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(54) **High-strength, wear-resistant aluminum alloy.**

(57) A high-strength, wear-resistant aluminum alloy consisting of an Al-Si-based alloy consisting of Al as a main metal element and, added thereto, additive elements and Si element, characterized in that the mean crystal grain size of a matrix of Al is 40 to 1000 nm, the mean particle size of particles of a stable phase or a metastable phase of various intermetallic compounds formed from Al and the additive elements including Si and/or various intermetallic compounds formed from the additive elements themselves is 10 to 800 nm, the size of elemental Si particles is 10 μ m or less, the intermetallic compound particles are distributed in a volume fraction of 18 to 35 % in the Al matrix, and the elemental Si particles are distributed in a volume fraction of 15 to 40% in the Al matrix. The aluminum alloy has an improved strength at room temperature and a large toughness and can maintain the properties inherent in a material produced by the rapid solidification process even when it undergoes a thermal influence during working.

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BACKGROUND OF THE INVENTION

1. Field of the Invention

5 The present invention relates to an Al-Si-based alloy produced by the rapid solidification process and excellent in strength and toughness as well as in wear resistance.

2. Description of the Prior Art

10 An aluminum-based alloy having a high strength and a high heat resistance has heretofore been produced by the liquid quenching process or the like. In particular, an aluminum alloy produced by the liquid quenching process disclosed in Japanese Patent Laid-Open No. 275732/1989 is in an amorphous or finely crystalline form and an excellent alloy having a high strength, a high heat resistance and a high corrosion resistance.

15 Although the aluminum-based alloy is an alloy having a high strength, a heat resistance and a high corrosion resistance, and is excellent in the workability as a high-strength material, there is a room for an improvement in the toughness as a material required to have a high toughness. In general, an alloy produced by the rapid solidification process is liable to undergo a thermal influence during working, and the thermal influence causes excellent properties such as strength to be rapidly lost. This is true for the above-
20 described alloy, so that there is a room for an improvement in this respect as well. Further, there is a room for a further improvement in the wear resistance.

SUMMARY OF THE INVENTION

25 In view of the above-described problem, the present inventors have paid attention to the volume fraction of various intermetallic compounds formed from a main metal element and additive elements and/or from the additive elements themselves present together in a matrix of Al and the volume fraction of Si particles, and an object of the present invention is to provide a high-strength, wear-resistant aluminum alloy which has an improved strength at room temperature and a high toughness, can maintain the properties inherent in a
30 material produced by the rapid solidification process even when it undergoes a thermal influence during working, and has an excellent wear resistance.

In order to solve the above-described problem, the present invention provides a high-strength, wear-resistant aluminum alloy consisting of an Al-Si-based alloy consisting of Al as a main metal element and, added thereto, additive elements and Si element characterized in that the mean crystal grain size of a
35 matrix of Al is 40 to 1000 nm, the mean size of particles of a stable phase or a metastable phase of various intermetallic compounds formed from Al and the additive elements including Si and/or various intermetallic compounds formed from the additive elements themselves is 10 to 800 nm, the size of elemental Si particles is 10 μ m or less, the intermetallic compound particles are distributed in a volume fraction of 18 to 35% in the Al matrix, and the elemental Si particles are distributed in a volume fraction of 15 to 40% in the
40 Al matrix.

BRIEF DESCRIPTION OF THE DRAWINGS

45 FIGS. 1 to 4 are each a graph showing the relationship between the volume fraction of fine particles in an alloy having a composition described in the Examples of the present invention and the tensile strength.

FIGS. 5 and 6 are each an explanatory view of the dimension of a sample subjected to the wear test in Example 3.

FIG. 7 is a graph showing the results of the above-described wear test.

DETAILED DESCRIPTION OF THE PREFERRED EMBODIMENTS

50 In the above-described alloy, the mean crystal grain size of the Al matrix is that of a matrix consisting of Al or a supersaturated solid solution of Al. The mean crystal grain size of the Al matrix is limited to 40 to 1000 nm because when it is less than 40 nm, the ductility is unsatisfactory although the strength is high, while when it exceeds 1000 nm, it becomes impossible to prepare a high-strength alloy due to a rapid lowering in the strength.

The mean particle size of particles of the intermetallic compound is the mean particle size of particles of a stable phase or a metastable phase of various intermetallic compounds formed from the above-

described matrix element and other alloying elements and/or various intermetallic compounds formed from other alloying elements themselves. The mean particle size is limited to 10 to 800 nm because when it is outside this range, the intermetallic compounds do not function as strengthening elements for the main metal element matrix. Specifically, when the mean particle size is less than 10 nm, the intermetallic compounds do not contribute to strengthening of the matrix. In this case, when the intermetallic compounds are excessively dissolved in the solid solution form in the matrix, there is a possibility that the material might become brittle. On the other hand, when the mean particle size exceeds 800 nm, the particle size becomes so large that the strength cannot be maintained and, at the same time, the intermetallic compounds do not function as strengthening elements.

When the mean crystal grain size of Al element and the mean particle size are in the above-described respective ranges, it becomes possible to improve the Young's modulus, high-temperature strength and fatigue strength. In order to attain the above-described object, it is necessary that particles of various intermetallic compounds should be dispersed and present together in a matrix of Al element.

The volume fraction of the particles of the intermetallic compounds to be incorporated into the Al matrix is limited to 18 to 35% because, in the Al-Si-base alloy, when the volume fraction is less than 18%, not only the strengthening of the strength at room temperature and the rigidity are unsatisfactory but also the wear resistance is unsatisfactory, whereas when the volume fraction exceeds 35%, the ductility at room temperature is so poor that the working of the resultant material is unsatisfactory, which makes it impossible to attain the object of the present invention.

The volume fraction of the elemental Si particles is limited to 15 to 40% because when it is less than 15%, the improvement in the wear resistance is unsatisfactory, whereas, when it exceeds 40%, the material becomes brittle. When the Si content is controlled so as to fall within this range, the coefficient of thermal expansion can be regulated. This leads to an advantage in respect of design when the resultant material is used as a light-weight, high-strength sliding member instead of iron steel materials.

The above-described additive element preferably consists of a first additive element consisting of at least one element selected from among rare earth elements (including Y), Zr and Ti and a second additive element consisting of at least one element selected from among transition elements exclusive of the elements belonging to the first additive element, Li and Mg. Further, Mm (mischmetal) which is a composite comprising La and Ce as major elements and further rare earth (lanthanoid) elements exclusive of La and Ce and unavoidable impurities (Si, Fe, Mg, Al, etc.) also belong to the rare earth element of the first additive element. The first additive element contributes to the stabilization of a microcrystalline structure, while the second additive element contributes to an improvement in the stabilization of the mechanical properties (hardness, strength, rigidity, heat resistance, etc.) and the stabilization of a microcrystalline structure.

Specific examples of the above-described aluminum alloy include (I) an alloy represented by the general formula $Al_{100-a-b-c}X_aM_bSi_c$ wherein X represents at least one element selected from among La, Ce, Mm, Zr, Ti and Y; M represents at least one metal selected from Ni and Co; and a, b and c are each an atomic %, provided that $0.3 \leq a \leq 4$, $3 \leq b \leq 8$ and $15 \leq c \leq 40$; and (II) an alloy represented by the general formula $Al_{100-a-b-c-d}X_aM_bQ_dSi_c$ wherein X represents at least one element selected from among La, Ce, Mm, Zr, Ti and Y; M represents at least one element selected from Ni and Co; Q represents at least one element selected from among Mg, Cu and Zn; and a, b, c and d are each an atomic %, provided that $0.3 \leq a \leq 4$, $3 \leq b \leq 8$, $15 \leq c \leq 40$ and $0.1 \leq d \leq 1.5$.

The values of a, b, c and d in the above-described general formulae are limited to 0.3 to 4%, 3 to 8%, 15 to 40% and 0.1 to 1.5%, respectively, in terms of atomic % because when they are in the above-described respective ranges, the strength of the alloys at a temperature in the range of from room temperature to 300 °C is higher than that of the conventional (commercially available) high-strength aluminum alloy and the alloys have a ductility sufficient to withstand practical fabrication. Further, the wear resistance can be enhanced mainly by the precipitated fine Si particles and intermetallic compounds.

Since an increase in the Si content does not affect the workability, warm working becomes possible and is less liable to cause the coarsening of the structure. Further, the coefficient of thermal expansion can be controlled by regulating the Si content, so that when the material is used, for example, as a sliding member, the coefficient of thermal expansion can be easily conformed to that of the mating material.

The X element is at least one element selected from among La, Ce, Mm, Ti and Zr. It has a small diffusibility in the Al matrix, forms various metastable or stable intermetallic compounds and contributes to the stabilization of a microcrystalline structure.

In the above general formulae, the M element is at least one element selected from Ni and Co. It has a relatively small diffusibility in the Al matrix. When it is finely dispersed as intermetallic compounds in the Al matrix, it has the effect of strengthening the matrix and, at the same time, regulating the growth of crystal grains. Specifically, it contributes to a remarkable improvement in the hardness, strength and rigidity of the

alloy and stabilizes the fine crystal phase not only at room temperature but also at high temperature, so that the heat resistance can be imparted to the alloy.

The combination of the above-described elements enables the ductility necessary for the existing working to be imparted.

5 The Q element is at least one element selected from among Mg, Cu and Zn, and combines with Al to form compounds or combines with another Q element to form compounds, thus strengthening the matrix and contributing to an improvement in the heat resistance. Further, the specific strength and specific modulus are improved.

10 The Si element particles are dispersed in the fine elemental form having a size of 10 μm or less and has the effect of enhancing the wear resistance and hardness of the alloy. The regulation of the amount (content) of dispersion of the Si particles enables the coefficient of thermal expansion of the alloy to be regulated.

15 In the alloys represented by the above-described general formulae as well, for the reasons set out above, the mean crystal grain size of a matrix of Al or a supersaturated solid solution of Al should be 40 to 1000 nm, the mean particle size of a stable phase or a metastable phase of various intermetallic compounds formed from the above-described matrix element and other alloying elements and/or various intermetallic compounds formed from other alloying elements themselves should be 10 to 800 nm, the volume fraction of the intermetallic compound particles incorporated into the Al matrix should be 18 to 35%, and the volume fraction of the elemental Si particles having a size of 10 μm or less should be 15 to 40%

20 Further, in the alloys represented by the general formulae, the volume fraction of the Al-X type compound is preferably 1 to 18%. When the volume fraction is less than 1%, the matrix is coarsened and the strength is lowered. On the other hand, when the volume fraction exceeds 18%, the ductility is extremely lowered. The volume fraction of the Al-M type compound is preferably 17 to 29%. When the volume fraction is less than 17%, the strength at room temperature lowers, while when the volume fraction exceeds 29%, the ductility lowers.

25 In particular, in the alloys represented by the above-described general formulae, preferred examples of the dispersed Al-M type compound include Al_3Ni and Al_3CO_2 and preferred examples of the Al-X type compound include $\text{Ce}_3\text{Al}_{11}$, Al_4Ce , $\text{La}_3\text{Al}_{11}$, $\text{Mm}_3\text{Al}_{11}$, Al_3Ti and Al_3Zr . In both Al_3Ti and Al_3Zr , a compound of a metastable phase has a higher effect of contribution to a fine dispersion.

30 The alloy of the present invention can be directly prepared in the form of a thin ribbon, powder, fine wire, etc., by the liquid quenching process such as the single-roller melt-spinning process, the gas or water atomization process or the in-rotating-water melt-spinning process through a proper regulation of the cooling rate of the ordinary solidification process to 10^7 to 10^2 K/sec. Further, it can be directly prepared in the form of a foil by vapor phase deposition means such as sputtering, ion beam sputtering, vapor deposition or the like. Similarly, the powder can be prepared also by the mechanical alloying process (MA process).

35 A consolidated material of the alloy according to the present invention can be directly prepared by two-stage solidification means as described in Japanese Patent Laid-Open No. 253525/1991 through a proper control of the cooling rate. When the alloy is prepared in the form of a consolidated material, a material in the form of a thin ribbon, powder, fine wire, foil or the like prepared by the above-described process may be compacted and worked by the conventional plastic deforming means.

40 In this case, a powder, flake or the like having a fine structure prepared by rapid solidification or the like is desirably subjected to plastic deformation at a temperature of preferably 50 to 500 °C, still preferably 320 to 440 °C. The heat history in this case provides a more suitable crystalline structure.

45 In the above-described process, when the mechanical alloying process is used, an oxide, nitride or the like is formed. A material prepared by compacting and consolidating the above material has a superior strength at high temperature.

The present invention will now be described in more detail by referring to the following Examples.

Example 1

50 An aluminum-based alloy powder ($\text{Al}_{\text{bal}}\text{Ni}_{4-5}\text{Ce}_{1-3}\text{Si}_{20-37}$) having a predetermined composition was prepared by a gas atomizing apparatus. The aluminum-based alloy powder thus produced was filled into a metallic capsule, and a billet for extrusion was prepared with degassing. This billet was extruded at a temperature of 320 to 440 °C on an extruder to prepare samples.

55 The relationship between the mechanical properties (tensile strength) at room temperature and 200 °C and the volume fraction of the precipitated intermetallic compounds was examined for individual samples (materials consolidated by extrusion) produced under the above-described production conditions. In this case, the examination was conducted with the volume fraction of the fine elemental Si particles being fixed

to 20%.

The results are shown in FIG. 1.

The volume fraction of the above-described intermetallic compound and the volume fraction of the Si particles were measured by subjecting the resultant consolidated material to an image analysis under a TEM. The intermetallic compounds precipitated in the above-described samples were mainly Al_3Ni , $\text{Ce}_3\text{Al}_{11}$, etc. Observation under a TEM revealed that the above-described samples each comprised a matrix consisting of aluminum or a supersaturated solid solution of aluminum and having a mean crystal grain size of 40 to 1000 nm, that particles consisting of a stable phase or a metastable phase of various intermetallic compounds formed from the matrix element and other alloying elements and/or various intermetallic compounds formed from other alloying elements themselves were homogeneously distributed in the matrix, that the mean particle size of the particles of the intermetallic compounds was 10 to 800 nm, and that the size of the fine elemental Si particles homogeneously distributed in the matrix was 10 μm or less.

As is apparent from FIG. 1, the strength at room temperature and the strength at 200 °C rapidly increased when the volume fraction exceeded 18% and rapidly decreased when the volume fraction exceeded 35%.

The ductility of the sample at room temperature decreased with an increasing volume fraction of the intermetallic compound particles and became lower than the lower limit (2%) of the ductility necessary for general working when the volume fraction exceeded 35%.

The relationship between the mechanical properties (tensile strength) at room temperature and 200 °C and the volume fraction of the precipitated fine elemental Si particles was examined for individual samples produced under the above-described production conditions. In this case, the examination was conducted with the volume fraction of the intermetallic compounds being fixed to 25% for Al_3Ni and 7% for $\text{Ce}_3\text{Al}_{11}$.

The results are shown in FIG. 2.

As is apparent from FIG. 2, the strength at room temperature and the strength at a high temperature of 200 °C decreased with an increasing amount of the fine elemental Si particles. Further, the ductility at room temperature was examined to find out that when the volume fraction exceeds 40%, the ductility becomes lower than the lower limit (2%) of the ductility necessary for general working. The effect of wear resistance was significant when the volume fraction was 15% or more.

Changes in the strength at room temperature and the strength at a high temperature of 200 °C with the variation in the volume fraction of individual intermetallic compound particles were examined for Al_3Ni and $\text{Ce}_3\text{Al}_{11}$ as main intermetallic compounds in individual samples produced under the above-described production conditions. In this case, the examination was conducted with the volume fraction of the fine elemental Si particles being fixed to 20%.

The results are shown in FIGS. 3 and 4.

In FIG. 3, a change in the strength with the variation in the volume fraction of the Al_3Ni intermetallic compound particles was examined through the use of a sample having a composition of $\text{Al}_{\text{bal}}\text{Ni}_{15}\text{Ce}_{1.2}\text{Si}_{20}$ with the volume fraction of the $\text{Ce}_3\text{Al}_{11}$ intermetallic compound particles being fixed to 5%. In FIG. 4, a change in the strength with the variation in the volume fraction of the $\text{Ce}_3\text{Al}_{11}$ intermetallic compound particles was examined through the use of a sample having a composition of $\text{Al}_{\text{bal}}\text{Ni}_{4.5-5}\text{Ce}_{1.5-3}\text{Si}_{20}$ with the volume fraction of the Al_3Ni intermetallic compound particles being fixed to 20%.

As is apparent from FIG. 3, the strength at room temperature and the strength at a high temperature of 200 °C rapidly increased when the volume fraction of the Al_3Ni intermetallic compound particles exceeded 17% and rapidly lowered when the volume fraction exceeded 29%.

As is apparent from FIG. 4, the strength at room temperature and the strength at a high temperature of 200 °C rapidly increased when the volume fraction of the $\text{Ce}_3\text{Al}_{11}$ intermetallic compound particles exceeded 1% and rapidly lowered when the volume fraction exceeded 18%. The ductility at room temperature of the above-described samples became lower than the lower limit (2%) of the ductility necessary for general working when the volume fraction exceeded 29% for the Al_3Ni intermetallic compound and exceeded 18% for the $\text{Ce}_3\text{Al}_{11}$ intermetallic compound.

Example 2

Extruded materials (consolidated materials) consisting of various ingredients specified in Table 1 were prepared in the same manner as that of Example 1 to examine the mechanical properties (tensile strength) of these materials at room temperature. In the table, precipitated main intermetallic compound phases and their volume fractions are specified.

The results are given in Table 1.

Table 1

Comp. Ex.	Alloy Composition (at.%)	Intermetallic compound phase	Vol.% of compound	Vol.% of Si	Tensile strength (MPa)
Comp. Ex.	Al ₃ Ni ₄ Si ₁₂	Al ₃ Ni	15		307
Invention Ex. 1	Al ₃ Ni ₇ Mm _{1.3} Si ₁₆	Al ₃ Ni, Mm ₃ Al ₁₁	29	18	762
Invention Ex. 2	Al ₃ Ni ₅ Ce _{1.5} Si ₂₀	Al ₃ Ni, Ce ₃ Al ₁₁	28	22	684
Invention Ex. 3	Al ₃ Ni _{5.5} Zr _{1.5} Si ₂₅	Al ₃ Ni, Al ₃ Zr	26	28	647
Invention Ex. 4	Al ₃ Ni _{5.5} Ti _{1.5} Cu _{0.5} Si ₃₀	Al ₃ Ni, Al ₃ Ti	23	33	622
Invention Ex. 5	Al ₃ Ni ₄ Mm _{0.4} Si ₃₅	Al ₃ Ni, Mm ₃ Al ₁₁	17	36	658
Invention Ex. 6	Al ₃ Ni ₅ La _{1.8} Si ₂₅	Al ₃ Ni, La ₃ Al ₁₁	26	26	689
Invention Ex. 7	Al ₃ Ni _{5.5} Mm _{1.2} Zr _{0.5} Si ₂₀	Al ₃ Ni, Mm ₃ Al ₁₁ , Al ₃ Zr	27	21	523
Invention Ex. 8	Al ₃ Ni ₄ Mm ₃ Y ₁ Mg ₁ Si ₁₈	Al ₃ Ni, Mm ₃ Al ₁₁ , Al ₃ Y	21	18	726
Invention Ex. 9	Al ₃ Ni ₆ Mm _{0.4} Zn ₁ Mg ₁ Si ₂₅	Al ₃ Ni, Mm ₃ Al ₁₁ , Al ₃ Zr	24	26	623
Invention Ex. 10	Al ₃ Ni ₅ Mm _{1.4} Mg _{1.5} Si ₃₀	Al ₃ Ni, Mm ₃ Al ₁₁ , Al ₃ Zr	26	31	589

Table 1 (continued)

	Alloy Composition (at.%)	Intermetallic compound phase	Vol.% of compound	Vol.% of Si	Tensile strength (MPa)
Invention Ex. 11	Al ₉ Co ₄ Ce _{1.2} Si ₂₅	Al ₉ Co ₂ , Ce ₃ Al ₁₁	25	26	576
Invention Ex. 12	Al ₉ Co ₃ Mm _{1.5} Si ₂₀	Al ₉ Co ₂ , Mm ₃ Al ₁₁	28	22	591
Invention Ex. 13	Al ₉ Co _{3.5} Ti _{1.6} Si ₁₈	Al ₉ Co ₂ , Al ₃ Ti	32	20	611
Invention Ex. 14	Al ₉ Co ₄ Zr _{2.8} Cu _{1.2} Si ₂₀	Al ₉ Co ₂ , Al ₃ Zr	21	20	627
Invention Ex. 15	Al ₉ Co ₃ Mm _{1.2} Si ₁₇	Al ₉ Co ₂ , Mm ₃ Al ₁₁	26	18	599
Invention Ex. 16	Al ₉ Co ₄ La _{0.5} Mg _{0.5} Si ₂₅	Al ₉ Co ₂ , La ₃ Al ₁₁	23	27	533
Invention Ex. 17	Al ₉ Co ₄ Mm _{1.2} Zr ₁ Si ₃₀	Al ₉ Co ₂ , Mm ₃ Al ₁₁ , Al ₃ Zr	31	32	524
Invention Ex. 18	Al ₉ Co ₄ Mm ₂ Y ₁ Si ₂₀	Al ₉ Co ₂ , Mm ₃ Al ₁₁ , Al ₃ Y	22	22	521
Invention Ex. 19	Al ₉ Co ₄ Mm _{0.5} Zr ₁ Mg ₁ Si ₂₄	Al ₉ Co ₂ , Mm ₃ Al ₁₁ , Al ₃ Zr	27	25	518
Invention Ex. 20	Al ₉ Co ₄ Mm ₁ Zn ₁ Mg ₁ Si ₃₀	Al ₉ Co ₂ , Mm ₃ Al ₁₁ , Al ₃ Zr	26	31	548

As is apparent from Table 1, the extruded materials (consolidated materials) of the present invention have an excellent tensile strength at room temperature.

All the extruded materials listed in the table exhibited an elongation exceeding the lower limit (2%) necessary for general working.

Also, the alloy of this Example comprised a matrix consisting of aluminum or a supersaturated solid solution of aluminum and having a mean crystal grain size of 40 to 1000 nm, and particles consisting of a stable phase or a metastable phase of various intermetallic compound formed from the matrix element and other alloying element and/or various intermetallic compound formed from other alloying elements themselves were homogeneously distributed in the matrix. The mean particle size of the intermetallic compound was 10 to 800 nm and the size of the fine elemental Si particles homogeneously distributed in the matrix was 10 μm or less.

Example 3

Alloys of Invention Examples 1, 2, 3 and 4 specified in Table 1, Comparative Example 1 ($\text{Al}_{\text{bal}}\text{Si}_5\text{Fe}_3\text{Ce}_{0.2}$) and an alloy having a composition corresponding to A390 were atomized into powders (mean particle size: 15 μm) by a high-pressure gas atomizer. These powders were packed into a copper container and a cap, subjected to vacuum deaeration (1×10^{-5} Torr) and pressed into billets.

These billets were put in the container of an extruder, subjected to warm extrusion at a temperature of 320 to 440 °C to give extruded round bars. Among the extruded round bars, those of the Invention Examples had the same structure as that described in Example 2.

Each extruded round bar were worked into a shape shown in FIG. 5 and brought into contact with a mating material disk (S45C) shown in FIG. 6 to conduct a test by the pin-on-disk method under the conditions of a pressure of 20 kgf/cm² and a speed of 0.84 m/sec. The dimensions of the test samples are shown in millimeter units in FIGS. 5 and 6.

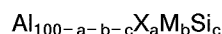
The results are shown in FIG. 7.

In the case of the A390 aluminum alloy, known as a wear-resistant aluminum alloy, and the rapidly solidified alloy of $\text{Al}_{\text{bal}}\text{Si}_5\text{Fe}_3\text{Ce}_{0.2}$ in Comparative Example 1, both the test material and the mating material underwent significant wear. By contrast, in the case of the materials of the present invention, both the material per se and the mating material were less liable to wear loss. Further, it has been found that the materials of the present invention have good compatibility with the mating material.

As is apparent from the foregoing description, the high-strength, wear-resistant aluminum alloy according to the present invention is excellent in the strength at room temperature and high temperature as well as in the toughness. Further, it can maintain excellent properties inherent in a material produced by the rapid solidification process even when it undergoes a thermal influence during working. Further, the alloy of the present invention not only is excellent in the wear resistance but also can reduce the wear loss of the mating material.

Claims

1. A high-strength, wear-resistant aluminum alloy consisting of an Al-Si-based alloy consisting of Al as a main metal element and, added thereto, additive elements and Si element, characterized in that the mean crystal grain size of a matrix of Al is 40 to 1000 nm, the mean particle size of particles of a stable phase or a metastable phase of various intermetallic compounds formed from Al and the additive elements including Si and/or various intermetallic compounds formed from the additive elements themselves is 10 to 800 nm, the size of elemental Si particles is 10 μm or less, the intermetallic compound particles are distributed in a volume fraction of 18 to 35 % in the Al matrix, and the elemental Si particles are distributed in a volume fraction of 15 to 40% in the Al matrix.
2. A high-strength, wear-resistant aluminum alloy according to claim 1, wherein the additive elements consist of a first additive element consisting of at least one element selected from among rare earth elements (including Y), Zr and Ti and a second additive element consisting of at least one element selected from among transition elements exclusive of the elements belonging to the first additive element, Li and Mg.
3. A high-strength, wear-resistant aluminum alloy according to claim 1 or 2, wherein said high-strength, wear-resistant aluminum alloy is represented by the general formula



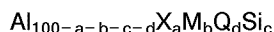
wherein X represents at least one element selected from among La, Ce, Mm, Zr, Ti and Y;

M represents at least one element selected from Ni and Co;

a, b and c are each an atomic %, provided that $0.3 \leq a \leq 4$, $3 \leq b \leq 8$ and $15 \leq c \leq 40$;
the volume fraction of an Al-X type intermetallic compound is 1 to 18%; and
the volume fraction of an Al-M type intermetallic compound is 17 to 29%.

5 4. A high-strength, wear-resistant aluminum alloy according to claim 3, wherein the Al-X type intermetallic compound comprises $\text{Ce}_3\text{Al}_{11}$, Al_4Ce , $\text{Mm}_3\text{Al}_{11}$, Al_3Ti and/or Al_3Zr and the Al-M type intermetallic compound comprises Al_3Ni and/or Al_3Co_2 .

10 5. A high-strength, wear-resistant aluminum alloy according to claim 1 or 2, wherein said high-strength, wear-resistant aluminum alloy is represented by the general formula



wherein X represents at least one element selected from among La, Ce, Mm, Zr, Ti and Y;

15 M represents at least one element selected from Ni and Co;

Q represents at least one element selected from among Mg, Cu and Zn;

a, b, c and d are each an atomic %, provided that $0.3 \leq a \leq 4$, $3 \leq b \leq 8$, $15 \leq c \leq 40$ and $0.1 \leq d \leq 1.5$;

20 the volume fraction of an Al-X type intermetallic compound is 1 to 18%; and

the volume fraction of an Al-M type intermetallic compound is 17 to 29%.

6. A high-strength, wear-resistant aluminum alloy according to claim 5, wherein the Al-X type intermetallic compound comprises $\text{Ce}_3\text{Al}_{11}$, Al_4Ce , $\text{Mm}_3\text{Al}_{11}$, Al_3Ti and/or Al_3Zr and the Al-M type intermetallic compound comprises Al_3Ni and/or Al_3Co_2 .

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FIG. 1

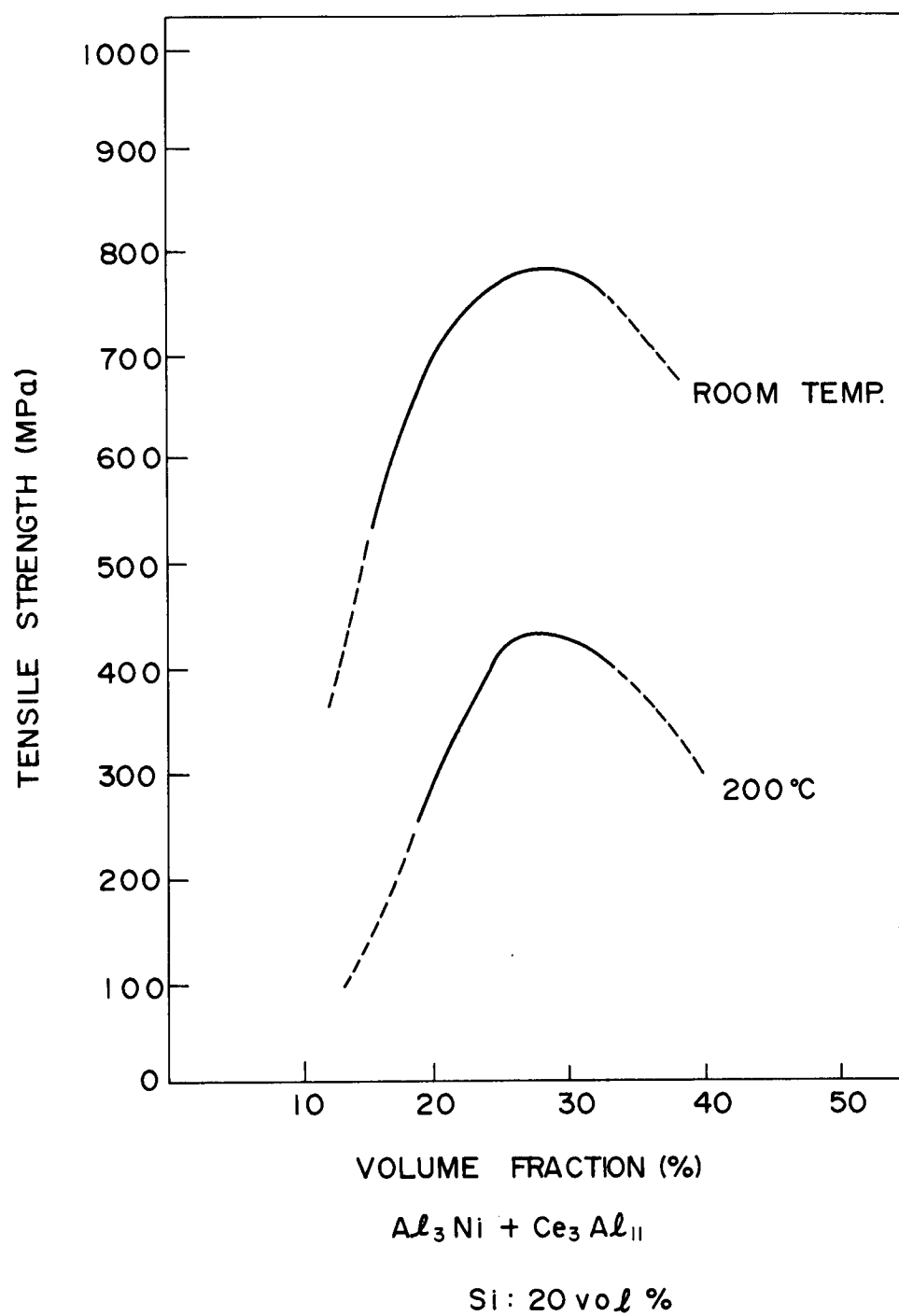
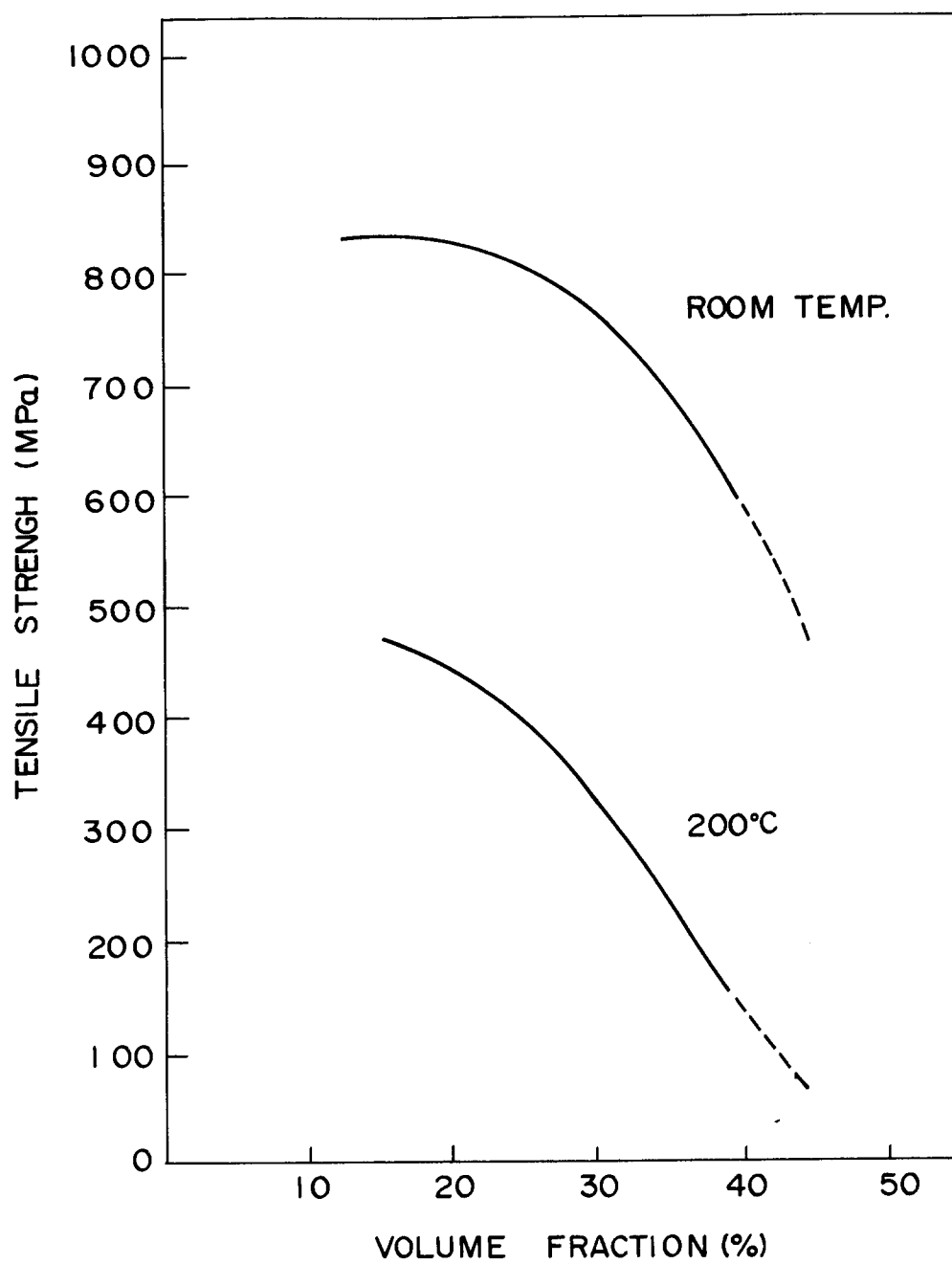


FIG. 2



Si

 Al_3Ni : 25vol % Ce_3Al_{11} : 7vol %

FIG. 3

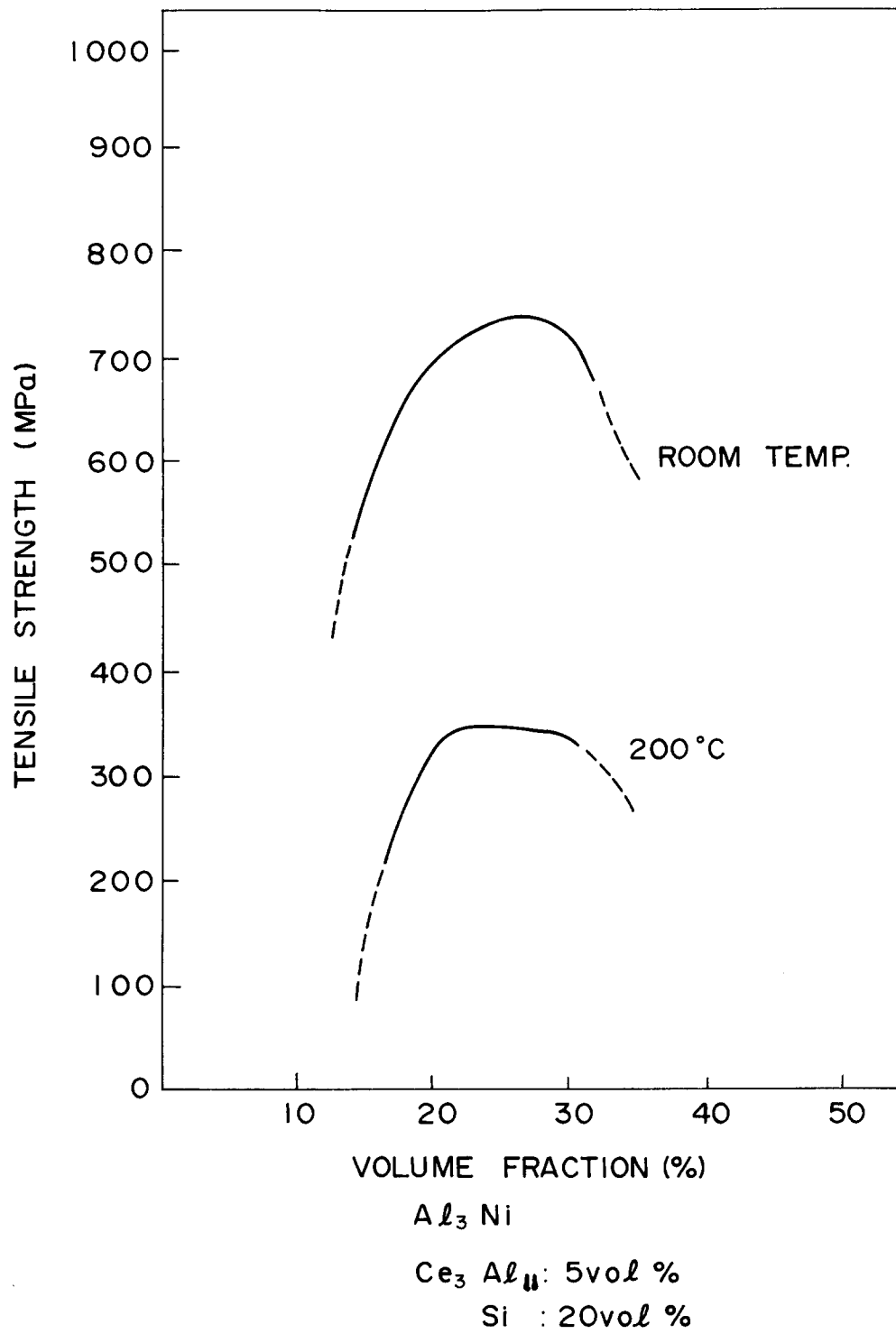
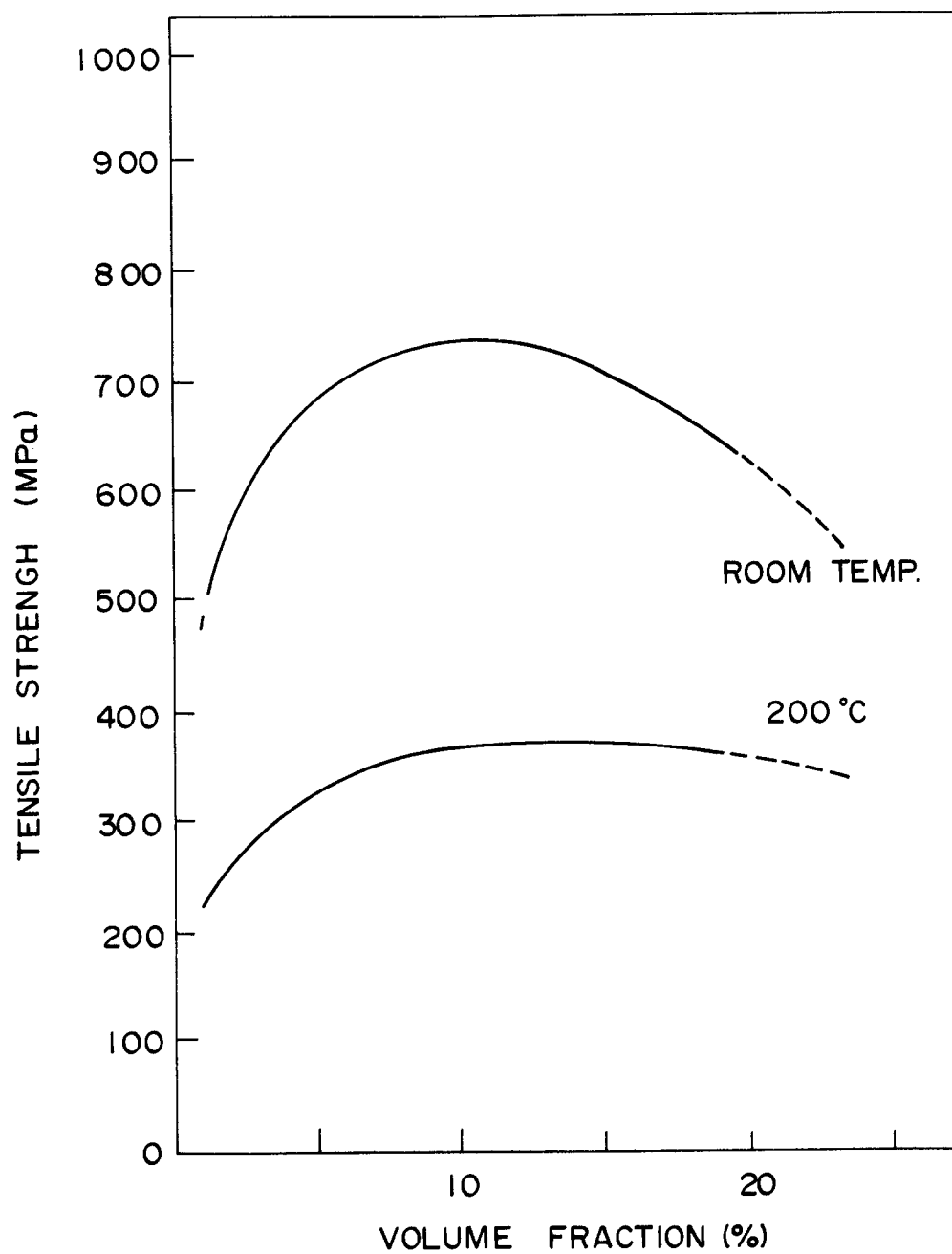


FIG. 4

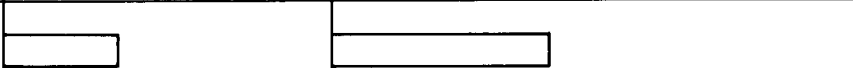
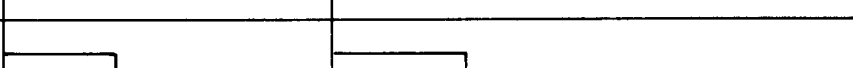
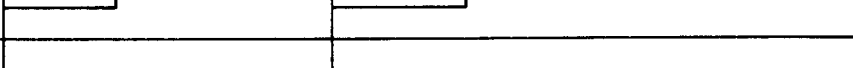
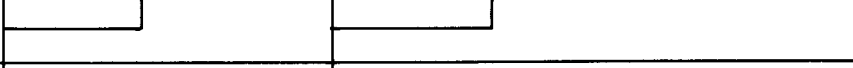
 $\text{Ce}_3\text{Al}_{11}$ Al_3Ni : 20vol %

Si : 20vol %

F I G. 7

PRESSURE : 20kgf / cm²

SPEED : 0.84m / sec

TEST MATERIAL	SPECIFIC WEAR. LOSS OF TEST MATERIAL $\times 10^{-7} \text{ mm}^2/\text{kg}$					SPECIFIC WEAR LOSS OF MATING MATERIAL (S45C) $\times 10^{-7} \text{ mm}^2/\text{kg}$									
	2	4	6	8	10	2	4	6	8	10	12	14	16	18	
INVENTION EX.1															
INVENTION EX.2															
INVENTION EX.3															
INVENTION EX.4															
COMP. MATERIAL, $\text{Al}_{0.01}\text{Si}_5\text{Fe}_3\text{Ce}_{0.2}$															
A390	