A COMPARATIVE STUDY ON LASER PROCESSING OF COMMERCIALLY AVAILABLE TITANIUM ALUMINIDE (TI-48AL-2CR-2NB) AND *IN-SITU* ALLOYING OF TITANIUM ALUMINIDE

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ABSTRACT

Titanium aluminides (TiAl) are acknowledged as promising high temperature structural materials due to their high melting point, high strength to density, high elastic modulus and high creep strength. Due to their low ductility, it is difficult to machine post manufacturing with conventional manufacturing techniques. TiAl components have been successfully produced using the cast methods, but prove to be very expensive and time consuming. *In situ* alloying, using laser processing, is a potential method of manufacture of these alloys; as elemental powders can be fed separately from different powder feeders and the feed rate can be controlled, independently, to control the composition and resulting microstructure.

1. INTRODUCTION

Dual phase gamma (γ) titanium aluminides (TiAl) is a promising candidate material in high temperature aerospace and automotive gas turbine engines to replace Ni-based superalloys for its high specific strength, high stiffness, good corrosion resistance, high creep resistance between 600 - 750 °C, and oxidation resistance [1,2,4]. Ni – based superalloys are found to possess superior properties than γ – TiAl however, the low density of γ – TiAl enhances its specific properties considerably in comparison to Ni – based superalloys [3]. Replacement of Ni – based superalloys parts with TiAl alloys is expected to reduce the structural weight of high performance gas turbine engines by 20 – 30% [1,2,3], however a major stumbling block to adoption is its low ductility.The ductility of gamma alloys is controlled by alloy chemistry and the microstructure [13]. Studies on Ti-(43-55) Al compositions showed that the lowest strength occurs at compositions around Ti-51Al and that room-temperature elongation varies with aluminium content, exhibiting a maximum around the two-phase composition of Ti-48Al [4,5,6,8]. The best ductility observed in Ti-48Al was hypothesised to be governed by an optimum volume ratio of the alpha and gamma phases, below which grain growth becomes pronounced when heat treated in the α+ γ phase field, and above which the brittle alpha-2 phase eliminates the beneficial effect of the refined microstructure [7,9,10]. Additions of chromium increase the ductility of two phase gamma alloys but have no obvious effect in single-phase alloys containing more than 50 at. % Al. Small additions of these elements in two-phase gamma compositions (e.g Ti-48Al base) are believed to increase metallic bonding by decreasing aluminium content in the gamma phase [8]. For a given two phase alloy composition, microstructural variations directly influence the room temperature tensile ductility, which ranges from 0.5% to 3.5% plastic elongation [8,11] and mechanical properties. For example equiaxed gamma materials provide good ductility and strength, but inferior fracture toughness and creep resistance. The coarse grained lamellar microstructure on the other hand possesses superior fracture toughness and creep resistance [6,7]. Near-gamma alloys containing a fully lamellar microstructure with a moderately small alpha grain size have been found to provide the desired mechanical properties [2,3,5,7]. Several processing techniques have been developed to obtain such microstructures. Casting metallurgy methods followed by hot isostatic pressing (HIP) are commonly accepted by industry for raw γ-TiAl material production before mechanical processing but prove to be very expensive and time consuming [1]. The brittleness and low fracture toughness of γ-TiAl makes it difficult to manufacture with conventional methods. To improve on their ductility, research and development make use of ternary and quaternary alloys amongst others, e.g. GKSS and GE Alloys in the manufacturing of durable, weight reduced structures [12,15]. As a structural alloy, machinability and fabricability represent important considerations in the application of the alloy. Due to the complexity of conventional manufacturing of these alloys and the expensive pre-alloyed powder, direct energy deposition, using laser processing, is a promising method of manufacture of these alloys as elemental powders can be fed separately and independently, to save on costs, to produce the desired microstructure, and eliminate the need for post processing [1]. Using this method, *in-situ* alloyed powder was processed on a 3kW, 1073 nm wavelength, IPG Nd-YAG fibre laser. This study looks into the effect of the processing parameters, specifically aluminium feed rate, on the *in-situ* alloy composition and microstructure and compares this to the commercial GE alloy (Ti-48Al-2Cr-2Nb).

2. METHODOLOGY

An IPG Nd:YAG 3kW 1073 nm fibre laser was utilised in these experiments for the deposition of single clad tracks onto Ti6Al4V substrates. Powder is carried to the workpiece by means of a GTV powder feeder together with a three-way cladding nozzle. The processed powders were pure aluminium and a master alloy consisting of Grade 1, commercially pure titanium, niobium and chromium which was subjected to satellite mixing. This master alloy powder was alloyed *in-situ* with the aluminium powder using two powder feeding hoppers. The feed rate of the powders were varied to evaluate its effect on composition and microstructure. Samples were cut, polished, and etched according to standard metallographic procedures for titanium alloys. Samples were also subjected to heat treatment at 1200°C for 20 minutes and then quenched in water. Parameters for the cladding are included in Table 1. The microstructures were investigated using optical microscopy and composition was studied with SEM-EDS. Zwick/Roell Indentec (ZHVµ) was used for measuring the micro-hardness of the produced alloy with a 0.5kg load.

Table 1: Summary of operating parameters for the IPG laser system

Sample	Power (kW)	Speed (m/min)	Shielding gas rate (L/min)	Master Alloy Feedrate (rpm)	Al federate (rpm)	Master alloy carrier gas (L/min)	Al carrier gas (L/min)
Commercial GE Alloy	n/a	n/a	n/a	n/a	n/a	n/a	n/a
1	1.25	0.5	12	1	1.6	1.8	2
2	1.25	0.5	12	1	1.8	1.8	2
3	1.25	0.5	12	1	2.2	1.8	2
4	1.25	0.5	12	1	2.5	1.8	2

3. RESULTS AND DISCUSSION

Figure 1: Microstructure of the commercial GE alloy Ti-48Al-2Cr-2Nb showing the lamellar type microstructure

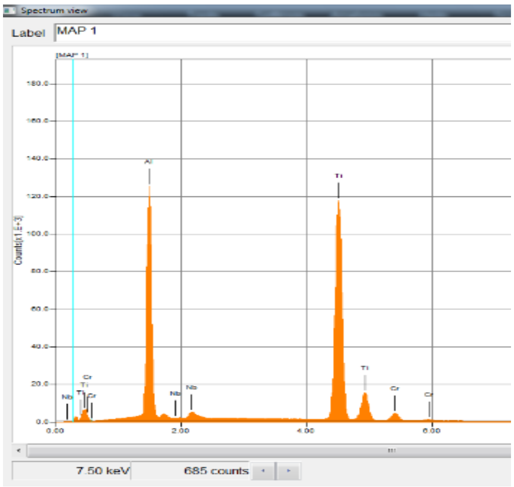
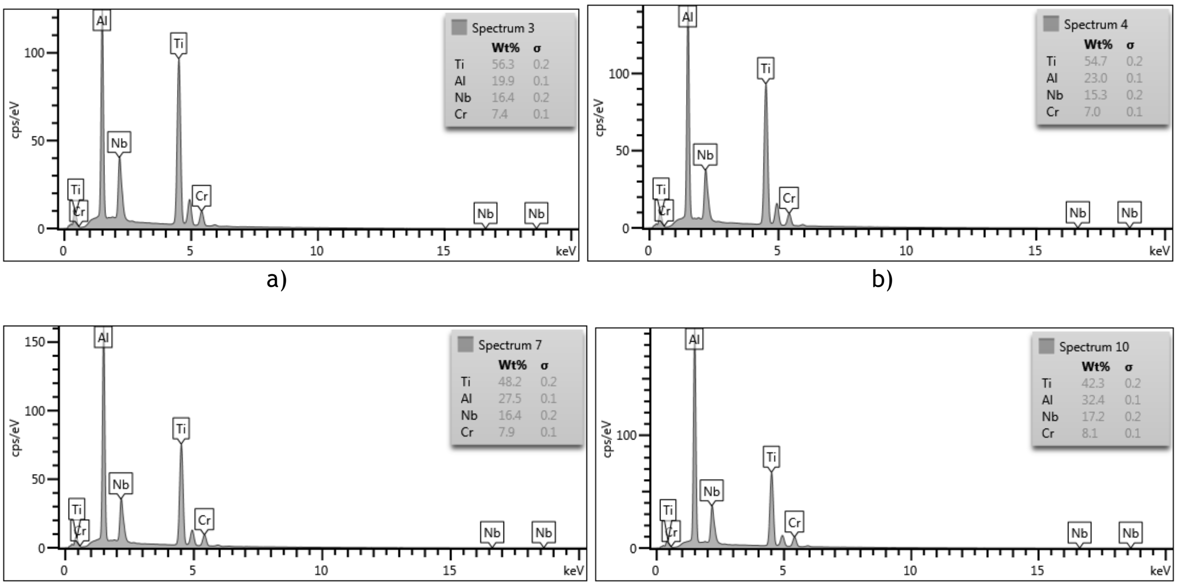


Figure 2: EDS spectrum of the commercial alloy Ti-48Al-2Cr-2Nb

Figures 1 and 2 report on the microstructure and composition of the GE plate that was purchased and analysed. The microstructure is fully lamellar as expected and EDS composition gave mass percentages of Al: 40.07, Ti: 55.44, Cr: 2.56, Nb: 1.93. These values correspond to a dual phase titanium aluminide. EDS results of the various *in-situ* produced alloys of samples 1,2,3 and 4 are shown in figure 3.

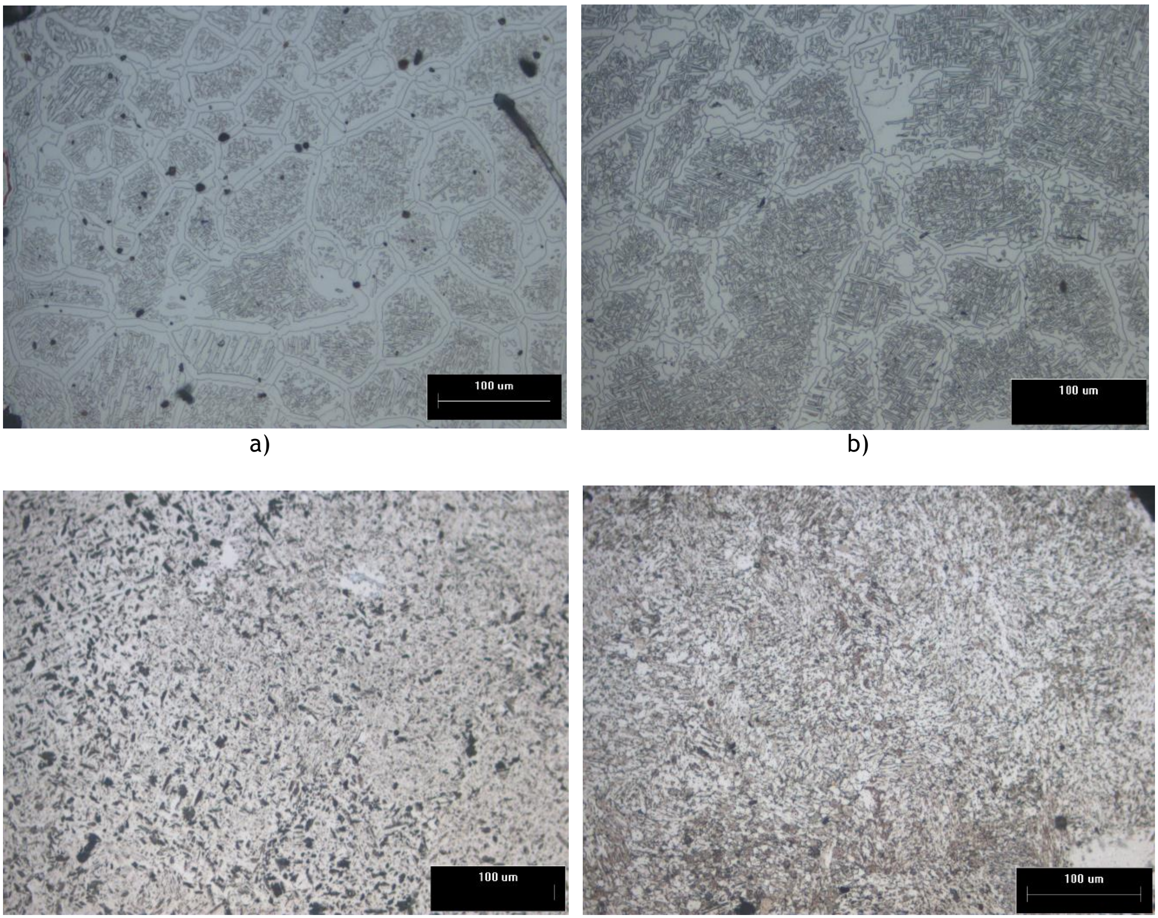


a) b)

c) d)

Figure 3: EDS results of a) Sample 1, b) Sample 2, c) Sample 3, d) Sample 4 showing the effects of the change in feed rate of aluminium on the Al composition of the in-situ alloy

The results shown in figure 3 show that it is possible to change and control the composition of the in-situ alloy by varying the feed rate of the powder feeders to achieve the desired composition. As the feed rate settings of the aluminium feeder is varied so too does the composition of the alloy change accordingly. The Nb and Cr content can be attributed to the inhomogeneous mixing of the powders in the master alloy. Corresponding microstructures are shown in figure 4. The microstructure of samples 1 and 2 (figure 4a and 4b) show hexagonal grains consisting of coarse lamellar predominantly of the alpha 2 phase which agrees with the EDS composition of samples 1 and 2 (figures 3a and 3b). Heat treatment at 1200ºC or above the alpha transus temperature at this Al composition and the subsequent rapid cooling results in this lamellar structure [7]. The dwell time of 20 minutes contributed to the grain coarsening [5,7,13]. Samples 3 and 4 in figures 4c and 4d show a finer duplex microstructure consisting of a dual phase of alpha 2 and the gamma phase. In this compositional range (figures 3c and 3d) heat treatment at 1200ºC takes place in the α + γ region and results in the duplex microstructure. Duplex microstructures typically contain fine equiaxed γ- TiAl grains with low volume fractions of relatively fine lamellar γ-TiAl/α_2-Ti_3Al grains and/or very fine equiaxed α_2-Ti_3Al grains and are generated by heat treatment in the α + γ phase region [7,15]. Residual Nb can also be expected in all these microstructures due to the high concentration of Nb. Table 2 shows the micro hardness of the various produced alloys. It is evident that the hardness is affected by the change in Al content and the resulting microstructure. The GE alloy showed a hardness of 377 HV (0.5kg). Samples 1 and 2 consists of predominantly the alpha 2 phase which is inherently brittle resulting in the higher hardness. [7,9,10]. The duplex microstructure of samples 3 and 4 contribute to the lower hardness of these samples [7,13].



a) b)

c) d)

Figure 4: Microstructure of a) sample 1, b) sample 2, c) sample 3, and d) sample 4

Table 2: Micro hardness of the produced alloys

Sample	Hardness (HV0.5)
GE alloy	377
1	608
2	556
3	470
4	378

4. CONCLUSION

The compositions achieved in this work were below the desired dual phase near gamma region of titanium aluminide as indicated by the EDS in figure 3. An increase in the feed rate of aluminium produced a dual phase duplex microstructure; however these structures (figures 4c and 4d) did not consist of the fine lamellar of the commercial alloy. Reasons for this include the EDS showing that the feed rates of aluminium need to be increased to produce the gamma titanium aluminide composition and microstructure. Further investigation is also required of the heat treatment profile. The current heat treatment temperature of 1200°C showed this temperature is below the alpha transus temperature required to produce the fine grain lamellar. This work shows promising results that it is possible to produce TiAl and to manipulate the microstructure using direct energy deposition laser processing via *in-situ* alloying.

5. REFERENCES

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