Available online at www.sciencedirect.com
ScienceDirect
Acta Materialia 58 (2010) 2014–2025
Acta MATERIALIA
www.elsevier.com/locate/actamat

Influence of high-temperature solution treatments on mechanical properties of an Al–Si–Cu aluminum alloy

Hiroyuki Toda [a,*], Takanori Nishimura [a], Kentaro Uesugi [b], Yoshio Suzuki [b], Masakazu Kobayashi [a]

[a] Department of Production Systems Engineering, Toyohashi University of Technology, Toyohashi, Aichi 441-8580, Japan
[b] Japan Synchrotron Radiation Research Institute, 1-1-1, Kouto, Mikazuki-cho, Sayo-gun, Hyogo 679-5198, Japan

Received 14 October 2009; received in revised form 11 November 2009; accepted 25 November 2009
Available online 23 December 2009

Abstract

It has been demonstrated that the strength of an Al–Si–Cu alloy is maximized by high-temperature solution treatment at 807 K, which is approximately 16 K higher than the ternary eutectic temperature. The dual-energy K-edge subtraction imaging technique has been employed to obtain the spatial distribution of copper and its change during its solution treatment in three dimensions quantitatively, providing interpretations of the improved mechanical properties in terms of age-hardenability and its spatial variation. It has been also confirmed that the occurrence of incipient local melting and the accompanying growth of micro pores adjacent to the melt regions lead to fractures caused by these defects. However, it can be inferred that the positive effects can outweigh the negative effects even above the eutectic temperature, thereby realizing the maximum strength at such relative high temperature levels.
© 2009 Acta Materialia Inc. Published by Elsevier Ltd. All rights reserved.

Keywords: Cast aluminum alloys; High-temperature solution heat treatment; X-ray tomography; K-edge subtraction imaging; Chemical concentration mapping

1. Introduction

When more than one soluble phase is formed during solidification, as found with Al–Si–Cu–Mg and Al–Zn–Cu–Mg alloys, it can be readily supposed that phases with lower melting points are melted locally during subsequent heating. In practical aluminum alloys, a uniform solute–atom concentration is not realized during solidification owing to non-equilibrium solidification [1]. Therefore, in most alloy systems, segregation also renders the alloys prone to the formation of eutectic compounds with a low melting point. This means that the homogenization and solution heat treatment temperatures can sometimes exceed the final solidification temperature at which eutectic compounds are formed during solidification. If the compounds do not dissolve before the final solidification temperature is reached during heating, local melting occurs during the heat treatments. In addition, other local melting mechanisms, which can occur even without segregation, have been reported for dilute Al–Mg–Si [2] and Al–Zn–Mg [3] alloys. For these reasons, industrial solution heat treatments are performed well below eutectic solidus temperatures. For example, a typical solution heat treatment temperature of 773–778 K has been specified for an Al–6Si–3.5Cu alloy, which is designated as A319 in ANSI or AC2B in JIS. The temperatures are roughly 50 K lower than the eutectic solidus temperature of the Al–6Si–3.5Cu alloy.

It might be reasonable and efficient to hold a material at a sufficiently high temperature long enough to allow constituents to dissolve into a solid solution during homogenization and solution heat treatments. Given that diffusion is accelerated as the temperature is increased, it would be natural to attempt to maximize the solution heat treatment

* Corresponding author. Tel.: +81 532446697; fax: +81 532446690.
E-mail address: toda@pse.tut.ac.jp (H. Toda).

1359-6454/$36.00 © 2009 Acta Materialia Inc. Published by Elsevier Ltd. All rights reserved.
doi:10.1016/j.actamat.2009.11.044

temperature. Shimizu et al. have demonstrated that toughness is maximized when an AC2B alloy is solution-heat-treated at 798 K [4]. They attributed the toughness enhancement to the spherodization of eutectic Si particles, the dissolution of non-equilibrium Cu-bearing compounds and the refinement of needle-shaped Fe-bearing compounds. Koga et al. have also shown that the tensile strength of an Al–7.2Si–0.37Mg alloy was maximized under a solution treatment condition of 843 K/0 ks [5]. They also attributed this to the spherodization and coarsening of eutectic silicon particles. Irinouchi et al. have investigated the effects of solution treatment temperature and time (813–843 K/0.9–18 ks) on the mechanical properties of an Al–6.7Si–0.31Mg alloy [6]. It has been clarified that a solution treatment at 833 K for 1.8 ks results in superior mechanical properties including 16% and 35% increases in ultimate tensile strength and 0.2% proof strength, respectively, together with a shorter solution treatment time (typically 1/10 of the standardized time). It has been confirmed with the X-ray tomography technique that micro pores and intermetallic compounds are grown during the solution treatment, thereby showing that the mechanical property degradation caused by further solution treatment might be attributable to these microstructural changes. Sokolowski et al. have attempted to increase the solution treatment temperature of A319 alloy to 788 K by introducing a preliminary solution treatment at 768 K [7].

It would be reasonable to assume that the positive effects of a high-temperature solution treatment are competent with its negative effects, and that the optimum condition is realized according to their trade-off relationship when the former exceeds the latter in terms of extent. For age-hardening aluminum alloys, the positive effects would consist of enhanced age-hardenability, homogenization and the suppression of the brittle fracture of Si and intermetallic compound particles due to refinement and morphological modification. On the other hand, the negative effects are incipient melting and the resultant microstructural defects, such as micro pores and brittle intermetallic compounds. Of these, the effects of the Si and intermetallic compound particles have been investigated relatively well, while the other factors remain poorly understood. Sampling a limited number of cross-sections using traditional metallographic observation might produce misleading information, especially for inhomogeneously and sparsely distributed microstructural features such as defects. To better understand the remaining negative effects, this study employed high-resolution three-dimensional (3-D) imaging [8–11]. Also, the combination of this approach with K-edge subtraction imaging enables the visualization of local age-hardenability [12,13], thus providing a unique opportunity to analyze the positive effects quantitatively. This is clearly an improvement on the procedures described in the available literature on high-temperature solution treatment, which allow only qualitative and indirect evaluations.

2. Experiments

2.1. Sample preparation

Aluminum with a purity of 99.99% and Al–26% Si and Al–40% Cu mother alloys were used to prepare an Al–Si–Cu pure ternary alloy. The alloy was melted in ambient laboratory air, kept at 983 K for 1.8 ks, and then poured directly into a steel mould about 153 × 270 × 100 mm in size that had been preheated at 473 K. Here, no modification was performed at all. The alloy cast had a chemical composition of 6.1 Si, 4.00 Cu, 0.09 Fe, 0.01 Ti, 0.0052 Ni, 0.0054 Ca and balance Al in mass%.

To determine the solution heat treatment temperatures, differential scanning calorimetry (DSC) was employed with a sample mass of approximately 0.03 g. A specimen was heated in pure argon between room temperature and 873 K with a scan rate of 9 K min^{-1} and a sampling interval of 1 s. In the DSC curve thus obtained, there were three endothermic peaks corresponding to ternary eutectic, unknown and binary eutectic reactions, respectively. In terms of the unknown second peak located at about 806 K, a much larger endotherm was observed in a practical AC2B alloy (Al–6.1Si–4.00Cu–0.17Fe–0.01Ti–0.01Mn–0.01Zn and balance Al in mass%) that has almost identical Si and Cu contents with more impurities. It is, therefore, reasonable to assume that the peak might be associated with the dissolution of some impurity-element-bearing phase. The extrapolated onset temperatures for the first and third peaks were located at 790.6 and 817.3 K, respectively. Four solution heat treatment temperatures of 773, 790, 807 and 824 K were selected considering the locations of the two peaks. 773 K is specified in the industrial standards. The samples were solution-heat-treated for 0–36 ks in a salt bath, quenched in iced water, and then artificially aged for 18 ks at 433 K in an oil bath to provide a T6 temper condition. The interval between the quenching and the onset of the aging treatment was controlled at 0.3 ks.

2.2. Tomographic imaging

X-ray tomography was performed using the undulator beamline BL20XU of the synchrotron radiation facility SPring-8. An experimental hutch is located about 245 m from the X-ray source in this beam line. First, an absorption spectrum was corrected in the 8.87–9.07 keV photon energy range around the Cu K-edge located at 8.977 keV as shown in Fig. 1. The monochromatized radiation was continuously monitored with ionization chambers. A 99.99% copper thin film with a thickness of 5 μm was used for this measurement. 8.938 and 9.038 keV were chosen so that the imaging was performed outside the positions of near-edge features seen in Fig. 1. 20 keV was also used because it provides the optimum condition in terms of S/N ratio. A monochromatic X-ray beam produced by a liquid-nitrogen-cooled Si (1 1 1) double crystal monochro-

Download English Version:

https://daneshyari.com/en/article/1447714

Download Persian Version:

https://daneshyari.com/article/1447714

Daneshyari.com