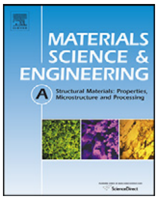
Materials Science and Engineering A 528 (2011) 4104–4110

Contents lists available at ScienceDirect

Materials Science and Engineering A

journal homepage: www.elsevier.com/locate/msea



Alloy design for Al addition on microstructure and mechanical properties of $Ni_3(Si,Ti)$ alloy

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ARTICLE INFO

Article history:
Received 28 June 2010
Received in revised form 20 January 2011
Accepted 21 January 2011
Available online 23 February 2011

Keywords:
Mechanical characterization
Intermetallics
Casting
Thermomechanical processing
Tensile properties

ABSTRACT

The effect of Al addition on the microstructure and tensile properties of $Ni_3(Si,Ti)$ alloys with an $L1_2$ ordered structure, which were fabricated through thermomechanical processing from arc-melted ingots, was investigated. Al was added to a $Ni_3(Si,Ti)$ alloy by using two methods such that Al substituted for (1) only Ti and (2) both Ni and Ti along a $Ni_3(Si,Ti)$–Ni_3Al pseudo-binary line. In the case of the alloys prepared by the former method, the addition of more than 4 at.% Al resulted in a two-phase microstructure consisting of disordered fcc Ni solid solution dispersions in the $L1_2$ matrix, while in the case of the alloys prepared by the latter method, the addition of 4 at.% Al retained the $L1_2$ single-phase microstructure. In the case of the 4 at.% Al-added alloys, the room-temperature tensile properties were similar and independent of the alloying methods, whereas the high-temperature yield stress was higher in the alloys prepared by the latter method than in the case of the alloys prepared by the former method. These results suggest that a single-phase microstructure consisting of an entire $L1_2$ structure is favorable for obtaining high-temperature tensile properties.

1. Introduction

An $L1_2$-type $Ni_3(Si,Ti)$ alloy, which has been developed by adding Ti to Ni_3Si [1], possesses many attractive properties such as those of high-temperature structural materials. For example, the $Ni_3(Si,Ti)$ alloy shows a positive temperature dependence of yield strength and exhibits higher strength and peak temperature in the strength versus temperature curve than Ni_3Si [1]. Further, the relatively low density ($\sim$8 g/mm^3) and the good corrosion resistance of the $Ni_3(Si,Ti)$ alloy are favorable properties for structural materials [2]. In addition, a high tensile ductility over a wide range of temperatures due to the prevention of the propensity for an intergranular fracture is a notable feature of the $Ni_3(Si,Ti)$ alloy [1]. Such high tensile ductility of the $Ni_3(Si,Ti)$ alloy is noteworthy because most intermetallic alloys have limited ductility, particularly at low temperatures. Recently, heavily cold-rolled thin sheets and foils of the $Ni_3(Si,Ti)$ alloy were fabricated from conventional polycrystalline ingots via a repeated warm-rolling and annealing process [3]. The $Ni_3(Si,Ti)$ thin sheet annealed at an intermediate temperature (e.g., around 900 K) exhibited an extremely high tensile strength and yield strength (more than 2 GPa) with a reasonable fracture strain at room temperature and was superior to the commercial nickel alloys and stainless steels below 873 K [3]. However, the tensile strength and the elongation of the annealed $Ni_3(Si,Ti)$ thin sheets were found to rapidly decrease at temperatures beyond 873 K with an increase in the temperature, and therefore, the improvement of the high-temperature mechanical properties is considerably desired.

Thus far, the improvement of high-temperature tensile properties for the $Ni_3(Si,Ti)$ alloy has been attempted by means of an alloying technique [4–6]. Transition metals, Hf, Nb, Cr, Fe and Mn, were added to the $Ni_3(Si,Ti)$ alloy at 1 at.% level below their solubility limits [4]. However, their addition was less effective in improving the high-temperature tensile properties of the $Ni_3(Si,Ti)$ alloy. Further, 2 at.% of Al, Cr, and Mo metals was added to the $Ni_3(Si,Ti)$ alloy [6]. The 2 at.% addition of Al was below the solubility limit in the $Ni_3(Si,Ti)$ matrix; therefore, the new alloy retained the $L1_2$ single-phase microstructure. On the other hand, the 2 at.% addition of Cr was beyond the solubility limit in the $Ni_3(Si,Ti)$ matrix and consequently resulted in a two-phase microstructure consisting of disordered fcc Ni solid solution dispersions in the $L1_2$ matrix. It was found that the addition of Al as well as Cr was slightly effective in improving both high-temperature ductility and oxidation resistance [6].

In the present study, we selected Al as an alloying element for the $Ni_3(Si,Ti)$ alloy because addition of Al is expected to yield a number of microstructural, mechanical, engineering, and economic advantages such as low density, low cost, the use of a ubiquitous element, high oxidation resistance, high corrosion resistance, and large solubility in the $L1_2$ matrix. In order to make Al largely soluble in the $Ni_3(Si,Ti)$ matrix, various alloying procedures were

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doi:10.1016/j.msea.2011.01.083

Table 1
Chemical compositions, microstructures, and hardness of the alloys used in this study.

Group	Alloy	Ni (at.%)	Si (at.%)	Ti (at.%)	Al (at.%)	B (wt ppm)	Microstructure	Hardness (HV)
Group A	Base	79.5	11.0	9.5	–	50	$L1_2$	385
	#A2	79.5	11.0	7.5	2.0	50	$L1_2$	376
	#A4	79.5	11.0	5.5	4.0	50	$L1_2 + Ni_{SS}$	375
	#A6	79.5	11.0	3.5	6.0	50	$L1_2 + Ni_{SS}$	367
	#A8	79.5	11.0	1.5	8.0	50	$L1_2 + Ni_{SS}$	372
Group B	#B4	78.9	11.0	6.1	4.0	500	$L1_2$	389
	#B8	78.3	11.0	2.7	8.0	500	$L1_2 + Ni_{SS}$	391

taken into consideration. In this study, the microstructure, cold workability, and high-temperature tensile properties of the Al-added $Ni_3(Si,Ti)$ alloys were investigated, and the alloy composition and the microstructure preferable for high-temperature structural materials were suggested.

2. Experimental procedure

The base composition of the $Ni_3(Si,Ti)$ alloy used in this study was 79.5 at.% Ni, 11.0 at.% Si, and 9.5 at.% Ti doped with 50 wt ppm B. Boron doping of the $Ni_3(Si,Ti)$ alloy is necessary to suppress the intergranular fracture due to environmental embrittlement at room temperature [7]. The quaternary element Al was added to the base alloy composition by using two methods: (1) Al substituted only for Ti and (2) Al substituted both for Ni and Ti. The samples prepared by the two methods are hereafter referred to as groups A and B, respectively (Table 1). In the case of group A, the addition of 2, 4, 6 and 8 at.% Al content was designed to reduce the amount of Ti that was harmful for the oxidation resistance of the $Ni_3(Si,Ti)$ alloy [8]. In the case of group B, the addition of 4 and 8 at.% Al content was designed to reduce the amount of Ni and Ti, forming a continuous $L1_2$ solid solution along the pseudo-binary $Ni_3(Si,Ti)$–Ni_3Al line, as shown in Fig. 1. In the alloys of group B, an increased amount of 500 wt ppm B was added because a higher boron content was required in Ni_3Al than in $Ni_3(Si,Ti)$ to suppress the intergranular fracture [9].

Alloy ingots were prepared by non-consumable arc melting in an argon gas atmosphere on a copper hearth. All the ingots were homogenized at 1323 K for 48 h in a vacuum and then cut into several plates with a thickness of approximately 10 mm using an electro-discharge machine (EDM). These plates were warm-rolled at 573 K and then annealed at 1273 K for 5 h in a vacuum. This procedure was repeated until the thickness of the plate became 2 mm. The sheets were finally annealed at 1273 K for 1 h in a vacuum before cold rolling. Cold rolling was conducted without intermediate annealing up to 90% reduction in thickness. Reduction in sheet thickness per rolling pass was approximately 0.1 mm both for warm and cold rolling. The thickness of the obtained cold-rolled sheets was approximately 0.2 mm. Microstructural observations were carried out by using a scanning electron microscope (SEM; JEOL, JSM-5600 operating at 15 kV) and a transmission electron microscope (TEM; JEOL, JEM-2000FX operating at 200 kV). Further, X-ray diffraction (XRD; RIGAKU, RINT2500HK) was performed to determine constituent phases and second-phase dispersions by using Ni-filtrated Cu Kα radiation at an accelerated voltage of 30 kV mostly with a scan speed of 1°. In the XRD measurement, bulk materials of the alloys were used as the specimens.

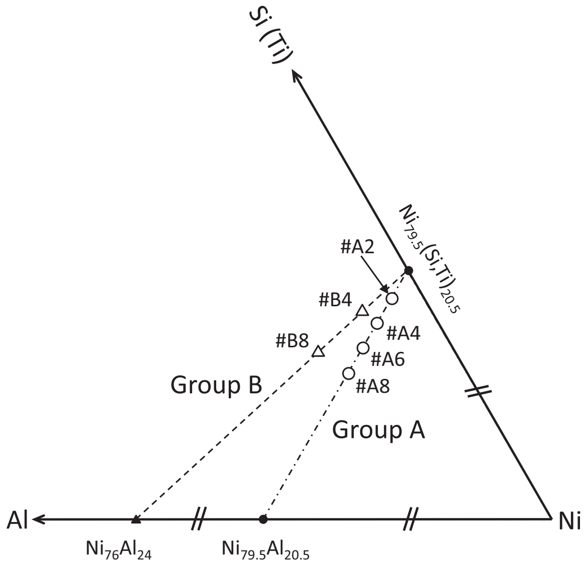

Fig. 1. Alloy compositions used in this study. Note that Al contents were changed along two pseudo-binary lines, $Ni_{79.5}(Si,Ti)_{20.5}$–$Ni_{79.5}Al_{20.5}$ (group A) and $Ni_{79.5}(Si,Ti)_{20.5}$–$Ni_{76}Al_{24}$ (group B).

Dog-bone-shaped tensile test pieces with a length of 10 mm, width of 4 mm, and thickness of 0.2 mm were prepared by using the EDM. Room-temperature tensile tests were carried out in air by using the cold-rolled and fully recrystallized (1173 K; 1 h) specimens. High-temperature tensile tests were conducted in a vacuum in a temperature range of room temperature to 973 K by using the fully recrystallized specimens. A strain rate of $8.4 \times 10^{-5}\,s^{-1}$ was used in both room-temperature and high-temperature tensile tests. After the tensile tests, the fracture surfaces of the test pieces were examined by using an SEM.

3. Results and discussions

3.1. Microstructure

Fig. 2 shows the SEM microstructures of the alloys in group A (i.e., alloys #A2, #A4, #A6 and #A8) after homogenization at 1323 K. Alloy #A2 as well as the base alloy (not shown here) showed an $L1_2$ single-phase microstructure, indicating that 2 at.% Al is soluble in the $L1_2$ matrix. On the other hand, the other alloys (i.e., alloys #A4, #A6 and #A8) exhibited a two-phase microstructure containing second-phase dispersions. At Al contents above 2 at.%, the volume fraction of the second-phase dispersions increased with an increase in the Al content. Bimodal microstructures were more or less observed in the alloys with a two-phase microstructure: eutectic-like two-phase microstructures consisting of $L1_2$ and Al (fcc) phases were mixed with a featureless $L1_2$ single-phase region. This eutectic-like microstructural feature may be explained such that the Al phase with Ni-rich content is formed as a primary phase and present in the dendritic core region during solidification and remains as a two-phase microstructure after homogenization.

Fig. 3 shows the XRD profiles of alloy #A4 with a two-phase microstructure, together with the alloys with a single-phase microstructure (i.e., the base alloy and alloy #B4 (see later)). The XRD profiles of all alloys shown in Fig. 3 were basically identical to each other, although the relative peak intensities of each diffraction were slightly different among the shown alloys because of the crystallographic textures of these alloys. The observed reflection