurred. The substrate shows a clear melted zone where melting occurred. In this zone there are no signs of Al_2O_3 particles. The absorption of the laser energy by Al_2O_3 particles resulted in the melting of these particles; so that, this led to the surface melting of Al_2O_3 instead of particle injection.

CHAPTER 5

DISCUSSION

5.1 LASER SURFACE ALLOYING

5.1.1 Operational Parameters

In the laser surface alloying of aluminium alloy (LM 25) substrates, operational parameters, microstructure, and hardness of alloyed tracks were investigated. The objective was to determine the suitability of laser alloying with silicon powder pneumatically blown into the melt pool, for producing a hard, usable deposit on the aluminium alloy substrate.

The results obtained in terms of operational parameters, bead quality, microstructure, hardness, and composition are presented in Chapter 4.

Operational parameters are scan speed, powder flow rate, and laser power density. Especially for copper and aluminium materials reflectivity is an important factor, and beam coupling efficiency is quite poor. Aluminium alloy surfaces in the machined condition reflect 99% of the incident laser energy at 10.6 μm (CO_2 laser) wave length. From these experiments, it has been observed that a machined surface of LM25 alloy treated at 3 mm beam diameter and 1.95 kW laser power has not been noticeably affected. However, on a grit blasted surface melting occurred and a black painted surface was melted more efficiently. But, this type of painting can change the composition of the material. Therefore, in experiments with 2.0 kW maximum power laser, it is necessary

to use the power as high as possible. Substrate melting does not occur with a lower power. Therefore, changes in laser power were not preferred in the experiments. Laser power density was changed by altering spot size. For 2.0 kW laser power with 5 mm spot diameter substrate melting did not occur without an optical feed back system. With an optical feed back system melting was obtained. This shows that a reflective dome or nozzle has a great effect on beam coupling efficiency. For a flat surface, the reflective nozzle recycles the reflected energy approaching that of reflective dome.

Walker reported that increasing time by reducing scan speed widens and deepens the melt track of substrates.(23) At rapid scan rates minimal silicon alloying occurs and substrate coupling and melting dominate. At low scan rates, critical coupling is achieved; the silicon melt path widens and silicon alloying increases. Walker used preplaced silicon on Al-10wt%Si alloy and had a problem of losing the preplaced silicon layer by shrouding gas erosion, but this problem was eliminated with the blown powder system.(23)

However, two parameters, scan speed and powder flow rate are variable in blown powder feeding systems. When scan speed is reduced for a constant powder flow rate, bead height will increase. Therefore, identification of operating parameters on bead width and height needs very accurate control. But, preplaced powder processing does not require this necessity, because the amount of powder on the substrate is constant and it does not change with changing scan speed. The effect of scan speed and powder deposition

rate on track geometry is shown in Fig.4.1.1 and Fig.4.1.2. Track height shows regular change with changing scan speed and for slower scan speed and higher powder deposition rate track height increases. However, track width does not show regular change.

At the present time suitably low powder feed rates are not possible due to the limitation of powder feed equipment. Especially when two powder feeders are coupled, uniform powder flow is not possible under a certain limit.

The size of the powder feeding tube at the outlet point is important together with feeding angle (see Fig.3.5). If the powder feed tube is larger than the clad track some powder will miss the melt pool. This powder may either stick to the track edge seriously affecting the quality of subsequent overlapped tracks or it may be on the substrate surface affecting the surface reflectivity and is again upsetting the control of the process.

5.1.2 Track Properties

Surface appearance of layers is usually of acceptable quality. Uniformity of powder feed greatly affects the surface roughness. If the powder flow is non-uniform, the track surface will be rough. The surface roughness of a track in the case of non-uniform feeding is about 400 μm due to the pulsing effect of fine silicon powder on a silicon alloyed track. However, in the case of uniform feeding this value becomes as low as 50 μm . Addition of aluminium powder makes the track surface smoother together with strong bonding to substrate. However, in all cases surface roughness is much

less than bead height. Suitable size of powder feed tube with convenient feed angle, and accurate control of powder flow rate promises applicable results.

Both large and small scale porosity is evident in the melted zone and as the largest volume fraction of this zone represents melted substrate material, we must assume that the porosity contained within the melt zone originates in the substrate material. The porosity visible within the alloy beads will be a mixture of porosity of the substrate material plus gases/vapours generated within the alloy bead itself from a variety of sources e.g adsorbed gases and/or moisture.(69) Especially the addition of aluminium powder increases porosity. This is probably due to the aluminium powder which may contain gases and moisture.

Megaw et al. reported that for shorter interaction time $T = 0.051$ sec. there has been insufficient energy input for the eutectic matrix to adequately form and encase the silicon particles. Angular voids result. Higher interaction time resulted in disappearance of the voids. These voids must inherently degrade mechanical properties.(60)

Better mixing is obtained for lower scan speeds, and deeper dilution by remelting the alloyed tracks without powder feed, mentioned in ref. (70). Convection currents in the melt pool cause mixing.

No cracks were observed in alloy layers. This shows that thermal conductivity and difference between coefficient of expansion of alloyed track and the substrate is small; therefore they do not cause a problem. The other point is no brittle inter-metallic compound occurs in the

aluminium-silicon system and a eutectic matrix encase primary silicon.

5.1.3 Microstructural Features

Laser surface alloying of aluminium alloy substrate produced a fine dendritic proeutectic aluminium and fibrous eutectic together with proeutectic silicon particles. Alloyed tracks produced hypereutectic structures.

Growth of the faceted silicon phase is very sensitive to solidification conditions and in the present work proeutectic silicon crystals showed various particle sizes and morphologies. Fig.4.1.5b-d. In particular, plate-like, star-like, polygonal, and dendritic morphologies were found. Walker (23) reported that plate-like crystals were found in regions containing high volume fractions of silicon ($> \sim 30\%$ wt Si) with a tendency to directional growth upwards from the base of the zone. The star-like, polygonal, and dendritic morphologies tend to occur in regions of lower fractions. (65,67)

Growth mechanisms of proeutectic silicon and eutectic structure have been studied in detail by several researchers (61-67), and some information about this subject was given in the literature survey (chapter 2).

Massive geometric particles of silicon develop by layer growth, and growth from an untwinned nucleus develops a faceted particle bounded by slow growth in $\{111\}$ facets.

Primary particles showing twins have been commonly observed in previous work. The twinning does not necessarily influence the growth process, but the twin plane re-entrant

edge mechanism (TPRE) can operate. The formation of star-like morphology has been reported. It has been explained as developing from a decahedron shaped nucleus formed from five twin related tetrahedra.(65-67)

Higher silicon composition levels which have higher liquidus temperatures and slower cooling rates produce plate-like morphology.

Eutectic regions nucleate from proeutectic silicon rods surrounded by the aluminium shell appear to develop from the proeutectic silicon, and show fan-like networks.(23)

5.1.4

Hardness

Silicon alloying of aluminium alloys by laser processing was investigated by several researchers and previous publications showed that increases in hardness were achieved. Megaw et al. reported that the hardness and volume fraction of primary silicon increase with increasing scan speed. Actually, increase in hardness is due to the presence of primary silicon particles.(60)

Walker et al.(31) reported that an approximately linear increase in hardness occurs with increasing silicon content, but substantial scatter occurred above ~30% wt Si. ~450 HV hardness level was reported with ~75 wt.% Si. In the present work over 750 HV was recorded with ~90% wt Si. In some regions upto 800 HV was recorded. The hardness of alloyed tracks against depth is shown in Fig. 4.1.12 to 4.1.15. Fig.5.1.1 shows the summary of the present results in comparison with Walker's results, indicating reasonable agreement.

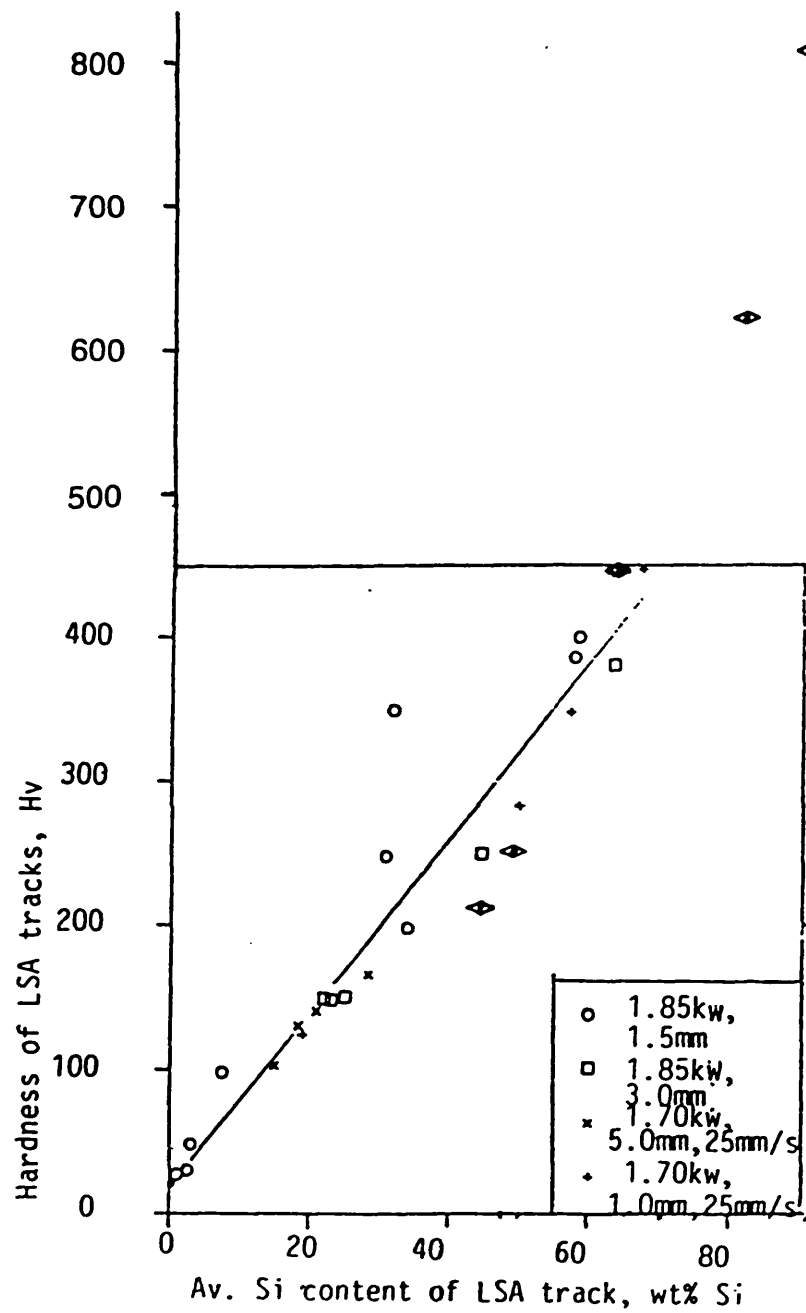


Fig.5.1.1 Average hardness of alloyed/clad tracks vs average Si content.(31)

◆ shows present results.

5.2 LASER SURFACE CLADDING

5.2.1 Operational Parameters

Laser surface cladding of aluminium alloy substrates with mild steel+ aluminium powder mixture and mild steel powders was carried out to find out the possibility of producing useful hard clad layers.

The effect of operating parameters on track geometry was explained in Chapter 4. For a constant powder flow rate increasing scan speed resulted in decreasing track height and at speeds of faster than 20 mm/sec for 3mm beam diameter cladding does not occur; only alloying occurs.

Dilution of clad layer is dependent on scan speed and powder flow rate. At a constant powder flow rate, which is sufficient to produce a clad layer, dilution increases with increasing scan speed until alloying occurs.

Powder flow rate has an opposite but similar effect on track geometry. For a constant scan speed, increasing powder flow rate increases track height within the limits of cladding operational parameters.

The change in track width with scan speed and powder flow rate is different from track height. Under conditions of alloying track height is low, and track width is almost as wide as beam diameter. When the powder deposition increases track width decreases. However, for very high powder deposition rate i.e. at slow scan speeds, track width shows some increase. This is a result of increasing interaction time.

5.2.2 Track Properties

Some properties related to the clad layers are mentioned below.

Porosity is not a problem in these clad layers. However, cracking was present in sections of all tracks. The reasons for this cracking are several, but the main reasons are the following:

a) Thermal expansion coefficient of aluminium and mild steel and intermetallic phases are quite different. Therefore, during cooling in the solid state stresses build up which may cause cracking.

b) Liquidus temperatures of the substrate and the clad layer are quite different.

c) Thermal conductivities of substrate and clad layer are very different. When the aluminium substrate cools down shrinkage occurs. On the other hand, the clad layer cools down much slower than the substrate does. This shrinkage puts a stress on the clad layer and cracking results.

Heating of the substrate upto 400 °C may reduce or eliminate the possibility of cracking.

d) Fe_3Al is a very brittle phases and will crack under small tensile loads.

High hardness levels were produced by this alloy system based on Fe-Al. Hardness of clad layers changes from track to track with different phases as a result of different powder ratio i.e. composition. Volume fraction of phases also affects hardness. Upto 650 HV was measured in some

inter-metallic clad layers. The hardness of two phase structures ($\text{Al}_3\text{Fe}-\text{Al}$) at the composition of ~78 at.% Fe is about 350 HV.

5.2.3 Microstructural Features

Virtually all the beads appear to contain a very high proportion of Fe-Al intermetallics except where the processing conditions produced alloying rather than cladding. From composition micro analysis and X-ray diffractometry, Fe_3Al , FeAl , FeAl_2 , $\text{Al}_{13}\text{Fe}_4$ (shown as Al_3Fe in Fig.4.2.11), Al, and alpha iron are expected according to the Al-Fe phase diagram.(Fig.4.2.11)

Generally, the solidification structure shows cellular and dendritic structure for Al- Al_3Fe region. Fe_3Al shows equiaxed grain structure. Structural and phase studies of the system need considerable work. Different phases with different compositions were obtained under different cooling rates. Therefore it is a very wide investigation area and further detailed work is needed.

Mild steel clad layers have ferrite+pearlite structures and a tempered martensitic structure was also found in one track as a result of quenching and probably the tempering during the processing of the next track.

The most important aim of this part of the investigation was to establish if there was a possibility of producing durable, strong, useful clad layers. The results were largely unsatisfactory due to excessive cracking.

5.3 LASER-MELT PARTICLE INJECTION

5.3.1 Operational Parameters

Laser melt particle injection processing of an aluminium alloy (LM 25) substrate was investigated. The particle injection technique was attempted with SiC and alumina. The results obtained were largely unsatisfactory, and showed that no actual powder injection occurred. Only some tracks show surface injection with 2 and 3 mm spot diameter, but no deep injection occurred in the melted zone where melting has occurred. In this zone there are no signs of SiC or alumina particles. It has been suggested that the particles float on the surface of the melt pool rather than sinking to form a mix with the substrate matrix. This would seem reasonable for a steel substrate where the injected material would have lower density; however, with an aluminium substrate this should not be a problem, because, the density of SiC particles ($\sim 3.2 \text{ g/cm}^3$) is higher than the density of the aluminium alloy substrate ($\sim 2.67 \text{ g/cm}^3$).

Ayers et al.(42) reported that particles must enter the melt pool under a sufficient particle velocity, especially for powder particles whose density is lower than the substrate material. It is reported that TiC particles (4.9 g/cm^3) on a steel surface (7.9 g/cm^3) need sufficient entry kinetic energy.

In another experiment, Ayers (41) injected 5052 and 6061 aluminium alloys with TiC and WC particles, and no difficulty appeared which would prevent the injection of

aluminium alloys with substantially these powders. However, difficulty was experienced with injecting uniform surface layers of 5052 aluminium alloy with -70 +100 mesh TiC powder if the processing was done at reduced pressure in a helium environment. When the processing was done in air using helium gas shielding, a uniform layer was produced. However it is not obvious why processing in air produced better results with the coarse particles.

In virtually all of the test tracks in the present work, a SiC clad has formed on top of the melted zone. This clad layer is normally variable in quality. The quality can be improved with shrouded remelting.

Coquorgelle et al. reported work on Ti-6Al-4V alloy injected with TiC 100 μ m particles involving a second remelting by which a more uniform alloy was obtained of hardness exceeding 700 HV.(71)

5.3.2 Microstructural Properties

SiC particles with optimum operational parameters in the present work show a strong bond to the substrate and carbide dissolution was observed in all cases. Dissolved carbides show curvilinear long needle-like morphology and solidifies oriented to undissolved SiC particles. The hardness level of this carbide dissolved region is about 130-140 HV, However, the hardness of SiC particles was greater than 1500 HV. Fig.4.3.4a shows dissolved carbide particles with undissolved carbide particles in an Al-Si matrix.

No carbide dissolution or melting is reported with TiC injected 5052 and 6061 aluminium alloys. However, carbide

dissolution and melting in TiC injected Ti-6Al-4V and WC injected Inconel 625 and WC injected Ti-6Al-4V were reported.(37)

In the present work, with a 5 mm spot diameter, at slow scan speed and high powder flow rate vaporization of aluminium substrate occurred, but no injection was observed. Due to aluminium vapour SiC particles stick to each other and deposit on each side of the gap which is formed by aluminium vaporization, with very weak adhesion to the substrate. This shows that particle deposition is high enough and a remarkable amount of laser energy is absorbed by carbide particles and it evaporates the substrate.

It is suggested that the absence of actual particle injection taking place is due to:

a) No stable or sufficient melt pool being formed on the substrate surface because of the bad irradiation absorbance and high thermal conductivity of the aluminium alloy substrate. The high conductivity causes rapid solidification of the melt zone not allowing time for injection of powder particles.

b) Other successful experiments which were carried out by other researchers were done at higher laser powers such as 5.0 kW or more.(37,40-42) This produces sufficient melting and stable melt pools.

c) The particles used in this work, especially alumina powder, were very small in size (resembling polishing powder). It was found difficult to blow these particles into the melt pool uniformly. They absorbed enough laser energy to raise their temperature to melting so that they formed a

surface melt rather than striking the substrate melt pool as solid particles.

For further investigation, remelting of shallow injected layer or fusing SiC particles to the surface may give more uniform mixing to substrate.

CHAPTER 6

CONCLUSIONS

6.1

LASER SURFACE ALLOYING

Laser surface alloying of aluminium-silicon alloy (LM25) substrates with silicon and silicon+aluminium powders were carried out. From the results of the alloying process:

1) Alloying of Al-Si alloy (LM25) substrates with silicon and Si+Al powders by blown powder system was achieved.

2) Due to agglomeration of silicon powder, design of powder feeder and powder particle size are important. Powders of coarser than 44 μm together with closer size distribution allow better control of the feeding process.

3) Track geometry is controlled by scanning speed and powder flow rate for a chosen beam diameter; decreasing scan speed at a constant powder flow rate increases the track height. With increasing bead height penetration of laser beam into the substrate becomes less and the melted zone in the substrate becomes shallower.

The effect of powder flow rate is similar. Increasing powder flow rate at a constant scan speed increases track height for scan speeds slower than 20 mm/sec.

The powder feeding angle was chosen as 35° - 40° for optimum operation.

4) Surface roughness and flatness depend mainly on uniform powder flow. Pulsing and changing flow rate result in discontinuity of flatness.

5) Excessive porosity is present in the majority of

tracks, but some tracks with little porosity were produced.

6) Cracking did not occur in silicon and Si+Al alloying.

7) Silicon and Si+Al alloyed tracks produced by laser processing consist of proeutectic silicon, fibrous eutectic and proeutectic aluminium. Proeutectic silicon is generally surrounded by a proeutectic Al halo and dendrites. Proeutectic silicon shows polyhedral, dendritic, star-like and plate-like morphologies changing with composition, and the eutectic silicon nucleates from proeutectic silicon.

8) Laser processing causes rapid cooling. Cooling rates are estimated from dendrite arm spacing as $\sim 10^4$ K/sec and increase with increasing scan speed.

The eutectic composition was shifted towards the silicon side of the Al-Si phase diagram (to ~ 18 wt.% Si) by rapid solidification.

9) Silicon alloying increases the hardness depending greatly on the proportion of proeutectic silicon. Hardness values up to 800 HV were obtained with ~ 94 wt.% Si.

10) The addition of aluminium changes composition of alloys. Therefore, volume fraction of silicon may be controlled by this way. The feeding of Al+Si mixture is an alternative method of controlling the volume fraction of silicon in the alloy layers.

6.2

LASER SURFACE CLADDING

Laser surface cladding of LM25 substrates with mild steel+aluminium and mild steel powders was carried out.

1) Track geometry is controlled by scan speed and powder

flow rate for a certain beam diameter. Dilution changes as a function of scan speed and/or powder flow rate. At a scan speed of faster than 20 mm/sec alloying occurs instead of cladding.

2) A narrow aluminium-iron inter-metallic zone between the clad layer and the substrate is always present. Different inter-metallic phases were produced in the tracks as a function of changing mild steel/aluminium ratio.

3) Surface quality is acceptably good.

4) Tracks are almost porosity free.

5) Cracking is the main problem, so that this cladding procedure does not give satisfactory results for production.

6) The solidification structure of clad layers consists of cellular, dendritic and equiaxed structures of the phases.

6.3 LASER PARTICLE INJECTION

From the results of the laser particle injection process the following conclusions are obtained.

1) No deep injection of SiC powder was found, but some shallow injection was obtained.

2) Deposit morphology is dependent on powder flow rate, scan speed, and energy density.

3) Silicon carbide dissolution occurs with some decomposition of SiC.

4) The hardness of the resolidified area shows only a small increase of ~60-65 HV.

5) The use of a higher power laser (for example 5 kW) may produce more successful and satisfactory results.

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APPENDIX.

TABLE 1 Operation Parameters for Silicon Alloying.
(Power= 2.0 kW)

Track No	Beam	Scanning	Powder
	Diameter (mm)	Speed (mm/sec)	Flow (g/sec)
1	3	3.54	0.019
2	3	5.18	0.019
3	3	6.94	0.019
4	3	8.90	0.019
5	3	12.37	0.019
6	3	20.04	0.019
7	3	4.48	0.027
8	3	7.43	0.027
9	3	10.48	0.027
10	3	13.77	0.027
11	3	20.04	0.027
12	5	5.60	0.019
13	5	6.94	0.019
14	5	9.09	0.019
15	5	13.93	0.019
16	5	32.89	0.019

TABLE 1a Effect of Operation Parameters on Track Geometry
for Si Alloying (Power= 2.0 kW)

Track No	Beam	Track Diameter	Track Height (mm)	Track Width (mm)	Track Depth (mm)
1	3	1.18	3.00	0.87	
2	3	0.80	3.15	0.85	
3	3	0.40	2.85	0.80	
4	3	0.25	2.95	0.85	
5	3	0.125	2.85	0.875	
6	3	0.05	2.70	0.80	
7	3	0.95	3.50	0.95	
8	3	0.55	2.95	0.75	
9	3	0.25	2.85	0.80	
10	3	0.175	2.70	0.825	
11	3	0.075	2.50	0.775	
12	5	1.00	3.70	1.05	
13	5	0.65	3.20	1.00	
14	5	0.43	4.00	1.15	
15	5	0.25	3.75	1.10	
16	5	0.10	3.20	0.90	

TABLE 2 Operation Parameter for Al+Si Alloying

Track No	Power (kW)	Beam Diameter(mm)	Scanning Speed (mm/sec)	Powder Flow(g/sec)
1	1.9	4.8	8.80	0.07
2	1.9	4.8	8.80	0.08
3	1.9	4.8	8.80	0.09
4	1.9	4.8	8.80	0.108
5	1.9	4.8	8.80	0.122
6	1.9	4.8	8.80	0.138
7	1.9	4.8	8.80	0.150
8	1.9	4.8	8.80	0.160
9	1.9	4.8	8.80	0.128
10	1.9	4.8	8.80	0.117
11	1.9	4.8	8.80	0.094
12	1.9	4.8	8.80	0.058
13	1.9	4.8	8.80	0.038
14	1.9	4.8	6.65	0.047
15	1.9	4.8	6.65	0.067
16	1.9	4.8	6.65	0.10
17	1.9	4.8	6.65	0.125
18	1.85	4.8	4.68	0.07
19	1.85	4.8	4.68	0.08
20	1.85	4.8	4.68	0.09
21	1.85	4.8	6.10	0.07
22	1.85	4.8	8.68	0.07
23	1.85	4.8	8.68	0.08

TABLE 3 Operation Parameters for aluminium+mild steel
Cladding

Power: 2.05 kW and Beam Diameter: 3 mm

Track No	Scan Speed (mm/sec.)	Powder Flow Rate (g/min.)	
		Fe	Al
1	4.44	3.35	1.05
2	7.55	"	"
3	14.30	"	"
4	4.73	"	"
5	4.59	2.10	2.00
6	7.12	"	"
7	10.50	"	"
8	19.57	"	"
9	4.86	2.10	3.10
10	7.41	"	"
11	10.58	"	"
12	17.60	"	"
13	4.67	2.10	3.50
14	7.45	"	"
15	10.87	"	"
16	19.38	"	"
17	4.50	4.80	1.40
18	7.29	"	"
19	10.42	"	"
20	19.53	"	"
21	10.64	6.60	1.05
22	14.45	"	"
23	7.47	"	"
24	4.45	"	"
25	4.45	7.60	1.05
26	6.32	"	"
27	9.93	"	"
28	18.35	"	"

TABLE 3a Track Geometry of aluminium+mild steel clad layers

Power: 2.05 kW and Beam Diameter: 3 mm

Track No	Track Height (mm)	Track Width (mm)	Track Depth (mm)
1	0.95	2.85	0.20
2	0.25	2.25	0.55
3	0.15	3.00	0.80
4	0.55	2.15	0.45
5	0.95	2.10	0.30
6	0.60	2.10	0.35
7	0.25	2.60	0.65
8	0.15	2.60	0.80
9	bead lost	1.95	0.275
10	0.95	2.10	0.30
11	0.55	2.20	0.45
12	0.15	2.50	0.75
13	1.80	1.80	0.20
14	1.35	1.85	0.20
15	0.90	2.10	0.45
16	0.30	2.25	0.65
17	1.00	2.05	0.35
18	0.65	2.00	0.35
19	0.30	2.10	0.40
20	0.10	2.50	0.75
21	0.45	2.10	0.40
22	0.15	2.00	0.45
23	0.70	2.05	0.40
24	1.25	2.30	0.35
25	1.40	2.35	0.40
26	0.90	2.25	0.45
27	0.45	2.10	0.45
28	0.125	2.25	0.575

TABLE 4 Operation Parameters for Laser-Melt Particle Injection

Track No	Power (kW)	Beam Diameter(mm)	Scanning Speed (mm/sec)	Powder Flow(g/sec)
1	2.0	3	10.11	0.037
2	2.0	3	10.11	0.028
3	2.0	3	12.5	0.028
4	2.0	3	20.04	0.028
5	2.0	3	20.04	0.019
6	2.0	3	9.44	0.019
7	2.0	3	4.83	0.019
8	2.0	3	9.77	0.019
9	2.0	3	20.20	0.019
10	2.0	3	3.46	0.019
11	2.0	3	7.97	0.037
12	2.0	3	12.48	0.037
13	2.0	3	21.83	0.037
14	2.0	3	33.90	0.037

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