

(19)



(11)

EP 2 835 434 A2

(12)

EUROPEAN PATENT APPLICATION

(43) Date of publication:
11.02.2015 Bulletin 2015/07

(51) Int Cl.:
C22C 19/05 (2006.01) C22F 1/10 (2006.01)

(21) Application number: **14179191.3**

(22) Date of filing: **30.07.2014**

(84) Designated Contracting States:
**AL AT BE BG CH CY CZ DE DK EE ES FI FR GB
GR HR HU IE IS IT LI LT LU LV MC MK MT NL NO
PL PT RO RS SE SI SK SM TR**
Designated Extension States:
BA ME

- Imai, Kiyoshi
Tokyo (JP)
- Nemoto, Kuniyoshi
Tokyo (JP)
- Oinuma, Shun
Tokyo (JP)
- Iwai, Shogo
Tokyo (JP)
- Suga, Takeo
Tokyo (JP)

(30) Priority: **07.08.2013 JP 2013164148**

(71) Applicant: **Kabushiki Kaisha Toshiba
Tokyo 105-8001 (JP)**

(72) Inventors:
• Miyashita, Shigekazu
Tokyo (JP)

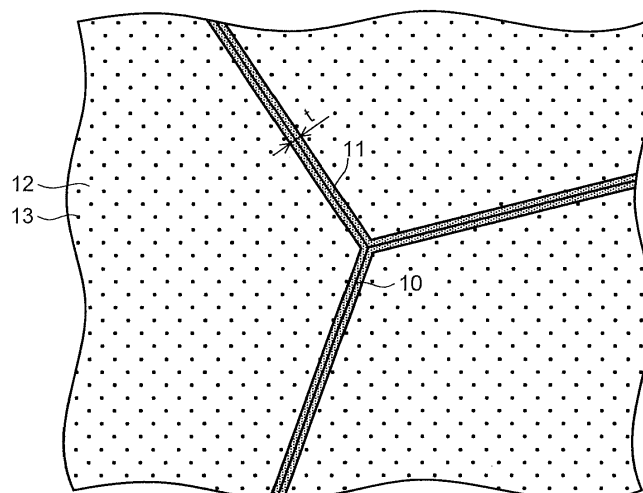
(74) Representative: **Granleese, Rhian Jane
Marks & Clerk LLP
90 Long Acre
London WC2E 9RA (GB)**

(54) **Ni-based alloy for forging, method for manufacturing the same, and turbine component**

(57) An Ni-based alloy for forging of an embodiment contains, in mass%, C: 0.01 to 0.07%, Cr: 14 to 26%, Co: 10 to 15%, Mo: 5 to 12%, Al: 0.8 to 3%, Ti: 0.8 to 3%, and B: 0.001 to 0.006%, the balance being made of

Ni and an unavoidable impurity, and satisfies a relation of $10 \text{ mass\%} \leq \text{Mo} + 0.176\text{Cr} + 0.037\text{Co} \leq 15 \text{ mass\%}$. Further, an average thickness of a carbide precipitated along a grain boundary is 250 nm or less.

FIG. 1



EP 2 835 434 A2

Description**FIELD**

5 **[0001]** Embodiments described herein relate generally to an Ni-based alloy for forging, a method for manufacturing the same, and a turbine component.

BACKGROUND

10 **[0002]** A thermal power generation plant has been made highly efficient in recent years in view of reduction of carbon dioxide emissions into the atmosphere. Thus, it is required to make a steam turbine and a gas turbine which are installed in the thermal power generation plant highly efficient. Further, it is also required to make a CO₂ turbine which is installable in the thermal power generation plant highly efficient. Here, the CO₂ turbine is one which drives a turbine, using CO₂ generated by combustion of fuel such as natural gas and oxygen as a working fluid. In the CO₂ turbine, great part of
15 generated CO₂ can be easily separated and recovered. Thus, the CO₂ turbine attracts attention in view of global environmental protection.

[0003] In order to raise an efficiency in each turbine described above, it is effective to raise an inlet temperature of a working fluid introduced into the turbine. For example, with regard to the steam turbine, an operation at a temperature of 700°C or higher of steam being the working fluid is expected in the future. With regard to the gas turbine or the CO₂
20 turbine, the inlet temperatures of the introduced working fluids tend to be higher.

[0004] Thus, a component constructing a hot section of each turbine is desirable to be constituted with an Ni-based alloy, which is used for a component of a power generation gas turbine or an air plane engine, and is time-proven in usage in a high-temperature place.

[0005] As a representative example of the Ni-based alloy, there can be cited Inconel 718 and Inconel 617 (manufactured by Special Metals Corporation). A strengthening mechanism of the Ni-based alloy can be broadly classified into a precipitation strengthening type and a solid solution strengthening type.

[0006] In the precipitation strengthening type Ni-based alloy, as a result that a precipitates called as a γ' (gamma prime: Ni₃(Al, Ti)) phase or a γ'' (gamma double prime: Ni₃Nb) phase is precipitated by adding Al, Ti, Ta, and Nb to Ni, a mechanical strength under a high temperature is improved. As the representative precipitation strengthening type Ni-based alloy, the above-described Inconel 718 can be cited.
30

[0007] On the other hand, in the solid solution strengthening type Ni-based alloy, a matrix itself is strengthened by adding Co, Mo, and the like to Ni. As the representative solid solution strengthening type Ni-based alloy, the above-described Inconel 617 can be cited.

[0008] As described above, as a material for a component of a turbine used under a high-temperature environment, application of an Ni-based alloy is discussed. As for the Ni-based alloy, a sufficient mechanical strength under the high-temperature environment is required, and further, productivity in manufacturing of a large-sized forged component and the like is required.
35

BRIEF DESCRIPTION OF THE DRAWINGS

40 **[0009]**

Fig. 1 is a diagram schematically showing a microstructure of an Ni-based alloy in an embodiment;
Fig. 2 is a view showing an electron micrograph of an Ni-based alloy, in order to explain a precipitation form of a carbide precipitated into a grain boundary by a condition of an aging treatment; and
45 Fig. 3 is a view showing an electron micrograph of an Ni-based alloy, in order to explain a precipitation form of a carbide precipitated into a grain boundary by a condition of the aging treatment.

DETAILED DESCRIPTION

50 **[0010]** In one embodiment, a nickel (Ni)-based alloy for forging contains, in mass%, carbon (C): 0.01 to 0.07%, chromium (Cr): 14 to 26%, cobalt (Co): 10 to 15%, molybdenum (Mo): 5 to 12%, aluminum (Al): 0.8 to 3%, titanium (Ti): 0.8 to 3%, and boron (B): 0.001 to 0.006%, the balance being made of nickel (Ni) and an unavoidable impurity, and satisfies a relation of $10 \text{ mass\%} \leq \text{Mo} + 0.176\text{Cr} + 0.037\text{Co} \leq 15 \text{ mass\%}$. Further, an average thickness of a carbide precipitated along a grain boundary is 250 nm or less.

[0011] Hereinafter, an embodiment according to the present invention will be described.

[0012] In the Ni-based alloy, a material strength at a room temperature and a high temperature is improved by solid solution strengthening by a solid solution strengthening element such as Mo and W and by precipitation strengthening

by fine precipitation of a γ' (gamma prime: $\text{Ni}_3(\text{Al}, \text{Ti})$) phase obtained by addition of Al, Ti and the like. On the other hand, excessive strengthening deteriorates hot workability of a material and reduces productivity.

[0013] For example, Inconel 617 in which a precipitation strengthening amount by a γ' phase is minor has better forgeability compared with Udimet 520 (manufactured by Special Metals Corporation) or the like in which a precipitation strengthening amount by a γ' phase is large. On the other hand, Inconel 738LC (manufactured by Special Metals Corporation) in which a precipitation amount of a γ' phase is large is not able to be formed by forging and is generally formed by casting.

[0014] As described above, a method for manufacturing an Ni-based alloy is determined mainly by a precipitation amount of a γ' phase. For example, in a case of an Ni-based alloy for forging, an alloy composition is set in a manner not to generate excessive precipitation of a γ' phase during forging process.

[0015] A large-sized forging such as a turbine rotor of a steam turbine or a CO_2 turbine installed in a thermal power generation plant is larger compared with a forging such as a gas turbine or a jet engine for which an Ni-based alloy is conventionally used. Thus, in order to manufacture such a large-sized forging, a forging of over 10 tons is necessary, for example.

[0016] In forging such a large-sized forging, there is a case where a good forged product is not obtained even with Inconel 617 or the like, which has been conventionally considered forgeable, due to a cause such as capacity lack of a forging press. As described above, with regard to an Ni-based alloy used for a large-sized forging, it is necessary to consider not only a precipitation amount of a γ' phase but also a solid solution strengthening amount which influences a deformation resistance at a high temperature.

[0017] Solid solution strengthening is obtained as a result that a different solute atom dissolves in a solvent atom constituting a matrix (solid solution) and that an internal strain occurring at that occasion prevents a motion of dislocation. Solid solution strengthening is theoretically construed by a model in which dislocation moves while disengaging an obstacle of the solute atom. According to Friedel et al., a solid solution strengthening amount of a lean solid solution is proportional to the one-half power of a solute atom concentration and is proportional to the three-thirds power of a misfit strain due to an atom size difference (Advances in Physics, vol. 3, Issue 12, p 446-507). Further, according to Labusch et al., in a highly concentrated solid solution, a solid solution strengthening amount is proportional to the two-thirds power of a solute atom concentration and is proportional to the four-thirds power of a misfit strain due to an atom size difference (Physica status solidi (b). Volume 41, Issue 2, p. 659-669).

[0018] Further, as a factor to influence a characteristic of a metal material substantially, a microstructure of a material can be cited. In an Ni-based alloy, a characteristic of a material is dependent on structures of not only within a grain but also in a grain boundary. In particular, it is known that an M_{23}C_6 type carbide precipitated on a grain boundary in a film shape reduces a toughness of a material. Therefore, in order to secure reliability of a material, it is necessary to control a microstructure properly by optimizing a heat treatment condition.

[0019] Under the circumstances, the present inventors find a parameter indicating a solid solution strengthening amount by quantitatively evaluating a solid solution amount of each additive element and a misfit strain which influence an Ni-based alloy. Further, the present inventors carry out various material testings about materials whose chemical compositions are changed, and find a chemical composition which has excellent forgeability while maintaining a sufficient material strength.

[0020] Further, as a result of investigation of grain boundary structures of Ni-based alloys to which various heat treatments are applied, the present inventors find "an average thickness of a carbide on a grain boundary" as a factor dominating a toughness of an Ni-based alloy. Besides, the present inventors clarify a range of the carbide thickness on the grain boundary by which the toughness can be secured.

[0021] Next, an Ni-based alloy for forging of an embodiment will be described concretely.

[0022] Fig. 1 is a diagram schematically showing a microstructure of an Ni-based alloy in the embodiment. Note that "%" indicating a composition component in the following explanation means "mass%" if not specifically mentioned.

[0023] The Ni-based alloy of the embodiment contains C: 0.01 to 0.07%, Cr: 14 to 26%, Co: 10 to 15%, Mo: 5 to 12%, Al: 0.8 to 3%, Ti: 0.8 to 3%, and B: 0.001 to 0.006%, the balance being made of Ni and an unavoidable impurity, and satisfying a relation of $10 \text{ mass}\% \leq \text{Mo} + 0.176\text{Cr} + 0.037\text{Co} \leq 15 \text{ mass}\%$.

[0024] Further, in the Ni-based alloy of the embodiment, as shown in Fig. 1, a carbide 11 is precipitated along a grain boundary 10. An average thickness t of the carbide 11 is preferable to be 250 nm or less. The carbide 11 is continuously precipitated along the grain boundary 10. Further, in a grain 12, a precipitate 13 is precipitated in a grain shape.

[0025] The carbide 11 is a carbide whose major constituents are Cr and Mo, and concretely, is an M_{23}C_6 type carbide. The reason why the average thickness t of the carbide 11 is preferable to be 250 nm or less is that, for example, such a thickness not reducing a toughness, a toughness for manufacturing a turbine component properly can be secured.

[0026] The precipitate 13 is constituted with a γ' (gamma prime: $\text{Ni}_3(\text{Al}, \text{Ti})$) phase. A diameter of the γ' phase is preferable to be small, in view of precipitation strengthening. An average diameter of the γ' phase is preferable to be 150 nm or less, for example.

[0027] Here, the Ni-based alloy in the embodiment can further contain 0.05 to 0.7% of Ta, in addition to the aforemen-

tioned chemical composition. Further, the Ni-based alloy in the embodiment can further contain 0.1 to 0.7% of Nb in addition to the aforementioned chemical composition. Further, the Ni-based alloy in the embodiment can further contain 0.05 to 0.7% of Ta and 0.1 to 0.7% of Nb, in addition to the aforementioned chemical composition.

[0028] Note that as the unavoidable impurity there can be cited Si, Mn, N, Cu, Fe, S, and the like, for example. A remaining content ratio of the unavoidable impurity as above is preferable to be approximated to 0% to the extent possible.

[0029] The above-described Ni-based alloy of the embodiment is suitable as a material constituting a turbine component constructed by forging, such as a power generation turbine, for example, which is used under a temperature of 650°C or higher, for example. As the turbine component, there can be cited a turbine rotor, a rotor blade, a stationary blade, a screwing member, a pipe, and the like, for example. Those forged components are each disposed in a high temperature and high pressure environment.

[0030] Here, as the screwing member, there can be exemplified a bolt and a nut fixing a turbine casing or various component parts inside the turbine, for example. Further, as the pipe, there can be exemplified a pipe which is disposed in a power generation turbine plant or the like and through which a high temperature and high pressure working fluid passes, for example.

[0031] Note that it is possible to construct all portions of the turbine components of the power generation turbine described above with the above-described Ni-based alloy. Further, it is also possible to construct a limited portion of the turbine component to be subjected to a high temperature in particular with the above-described Ni-based alloy.

[0032] The Ni-based alloy for forging of the embodiment described above is superior to a conventional Ni-based alloy for forging in a strength characteristic and is superior in forgeability. Thus, the turbine components fabricated by using the Ni-based alloy for forging of the embodiment, such as a turbine rotor, a rotor blade, a stationary blade, a screwing member, and a pipe, have high reliability even under a high temperature environment.

[0033] Next, there will be explained a reason for limiting a range of each composition component in the Ni-based alloy for forging of the embodiment described above.

(1) C (carbon)

[0034] C is effective as a constituent element of a carbide being a strengthening phase. Further, C has a function to suppress coarsening of a grain under a high temperature, by a pinning effect of the carbide to prevent movement of a grain boundary. When a content ratio of C is less than 0.01%, strengthening by the carbide is not sufficient. Further, when the content ratio of C is less than 0.01%, there is a possibility that failure in securing of a sufficient precipitation amount of the carbide causes coarsening of the grain. On the other hand, when the content ratio of C is over 0.07%, forgeability is reduced. Thus, the content ratio of C is set to be 0.01 to 0.07%. Further, the more preferable content ratio of C is 0.03 to 0.07%.

(2) Cr (chromium)

[0035] Cr is an element indispensable for heightening an oxidation resistance, a corrosive resistance, and a high temperature strength characteristic of an Ni-based alloy. When a content ratio of Cr is less than 14%, the oxidation resistance and the corrosive resistance are reduced. On the other hand, when the content ratio of Cr is over 26%, precipitation of a σ phase which causes reduction of a creep strength becomes prominent and forgeability is deteriorated. Thus, the content ratio of Cr is set to be 14 to 26%. Further, the more preferable content ratio of Cr is 16 to 20%.

(3) Co (cobalt)

[0036] Co solid-dissolves in a matrix in an Ni-based alloy and improves a creep strength and a tensile strength. When a content ratio of Co is less than 10%, a sufficient mechanical strength cannot be obtained. On the other hand, when the content ratio of Co is over 15%, forgeability is reduced. Thus, the content ratio of Co is set to be 10 to 15%. Further, the more preferable content ratio of Co is 11 to 14%.

(4) Mo (molybdenum)

[0037] Mo solid-dissolves in an Ni matrix and improves a creep strength and a tensile strength. Further, by part of Mo substituting in an $M_{23}C_6$ type carbide, stability of the carbide is heightened. When a content ratio of Mo is over 12%, hot workability is reduced. On the other hand, when the content ratio of Mo is less than 5%, improvement of a mechanical strength cannot be obtained. Thus, the content ratio of Mo is set to be 5 to 12%. Further, the more preferable content ratio of Mo is 7 to 10%.

(5) Al (aluminum)

[0038] Al generates a γ' (Ni_3Al) phase with Ni and improves a mechanical strength of an Ni-based alloy by precipitation. When a content ratio of Al is less than 0.8%, an effect by precipitation of the γ' phase is not exhibited. On the other hand, when the content ratio of Al is over 3%, precipitation of a σ phase is promoted, and the mechanical strength is reduced. Further, when the content ratio of Al is over 3%, hot workability is substantially reduced. Thus, the content ratio of Al is set to be 0.8% to 3%. Further, the more preferable content ratio of Al is 1 to 2%.

(6) Ti (titanium)

[0039] Ti, similarly to Al, generates a γ' ($\text{Ni}_3(\text{Al}, \text{Ti})$) phase with Ni, and improves a mechanical strength of an Ni-based alloy. When a content ratio of Ti is less than 0.8%, an effect by precipitation of the γ' phase is not exhibited. On the other hand, when the content ratio of Ti is over 3%, precipitation of a σ phase or a η phase is promoted, and the mechanical strength is reduced and hot workability is reduced. Thus, the content ratio of Ti is set to be 0.8 to 3%. Further, the more preferable content ratio of Ti is 1 to 2%.

(7) B (boron)

[0040] B segregates in a grain boundary and improves a high temperature strength characteristic. When a content ratio of B is less than 0.001%, such an effect to improve the high temperature strength characteristic is not exhibited. On the other hand, when the content ratio of B is over 0.006%, intergranular embrittlement occurs. Thus, the content ratio of B is set to be 0.001 to 0.006%. Further, the more preferable content ratio of B is 0.002 to 0.004%.

(8) Ta (tantalum)

[0041] Ta solid-dissolves in a γ' ($\text{Ni}_3(\text{Al}, \text{Ti})$) phase and stabilizes the γ' phase. When a content ratio of Ta is less than 0.05%, the above-described effect is not exhibited. On the other hand, when the content ratio of Ta is over 0.7%, forgeability is reduced. Thus, the content ratio of Ta is set to be 0.05% to 0.7%. Further, the more preferable content ratio of Ta is 0.08 to 0.12%.

(9) Nb (niobium)

[0042] Nb, similarly to Ta, solid-dissolves in a γ' ($\text{Ni}_3(\text{Al}, \text{Ti})$) phase and stabilizes the γ' phase. When a content ratio of Nb is less than 0.1%, the above-described effect is not exhibited. On the other hand, when the content ratio of Nb is over 0.7%, segregation occurs at a time of dissolving or forging, and forgeability is reduced. Thus, the content ratio of Nb is set to be 0.1 to 0.7%. Further, the more preferable content ratio of Nb is 0.2 to 0.5%.

(10) $\text{Mo} + 0.176\text{Cr} + 0.037\text{Co}$

[0043] As stated above, it is considered that a solid solution strengthening amount in a highly concentrated solid solution is proportional to the two-thirds power of a solute atom concentration and is proportional to the four-thirds power of a misfit strain due to an atom size difference. Thus, in the present embodiment, for Mo, Cr, and Co, which are considered to contribute to solid solution strengthening, a parameter representing solid solution strengthening is defined from the number of atoms per one mass% and each atomic radius. Note that since the content ratio of C (carbon) is small in the present embodiment, C is excluded from the parameter.

[0044] Atomic weights of Mo, Cr, and Co are 95.9, 52.0, and 58.9, respectively. A ratio of the number of atoms in a case where the same amounts of the respective elements are added is, in a case of Mo being 1, Cr and Co being 1.84 and 1.62, respectively. Values of the two-thirds power of this ratio are 1, 1.50, and 1.38, respectively.

[0045] Further, a misfit strain occurring at a time that each element is added is determined by an atomic size difference from an Ni atom. Atomic radius differences between the Ni atom and Mo, Cr, Co atoms is 0.15 Å (angstrom), 0.03 Å, and 0.01 Å, respectively. Thus, a ratio of the misfit strain amounts in a case where the respective elements are added is, in a case of Mo being 1, Cr and Co being 0.200 and 0.067, respectively. Values of the four-thirds power of this ratio are 1, 0.117, and 0.027, respectively.

[0046] Therefore, a ratio of solid solution strengthening amount per one mass% of the respective elements is, in a case of Mo being 1, Cr being 0.176 ($1.50 \times 0.117 = 0.176$), and Co being 0.037 ($1.38 \times 0.027 = 0.037$). From those results, as a parameter representing a solid solution strengthening amount, " $\text{Mo} + 0.176\text{Cr} + 0.037\text{Co}$ " is set.

[0047] When a value (content ratio) of this parameter is over 15%, the solid solution strengthening amount becomes excessive, and deteriorates ductility during forging process. On the other hand, when the value of the parameter is less

than 10%, the solid solution strengthening amount becomes substantially low, and a sufficient strength cannot be obtained. Thus, the value of the above-described parameter is set to be 10 to 15%. Further, the more preferable value of the above-described parameter is 11 to 13.5%.

[0048] Note that a misfit strain by addition of an element is considered, in a strict sense, to be influenced not only by an atomic size but also by interaction or the like with Ni and other atoms. However, here, for the sake of simplicity, a misfit strain value is determined from a difference between each solute atom and an Ni atom. Further, though it is known that Mo and Cr in combination with C form carbides, consumption of Mo and Cr by the carbide is ignored since the content ratio of C is low.

(11) Si (silicon), Mn (manganese), N (nitrogen), Cu (copper), Fe (iron), and S (sulfur)

[0049] Si, Mn, N, Cu, Fe, and S are classified into unavoidable impurities in the Ni-based alloy for forging of the embodiment. Remaining content ratios of those unavoidable impurities are desirable to be approximated to 0% to the extent possible. Further, it is preferable that among those unavoidable impurities, at least Si and Mn are restricted to be 0.1% or less and that N is restricted to be 0.01% or less.

[0050] Si is added in order to supplement a corrosive resistance in a case of ordinary steel. However, an Ni-based alloy has a large Cr content and a sufficient corrosive resistance can be secured. Thus, it is desirable that a remaining content ratio of Si is set to be 0.1% or less and that the remaining content ratio thereof is approximated to 0% to the extent possible.

[0051] Mn makes S (sulfur) causing brittleness into MnS, to prevent brittleness, in a case of ordinary steel. However, a content of S in an Ni-based alloy is quite low and it is not necessary to add Mn. Thus, it is desirable that a remaining content ratio of Mn is set to be 0.1% or less and that the remaining content ratio thereof is approximated to 0% to the extent possible.

[0052] N forms TiN by reacting with Ti in a material and decreases Ti which contributes to generation of a γ' phase. Consequently, a mechanical strength is reduced. Thus, it is desirable that a remaining content ratio of N is set to be 0.01% or less and that the remaining content ratio thereof is approximated to 0% to the extent possible.

[0053] Here, there will be described a method for manufacturing an Ni-based alloy for forging of the embodiment and a turbine component manufactured by using the Ni-based alloy for forging.

[0054] The above-described Ni-based alloy for forging of the embodiment is manufactured as follows, for example.

[0055] First, composition components to constitute the Ni-based alloy are vacuum induction melted (VIM) and molten metal thereof is poured into a predetermined mold form, to form an ingot. Then, the ingot is soaking treated and hot forged, and subjected to a solution treatment, an ageing treatment, and the like, so that the Ni-based alloy for forging is fabricated.

[0056] A turbine rotor being a turbine component is fabricated as below, for example.

[0057] For example, as one method (double melt), composition components to constitute the Ni-based alloy for forging of the embodiment are vacuum induction melted (VIM), electroslag remelted (ESR), and poured into a predetermined mold form. Subsequently, a soaking treatment, a forging treatment, a solution treatment, an aging treatment, and the like are carried out, so that the turbine rotor is fabricated.

[0058] As another method (double melt), composition components to constitute the Ni-based alloy for forging of the embodiment are vacuum induction melted (VIM), vacuum arc remelted (VAR), and poured into a predetermined mold form. Subsequently, a soaking treatment, a forging treatment, a solution treatment, an aging treatment, and the like are carried out, so that the turbine rotor is fabricated.

[0059] Further, as still another method (triple melt), composition components to constitute the Ni-based alloy for forging of the embodiment are vacuum induction melted (VIM), electroslag remelted (ESR), vacuum arc remelted (VAR), and poured into a predetermined mold form. Subsequently, a soaking treatment, a forging treatment, a solution treatment, an aging treatment, and the like are carried out, so that the turbine rotor is fabricated.

[0060] At least a predetermined portion of the turbine rotor is manufactured by the above-described method for manufacturing the turbine rotor. As the predetermined portion, there can be cited a portion exposed to a high temperature of 700°C or higher, for example, of the turbine rotor. In this case, a portion exposed to a temperature of about 600°C, for example, of the turbine rotor is manufactured with a conventional heat resistant alloy. Then, a component made of the Ni-based alloy for forging of the embodiment manufactured by the above-described manufacturing method and a component made of the conventional heat resistant alloy are joined by welding, for example, to construct a turbine rotor. Note that a method for joining the component made of the Ni-based alloy for forging of the embodiment and the component made of the conventional heat resistant alloy is not limited to welding, but the components can be fastened by a bolt and a nut, for example.

[0061] As a result that components to construct the turbine rotor are fabricated separately as above, it is possible to manufacture a turbine rotor usable under a high temperature environment of 700°C or higher with a small steel ingot of an Ni-based alloy. Note that depending on a temperature condition to be used, it is possible to manufacture the whole

turbine rotor by the above-described method for manufacturing the turbine rotor.

[0062] A rotor blade, a stationary blade, and a screwing member being turbine components are fabricated as below, for example.

[0063] First, composition components to constitute the Ni-based alloy for forging of the embodiment are vacuum induction melted (VIM) and electroslag remelted (ESR). Subsequently, a molten alloy is poured into a predetermined mold form under a reduced pressure atmosphere to fabricate an ingot, and a soaking treatment is performed. Then, the ingot is disposed in a mold form corresponding to a shape of the above-described turbine component and a forging treatment, a solution treatment, an aging treatment, and the like are carried out, so that the rotor blade, the stationary blade, and the screwing member are fabricated. In other words, the rotor blade, the stationary blade, and the screwing member are fabricated by die forging.

[0064] Further, as another method (double melt), composition components to constitute the Ni-based alloy for forging of the embodiment are vacuum induction melted (VIM) and vacuum arc remelted (VAR), for example. Subsequently, a molten alloy is poured into a predetermined mold form under a reduced pressure atmosphere to fabricate an ingot. Then, a soaking treatment is applied to the ingot, and similarly to the above, a forging treatment, a solution treatment, an aging treatment, and the like are carried out, so that the rotor blade, the stationary blade, and the screwing member can be fabricated.

[0065] Further, as another method (triple melt), composition components to constitute the Ni-based alloy for forging of the embodiment are vacuum induction melted (VIM), electroslag remelted (ESR), and vacuum arc remelted (VAR), for example. Subsequently, a molten alloy is poured into a predetermined mold form under a reduced pressure atmosphere to fabricate an ingot. Then, a soaking treatment is applied to the ingot, and similarly to the above, a forging treatment, a solution treatment, an aging treatment, and the like are carried out, so that the rotor blade, the stationary blade, and the screwing member can be fabricated.

[0066] A pipe being a forged component of the embodiment is fabricated as below, for example.

[0067] First, composition components to constitute the Ni-based alloy for forging of the embodiment are electric furnace melted (EF) and argon-oxide decarburization (AOD) is performed, so that an ingot is fabricated. Subsequently, a soaking treatment is applied to the ingot. The ingot is drilled by a vertical press, so that an element pipe of a cup shape is fabricated. Then, treatment and reheating by a mandrel and a die are repeated by a horizontal press, so that the element pipe is formed in a shape of a pipe. This treatment method is an Ehrhardt push bench pipe manufacturing method. Then, a solution treatment, an aging treatment, and the like are carried out, so that the pipe is fabricated.

[0068] Note that the methods for fabricating the turbine rotor, the rotor blade, the stationary blade, the screwing member, and the pipe are not limited to the methods described above. Further, the above-described forged components such as a turbine rotor, a rotor blade, a stationary blade, a screwing member, and a pipe can be applied to power generation turbines, such as a steam turbine, a gas turbine, and a CO₂ turbine, for example.

[0069] Here, the above-described each heat treatment performed in manufacturing the Ni-based alloy for forging and the turbine component will be explained. Note that a temperature in each heat treatment is set in each range described below, in correspondence with the Ni-based alloy for forging, the turbine component, and the like to be treated. Further, a time for each treatment is set properly in correspondence with the Ni-based alloy for forging, the turbine component, and the like to be treated.

[0070] In the soaking treatment, it is necessary to heat the alloy at a high temperature for a sufficient time, in order to decrease segregation of a chemical component by thermal diffusion. Thus, the soaking treatment is preferable to be performed in a temperature range of 1000 to 1200°C.

[0071] Forging is required to be performed in a range of temperature where sufficient ductility of a material can be obtained to a zero ductility temperature. Thus, forging is preferable to be performed in a temperature range of 950 to 1100°C.

[0072] In the solution treatment, a temperature range of 1050 to 1200°C is preferable to be maintained for 1 to 24 hours. Here, the solution treatment is performed for the purpose of solid dissolving an alloy element into a matrix sufficiently to obtain an effect of solid solution strengthening sufficiently, and of enabling precipitation control of a precipitate by a heat treatment thereafter. Further, the solution treatment is sometimes performed for the purpose of adjusting a grain size.

[0073] When a temperature of the solution treatment is lower than 1050°C, an alloy element is not solid dissolved into the matrix completely, and strengthening by a solid solution strengthening element is not carried out sufficiently. Further, it also becomes difficult to control a precipitated form of a precipitates by the heat treatment after the solution treatment. On the other hand, when the temperature of the solution treatment is over 1200°C, coarsening of the grain size is brought about, and a mechanical strength is reduced. Thus, the temperature of the solution treatment is set to be 1050 to 1200°C. Further, it is further preferable that the temperature of the solution treatment is set to be 1050 to 1150°C. Note that the Ni-based alloy and the turbine component having been subjected to the solution treatment are cooled to a room temperature by water cooling, forced air cooling, or the like, for example.

[0074] Next, there will be described an aging treatment applied to the Ni-based alloy and the turbine component which

have been cooled to the room temperature after the solution treatment.

[0075] In the aging treatment, a temperature range of 700 to 800°C is preferable to be maintained for 5 to 50 hours. The aging treatment can be performed in multiple stages. Note that after the aging treatment the Ni-based alloy and the turbine component are cooled to the room temperature by water cooling or furnace cooling, for example.

[0076] Here, the reason why the temperature and the time in the aging treatment are set in the above-described range will be explained.

[0077] A major object of the aging treatment is control of a precipitation form of a γ' phase precipitated in a grain. Further, the aging treatment influences a property of a grain boundary. Therefore, for the aging treatment, the temperature and the time condition are required to be determined in consideration of structures in the grain and the grain boundary.

[0078] Fig. 2 and Fig. 3 are views showing electron micrographs of Ni-based alloys in order to explain precipitation forms of carbides precipitated into grain boundaries by conditions of the aging treatments. A composition of the Ni-based alloy shown here is 0.04% of C, 18% of Cr, 12% of Co, 9% of Mo, 1.3% of Al, 1.4% of Ti, 0.003% of B, 0.1% of Ta, 0.3% of Nb, and the balance being Ni. Fig. 2 shows the microstructure having been subjected to the aging treatment of 850°C for 10 hours, and Fig. 3 shows the microstructure having been subjected to the aging treatment of 750°C for 10 hours. Further, the soaking treatment and the solution treatment are performed in the above-described ranges. Note that Fig. 2 and Fig. 3 also show precipitates 13 (γ' phase).

[0079] In an ordinary aging treatment, as shown in Fig. 2, a film-shaped carbide 11 is precipitated in a manner to cover the grain boundary of the Ni-based alloy. The film-shaped carbide 11 is a brittle carbide ($M_{23}C_6$ type carbide) whose major constituents are Cr and Mo. The carbide 11 promotes destruction of the grain boundary and substantially reduces a toughness of the material. Therefore, it has been considered that it is necessary to perform the aging treatment preventing precipitation of such a film-shaped carbide 11 covering the grain boundary.

[0080] However, as shown in Fig. 3, a thickness of the film-shaped carbide covering the grain boundary becomes small, depending on the condition of the aging treatment. Note that the carbide is continuously precipitated along the grain boundary. As a result of a material testing, the inventors clarify that reduction of ductility/toughness does not occur in a case where a thickness of a carbide is sufficiently small. The above-described temperature and time are specified to a range satisfying both fine precipitation of a γ' phase and suppression of coarsening of the carbide covering the grain boundary.

[0081] When the temperature of the aging treatment is lower than 700°C, coarsening of the carbide covering the grain boundary can be suppressed, but growth of the γ' phase is quite slow. Thus, improvement of a mechanical strength by precipitation of the γ' phase cannot be obtained. On the other hand, when the temperature of the aging treatment is over 800°C, fine precipitation of the γ' phase is achieved, and a sufficient strength is obtained. However, coarsening of the carbide covering the grain boundary is significant, and a toughness is reduced.

[0082] Under the circumstances, the temperature of the aging treatment is set to be 700 to 800°C. Here, for the purpose of early precipitation of the γ' phase, heat treatments of multiple stages, for example, two stages, can be performed in the aging treatment. In also such a case, the temperature is set within the above-described temperature range of the aging treatment. Further, the whole heat treatment time in multiple stages is also set within the above-described time range of the aging treatment. For example, there can be exemplified a treatment in which a temperature of 800°C is maintained for 10 hours and thereafter a temperature of 750°C is maintained for 20 hours. Note that temperature reduction from 800°C to 750°C is performed by furnace cooling, for example.

[0083] Cooling after the aging treatment is performed by furnace cooling or air cooling, for example. When the aging treatment is performed in multiple stages, cooling between each aging treatment is performed by furnace cooling, for example, as described above. Then, cooling is not performed to reach a room temperature but the aging treatment is performed continuously.

[0084] Here, an intermediate heat treatment can be applied, before the aging treatment is performed, to the Ni-based alloy and the turbine component which are cooled to the room temperature after the solution treatment. An object of the intermediate heat treatment is to form a block-shaped carbide intermittently along a grain boundary, first, before the aging treatment, in order to suppress precipitation or coarsening of a film-shaped carbide covering the grain boundary. This carbide is also a carbide whose major constituents are Cr and Mo.

[0085] The intermediate heat treatment is preferable to be performed in a temperature range of 1000 to 1050°C. In cases where the intermediate heat treatment temperature is lower than 1000°C and where the intermediate heat treatment temperature is over 1050°C, the block-shaped carbide is not precipitated. A time of the intermediate treatment is set properly in correspondence with the Ni-based alloy and the turbine component to be treated.

[0086] Note that in a case where a content ratio of C (carbon) is sufficiently small, precipitation of a film-shaped carbide on a grain boundary is not prominent, and such an intermediate heat treatment can be omitted. The case where the content ratio of C is sufficiently small is, though it varies depending on a grain size or the like, a case where the content ratio of C is 0.04% or less, for example. Cooling after the intermediate heat treatment is performed by furnace cooling, water cooling, or forced air cooling, for example. Then, the Ni-based alloy and the turbine component are cooled to the room temperature.

(Influence of Chemical Composition)

[0087] Hereinafter, it will be described that the Ni-based alloy for forging of the embodiment is excellent in a strength characteristic and forgeability.

[0088] Table 1 shows chemical compositions of a sample 1 to a sample 21 used for evaluation of a strength characteristic, forgeability, and the like. Note that the sample 1 to the sample 13 shown in Table 1 are Ni-based alloys within a chemical composition range of the Ni-based alloy for forging of the embodiment, and the sample 14 to the sample 21 are Ni-based alloys whose compositions are not within the chemical composition range of the Ni-based alloy for forging of the embodiment, and are comparative examples.

[Table 1]

		Mass%									
	Ni	C	Cr	Co	Mo	Al	Ti	B	Ta	Nb	Mo+0.176Cr +0.037Co
Sample 1	Balance	0.04	18.5	12.3	9.0	1.3	1.4	0.003	0.11	0.3	12.7
Sample 2	Balance	0.012	15.1	10.5	11.7	2.5	0.9	0.003	0.65	0.11	14.7
Sample 3	Balance	0.03	22.1	14.5	5.8	1.4	2.5	0.003	0.06	0.66	10.2
Sample 4	Balance	0.06	16.5	12.5	9.2	1.0	1.0	0.005	0.64	0.66	12.6
Sample 5	Balance	0.02	18.1	10.2	9.5	1.4	1.4	0.004	0	0.12	13.1
Sample 6	Balance	0.03	25.1	12.6	9.5	1.8	1.0	0.003	0	0.35	14.4
Sample 7	Balance	0.05	24.1	14.3	10.0	1.3	1.3	0.003	0	0.65	14.8
Sample 8	Balance	0.05	18.2	11.2	8.1	1.1	1.2	0.003	0.06	0	11.7
Sample 9	Balance	0.02	20.1	12.6	6.3	2.0	0.86	0.003	0.1	0	10.3
Sample 10	Balance	0.02	18.2	11.2	11.0	1.2	1.7	0.003	0.66	0	14.6
Sample 11	Balance	0.05	19.0	10.8	11.0	0.85	1.5	0.003	0	0	14.7
Sample 12	Balance	0.03	22.5	13.5	9.5	1.5	1.4	0.003	0	0	14.0
Sample 13	Balance	0.05	24.1	12.6	7.1	1.5	1.0	0.003	0	0	11.8
Sample 14	Balance	0.05	18.3	12.6	9.0	0.6	0.6	0.003	0.12	0.29	12.7
Sample 15	Balance	0.05	26.5	15.3	7.8	1.3	1.4	0.003	0.11	0.3	13.0
Sample 16	Balance	0.01	13.8	9.0	12.1	2.4	1.8	0.003	0.11	0.33	14.9
Sample 17	Balance	0.01	19.5	10.0	11.1	3.1	0.8	0.003	0	0	14.9
Sample 18	Balance	0.05	20.1	12.3	10.1	1.0	3.2	0.003	0	0	14.1
Sample 19	Balance	0.05	24.1	14.3	11.5	1.3	1.3	0.003	0.12	0.31	16.3
Sample 20	Balance	0.05	18.2	12.6	11.8	1.2	1.3	0.003	0.11	0.28	15.5
Sample 21	Balance	0.05	17.0	12.6	6.3	1.1	0.9	0.003	0.1	0.31	9.8

[0089] A strength characteristic was evaluated by a tensile test, a toughness was evaluated by a Charpy impact test, and forgeability was evaluated by visual observation. Further, a thickness of a film-shaped carbide covering a grain boundary was measured by microstructure observation.

[0090] A test piece used in each test was fabricated as follows.

[0091] Each of the Ni-based alloys of the sample 1 to the sample 21 having the chemical compositions shown in Table 1 was melted in a vacuum induction melting furnace, to fabricate an ingot.

[0092] Subsequently, a soaking treatment was applied to the ingot at 1050°C for 5 hours. Thereafter, forging was performed in a temperature range of 950 to 1100°C (reheating temperature was 1100°C) by a 500 kgf hammer forging machine. After forging, a solution treatment was performed at a temperature of 1100°C for 4 hours, and thereafter, cooling to a room temperature was carried out by air cooling. After cooling, an intermediate heat treatment was performed at a temperature of 1025°C for 10 hours, and thereafter, cooling to the room temperature was carried out by furnace cooling. After cooling, a two-stage aging treatment at a temperature of 800°C for 10 hours and subsequently at a

temperature of 750°C for 20 hours was performed continuously. Thereafter, cooling to the room temperature was performed by air cooling, so that a forged product was made.

[0093] Then, from the above forged product, the test piece for the tensile test and the Charpy impact test was obtained.

[0094] The tensile test was performed in accordance with JIS Z 2241, and measurement of 0.2% proof stress and a tensile strength at the room temperature was performed. The Charpy impact test was performed in accordance with JIS Z 2242, and measurement of a Charpy impact value was performed.

[0095] For evaluation of forgeability, the sample after the soaking treatment described above was forged by a 500 kgf hammer forging machine, to fabricate a test piece of a solid columnar shape with a diameter of 125 mm and a length of 210 mm. Further, the forging treatment was performed until a forging ratio (a forging ratio based on JIS G 0701 (representation of a forging ratio of a steel product forging operation) became 3. Note that the forging treatment was performed in a range of 950 to 1100°C. When a temperature of the test piece being an object to be forged was reduced, that was, when the object to be forged was being cured, reheating to the reheating temperature of 1100°C was performed and the forging treatment was repeated. The evaluation of forgeability was performed by visual observation of existence/absence of a forging crack after the test piece was cooled.

[0096] Here, the forging ratio is obtained by dividing a cross-sectional area of the object to be forged before application of the forging treatment, the cross-sectional area being vertical in a direction where the object to be forged is to be expanded, by a cross-sectional area of the object to be forged after the forging treatment, the cross-sectional area being vertical in a direction where the object to be forged has been expanded.

[0097] In measurement of the thickness of the film-shaped carbide covering the grain boundary, the forged product cooled to the room temperature after the aging treatment was used. The thickness of the carbide was obtained by image-analyzing an electron micrograph photographed at 20000 magnification by using a field-emission scanning electron microscope. In each forged product, 5 representative grain boundaries were selected, and thicknesses of 20 points of the carbide were measured per each grain boundary. Then, the above thicknesses were arithmetically averaged, to obtain an average thickness of the carbide.

[0098] Test results and observation results are shown in Table 2. In Table 2, a case where a forging crack does not exist is denoted as "No", and further, evaluation of forgeability is denoted as "○" in order to indicate that forgeability is excellent. On the other hand, a case where a forging crack exists is denoted as "Yes", and evaluation of forgeability is denoted as "x" to indicate that forgeability is inferior.

[Table 2]

	Forging state		Room temperature tensile test		Charpy impact test	Structure observation
	Forging crack	Forgeability	0.2% proof stress, MPa	Tensile strength, MPa	Impact value, J/cm ²	Average thickness of carbide, nm
Sample 1	No	○	585	1042	78	182
Sample 2	No	○	668	1112	68	63
Sample 3	No	○	704	1187	66	125
Sample 4	No	○	570	1038	55	191
Sample 5	No	○	588	1095	83	80
Sample 6	No	○	632	1098	56	85
Sample 7	No	○	648	1121	54	135
Sample 8	No	○	591	1010	60	148
Sample 9	No	○	641	1084	67	95

EP 2 835 434 A2

(continued)

	Forging state		Room temperature tensile test		Charpy impact test	Structure observation
	Forging crack	Forgeability	0.2% proof stress, MPa	Tensile strength, MPa	Impact value, J/cm ²	Average thickness of carbide, nm
Sample 10	No	○	549	1000	64	84
Sample 11	No	○	555	1022	61	191
Sample 12	No	○	591	1045	54	170
Sample 13	No	○	625	1048	61	210
Sample 14	No	○	460	906	112	189
Sample 15	Yes	×	584	1001	51	221
Sample 16	Yes	×	742	1167	44	58
Sample 17	Yes	×	768	1144	36	64
Sample 18	Yes	×	750	1058	30	204
Sample 19	Yes	×	741	1120	67	206
Sample 20	Yes	×	1017	1100	61	175
Sample 21	No	○	482	950	56	168

[0099] As shown in Table 2, the sample 1 to the sample 13 are higher in both 0.2% proof stresses and tensile strengths compared with the sample 14. It is considered that in the sample 1 to the sample 13, the 0.2% proof stresses and the tensile strengths are high in values because sufficient solid solution strengthening and precipitation strengthening are enhanced. Further, the sample 1 to the sample 13 are excellent in forgeability, and thicknesses of the carbides are 250 nm or less. Further, results of Charpy impact values of the sample 1 to the sample 13 each indicate a value of 50 J/cm² or more. Therefore, it is confirmed that the sample 1 to the sample 13 have practically sufficient toughnesses.

[0100] On the other hand, when a value of "Mo + 0.176Cr + 0.037Co" is less than 10 mass% as in the sample 21, a sufficient 0.2% proof stress or a tensile strength are not obtained even in a case where each alloy component is within a chemical composition range prescribed in the present embodiment. The sample 15 to the sample 20 indicate high values in 0.2% proof stresses and the tensile strengths, but are inferior in forgeability. This is considered to be a result of excessive addition of a strengthening element.

[0101] As described above, in the Ni-based alloy departing from the chemical composition range prescribed in the present embodiment or a range of "Mo + 0.176Cr + 0.037Co", a result which is excellent in both the strength characteristic and the forgeability is not obtained.

(Influence of Heat Treatment)

[0102] Here, in the sample 1, while conditions of the intermediate heat treatment and the aging treatment were changed, there were performed tensile tests, Charpy impact tests, evaluation of forgeability, and measurement of thicknesses of film-shaped carbides covering grain boundaries. Note that methods of respective test, evaluation of forgeability, measurement of the thickness of the carbide were the same as the aforementioned methods.

[0103] By using the sample 1 shown in Table 1, heat treatments were performed under respective conditions of the intermediate heat treatment and the aging treatment shown in Table 3. Note that process steps other than the intermediate heat treatment and the aging treatment are the same as those in the method for fabricating the test piece. In Table 3, for example, "800°C × 10h" means that the heat treatment is performed while a temperature of 800°C is maintained for 10 hours. Further, in the aging treatment, when the two-stage heat treatment is performed, heat treatment conditions are indicated in columns of a first stage and a second stage.

[Table 3]

	Intermediate heat treatment condition	Aging treatment condition	
		First stage	Second stage
Sample 1	1025°C×10h	800°C×10h	750°C×20h
Sample 22	-	750°C×10h	-
Sample 23	-	750°C×30h	-
Sample 24	-	750°C×48h	-
Sample 25	-	800°C×10h	-
Sample 26	-	800°C×30h	-
Sample 27	-	800°C×40h	-
Sample 28	-	800°C×10h	780°C×20h
Sample 29	-	800°C×10h	780°C×30h
Sample 30	1025°C×10h	750°C×10h	-
Sample 31	1025°C×10h	800°C×10h	-
Sample 32	-	675°C×10h	-
Sample 33	-	675°C×30h	-
Sample 34	-	850°C×10h	-
Sample 35	-	850°C×30h	-
Sample 36	-	850°C×10h	780°C×20h
Sample 37	-	850°C×10h	780°C×30h
Sample 38	1025°C×10h	675°C×30h	-
Sample 39	1025°C×10h	850°C×30h	-

[0104] The sample 1, and a sample 22 to a sample 31 shown in Table 3 are heat-treated under the heat treatment condition of the present embodiment, and the other samples are comparative examples heat-treated under a condition departing from a range of the heat treatment condition of the present embodiment. Test results and observation results are shown in Table 4.

[Table 4]

	Room temperature tensile test		Charpy impact test	Structure observation
	0.2% proof stress, MPa	Tensile strength, MPa	Impact value, J/cm ²	Average thickness of carbide, nm
Sample 1	585	1042	78	182
Sample 22	606	1052	100	87
Sample 23	629	1081	74	112
Sample 24	624	1082	68	138

(continued)

5		Room temperature tensile test		Charpy impact test	Structure observation
		0.2% proof stress, MPa	Tensile strength, MPa	Impact value, J/cm ²	Average thickness of carbide, nm
	Sample 25	581	1030	72	191
	Sample 26	589	1049	62	215
10	Sample 27	578	1042	64	201
	Sample 28	574	1025	60	180
	Sample 29	566	1011	58	225
	Sample 30	620	1047	62	99
	Sample 31	571	1055	60	148
20	Sample 32	468	920	79	170
	Sample 33	488	955	72	210
	Sample 34	531	989	38	260
30	Sample 35	508	959	33	425
	Sample 36	541	1000	35	320
	Sample 37	513	968	33	468
35	Sample 38	495	940	61	204
40	Sample 39	524	978	28	380

[0105] The sample 1 and the sample 22 to the sample 31 are higher in both 0.2% proof stresses and tensile strengths compared with samples 32 to samples 39. Average thicknesses of film-shaped carbides covering grain boundaries in the sample 1 and the sample 22 to the sample 31 are each 250 nm or less. The sample 1 and the sample 22 to the sample 31, having thin average thicknesses of the carbides, exhibit higher Charpy impact values compared with the sample 34 to the sample 37 and the sample 39.

[0106] As described above, under the aging treatment condition prescribed in the present embodiment, it is possible to simultaneously attain fine precipitation of the γ' phase in the grain and suppression of coarsening of the carbide covering the grain boundary. Thereby, high values are obtained in both the tensile strength and the Charpy impact value.

[0107] On the other hand, in the sample departing from the aging treatment condition prescribed in the present embodiment, a result which is excellent in both the tensile strength and the Charpy impact value is not obtained.

[0108] According to the embodiment described hereinabove, it becomes possible to have excellent strength characteristic and forgeability.

[0109] While certain embodiments have been described, these embodiments have been presented by way of example only, and are not intended to limit the scope of the inventions. Indeed, the novel embodiments described herein may be embodied in a variety of other forms; furthermore, various omissions, substitutions and changes in the form of the

embodiments described herein may be made without departing from the spirit of the inventions. The accompanying claims and their equivalents are intended to cover such forms or modifications as would fall within the scope and spirit of the inventions.

5

Claims

1. A nickel (Ni)-based alloy for forging containing, in mass%, carbon (C): 0.01 to 0.07%, chromium (Cr): 14 to 26%, cobalt (Co): 10 to 15%, molybdenum (Mo): 5 to 12%, aluminum (Al): 0.8 to 3%, titanium (Ti): 0.8 to 3%, and boron (B): 0.001 to 0.006%, the balance being made of nickel (Ni) and an unavoidable impurity, and satisfying a relation of $10 \text{ mass\%} \leq \text{Mo} + 0.176\text{Cr} + 0.037\text{Co} \leq 15 \text{ mass\%}$, wherein an average thickness of a carbide precipitated along a grain boundary is 250 nm or less.
2. The nickel (Ni)-based alloy for forging according to claim 1, further containing 0.05 to 0.7 mass% of tantalum (Ta).
3. The nickel (Ni)-based alloy for forging according to claim 1, further containing 0.1 to 0.7 mass% of niobium (Nb)
4. The nickel (Ni)-based alloy for forging according to claim 1, further containing 0.05 to 0.7 mass% of tantalum (Ta) and 0.1 to 0.7 mass% of niobium (Nb).
5. The nickel (Ni)-based alloy for forging according to any one of claim 1 to claim 4, wherein the carbide is precipitated by performing a solution treatment at a temperature of 1050 to 1200°C and performing an aging treatment at a temperature of 700 to 800°C.
6. A method for manufacturing a nickel (Ni)-based alloy for forging, comprising:
 - forming a structure of a predetermined shape by melting a nickel (Ni)-based alloy material containing, in mass%, carbon (C): 0.01 to 0.07%, chromium (Cr): 14 to 26%, cobalt (Co): 10 to 15%, molybdenum (Mo): 5 to 12%, aluminum (Al): 0.8 to 3%, titanium (Ti): 0.8 to 3%, and boron (B): 0.001 to 0.006%, the balance being made of nickel (Ni) and an unavoidable impurity, and satisfying a relation of $10 \text{ mass\%} \leq \text{Mo} + 0.176\text{Cr} + 0.037\text{Co} \leq 15 \text{ mass\%}$;
 - solution-treating the structure at a temperature of 1050 to 1200°C; and
 - age-treating the structure having been solution-treated, at a temperature of 700 to 800°C,
 - wherein a carbide with an average thickness of 250 nm or less is precipitated along a grain boundary by the solution-treating and the age-treating.
7. A turbine component comprising
 - at least a predetermined portion fabricated by using the nickel (Ni)-based alloy for forging according to any one of claim 1 to claim 5.

40

45

50

55

FIG. 1

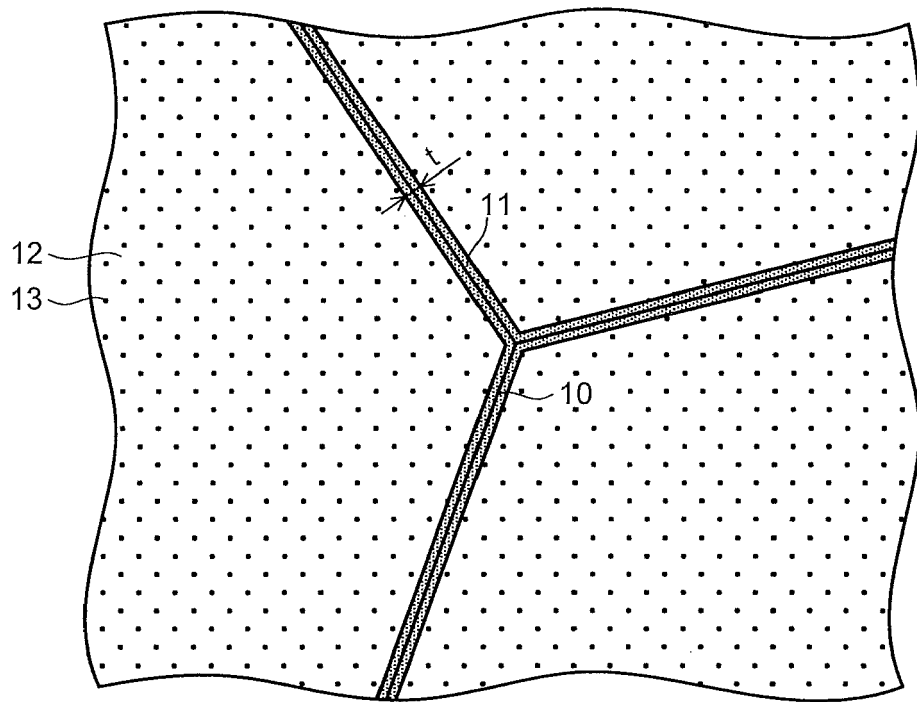


FIG. 2

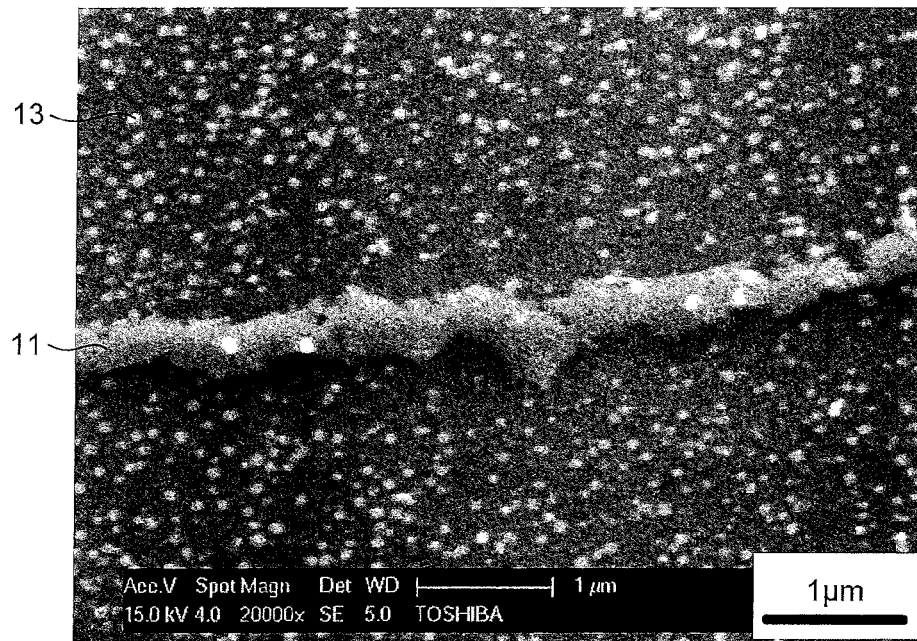
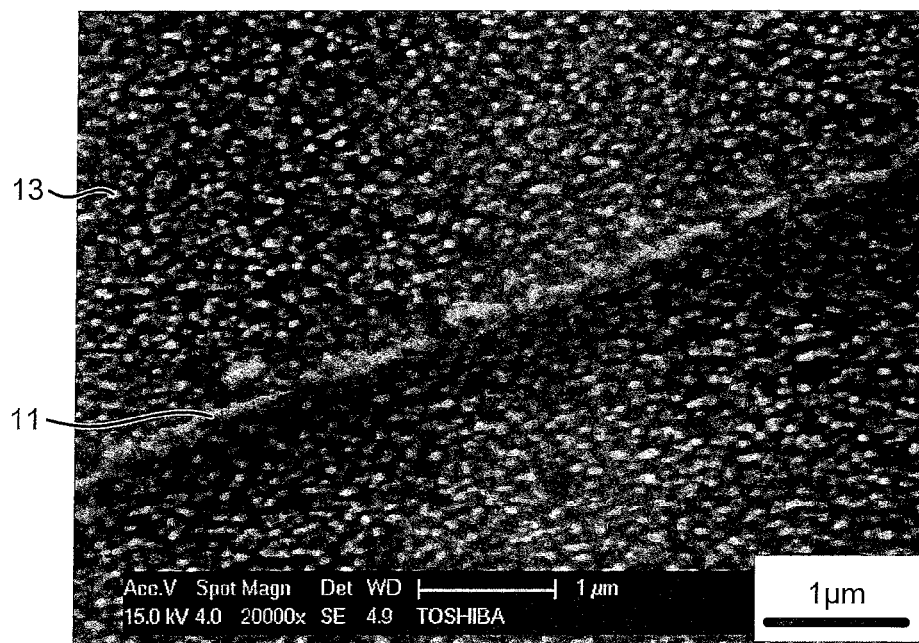


FIG. 3



REFERENCES CITED IN THE DESCRIPTION

This list of references cited by the applicant is for the reader's convenience only. It does not form part of the European patent document. Even though great care has been taken in compiling the references, errors or omissions cannot be excluded and the EPO disclaims all liability in this regard.

Non-patent literature cited in the description

- **LABUSCH.** *Advances in Physics*, vol. 3 (12), 446-507 **[0017]**
- *Phisica status solidi (b)*, vol. 41 (2), 659-669 **[0017]**