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The influence of Sc solute partitioning on the microalloying effect and mechanical properties of Al-Cu alloys with minor Sc addition

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ABSTRACT

A transmission electron microscope and an atom probe tomography were used to quantitatively characterize the microstructural evolution of Al-XCu alloys (X = 1.0, 1.5, and 2.5 wt%) with 0.3 wt% Sc addition. A dual solute alloying/microalloying effect on the microstructural evolution was demonstrated. On the one hand, the nucleation and coarsening of Al_3Sc dispersoids displayed a *Cu alloying effect*. By increasing the Cu content, both the Al_3Sc disperoid size and the volume fraction decreased after solution treatment. On the other hand, the precipitation of θ'-Al_2Cu strengthening particles during aging treatment was promoted by Sc segregation at the θ'/matrix interfaces, showing a notable *Sc microalloying effect*. The strongest interfacial Sc segregation was generated in the Al-2.5 wt%Cu-Sc alloy, resulting in the most promoted θ' precipitation. The Sc partitioning between Al_3Sc dispersoids and Sc segregation at the θ'/matrix interfaces, tailored by the Cu content, impacted the mechanical properties and deformation behavior at both room temperature and high temperature. The Al-2.5 wt%Cu-Sc alloy had a room temperature yield strength of approximately 2.2 times that in its Sc-free counterpart and approximately 1.8 times that in the Al-1.5 wt%Cu-Sc alloy, which is rationalized by strengthening models. In addition, the improvement in the high-temperature mechanical properties after Sc addition was discussed in terms of the Sc segregation-induced high coarsening resistance of θ' precipitates.

1. Introduction

Due to their attractive properties of low density and high strength, aluminum (Al) alloys are the primary material of choice for applications such as aircraft and automobiles, in which specific strength (strength-to-weight ratio) is a major design consideration. The high strength of heat-treatable aluminum alloys results from the precipitation of excess alloying elements from supersaturated solid solution to form fine second-phase particles (termed precipitates) that act as obstacles to moving dislocations [1,2]. The composition, manufacture technology, and heat treatment all affect the precipitation evolution and, concomitantly, the strengthening response [3–5]. Artificially controlling the precipitates, including size, number density, and distribution, is central for developing advanced Al alloys that could achieve the desired level of performance.

Microalloying is an effective approach to enhance the performance of Al alloys [6], in which the addition of minor microalloying elements will greatly improve the mechanical properties. Generally, microalloying elements can be divided into two types: one promotes the precipitation of hardening particles that are inherently precipitated in the Al alloys, and the other forms new precipitates that are additionally introduced. The former can be exemplified by some microalloying effects in Al-Cu alloys [6–14]. Trace elements of Sn, In, or Cd added into the Al-Cu alloys have been well known to suppress low-temperature aging and enhance age hardening at elevated temperatures by promoting the formation of θ'-Al_2Cu precipitates at the expense of GP zones and θ'' [7,8]. The possible mechanisms proposed for these microalloying effects include (i) the Sn (In or Cd) atoms segregate to the precipitate/

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matrix interface and reduce the interfacial energy [8–10], or (ii) facilitating heterogeneous nucleation of θ' either directly at Sn (In or Cd)-rich particles [11] or indirectly at the dislocation loops present in the as-quenched microstructure [12]. In a later study [13], it was clearly revealed via atom probe field ion microscopy that heterogeneous nucleation of θ'-Al_2Cu is the predominant microalloying mechanism in the Sn (In or Cd)-modified Al—Cu alloys. Similarly, this mechanism is also responsible for the recent findings [14,15] that minor additions of Si and Ge in Al-Cu increased the hardness and produced a fast hardening rate when aged at elevated temperature, where the first formed Si-Ge precipitates provide preferential nucleation sites for metastable θ'' and then for θ' phases.

A typical example of the second type of microalloying elements is Sc. The addition of minor Sc atoms has been found to significantly increase the strength and preserve the recrystallization of Al alloys [16–19]. The improvements are related to the formation of Al_3Sc particles, including Al_3Sc dispersoids and Al_3Sc precipitates [16]. Here, the Al_3Sc dispersoids specially refer to the Al_3Sc particles formed during high-temperature processing of the alloy, such as homogenization, hot-rolling or extrusion, whereas the Al_3Sc precipitates refer to those formed during the aging process after solution treatment. Al_3Sc dispersoids are very effective in preventing recrystallization up to temperatures of approximately 350–400 °C [20]. By adding Sc and Zr in combination, the recrystallization resistance of Al alloys can be further improved owing to the formation of more stable core-shell $Al_3(Sc_xZr_{1-x})$ dispersoids [21–23]. The Al_3Sc precipitates, coherent with the Al matrix, can induce a significant strength increase in the Al alloys [17,18,24]. Because the Al_3Sc particles (both dispersoids and precipitates) have excellent thermal stability and the diffusion coefficient of Sc solutes in Al is relatively low (several orders of magnitude less than that of Cu and Mg [16]), it is increasingly believed [25,26] that Sc microalloying can extend the applications of Al alloys favorably at elevated temperatures. In contrast, the conventional Al-Cu- and Al-Mg-Si-based series age-hardenable Al alloys usually have limited use at temperatures below 250 °C, because the coarsening of precipitates is very rapid at temperatures above 250 °C [27–29].

Most recently, some of the present authors experimentally found [30–33] that the Sc atoms can also promote the precipitation of θ' phases in Al-Cu alloys. On the one hand, the high Sc-vacancy interaction [34] facilitates the formation of Sc-vacancy pairs that would be arranged in clusters with vacancies and multiple solute atoms. The strong Sc-Cu interaction [35] then drives Sc-vacancy clusters to seize Cu atoms. The Cu/Sc/vacancy complex clusters are expected to act as preferential nucleation sites for θ' particles, accelerating the formation of high-density precipitates. On the other hand, the Sc atoms are apt to segregate at the coherent θ' precipitate/matrix interfaces [31,32]. The interfacial Sc segregation, similar to previous findings of Si and Mg atoms segregated at θ' precipitate/matrix interfaces in Al-Cu alloys [36,37] and Ag atoms segregated at Ω precipitate/matrix interfaces in Al-Cu-Mg-Ag alloys [38], reduces the interface energy and concomitantly reduces the driving force for precipitate coarsening. Solute segregation at precipitate/matrix interfaces is an important microalloying method to tailor the precipitation and improve the hardening response and the thermal stability of precipitates at elevated temperatures [39,40].

Because the minor Sc addition into the Al-Cu alloys has two microalloying effects, *i.e.*, promoting the precipitation of inherent θ' hardening particles in particular by interfacial Sc segregation and forming additional Al_3Sc hardening particles, some questions are naturally raised: (i) which is the dominant Sc microalloying effect in the Al-Cu-Sc alloys? (ii) how does the Sc partitioning between Al_3Sc particles and segregation at the θ' precipitate/matrix interfaces influence the final microalloying effect? and (iii) what are the possible easy approaches to tailor the Sc partitioning in the microalloyed Al-Cu-Sc alloys? Understanding these questions is necessary for developing a fundamental framework for the material design of advanced Al alloys microalloyed with Sc.

In this paper, systematical studies were performed to investigate the effect of Sc addition on microstructural evolution and tensile mechanical properties of Al-Cu alloys at room temperature (RT) and elevated temperature. Both nanosized Al_3Sc dispersoids and θ' precipitates were produced in the alloys after heat treatments, which allowed the two Sc microalloying effects to coexist. The Cu content was varied to change the Sc solute partitioning between the Al_3Sc dispersoids and Sc segregation at the θ' precipitate/matrix interfaces. Influences of the two particles (Al_3Sc dispersoids and θ' precipitates) on the RT mechanical properties were quantitatively assessed to demonstrate their separate microalloying effects and combined response. Attention was also paid to the microstructural evolution at elevated temperatures and the possible cooperation or competition between the two particles on the tensile mechanical properties at elevated temperatures.

2. Experimental procedures

2.1. Material preparation and heat treatments

Alloys with compositions of Al-X wt% Cu alloys (X = 1.0, 1.5, and 2.5, abbreviated Al-XCu alloys), Al-X wt% Cu-0.3 wt% Sc alloys (Al-XCu-Sc), and Al-0.3 wt% Sc alloys (Al-Sc) were melted and cast in an argon stream; 99.99 wt % pure Al, 99.99 wt % pure Cu, and Al-2.0 wt % Sc master alloy were used. The cast ingots were homogenized at 723 K for 5 h and hot extruded at 723 K into plates 14 mm in thickness and 60 mm in width. All the plates were subsequently solutionized at 853 K for 3 h, quenched in cold water, and finally aged at 523 K for 8 h. The maximum error of all temperature measurements in the present experiments was ±1 K.

2.2. Microstructural characterizations

Microstructures of the alloys were characterized by using transmission electron microscopy (TEM) and high-resolution TEM (HRTEM). TEM foils were prepared following standard electropolishing techniques for Al alloys. Quantitative measures of the number density and size of the second phase particles (Al_3Sc dispersoids and θ' precipitate) were reported as average values of more than 300 measurements. The volume fraction of the second phase particles was determined by employing a corrected projection method [5,41], in which the foil thickness of each captured region was obtained through convergent beam electron diffraction patterns [42]. Using the raw radius data of particles and the foil thickness in each area, the radius distributions were calculated after correction for the truncation effects based on a method by Crompton et al. [43]. Details about the measurements can be found in our previous publications [30,31,44,45].

To visibly reveal the microalloying mechanism at the atomic level, atom probe tomography (APT) was performed using an Imago Scientific Instruments 3000HR local electrode atom probe (LEAP). APT sample blanks with a square cross-sectional area of approximately $300 \times 300\ \mu m^2$ and a length of 1 cm were prepared by a combination of slicing and mechanical grinding. A two-step electropolishing procedure was used for making tips from these blanks [46,47]. A 10 vol% perchloric acid in methanol solution was used for coarse polishing, and the final polishing was performed using a solution of 2 vol% perchloric acid in butoxyethanol. APT data collection using the electrical pulsing mode was performed at a specimen temperature of 30 ± 0.3 K, with a voltage pulse fraction