Mechanical Properties of an $\text{Al}_{88.5}\text{Ni}_8\text{Mm}_{3.5}$ (Mm: Misch Metal) Alloy Produced by Extrusion of Atomized Amorphous plus fcc-Al Phase Powders

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A bulky $\text{Al}_{88.5}\text{Ni}_8\text{Mm}_{3.5}$ alloy was produced by extrusion of atomized powders with a mixed structure of amorphous and fcc-Al phases at an extrusion ratio of 10 and at 633 K and was found to exhibit high tensile strengths (σ_t) of 940 MPa at room temperature (RT), 700 MPa at 423 K and 520 MPa at 473 K. Similarly, the yield strength ($\sigma_{0.2}$) is also as high as 845 MPa at RT and 420 MPa at 473 K. The high values of σ_t and $\sigma_{0.2}$ remain unchanged even after annealing for 100 h at these testing temperatures. The fatigue limit after the cycles of 10^7 is also as high as 330 MPa at 293 K and 220 MPa at 473 K. The fatigue limit at RT is also 1.2 times as high as the highest value for conventional Al-based alloys. The high mechanical strengths and their good heat resistance are presumably due to the formation of a finely mixed structure consisting of Al_3Ni and $\text{Al}_{11}(\text{La}, \text{Ce})_3$ particles with a size of $0.05\text{ }\mu\text{m}$ homogeneously embedded in an Al matrix with a grain size of 0.1 to $0.2\text{ }\mu\text{m}$. In addition, the Al-Ni-Mm alloy also exhibits a low coefficient of thermal expansion and good wear resistance because of the homogeneous dispersion of a large amount of the intermetallic compounds. The simultaneous achievement of good mechanical and physical properties allows an expectation for practical use of the new type Al-based alloy in various fields.

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I. Introduction

Since the finding of amorphous alloys with good bending ductility in the Al-Ni-Si system⁽¹⁾, a number of Al-based amorphous alloys with good ductility have been found in a rapidly solidified state in Al-Ln⁽²⁾, Al- $\text{M}_1\text{-M}_2$ ⁽³⁾ and Al-Ln- M_1 ⁽⁴⁾⁽⁵⁾ (Ln=lanthanide metal, M_1 and M_2 : transition metal) systems. Furthermore, it has been reported that these Al-based amorphous alloys exhibit high tensile strength above 1000 MPa even at Al-rich compositions of 84 to 86 at%⁽⁶⁾ and the highest strength reaches as high as 1250 MPa⁽⁷⁾. It has subsequently been found⁽⁸⁾⁽⁹⁾ that the homogeneous dispersion of nanoscale fcc-Al particles into an Al-based amorphous matrix causes a significant increase in tensile strength to 1560 MPa which is about 1.5 times as high as that of the corresponding amorphous single phase alloy. However, these data were obtained from the ribbon samples with a cross section of about $0.02 \times 1.0\text{ mm}^2$ produced by the melt spinning technique. In order to develop Al-based alloys with a higher specific strength in practical use by utilizing the amorphous phase with high strength and good ductility, it is essential to accumulate the information on the microstructure and mechanical properties of Al-based alloys in a bulk form produced by extrusion or forging of the amorphous alloy powders. In a series of studies on the microstructure and mechanical properties of Al-bas-

ed alloys in bulk form, we have found that high tensile strength above 900 MPa and high fatigue strength above 300 MPa are simultaneously obtained for a bulky Al-Ni-Mm alloy produced by extrusion of the alloy powders containing an amorphous phase as a main phase at temperatures above crystallization temperature. This paper is intended to present the microstructure, mechanical properties including tensile strength, fatigue strength and wear resistance, and coefficient of thermal expansion for the bulky $\text{Al}_{88.5}\text{Ni}_8\text{Mm}_{3.5}$ alloy.

II. Experimental Procedure

An alloy with nominal composition $\text{Al}_{88.5}\text{Ni}_8\text{Mm}_{3.5}$ (at%) was used because the alloy composition was located at the most Al-rich side in the composition range where an amorphous phase in the Al-Ni-Mm and Al-Ni-Ln systems is formed by melt spinning⁽⁴⁾⁽¹⁰⁾. A mixture of pure Al, Ni and Mm metals was melted by arc melting in a purified argon atmosphere. The Mm metal is composed of 52 mass% Ce, 29 mass% La, 14 mass% Nd and 5 mass% Pr. Three kg of the pre-alloyed ingot was charged into the atomization crucible and remelted under an argon atmosphere. The pre-alloyed ingot was atomized from a temperature of 1473 K with helium gas at an applied pressure of 5.88 MPa in a high-pressure gas atomization equipment. The powders were allowed to cool in the inert atmosphere but were collected and

screened in a nitrogen atmosphere. Fine powders below $25\text{ }\mu\text{m}$ in size were put in a cylindrical Al can with an inner diameter of 24 mm. The tap density of the powder was 60%. The powder in the can was outgassed, precompact to 90% density at 673 K by uniaxial hot-pressing and sealed by welding. The canned powder was heated in a resistance furnace at an extrusion temperature of 633 K for 60 s and then extruded to an area reduction of 10:1 in a 150-ton extrusion press. The as-quenched and extruded phases of the powders were examined by X-ray diffraction, optical microscopy and differential scanning calorimetry. The tensile strength and elongation were measured in a temperature range from room temperature to 573 K using an Instron-type testing machine under an argon atmosphere. The sample for the tensile test has a cylindrical shape with the gauge dimensions of $3\text{ mm} \times 50\text{ mm}$. The fatigue strength was measured for the cylindrical sample with the gauge dimensions of $8\text{ mm} \times 15.4\text{ mm}$ at room temperature and 473 K in air using a rotating bending machine. The coefficient of thermal expansion was measured with a dilatometer at a heating rate of 0.33 K/s in the temperature range of 293 to 623 K. The relative wear resistance against plain carbon steel (S45C) was evaluated at room temperature in air by the pin-on disc method.

III. Results

The atomized powder had a fine particle size and 50 wt% of the particles had diameters less than $26\text{ }\mu\text{m}$. Powder particles in this size range have previously been reported⁽¹¹⁾ to exhibit the cooling rates above 105 K/s. The powder was spherical and was relatively free from satellites. The internal porosity was also very low. Figure 1(a) and (b) show the etched cross-sections of two size fractions of powder particles. No distinct contrast can be seen in the optical micrograph (a) of the powder with a particle size fraction below $26\text{ }\mu\text{m}$ and the microstructure of the $<26\text{ }\mu\text{m}$ powder is below the resolution limit of the microscope.

Figure 2 shows the change with powder size in the X-ray diffraction pattern of the atomized Al-Ni-Mm powders. The structure is composed of amorphous and fcc-Al phases for the powders smaller than $26\text{ }\mu\text{m}$ in diameter and Al + Al_3Ni + $\text{Al}_{11}(\text{La}, \text{Ce})_3$ for the powders above $26\text{ }\mu\text{m}$. It has previously been reported⁽¹²⁾ that the melt-spun structure of the $\text{Al}_{88.5}\text{Ni}_8\text{Mm}_{3.5}$ alloy is an amorphous single phase, when the rotation speed of wheel is 41.8 m/s (4000 rpm), and changes to a mixed structure consisting of amorphous and fcc-Al phases with decreasing rotation speed of wheel. Furthermore, the precipitation of the Al phase at the Al-rich composition has been presumed to occur during cooling after the formation of the amorphous phase. Consequently, the cooling rate in the present gas atomization is not high enough to suppress the precipitation of the fcc-Al phase in a rather low temperature range during cooling, though the present equipment has a high cooling capability leading to the formation of the amorphous phase. The volume frac-

Fig. 1 Etched cross sections of as-atomized $\text{Al}_{88.5}\text{Ni}_8\text{Mm}_{3.5}$ powder particles: (a) $<26\text{ }\mu\text{m}$ size fraction, (b) $38\text{--}45\text{ }\mu\text{m}$ size fraction.

Fig. 2 X-ray diffraction patterns of as-atomized $\text{Al}_{88.5}\text{Ni}_8\text{Mm}_{3.5}$ powder particles: (a) $<26\text{ }\mu\text{m}$ size fraction, (b) $\geq 26\text{ }\mu\text{m}$ size fraction.

tions of the amorphous and Al phases in the atomized powders were evaluated by measuring the exothermic amount due to crystallization in comparison with that for an amorphous single phase produced by melt spinning. As shown in Fig. 3, the amorphous alloy crystallizes through two stages of the precipitation of fcc-Al phase from the amorphous matrix in the temperature range of 380 to 500 K, followed by the simultaneous precipitation

Fig. 3 Differential scanning calorimetric curves of as-atomized $\text{Al}_{88.5}\text{Ni}_8\text{Mm}_{3.5}$ powder particles: (a) $<26\ \mu\text{m}$ size fraction, (b) $\geq 26\ \mu\text{m}$ size fraction.

of $\text{Al} + \text{Al}_3\text{Ni} + \text{Al}_{11}(\text{La}, \text{Ce})_3$ from the remaining amorphous phase in the range of 600 to 650 K. No distinct change in the second exothermic peak corresponding to the precipitation of the compounds was seen between the powders with a particle size fraction below $26\ \mu\text{m}$ and the melt-spun ribbon. The exothermic amount of the first-exothermic peak due to the precipitation of fcc-Al phase was compared with that for the melt-spun $\text{Al}_{88.5}\text{Ni}_8\text{Mm}_{3.5}$ ribbon with an amorphous single phase and the volume fraction of the fcc-Al phase was evaluated to be 17% for the powders below $26\ \mu\text{m}$.

Figure 4 shows the optical micrographs of a bulky $\text{Al}_{88.5}\text{Ni}_8\text{Mm}_{3.5}$ alloy extruded at an extrusion ratio of 10 and at 633 K. The featureless contrast indicates that the microstructure after the hot extrusion treatment is also below the resolution limit of the microscope. The X-ray diffraction pattern taken from the extruded bulk shown in Fig. 5 indicates that the structure consists of Al, Al_3Ni and $\text{Al}_{11}(\text{La}, \text{Ce})_3$. In order to clarify the microstructure of the extruded bulk, TEM observation was made for the thin foil made from the bulk material. As shown in Fig. 6, the structure consists of equiaxed fine grains with grain sizes of 0.1 to $0.2\ \mu\text{m}$ for the Al matrix and about $0.05\ \mu\text{m}$ for the compounds. Judging from the previous result⁽¹³⁾ that the grain size of a metastable Al-base solid solution in a rapidly solidified $\text{Al}_{95}\text{Ce}_3\text{Fe}_2$ alloy is about $2\ \mu\text{m}$, it is concluded that the present process consisting of hot extrusion of the amorphous plus Al phase powders is very useful for the formation of the ultrafine mixed structure consisting of fine Al_3Ni and $\text{Al}_{11}(\text{La}, \text{Ce})_3$ particles embedded in the Al matrix.

Figures 7 and 8 show the temperature dependence of tensile yield strength ($\sigma_{0.2}$) and tensile fracture strength (σ_f), respectively, for the bulky Al-Ni-Mm alloy extruded at 633 K, along with the data⁽¹⁴⁾ for the A7075 alloy. The

Fig. 4 Etched sections of an as-extruded $\text{Al}_{88.5}\text{Ni}_8\text{Mm}_{3.5}$ alloy: (a) section perpendicular to the extrusion direction, (b) section parallel to the extrusion direction.

Fig. 5 X-ray diffraction pattern of an as-extruded $\text{Al}_{88.5}\text{Ni}_8\text{Mm}_{3.5}$ alloy.

open and closed circles in the figure represent the data after annealing at each temperature for 1 h and 100 h, respectively. $\sigma_{0.2}$ and σ_f of the extruded Al-Ni-Mm bulk are as high as 845 and 940 MPa, respectively, at room temperature which are much higher than those ($\sigma_{0.2}=590\ \text{MPa}$, $\sigma_f=700\ \text{MPa}$) for the A7075 alloy. Although $\sigma_{0.2}$ and σ_f after 1 h annealing decrease monotonously with increasing temperature, they keep rather high values of 550 and 700 MPa at 423 K and 420 and 520 MPa at 473 K. It should be further noticed that

Fig. 6 Bright-field electron micrograph of an as-extruded $\text{Al}_{88.5}\text{Ni}_8\text{Mm}_{3.5}$ alloy.

Fig. 7 Temperature dependence of tensile yield strength ($\sigma_{0.2}$) for as-extruded $\text{Al}_{88.5}\text{Ni}_8\text{Mm}_{3.5}$ alloy annealed for 1 and 100 h at each testing temperature. The data for the A7075 alloy are also shown for comparison.

no appreciable decrease in $\sigma_{0.2}$ and σ_f is seen even after annealing for 100 h, though the $\sigma_{0.2}$ and σ_f of the A7075 alloy decrease by about 5% at 423 K and by about 45% at 473 K. Consequently, in addition to the high tensile strengths, the Al-Ni-Mm alloy also has high thermal stability of mechanical strengths. The Young's modulus of the extruded Al-Ni-Mm alloy is also as high as 91 GPa which is about 30% higher than that⁽¹⁵⁾ of conventional Al-based alloys. On the other hand, a plastic elongation is about 1.0% at room temperature. The achievement of the extremely high tensile strength is presumably because of the formation of the finely mixed structure consisting of a large amount of Al_3Ni and $\text{Al}_{11}(\text{La}, \text{Ce})_3$ compounds homogeneously embedded in the fcc Al matrix which cannot be obtained by conventional thermo-mechanical treatments. Furthermore, the

Fig. 8 Temperature dependence of tensile fracture strength (σ_f) for as-extruded $\text{Al}_{88.5}\text{Ni}_8\text{Mm}_{3.5}$ alloy annealed for 1 and 100 h at each testing temperature. The data for the A7075 alloy are also shown for comparison.

good elevated temperature strength for the present alloy is presumably because the strengthening mechanism is due to the dispersion hardening of the intermetallic compounds, being different from the result⁽¹⁴⁾ that the strengthening mechanism for the A7075 alloy is due to the age-hardening mechanism. The details of the strengthening mechanism for the present alloy will be described in the following section.

Fatigue strength of the extruded $\text{Al}_{88.5}\text{Ni}_8\text{Mm}_{3.5}$ alloy at room temperature and 473 K was measured by the rotating bending test method. Figure 9 shows the maximum stress-cycle fatigue curve at room temperature. The fatigue failure under a maximum stress below 330 MPa does not occur even after the cycles of 3.6×10^7 , indicating that the fatigue limit lies in the vicinity of 330 MPa. Similarly, no fatigue failure at 473 K is observed in the applied stress range below 225 MPa and hence the fatigue limit at 473 K is determined to be about 220 MPa. In order to compare the present fatigue limit with the data for other Al-based alloys in fully annealed (O), as-quenched (T4) and age-hardened (T6) states, the fatigue limit is plotted in the relation between fatigue limit and tensile strength for Al-based alloys shown in Fig. 10. As marked with a solid square, the fatigue limit of the present Al-Ni-Mm alloy is about 1.2 times as high as the highest fatigue limit⁽¹⁶⁾ for newly developed Al-based alloys made from rapidly solidified powders. Also, the fatigue ratio defined by the ratio of fatigue limit to $\sigma_{0.2}$ is 0.39 at room temperature and 0.52 at 473 K which are considerably higher than that (≈ 0.35 at room temperature)⁽¹⁷⁾ for conventional Al-based alloys. A similar high fatigue ratio has also been reported⁽¹⁸⁾ for the Al-25Si-2024 alloy produced by hot-forging of rapidly solidified powders. Accordingly, the sequent processing of hot extrusion or hot forging of rapidly solidified Al-based alloy powders seems to be useful for the production of Al-based bulky materials with improved fatigue strengths, though accumulation of a similar kind of data is required

Fig. 9 Relation between maximum stress and number of cycles up to failure at room temperature for an extruded $\text{Al}_{88.5}\text{Ni}_8\text{Mm}_{3.5}$ alloy.

Fig. 10 Relation between fatigue limit after the cycles of 10^7 and tensile fracture strength for an as-extruded $\text{Al}_{88.5}\text{Ni}_8\text{Mm}_{3.5}$ alloy. The data for conventional Al-based alloys are also shown for comparison.

hereafter. At any event, it is concluded that both the tensile strengths and the fatigue limit for the extruded Al-Ni-Mm alloy are much superior to those for the conventional Al-based alloys.

It is expected that a low coefficient of thermal expansion and a high wear resistance are obtained for the Al-Ni-Mm alloy because of the precipitation of a large amount of Al_3Ni and $\text{Al}_{11}(\text{La}, \text{Ce})_3$ compounds. Tables 1 and 2 summarize the coefficient of thermal expansion (α) in the temperature range of 423 to 473 K and the relative wear resistance against S45C, respectively, along with the data⁽¹⁵⁾ for the A6061, A5056 and Al-17 mass% Si alloys. The α value is about 20% smaller than that for the conventional Al-Si-Mg and Al-Mg base alloys. The wear loss is also about 25% smaller than that for the Al-Si alloy with the highest wear resistance in the conventional Al base alloys. It is thus concluded that the Al-Ni-Mm alloy has the high wear resistance as well as the low coefficient of thermal expansion.

The extruded Al-Ni-Mm alloy has already been used⁽¹⁹⁾ as machine parts which are required to have

Table 1 Coefficient of thermal expansion of extruded Al-Ni-Ln bulks and conventional aluminum alloys.

Alloy	Thermal expansion (K^{-1})
$\text{Al}_{88.5}\text{Ni}_8\text{Mm}_{3.5}$	20.2×10^{-6}
$\text{Al}_{88.5}\text{Ni}_8\text{Y}_{3.5}$	19.6×10^{-6}
A6061	24.4×10^{-6}
A5056	25.4×10^{-6}

Table 2 Wear resistance of the extruded Al-Ni-Ln bulks and a conventional Al-17 mass% Si alloy against plain carbon steel (S45C).

Alloy	Wear loss ($10^{-7} \text{ mm}^2/\text{kg}$)
$\text{Al}_{88.5}\text{Ni}_8\text{Mm}_{3.5}$	4.0
$\text{Al}_{85}\text{Ni}_5\text{Y}_{10}$	3.2
Al-17 mass%Si	5.2

simultaneously high σ_f , high fatigue limit, low coefficient of thermal expansion and high wear resistance, as exemplified in Fig. 11. The machine parts were produced by the process of extrusion, forging and mechanical polishing. In addition to such machine parts, the new Al-based alloys produced by extrusion of nonequilibrium phase powders consisting mainly of an amorphous phase are expected to be used in various fields where the simultaneous achievement of high strength with light weight, high heat-resistance of strength, high fatigue strength, low coefficient of thermal expansion and wear resistance is required.

IV. Discussion

It is shown in Section III that the yield strength ($\sigma_{0.2}$) and tensile fracture strength (σ_f) of the extruded $\text{Al}_{88.5}\text{Ni}_8\text{Mm}_{3.5}$ alloy are as high as 845 and 940 MPa, respectively, at room temperature and 420 and 520 MPa, respectively, at 473 K. It is also demonstrated that the structure of the extruded alloy consists of homogeneously dispersed Al_3Ni and $\text{Al}_{11}(\text{La}, \text{Ce})_3$ compounds with a size of about 50 nm embedded in an fcc-Al matrix with a grain size of 100 to 200 nm. Accordingly, it is interpreted that the high mechanical strengths are due to a combined effect of grain size refinement and dispersion strengthening. We shall investigate first the contribution of dispersion strengthening ($\Delta\tau$) to the yield strength. The contribution for the present mixed structure can be evaluated using the following Orowan equation⁽²⁰⁾:

$$\Delta\tau = \frac{0.8Gb}{2\pi(1-\nu)^{1/2}} \cdot \frac{\ln(2r_s/r_0)}{\lambda_s - 2r_s} \quad (1)$$

Here G , b and ν are the shear modulus, Burger's vector and Poisson's ratio for Al, respectively. r_s is $2r/3$ (r =particle radius) and r_0 is the core dislocation radius for Al. λ_s is expressed by $1.25r(2\pi/3f)^{1/2}$ (f =volume fraction of dispersoids). The G , b , ν and r_0 for the Al matrix are 27 GPa, 2.86×10^{-10} m, 0.345 and 2.86×10^{-10} m⁽²¹⁾, respectively. Additionally, the r and f

Fig. 11 External appearance of machine parts made from $\text{Al}_{88.5}\text{Ni}_8\text{Mm}_{3.5}$ powders. The machine parts were produced by extrusion, forging and mechanical polishing.

values were measured to be about 25 nm and 0.34, respectively, for the present mixed structure. The $\Delta\tau$ value evaluated from eq. (1) using these values is 129 MPa and hence the contribution to the tensile yield strength ($\Delta\sigma$) is evaluated to be 395 MPa which is about 47% as high as the yield strength of the Al-Ni-Mm alloy. The significant difference between $\Delta\sigma$ and the yield strength is presumed to originate from the contribution of the grain size refinement.

The contribution of the grain size refinement on the yield strength ($\sigma_{0.2}$) is known to be evaluated through the Hall-Petch equation⁽²²⁾⁽²³⁾:

$$\sigma_{0.2} = \sigma_0 + kd^{-n} \quad (2)$$

Here, σ_0 is the yield strength of a single grain due to all strengthening mechanisms except grain boundary contributions⁽²⁴⁾ and k is the constant relating to how effective the grain boundaries are in increasing yield strength. d is the grain size and n is the exponent in the range 0.5 to 1.0 and has usually assumed to be 0.5⁽²⁴⁾. There are no experimental data on an exact solute composition in the Al matrix for the extruded Al-Ni-Mm alloy. However, considering that the extrusion temperature is 633 K corresponding to the temperature just above the offset temperature of crystallization, a small amount of Ni and/or Mm elements seems to be left in the Al matrix. No data on the k and n values are obtained for the present Al matrix. According to a number of previous data on conventional Al-based crystalline alloys, the k value is 75 $\text{MPa}\mu\text{m}^{1/2}$ for commercially pure Al⁽²⁵⁾, 120 $\text{MPa}\mu\text{m}^{1/2}$ for optimally aged 7075 alloy⁽²⁶⁾, 125 $\text{MPa}\mu\text{m}^{1/2}$ for Al-Cu alloys⁽²⁶⁾ and 220 $\text{MPa}\mu\text{m}^{1/2}$ for underaged 7091 alloy⁽²⁷⁾. Based on these previous data, the k value for the present Al matrix seems to be in the range of 75 to 220 $\text{MPa}\mu\text{m}^{1/2}$. When the n and k values are assumed to be 0.5 and 135 $\text{MPa}\mu\text{m}^{1/2}$, respectively, the increase in yield strength resulting from the contribution of the grain size refinement (kd^{-n}) is evaluated to be 426 MPa for $d=100$ nm and 302 MPa for $d=200$ nm. The use of the k value corresponding to a mean value of the above-described values appears to be appropriate because the grain

boundaries containing the fine intermetallic compounds can act as an effective barrier in increasing yield strength. Accordingly, the increase in yield strength by the contribution of both the grain size refinement and dispersion strengthening is estimated to be 821 MPa for $d=100$ nm [$=395 (\Delta\sigma) + 426 \text{ MPa } (kd^{-1/2})$] and 697 MPa for $d=200$ nm. These values are slightly smaller than the experimentally measured $\sigma_{0.2}$ (845 MPa) for the extruded Al-Ni-Mm alloy. The slight difference is presumably due to the neglect of the solid solution strengthening resulting from the survival of Ni and Mm elements in the Al matrix and the strain hardening caused by the allowance of 0.2% elongation. However, the approximate agreement between the experimentally determined and the theoretically estimated $\sigma_{0.2}$ values allows us to conclude that the high mechanical strengths for the present Al-Ni-Mm alloy are due to the combined effect of the dispersion strengthening and grain size refinement resulting from the homogeneous dispersion of the fine intermetallic compounds. Furthermore, the contribution to the increase in yield strength is judged to be almost equivalent for both the mechanisms. It is therefore concluded that the most dominant factor for the achievement of the high mechanical strength for the present mixed alloy is due to the extremely fine grain size of Al matrix as well as the homogeneous dispersion of the small intermetallic compounds which can be obtained only through the use of the amorphous precursor.

V. Summary

The microstructure and mechanical properties were examined for a bulky $\text{Al}_{88.5}\text{Ni}_8\text{Mm}_{3.5}$ alloy produced by extrusion of the atomized powders with a particle size fraction below $26 \mu\text{m}$ at an extrusion ratio of 10 and at 633 K. The extruded alloy consists of a mixed structure in which fine Al_3Ni and $\text{Al}_{11}(\text{La}, \text{Ce})_3$ compounds with a particle size of about 50 nm are dispersed homogeneously in an fcc-Al matrix with a grain size of about 100 to 200 nm. The tensile fracture strength (σ_f) of the extruded alloy is as high as 940 MPa at room temperature and keeps high values of 700 MPa at 423 K and 520 MPa at 473 K. The high-strength level is also recognized for the tensile yield strength ($\sigma_{0.2}$) and the values at room temperature and 473 K are 845 and 420 MPa, respectively. The high σ_f and $\sigma_{0.2}$ values remain constant even after annealing for 100 h at 423 and 473 K. The fatigue limit after the cycles of 10^7 is also as high as 330 MPa at room temperature and 220 MPa at 473 K. Based on the discussion of the mechanism for the achievement of the high mechanical strengths, it is interpreted that the strengthening mechanism is contributed to the combined effect of grain size refinement and dispersion hardening resulting from the formation of the extremely fine mixed structure and the contribution is almost equivalent for both the mechanisms. In addition, the homogeneous dispersion of the intermetallic compounds was also found to cause the low coefficient of thermal expansion as well as good wear resistance. The simultaneous achievement of high

mechanical strengths and good physical properties allows us to expect the development of the present alloy as a practical material.

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