



Precipitation behavior of aluminum alloy 2139 fabricated using additive manufacturing



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ARTICLE INFO

Article history:

Received 26 June 2015

Received in revised form

26 August 2015

Accepted 27 August 2015

Available online 2 September 2015

Keywords:

Aluminum

Additive manufacturing

Characterization

ABSTRACT

Additive manufacturing (AM) is an emerging technology capable of producing near net shape structures in a variety of materials directly from a computer model. Standard metallic alloys that were developed for cast or wrought processing have largely been adopted for AM feedstock. In many applications, these legacy alloys are quite acceptable. In the aluminum alloy family, however, there is a significant performance gap between the casting alloys currently being used in AM processes and the high strength/toughness capability available in certain wrought alloys. The precipitation hardenable alloys, most often used in high performance structures, present challenges for processing by AM. The near net shape nature of AM processes does not allow for mechanical work prior to the heat treatment that is often necessary to develop a uniform distribution of precipitates and give peak mechanical performance. This paper examines the aluminum (Al) alloy 2139, a composition that is strengthened by homogeneous precipitation of Ω (Al_2Cu) plates and thus ideally suited for near net shape processes like AM. Transmission electron microscopy, microhardness, and tensile testing determined that, with proper processing conditions, Al 2139 can be additively manufactured and subsequently heat treated to strength levels comparable to those of peak aged wrought Al 2139.

Published by Elsevier B.V.

1. Introduction

Additive manufacturing (AM) is gaining widespread attention for its ability to produce high quality structural metallic components with greatly reduced cost and improved lead-time [1–3]. There are a wide variety of AM processes currently available though they all share similar attributes [4]. In general, an energy source is used to melt or consolidate a feedstock material in an incremental, layered manner using computer driven motion control. The result is a three-dimensional part created directly from a computer model without the need for molds or tooling.

For fusion-based AM processes, the applicability of an alloy is generally determined by its weldability. The easier an alloy is to weld, the better that alloy is for AM processing. Functional AM parts for a variety of applications have been demonstrated across most engineering alloy families [5]. As AM processing reduces the amount of material used and the amount of post processing (e.g. machining) required, most research and development has focused on expensive and difficult to work materials such as titanium and

nickel alloys [6]. Relatively little work has focused on aluminum alloys due largely to the fact that they are comparatively inexpensive and can be quickly and affordably machined to final dimensions. There are, however, certain application areas that can benefit from AM fabricated aluminum alloy structures. One such area is spaceflight structures where large component size and low production rates combine to make AM processed aluminum alloy parts particularly attractive.

Aluminum alloys present some unique constraints for AM processing. First, coupling of the heat source to the feedstock is challenging for laser-based AM processes due to the high spectral reflectance of aluminum in the most commonly employed range of laser wavelengths (e.g. Nd:YAG at 1064 nm) [7]. This high reflectance makes aluminum alloys better suited for electron beam-based AM processes. Second, certain aluminum alloy classes are not fusion weldable. The Al–Zn 7XXX series alloys, which are particularly important for aerospace applications, fall into this category [8]. The zinc in these alloys is highly volatile and creates excessive splatter and porosity when they are fusion welded. Additionally, the highest performance aluminum alloys usually rely on precipitation hardening, which can cause challenges related to micro and macro segregation during AM fabrication. Finally, many

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Table 1
Chemical analysis of EBF³ deposit using conventional wire and modified wire.

Element	Baseline wire	Deposit top	Deposit middle	Deposit bottom	HAZ under deposit	Base plate	Modified wire	Modified wire deposit
Al	93.8	94.1	93.7	93.6	93.5	93.7	92.3	93.2
Cu	4.7	4.7	5.0	5.1	5.1	4.7	5.0	5.3
Mg	0.52	0.11	0.16	0.19	0.47	0.47	1.69	0.52
Ag	0.38	0.59	0.55	0.53	0.42	0.55	0.50	0.50
Mn	0.36	0.36	0.42	0.38	0.28	0.32	0.33	0.31
Zr	< 0.002	0.002	0.002	0.003	0.024	0.025	0.011	0.009
Fe	0.062	0.054	0.060	0.064	0.054	0.053	0.075	0.080
Si	0.008	0.030	0.032	0.033	0.026	0.030	0.054	0.051
Ti	0.051	0.049	0.051	0.052	0.100	0.110	0.061	0.064

aluminum alloys contain solute additions with vastly different vapor pressures relative to the aluminum base (e.g. Mg in 5XXX series alloys, Li in 2X9X series alloys). AM processing, particularly electron beam-based processes requiring a vacuum chamber, can cause preferential vaporization of these solute species and significantly alter the resultant chemistry [9,10]. Despite these issues, aluminum alloy feedstock is currently commercially available in powder or welding wire form for a variety of AM processes. These feedstock materials are generally based on casting alloy compositions (e.g. Al–Si–Mg), as the AM process is essentially an incremental casting process.

Although aluminum casting alloys are well suited to AM processing, they suffer from lower mechanical performance relative to precipitation hardened wrought aluminum alloys [11]. In the wrought alloy family, the 2XXX series alloys are readily weldable and good candidates for AM processing. The alloy 2219 has been successfully deposited using additive manufacturing [12]. For many alloys in this series (and in others), peak strength can only be achieved using a T8 temper. This temper involves imparting mechanical deformation to the material (e.g. stretch) between the solution heat treatment and ageing steps to develop a dislocation network in the aluminum matrix that provides nucleation sites for precipitates. This mechanical working step is contrary to the near net shape capability provided by AM. Thus, it is desirable to have a high strength, precipitation hardened alloy that does not require a T8 temper to achieve peak properties.

A candidate alloy that can achieve high strength without the T8 temper is the aluminum alloy 2139. Alloy 2139 is an Al–Cu–Mg–Ag alloy that is strengthened by Al₂Cu precipitates [13–18]. The combined effect of Mg and Ag alters the orientation relationship between the precipitate and the aluminum matrix. In Al–Cu alloys without Mg and Ag, the Al₂Cu phase (identified as θ) forms on the $\langle 100 \rangle_{\alpha}$ planes and greatly benefits from the T8 temper, as homogeneous nucleation is very difficult [19–21]. In the alloy 2139, the Al₂Cu phase (identified as Ω) forms on the $\langle 111 \rangle_{\alpha}$ planes. The Ω phase is able to homogeneously nucleate off of Mg–Ag co-clusters and the alloy does not require mechanical deformation to achieve peak strength during ageing [22–28]. This paper examines AM processed Al 2139 alloy material with the goal of demonstrating peak aged properties comparable to wrought processed Al 2139.

2. Experimental procedure

Standard grade aluminum alloy 2139 wire (1.6 mm diameter) was used for the initial trials in this study. The base plate (152 mm width, 305 mm length, 6.4 mm thick) was also standard grade Al 2139 supplied in the T8 temper condition. The chemical compositions determined with direct current plasma emission spectroscopy are presented in Table 1. Additive manufacturing of test coupons was conducted using the electron beam freeform fabrication system (EBF³) located at the NASA Langley Research Center,

Hampton, Virginia, USA. The EBF³ process is conducted in a vacuum (approximately 1.6×10^{-3} Pa) and uses an electron beam heat source with wire feedstock to create additively manufactured structures. The beam was operated at sharp focus and rastered in a circular pattern approximately 3 mm in diameter with an incident power of 1350 W. The beam travel speed was 508 mm/min and the wire feed rate was 1270 mm/min. The initial deposited sample was five beads wide by five layers thick and 250 mm long.

Three slices of material were taken parallel to the deposition layers for chemical analysis. These slices were approximately 2.5 mm thick and were taken at the very bottom of the deposit, at the midpoint, and at the very top.

Solution annealing of the wrought baseplate and EBF³ deposited material was conducted at $800 \text{ K} \pm 3 \text{ K}$ for two hours. Isothermal ageing was conducted at $433 \text{ K} \pm 0.5 \text{ K}$ for 12, 18, 24, 36, 48, and 72 h.

Analytical transmission electron microscopy (TEM) of the baseplate and deposited material was carried out at the University of Canterbury using a Philips¹ CM200 operated at 200 keV and equipped with an Oxford² INCA x-sight EDS system and Gatan³ Orius™ SC 200 digital camera. Three-millimeter diameter discs were mechanically punched from 0.3 mm thick slices that were cut parallel to the deposition layers. The discs were subsequently ground to approximately 100 μm thick using 1200 and 4000 grit silicon carbide paper. Prior to ion milling, the specimens were electropolished in a solution of 6% (by volume) perchloric acid, 35% n-butyl alcohol, and 59% methanol at approximately 233 K, a voltage of 40 V and current of 25 mA using a E.A. Fishione⁴ automatic twin-jet electropolishing apparatus. Optimally, electropolishing was stopped just prior to perforation of the foil. Subsequent perforation and thinning to electron transparency was carried out using the Gatan Precision Ion Polishing system with an initial accelerating voltage of 3.5 keV and sputtering angle of 4.5°. Milling was halted once the perforation began to noticeably increase in size. The accelerating voltage was reduced to 1 keV to grow the perforation to approximately 0.5 mm in diameter. Final cleaning of the sample at 0.5 keV and sputtering angle of 4.5° was carried out for thirty minutes.

Microhardness traverses through the cross-sections of the base plate and the EBF³ deposits were taken to assess the compositional homogeneity (and hence the microstructural variability) along with the response of the material to precipitation hardening heat treatment. Testing was performed on slices taken parallel to the deposition layers, which were polished to a 0.06 μm finish. Two parallel rows of indents, spanning from the bottom of the baseplate to the top of the deposited material and spaced at 0.5–1 mm intervals, were made using a testing load of 300 g and dwell time

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