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(54) **STEEL**

(57) A steel according to an aspect of the present invention includes, as a chemical composition, by unit mass%, C: 0.15% to 0.40%, Mn: 0.10% to 1.50%, S: 0.002% to 0.020%, Ti: 0.005% to 0.050%, B: 0.0005% to 0.0050%, Bi: 0.0010% to 0.0100%, P: 0.020% or less,

N: 0.0100% or less, Si: 0% or more and less than 0.30%, Cr: 0% to 1.50%, Al: 0% to 0.050%, Mo: 0% to 0.20%, Cu: 0% to 0.20%, Ni: 0% to 0.20%, Nb: 0% to 0.030%, and a remainder including Fe and impurities.

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**Description**

[Technical Field of the Invention]

5 **[0001]** The present invention relates to a steel.

[Related Art]

10 **[0002]** Compared to hot forging, cold forging (including form rolling) can ameliorate surface conditions, dimensional accuracy, and the like of a product and realizes a favorable yield. Therefore, cold forging has been widely applied as a method of manufacturing relatively small-sized machine parts such as bolts. In a case where a machine part is manufactured by performing cold forging, as a material, for example, a carbon steel (medium carbon) for machine structural use or an alloy steel regulated in JIS G 4051, JIS G 4052, JIS G 4104, JIS G 4105, JIS G 4106, and the like are used. For example, a final product is often acquired through a manufacturing process of hot wire rod rolling, annealing (or spheroidizing annealing), wire-drawing, cold forging, quenching/tempering. The foregoing general manufacturing process is characterized by applying a process of annealing or spheroidizing annealing before cold forging. Annealing or spheroidizing annealing is applied before cold forging due to manufacturing problems as follows. Hardness of a rolled material is high in a carbon steel (medium carbon) or an alloy steel in a state of being hot-rolled with no additional treatment (that is, in a case of being air-cooled without performing heat treatment after hot rolling). The manufacturing cost increases due to significant wear on a die at the time of cold forging. Cracking is likely to occur at the time of cold forging due to insufficient ductility of a material in a state of being hot-rolled with no additional treatment, so that the yield is degraded.

20 **[0003]** However, since annealing is significantly costly, in order to reduce the manufacturing cost of components, development of steels which enables omission of an annealing process has been demanded. In response to such request, a so-called boron steel for bolts realized by adding a slight amount of B to a steel has been developed (for example, Patent Document 1 and Patent Document 3). A boron steel is characterized by the factors as follows. Hardness of a wire rod in a state of being hot-rolled with no additional treatment is decreased and ductility is improved by reducing the carbon content in the steel and addition amounts of alloying elements such as Cr and Mo, so that annealing is no longer necessary. Degradation in hardenability due to reduction of the addition amounts of alloying elements is supplemented with an effect of improving hardenability by adding a slight amount of B, which does not increase the hardness of a rolled material.

25 **[0004]** In order to manifest the effect of improving hardenability by adding a slight amount of B, B is required to be solid-solubilized in austenite. Meanwhile, in a case where solid-solubilized nitrogen is present in a steel, BN is generated and the amount of solute B (B solid-solubilized in a steel) is reduced, so that the effect of B improving hardenability is lost. Therefore, in a boron steel, generation of BN is generally suppressed by adding Ti having strong chemical attraction to N and fixing N in the steel in advance as TiN. For example, Patent Document 4 discloses that precipitation of BN is suppressed by setting Ti/N (mass% ratio) to 4 or more. In principle, if Ti/N is 3.42 or more, precipitation of BN can be suppressed.

35 **[0005]** However, in such general boron steels, compared to steels in the related art, so-called coarse grains are likely to be generated, that is, a portion of austenite grains causes abnormal grain growth at the time of heating for quenching and is thereby coarsened. In a component in which coarse grains are generated, deterioration in dimensional accuracy due to an increase in thermal strain generated at the time of quenching, and degradation in properties of the component, such as impact values, fatigue strength, and delayed fracture properties are caused. Therefore, particularly in high strength bolts of which tensile strength is 800 MPa or more, prevention of generation of coarse grains is a major practical problem. In order to suppress generation of such coarse grains generated due to abnormal grain growth, it is effective to disperse many pinning particles (precipitates or the like), that is, to disperse a large amount of fine particles in structures to realize pinning of the grain boundaries of austenite grains.

40 **[0006]** Two main reasons that coarse grains are likely to be generated in a boron steel are as follows.

50 (1) In a case where a boron steel is adopted as a component material, the annealing process for the boron steel after cold forging is omitted. Therefore, the boron steel is directly heated to an austenite region from a cold working structure. In this case, austenite grains are excessively refined and the grain size becomes partially inhomogeneous due to the influence of cold working. Therefore, a portion of the grains is in a state where abnormal grain growth is likely to be caused.

55 (2) In the boron steel described above, N in a steel is fixed as TiN due to the added Ti. Therefore, AlN which effectively acts as pinning particles is not generated in a carbon steel or an alloy steel (a steel in the related art). Since TiN is coarser than AlN, TiN cannot be finely dispersed, so that it is difficult to ensure the number of pinning particles required to prevent coarse grains.

**[0007]** In order to omit the annealing process, the factor of (1) is unavoidable. Therefore, in order to ameliorate the factor of (2), the way of ensuring the number of pinning particles in a boron steel has been regarded as the point of preventing generation of coarse grains.

**[0008]** In such circumstances, technologies for preventing generation of coarse grains in a boron steel have been proposed. For example, Patent Document 5 and Patent Document 6 disclose that TiC and Ti(CN) which are precipitates finer than TiN are utilized as pinning particles, instead of AlN or TiN. In these technologies, in order to ensure the number of pinning particles required to prevent coarse grains, dispersion of TiC and Ti(CN) of 20 pieces /100  $\mu\text{m}^2$  or more in total number having a diameter of 0.2  $\mu\text{m}$  or smaller in a steel before heating for quenching and after hot rolling is regulated. A large amount of such fine precipitates are dispersed in advance before heating for quenching, so that these precipitates function as pinning particles for pinning austenite grain boundaries at the time of heating for quenching. According to this technology, coarse grains can be stably prevented from being generated in a boron steel. Therefore, steels in which this technology is applied are currently in wide use as inexpensive steels for bolts in which the annealing process can be omitted.

**[0009]** However, the foregoing technology has a disadvantage. That is, in a case where a large amount of fine TiC or Ti(CN) is dispersed in a structure after hot rolling, there is an adverse reaction in which hardness of ferrite is increased due to precipitation strengthening by fine precipitate particles, thereby resulting in a problem that the effect of a boron steel softening a hot rolled material is decreased. That is, in a case where the amount of fine TiC or Ti(CN) is increased, generation of coarse grains can be suppressed, but the service life of a die for cold forging is degraded by increasing the hardness of a rolled material due to precipitation strengthening. On the contrary, if the amount of fine TiC or Ti(CN) is suppressed, the hardness of a rolled material can be suppressed, but coarse grains are generated. That is, in a case where fine TiC or Ti(CN) is utilized, there is an incompatible relationship between suppression of generation of coarse grains and suppression of hardness of a rolled material before cold forging. Therefore, it is difficult to completely achieve both softening of a rolled material and suppression of stable coarse grains by only the foregoing technology.

**[0010]** Patent Document 7 also discloses a technical idea similar to that in the foregoing technology of preventing generation of coarse grains in a boron steel. That is, in the technology, the relationship among the amounts of Ti, Nb, Al, and N is set to be within certain ranges, so that carbonitride of these elements is dispersed in a steel, and coarsening of grains is prevented. Moreover, Patent Document 7 discloses an effect of enhancing machinability by adding 0.01 % or more of Bi. However, Patent Document 7 discloses only the effect of enhancing machinability, as the effect of Bi. There is no disclosure related to a relationship between Bi and the properties of coarsening of grains at all. Since Bi is added for the purpose of the effect of improving machinability, only a case of adding a relatively large amount of Bi is examined in Patent Document 7. In this case, as disclosed in Patent Document 7, there is concern that hot workability will be degraded due to added Bi.

**[0011]** Patent Document 8 discloses a case hardening steel. An object thereof is to provide a case hardening steel which exhibits excellent grain coarsening resistance even in a case where carburizing is performed at a higher temperature than an example in the related art, and has excellent cold workability even if softening annealing is not performed. However, in Patent Document 8 as well, only utilization of fine Ti carbide, Ti-containing composite carbide, and the like is proposed as a way of ensuring grain coarsening resistance. In Patent Document 8, the hot rolling temperature is set to be extremely low in order to ensure cold workability. Therefore, productivity of a case hardening steel is impaired.

[Prior Art Document]

[Patent Document]

**[0012]**

[Patent Document 1] Japanese Unexamined Patent Application, First Publication No. H5-339676

[Patent Document 2] Japanese Examined Patent Application, Second Publication No. H5-63524

[Patent Document 3] Japanese Unexamined Patent Application, First Publication No. S61-253347

[Patent Document 4] Japanese Unexamined Patent Application, First Publication No. H3-47918

[Patent Document 5] Japanese Patent No. 3443285

[Patent Document 6] Japanese Patent No. 3490293

[Patent Document 7] Japanese Unexamined Patent Application, First Publication No. 2000-328189

[Patent Document 8] Japanese Unexamined Patent Application, First Publication No. 2006-265704

[Disclosure of the Invention]

[Problems to be Solved by the Invention]

**[0013]** One of problems of steels for cold forging is softly maintaining a steel without performing annealing after hot rolling and before cold forging and without adopting manufacturing conditions impairing productivity in order to improve cold forgeability of the steel and productivity of the steel. Another problem of steels for cold forging is exhibiting high hardenability after cold forging in order to apply high strength to machine parts. Then, further another problem of steels for cold forging is suppressing generation of coarse grains during quenching after cold forging in order to prevent deterioration in dimensional accuracy of machine parts, impact values, fatigue strength, delayed fracture properties, and the like. As described above, technologies in the related art cannot solve all of these problems at the same time. In utilization of TiC and Ti(CN) proposed in a technology in the related art as a way of suppressing generation of coarse grains, a steel after hot rolling and before cold forging is fully-hardened through precipitation strengthening, so that cold forgeability and productivity of the steel are impaired.

**[0014]** The present invention has been made in consideration of the foregoing problems. That is, an object of the present invention is to provide a steel in which generation of coarse grains at the time of quenching is suppressed without using Ti carbide and Ti carbonitride, such as TiC and Ti(CN), so that all of manufacturability, cold forgeability, and mechanical properties after quenching are excellent.

[Means for Solving the Problem]

**[0015]** The gist of the present invention is as follows.

**[0016]**

(1) According to an aspect of the present invention, there is provided a steel including, as a chemical composition, by unit mass%, C: 0.15% to 0.40%, Mn: 0.10% to 1.50%, S: 0.002% to 0.020%, Ti: 0.005% to 0.050%, B: 0.0005% to 0.0050%, Bi: 0.0010% to 0.0100%, P: 0.020% or less, N: 0.0100% or less, Si: 0% or more and less than 0.30%, Cr: 0% to 1.50%, Al: 0% to 0.050%, Mo: 0% to 0.20%, Cu: 0% to 0.20%, Ni: 0% to 0.20%, Nb: 0% to 0.030%, and a remainder including Fe and impurities.

(2) The steel according to (1), may include, as the chemical composition, by unit mass%, one or more selected from the group consisting of Si: 0.01% or more and less than 0.30%, Cr: 0.01% to 1.50%, and Al: 0.001% to 0.050%.

(3) The steel according to (1) or (2), may include, as the chemical composition, by unit mass%, one or more selected from the group consisting of Mo: 0.02% to 0.20%, Cu: 0.02% to 0.20%, Ni: 0.02% to 0.20%, and Nb: 0.002% to 0.030%.

(4) In the steel according to any one of (1) to (3), a N fixation index  $I_{FN}$  defined by the following Expression 1 may be 0 or more.

$$I_{FN}=[Ti]-3.5\times[N] \dots (\text{Expression 1})$$

Here, [Ti] indicates a Ti content by unit mass% and [N] indicates a N content by unit mass%.

(5) In the steel according to any one of (1) to (4), a Ti-Nb-based precipitate generation index  $I_p$  defined by the following Expression 2 may be 0.0100 or less.

$$I_p=0.3\times[Ti]+0.15\times[Nb]-[N] \dots (\text{Expression 2})$$

Here, [Ti] indicates a Ti content by unit mass%, [Nb] indicates a Nb content by unit mass%, and [N] indicates a N content by unit mass%.

[Effects of the Invention]

**[0017]** According to the present invention, it is possible to provide a steel in which both softening before cold forging and suppressing of generation of coarse grains at the time of quenching after cold forging can be achieved. In addition, the steel according to the present invention can be manufactured under conditions within a range in which cracking is not generated at the time of casting, at the time of rolling, and the like and no load is applied to manufacturing equipment, thereby being excellent in manufacturability. When the steel according to the present invention is applied to cold-forged

components, wear on a die at the time of cold forging is suppressed and the service life of a die can be improved. In addition, when the steel according to the present invention is applied to cold-forged components, the cost of expensive dies can be reduced. Therefore, it is possible to particularly contribute to reduction of the manufacturing cost of high strength bolts of which tensile strength is 800 MPa or more. Moreover, the steel according to the present invention also has excellent machinability. Therefore, industrial contribution of the present invention is extremely significant.

[Embodiment of the Invention]

**[0018]** A steel according to an embodiment of the present invention will be described. The steel according to the present embodiment has the following features.

**[0019]**

(a) The steel according to the present embodiment includes, as a chemical composition, by unit mass%, C: 0.15% to 0.40%, Mn: 0.10% to 1.50%, S: 0.002% to 0.020%, Ti: 0.005% to 0.050%, B: 0.0005% to 0.0050%, Bi: 0.0010% to 0.0100%, P: 0.020% or less, N: 0.0100% or less, Si: 0% or more and less than 0.30%, Cr: 0% to 1.50%, Al: 0% to 0.050%, Mo: 0% to 0.20%, Cu: 0% to 0.20%, Ni: 0% to 0.20%, Nb: 0% to 0.030%, and a remainder including Fe and impurities.

(b) The steel according to (a) may include, as the chemical composition, by unit mass%, one or more selected from the group consisting of Si: 0.01 % or more and less than 0.30%, Cr: 0.01% to 1.50%, and Al: 0.001% to 0.050%.

(c) The steel according to (a) or (b) may include, as the chemical composition, by unit mass%, one or more selected from the group consisting of Mo: 0.02% to 0.20%, Cu: 0.02% to 0.20%, Ni: 0.02% to 0.20%, and Nb: 0.002% to 0.030%.

(d) In the steel according to any one of (a) to (c), a N fixation index  $I_{FN}$  defined by the following Expression 1 may be 0 or more.

$$I_{FN}=[Ti]-3.5\times[N] \dots (\text{Expression 1})$$

Here, [Ti] indicates a Ti content by unit mass% and [N] indicates a N content by unit mass%.

(e) In the steel according to any one of (a) to (d), a Ti-Nb-based precipitate generation index  $I_P$  defined by the following Expression 2 may be 0.0100 or less.

$$I_P=0.3\times[Ti]+0.15\times[Nb]-[N] \dots (\text{Expression 2})$$

Here, [Ti] indicates a Ti content by unit mass%, [Nb] indicates a Nb content by unit mass%, and [N] indicates a N content by unit mass%.

**[0020]** In addition, it is possible to obtain bolts which have excellent productivity and in which no coarse grains are generated by performing bolt working, quenching, and tempering through a known method with respect to the steel according to the present embodiment.

**[0021]** The inventors have examined a technology of suppressing generation of coarse grains, separately from a technology in the related art, in which hardness of ferrite is noticeably increased due to precipitation strengthening such that hardness of a steel is increased, and TiC, Ti(CN), and the like which are particles impairing cold workability of a steel are finely dispersed. The foregoing features are based on the following knowledge obtained by the inventors who have intensely investigated a technology of suppressing abnormal grain growth of austenite grains at the time of heating for quenching of a steel.

(1) With an extremely slight amount of Bi, such as 0.0100% or less, abnormal grain growth of austenite grains at the time of heating for quenching is suppressed, and cold-worked parts excellent in dimensional accuracy, mechanical properties, and the like can be obtained.

(2) Due to the effect of Bi described above, without depending on precipitates (TiC, Ti(CN), NbC) which have been utilized as pinning particles in the related art (that is, without impairing cold workability of a steel), abnormal grain growth of austenite grains can be suppressed. Accordingly, hardness of a rolled material after hot rolling is suppressed, and cold workability of a steel can be enhanced.

(3) Meanwhile, it is ascertained that if the Bi content exceeds 0.0100%, hot ductility of a steel is degraded, so that cracking and imperfections are likely to be generated during manufacturing processes of the steel (processes of casting, rolling, and the like), and thus the yield of the steel is degraded. Moreover, it is also ascertained that if the

Bi content exceeds 0.0100%, intergranular embrittlement occurs in a steel after quenching, so that mechanical properties of the steel are impaired. Therefore, it is also ascertained that although it is essential to include Bi in the steel according to the present embodiment, the content thereof is required to be suppressed to an extremely low level.

**[0022]** Hereinafter, the steel according to the present embodiment will be described in detail.

**[0023]** First, the chemical composition of the steel of the present invention will be described. Hereinafter, the unit "%" regarding the chemical composition indicates "mass%".

[C: 0.15% to 0.40%]

**[0024]** C is an element which is necessary to enhance strength of a steel having a tempered martensite structure. In order to set tensile strength after quenching to 800 MPa or more, the C content is required to be set to 0.15% or more. A preferable lower limit for the C content is 0.17%, 0.19%, or 0.23%.

**[0025]** On the other hand, if the C content exceeds 0.40%, hardness of a rolled material after hot rolling is excessively increased, so that the service life of a die for cold forging is significantly degraded. Therefore, the upper limit for the C content is set to 0.40%. A preferable upper limit for the C content is 0.35%, 0.34%, 0.33%, or 0.30%.

[Mn: 0.10% to 1.50%]

**[0026]** Mn is an element which is effective in improving hardenability of a steel. In order to ensure hardenability necessary to obtain martensite through quenching, the Mn content is required to be set to 0.10% or more. A preferable lower limit for the Mn content is 0.20%, 0.35%, or 0.40%.

**[0027]** On the other hand, if the Mn content exceeds 1.50%, hardness of a rolled material after hot rolling and before cold forging is excessively increased, so that the service life of a die for cold forging is significantly degraded. Therefore, the upper limit for the Mn content is set to 1.50%. A preferable upper limit for the Mn content is 1.30%, 1.00%, or 0.80%.

[S: 0.002% to 0.020%]

**[0028]** S is present in a steel as MnS, TiS, and  $Ti_2C_2S$  and has an effect of suppressing abnormal grain growth of austenite grains by serving as pinning particles at the time of heating for quenching. Therefore, the S content is required to be set to 0.002% or more. A preferable lower limit for the S content is 0.003%.

**[0029]** However, in the steel according to the present embodiment, since abnormal grain growth is suppressed by using Bi, the S content may be less than that in the technology in the related art. Moreover, if the S content exceeds 0.020%, S embrittles prior austenite grain boundaries of a steel after quenching, so that delayed fracture resistance (hydrogen embrittlement resistance) are degraded. Furthermore, since  $Ti_2C_2S$  described above is particles impairing machinability of a steel, there is concern that if the S content exceeds 0.020%, machinability of a steel will be deteriorated. Therefore, the S content is required to be limited to 0.020% or less. Preferably, the upper limit value for the S content is 0.015%, 0.010%, or 0.005%.

[Ti: 0.005% to 0.050%]

**[0030]** Ti forms compounds with C, N, and S in a steel and is present in the steel as Ti-based inclusions such as TiN, Ti(CN), TiC, TiS, and  $Ti_2C_2S$ . Ti has the effect of suppressing abnormal grain growth of austenite grains by serving as pinning particles at the time of heating for quenching. In addition, Ti has strong chemical attraction to a solute N in a steel. Therefore, Ti is an element extremely effective in fixing the solute N in a steel in advance as TiN and suppressing generation of BN. In a boron steel, in order to ensure the amount of a solute B which is effective in improving hardenability, there is a need to suppress generation of BN. Thus, the Ti content is required to be set to 0.005% or more. A preferable lower limit for the Ti content is 0.010%, 0.015%, or 0.020%.

**[0031]** However, in the steel according to the present embodiment, since abnormal grain growth is suppressed by using Bi, the Ti content may be less than that in the technology in the related art. Moreover, if the Ti content exceeds 0.050%, Ti-based inclusion particles cause precipitation strengthening, and hardness of a rolled material after hot rolling excessively is increased, so that the service life of a die for cold forging is significantly degraded. In order to suppress hardness of a rolled material after hot rolling while increasing the amount of Ti-based inclusion particles, the hot rolling temperature is required to be lowered. However, it is not preferable in regard to productivity, the service life of equipment, and the like. Moreover, in a case where the Ti content is increased, a large amount of  $Ti_2C_2S$  which is particles impairing machinability of a steel is generated, and machinability is deteriorated, so that it is difficult to apply machining to the steel according to the present embodiment. Therefore, the upper limit for the Ti content is set to 0.050%. A preferable Ti content is 0.040% or less, 0.030% or less, less than 0.030%, or 0.025% or less.

[B: 0.0005% to 0.0050%]

**[0032]** B is an element which contributes to improvement of hardenability of a steel in a case where a slight amount thereof is contained. B achieves an effect of improving hardenability and can increase hardness after cold forging and quenching, without increasing hardness of a rolled material after hot rolling and before cold forging. B is an essential element particularly for a boron steel for bolts. In addition, B has an effect of suppressing fractures at grain boundaries by being segregated in prior austenite grain boundaries and strengthening the prior austenite grain boundaries. In a case of achieving the foregoing effect, the B content is required to be set to 0.0005% or more. Preferably, the lower limit value for the B content is 0.0010%, 0.0012%, or 0.0015%.

**[0033]** On the other hand, if the B content exceeds 0.0050%, the effect thereof is saturated. Therefore, the B content is set to 0.0050% or less. Preferably, the upper limit value for the B content is 0.0030%, 0.0025%, 0.0020%, or 0.0018%.

[Bi: 0.0010% to 0.0100%]

**[0034]** There has been no precedent examined in detail so far regarding the influence of a slight amount of Bi within a range of approximately 0.0010% to 0.0100% on the structure during quenching of a steel. The inventors have found that a slight amount of Bi exhibits an effect of preventing generation of coarse grains by suppressing abnormal grain growth of austenite grains at the time of heating for quenching. In addition, since Bi content necessary to suppress abnormal grain growth is a slight amount, the inventors have also found that the foregoing effect of Bi suppressing generation of coarse grains at the time of heating for quenching can be achieved without increasing hardness of a rolled material after hot rolling. In a case of achieving the foregoing effect, the Bi content is required to be set to 0.0010% or more. A preferable lower limit value for the Bi content is 0.0020%, 0.0025%, or 0.0030%.

**[0035]** On the other hand, if the Bi content exceeds 0.0100%, not only the effect thereof is saturated, but also hot ductility of a steel is degraded, so that cracking and imperfections are likely to be generated during manufacturing processes of the steel (processes of casting, rolling, and the like), and thus the yield is degraded. Moreover, if the Bi content exceeds 0.0100%, intergranular embrittlement occurs in a steel after quenching, so that mechanical properties of the steel are impaired. Therefore, the Bi content is set to 0.0100% or less. The Bi content is preferably less than 0.0100%, 0.0080% or less, or 0.0060% or less.

[P: 0.020% or less]

**[0036]** P is impurities and is an element which embrittles prior  $\gamma$  grain boundaries and degrades delayed fracture resistance (hydrogen embrittlement resistance) of a steel. Therefore, the P content is required to be limited to 0.020% or less. Preferably, the upper limit value for the P content is 0.015%, 0.013%, or 0.010%.

**[0037]** P is not necessary to solve the problems of a steel according to the present embodiment. Therefore, the lower limit value for the P content is 0%. However, in order to suppress the cost of a refining process for reducing the P content, the lower limit value for the P content may be set to 0.001%.

[N: 0.0100% or less]

**[0038]** In a case where N forms compounds with B and is present in a steel as BN, the amount of the solute B is reduced, and the effect of B improving hardenability is impaired. Since N is harmful in the steel according to the present embodiment, the lower limit value for the N content is 0%. However, in order to suppress the cost of a refining process for reducing the N content, the lower limit value for the N content may be set to 0.0001%, 0.0005%, or 0.0010%.

**[0039]** In a case where the N content is large, the Ti content necessary to fix N in a steel as TiN is increased, it is desirable that the N content is reduced as much as possible. Therefore, the N content is required to be limited to 0.0100% or less. Preferably, the upper limit value for the N content is 0.0070%, 0.0050%, or 0.0040%.

**[0040]** As necessary, a steel for springs according to the present embodiment may further include one or more selected from the group consisting of Si, Cr, and Al, within the range described below. However, since Si, Cr, and Al are not essential, the lower limit for each of the Si content, the Cr content, and the Al content is 0%.

[Si: 0% or more and less than 0.30%]

**[0041]** As described above, in the steel according to the present embodiment, the lower limit value for the Si content is 0%. However, Si is an element which is effective in improving hardenability of a steel and improving resistance to temper softening of martensite. In a case of achieving the foregoing effect, it is preferable that the Si content is set to be more than 0% or 0.01% or more. The lower limit value for the Si content may be set to 0.05% or 0.15%.

**[0042]** However, if the Si content is 0.30% or more, the degree of increase in hardness of a steel (rolled material) after

hot rolling and before cold forging is increased, so that the service life of a die for cold forging is degraded. Therefore, the Si content is set to less than 0.30%. A preferable upper limit for the Si content is 0.27%, 0.25%, or 0.20%.

[Cr: 0% to 1.50%]

**[0043]** As described above, in the steel according to the present embodiment, the lower limit value for the Cr content is 0%. However, Cr is an element which is effective in improving hardenability of a steel and improving resistance to temper softening of martensite. In a case of achieving the foregoing effect, it is preferable that the Cr content is set to be more than 0% or 0.01% or more. The lower limit value for the Cr content may be set to 0.10%, 0.20%, or 0.30%.

**[0044]** On the other hand, if the Cr content exceeds 1.50%, hardness of a rolled material after hot rolling and before cold forging is excessively increased, so that the service life of a die for cold forging is significantly degraded. Therefore, the upper limit for the Cr content is set to 1.50%. A preferable upper limit for the Cr content is 1.20%, 1.00%, or 0.80%.

[Al: 0% to 0.050%]

**[0045]** Al is an element which is effective in deoxidation of a steel. However, in a case where deoxidation is performed by using other elements (Si, Ti, and the like), Al is not necessarily contained. Therefore, the lower limit value for the Al content is 0%. However, in order to achieve the effect of Al performing deoxidation, it is preferable that the Al content is 0.001% or more, 0.005% or more, or 0.010% or more.

**[0046]** On the other hand, if the Al content exceeds 0.050%, coarse inclusions are generated, so that a problem such as degradation in toughness of a steel becomes noticeable. Therefore, even in a case where Al is contained, the upper limit for the Al content is set to 0.050%. The upper limit for the Al content is preferably 0.040%, 0.030%, or 0.025%.

**[0047]** As necessary, the steel for springs according to the present embodiment may further include one or more selected from the group consisting of Mo, Cu, Ni, and Nb, within the range described below. However, since Mo, Cu, Ni, and Nb are not essential, the lower limit for each of the Mo content, the Cu content, the Ni content, and the Nb content is 0%.

[Mo: 0% to 0.20%]

**[0048]** As described above, in the steel according to the present embodiment, the lower limit value for the Mo content is 0%. However, Mo is an element which contributes to improvement of hardenability of a steel even if the content thereof is small. In a case of achieving the foregoing effect, it is preferable that the Mo content is set to 0.02% or more. More preferably, the lower limit value for the Mo content is 0.03%, 0.04%, or 0.05%.

**[0049]** On the other hand, since Mo is an expensive alloying element, if the Mo content is more than 0.20%, it is disadvantageous to the manufacturing cost. Therefore, even in a case where Mo is contained, the Mo content is set to 0.20% or less. Preferably, the upper limit value for the Mo content is 0.16%, 0.13%, or 0.10%.

[Cu: 0% to 0.20%]

**[0050]** As described above, in the steel according to the present embodiment, the lower limit value for the Cu content is 0%. However, Cu is an element which improves corrosion resistance of a steel. In a case of achieving the foregoing effect, it is preferable that the Cu content is set to 0.02% or more. More preferably, the lower limit value for the Cu content is 0.05%.

**[0051]** On the other hand, if the Cu content exceeds 0.20%, hot ductility of a steel is degraded, so that a problem such as impairment of manufacturability at the time of continuous casting becomes noticeable. Therefore, even in a case where Cu is contained, the Cu content is set to 0.20% or less. Preferably, the upper limit value for the Cu content is 0.15%, 0.10%, or 0.08%.

[Ni: 0% to 0.20%]

**[0052]** As described above, in the steel according to the present embodiment, the lower limit value for the Ni content is 0%. However, Ni is an element which improves corrosion resistance of a steel and is an element which is also effective in improvement of toughness of a steel. In a case of achieving the foregoing effect, it is preferable that the Ni content is set to 0.02% or more. More preferably, the lower limit value for the Ni content is 0.03%, 0.04%, or 0.05%.

**[0053]** On the other hand, since Ni is an expensive alloying element, if the Ni content exceeds 0.20%, it is disadvantageous to the manufacturing cost. Therefore, even in a case where Ni is contained, the Ni content is set to 0.20% or less. Preferably, the upper limit value for the Ni content is 0.15%, 0.12%, 0.10%, or 0.08%.



[Nb: 0% to 0.030%]

**[0054]** As described above, in the steel according to the present embodiment, the lower limit value for the Nb content is 0%. However, Nb forms compounds with C in a steel and is present in the steel as Nb-based inclusions such as NbC or TiNb(CN). Nb has the effect of suppressing abnormal grain growth of austenite grains, as pinning particles at the time of heating for quenching. In a case of achieving the foregoing effect, it is preferable that the Nb content is set to 0.002% or more. More preferably, the lower limit value for the Nb content is 0.003%, 0.005%, or 0.006%.

**[0055]** On the other hand, if the Nb content exceeds 0.030%, not only the effect thereof is saturated, but also Nb-based inclusions cause precipitation strengthening, so that manufacturability at the time of continuous casting is impaired. Otherwise, in this case, since Nb-based inclusions cause precipitation strengthening, hardness of a rolled material after hot rolling is excessively increased. Therefore, if the Nb content exceeds 0.030%, problems such as degradation in manufacturability and significant degradation in the service life of a die for cold forging become noticeable. Therefore, even in a case where Nb is contained, the Nb content is set to 0.030% or less. Preferably, the upper limit value for the Nb content is 0.015%, 0.013%, or 0.010%.

**[0056]** The steel according to the present embodiment includes the foregoing alloy composition, and the remainder of the chemical composition includes Fe and impurities. In the present embodiment, impurities are compositions which are incorporated due to raw materials such as ores and scraps, and other factors when a steel is industrially manufactured. Impurities mean elements of an amount at a level at which the operational effects of the steel according to the present embodiment are not impaired.

[N fixation index  $I_{FN}$ : preferably 0 or more]

**[0057]** In order to achieve the effect of containing B as described above, there is a need to suppress generation of BN by reducing N (solute N) solid-solubilized in a steel. Therefore, it is desirable that the amount of N in a steel is reduced and Ti is contained in the steel, so that N is stably fixed as a form of TiN and the amount of the solute N is thereby reduced. In order to achieve the foregoing effects by fixing N due to Ti, it is preferable that the N fixation index  $I_{FN}$  defined by the following Expression 1 is set to 0 or more. The lower limit value for the N fixation index  $I_{FN}$  may be set to 0.0005, 0.0010, 0.0014, or 0.0050. However, even if the N fixation index  $I_{FN}$  is not particularly limited, as long as the Ti content and the N content are controlled to be within the range described above, the steel according to the present embodiment is softened before cold forging, and generation of coarse grains at the time of quenching can be suppressed.

$$I_{FN}=[Ti]-3.5\times[N] \dots (\text{Expression 1})$$

[Ti] and [N] in the foregoing Expression 1 indicate the Ti content and the N content in a steel by unit mass%, and in a case where these elements are not contained, the contents thereof are set to 0%.

[Ti-Nb-based precipitate generation index  $I_P$ : preferably 0.0100 or less]

**[0058]** As described above, it is preferable that N is fixed as TiN by using Ti, and the amount of the solute N is reduced. However, it is not preferable to contain an amount of Ti exceeding the amount necessary to fix TiN. As described above, there is concern that Ti will bond with C, S, and the like and form fine precipitates, so that these fine precipitates will adversely affect the properties of the steel according to the present embodiment. In addition, the inventors have found that Nb also has a similar operation as Ti.

**[0059]** Specifically, fine Ti-Nb-based precipitates, such as TiC, Ti(CN), NbC, TiNb(CN), and  $Ti_2C_2S$  which are precipitates present in a steel, have the effect of suppressing generation of coarse grains by suppressing abnormal grain growth of austenite grains at the time of heating for quenching as pinning particles. However, in a case where a large amount of these Ti-Nb-based precipitate particles are dispersed in a structure after hot rolling, there is adverse reaction such as an increase in hardness of ferrite caused by precipitation strengthening due to fine precipitate particles. Therefore, in a case where an excessively large amount of these Ti-Nb-based precipitate particles are dispersed in a steel, hardness of a rolled material after hot rolling is excessively increased, so that a problem such as significant degradation in the service life of a die for cold forging becomes noticeable. Moreover, as described above,  $Ti_2C_2S$  causes deterioration in machinability. Therefore, in the steel according to the present embodiment, it is preferable that the amounts of these Ti-Nb-based precipitate particles are limited.

**[0060]** In order to suppress hardness after rolling, that is, after hot rolling, it is desirable that the Ti-Nb-based precipitate generation index  $I_P$  calculated by using the following Expression 2 is set to 0.0100 or less. The Ti-Nb-based precipitate generation index  $I_P$  may be set to 0.0075 or less, less than 0.0050, 0.0045 or less, 0.0040 or less, or 0.0035 or less.

However, even if the Ti-Nb-based precipitate generation index  $I_p$  is not particularly limited, as long as the Ti content, the Nb content, and the N content are controlled to be within the range described above, the steel according to the present embodiment is softened before cold forging, and generation of coarse grains at the time of quenching can be suppressed.

$$I_p = 0.3 \times [\text{Ti}] + 0.15 \times [\text{Nb}] - [\text{N}] \dots (\text{Expression 2})$$

[Ti], [N], and [Nb] in the foregoing Expression 2 indicate the Ti content, the N content, and the Nb content in a steel by unit mass%, and in a case where these elements are not contained, the contents thereof are set to 0%.

**[0061]** Next, a suitable method of manufacturing the steel of the present embodiment will be described.

**[0062]** In order to manufacture the steel of the present embodiment, a steel having the chemical composition described above is smelted in a converter. As necessary, the steel passes through a secondary refining process and is formed into a cast slab by performing continuous casting. This cast slab is reheated and is subjected to blooming, thereby obtaining a material for wire-rod rolling (steel piece) having a cross section of 162 square mm (the length of 162 mm and the width of 162 mm), for example. Next, the steel piece is heated at a temperature of approximately 1,000°C to 1,280°C. Subsequently, wire-rod rolling is performed to have a wire rod shape with a diameter of 6 mm to 20 mm. Thereafter, the wire rod is wound into a coil shape through hot rolling by a winding device, and then is cooled to room temperature. In this manner, the steel of the present embodiment is obtained.

**[0063]** In the steel according to the present embodiment, the amount of Ti-based precipitate particles causing precipitation strengthening is suppressed. Therefore, in the method of manufacturing the steel according to the present embodiment, there is no need to apply a load to hot rolling equipment by lowering the hot rolling temperature in order to suppress hardness of a steel. In addition, imperfections such as cracking and defects due to an increase in hardness are unlikely to be generated in a steel. Moreover, in the steel according to the present embodiment, hardness thereof is suppressed without performing annealing after hot rolling. Therefore, the steel according to the present embodiment is also excellent for having high productivity.

**[0064]** According to the steel of the present embodiment, softening before cold forging and suppression of generation of coarse grains at the time of quenching can be compatible. In addition, since cracking is not generated at the time of casting or at the time of rolling, the steel of the present embodiment is excellent in manufacturability.

**[0065]** Since the hardness of the steel according to the present embodiment can be suitably adjusted in accordance with its usage, it is not particularly limited. However, in a case where there is a need to ensure cold forgeability, it is suitable for the hardness of the steel according to the present embodiment to be set to Hv 180 or less, and it is more suitable to be set to Hv 170 or less, or Hv 160 or less. The lower limit value for the hardness of the steel according to the present embodiment is not particularly limited. However, in consideration of the chemical composition thereof, it is assumed that the lower limit value therefor is substantially set to approximately Hv 130 or approximately Hv 140. The hardness of the steel according to the present embodiment can be within the suitable range described above even if annealing is not performed after hot rolling. In addition, the steel according to the present embodiment is also excellent in machinability.

**[0066]** In addition, for example, in a case where the steel according to the present embodiment is heated to a temperature of 840°C to 1,100°C and is retained for 30 minutes, and then quenching is performed under a condition of performing water cooling or oil cooling and tempering treatment in which the steel according to the present embodiment is heated and retained at a temperature within a range of 150°C to 450°C is further performed, tensile strength thereof can be set to 800 MPa or more. Therefore, the steel according to the present embodiment becomes suitable for a material of components which require high strength. However, in a case where the steel according to the present embodiment is used as a steel for quenching, heat treatment conditions are not particularly limited and can be suitably selected in accordance with its usage.

**[0067]** The usage of the steel according to the present embodiment is not particularly limited. It is suitable to be applied to high strength machine parts manufactured by performing cold forging and quenching, and is particularly suitable to be applied to high strength bolts. In a case where the steel according to the present embodiment having high cold forgeability is used as a material of high strength machine parts, wear on a die at the time of cold forging is suppressed and the service life of a die can be improved. In addition, since the cost of expensive dies can be reduced, it is possible to particularly contribute to reduction of the manufacturing cost of high strength bolts of which tensile strength is 800 MPa or more.

[Examples]

**[0068]** Next, the present invention will be described by using Examples. However, the present invention is not limited to the following examples.

**[0069]** First, steels having the chemical composition as shown in Table 1-1 and Table 1-2 were smelted in a converter and were made into cast slabs by performing continuous casting. In Table 1-1 and Table 1-2, regarding the elements of which the amount was equivalent to or less than the level of impurities, the spaces for displaying amounts were left blank. When the N fixation index  $I_{FN}$  and the Ti-Nb-based precipitate generation index  $I_p$  were calculated, the values thereof were regarded as "0 mass%". In addition, in Table 1-1 and Table 1-2, the values beyond the regulation range of the present invention were underlined. The cast slabs obtained as described above were checked whether or not cracking was generated on the surface of the cast slab. When cracking on the surface of the cast slab was checked for, scale on the surface of the cast slab was eliminated by using a check scarf. Thereafter, the cast slab was observed, and the depth of cracking was investigated. The surface of the cast slabs in which cracking having a depth of 1 mm or more was detected on its surface were determined that cracking on the surface of the cast slab was "Present" at the time of continuous casting and were determined to be "Failed" in manufacturability. Table 2-1 and 2-2 show the evaluation results of manufacturability.

**[0070]** As necessary, these cast slabs were subjected to soaking and blooming, and materials for wire-rod rolling (steel pieces) having a cross section of 162 square mm (the length of 162 mm and the width of 162 mm) were obtained. Next, the steel pieces were heated at a temperature of approximately 1,000°C to 1,280°C. Subsequently, wire-rod rolling was performed to have wire rods (steels for springs) having a diameter of 10 mm.

**[0071]** A test piece for measuring Vickers hardness was cut out from the wire rod after rolling. Specifically, the test piece having a cross section including the central axis of the wire rod was cut out in a direction parallel to the rolling direction. After the cut-out cross section was polished, Vickers hardness of a region of the wire rod at the depth of 1/4 (1/4 portion) in a diameter from the surface of the wire rod was measured. The test load was set to 10 kgf, and the average value of four measurements was recorded in Table 2-1 and Table 2-2 as "Hardness after rolling", and this was regarded as an index for estimating the service life of a die for cold forging. Since a sufficient effect of improving the service life of a die for cold forging could not be achieved in the test pieces of which hardness of a rolled material exceeded HV 180, the test pieces were determined to be "Failed" in "cold forgeability". Tables 2-1 and 2-2 show the evaluation results of cold forgeability.

**[0072]** In addition, in order to simulate the influence on wire-drawing or cold forging (cold working) when the wire rods were worked into a bolt shape, the wire rod was subjected to cold drawing with the area reduction of 70%. Thereafter, the wire rod was heated to a temperature of 840°C to 1,100°C for 30 minutes, and quenching (water cooling) was performed. Then, the austenite structures were fixed as prior austenite grain boundaries of martensite structures. Thereafter, as necessary, the quenched test piece was subjected to tempering within a temperature range of an A1 point or less, and the test piece having a cross section including the center of the drawing was cut out in a direction parallel to the rolling/drawing direction. After the cut-out cross section of the test piece was polished, prior austenite grain boundaries were manifested through corrosion, and the grain size of prior austenite after quenching and tempering was measured through observation using an optical microscope. The grain size of prior austenite was measured based on JIS G 0551. Regarding the visual field of measurement, 10 or more visual fields were adopted at magnification of 400 times. The test pieces, in which even one large grain with No. 5 or smaller as the grain size of prior austenite was present, were determined that coarse grains were generated. The minimum (lowest) heating temperature, at which coarse grains were generated and which was clarified by performing observation and measurement of the grain size of prior austenite with respect to the test pieces heated to various temperatures, was defined as the grain coarsening temperature of the test piece and was regarded as the index for grain coarsening resistance. Since the test piece of which the grain coarsening temperature was 900°C or less were poor in grain coarsening resistance, the test piece was determined to be "Failed". Table 2-1 and Table 2-2 show the measurement results of the grain coarsening temperature.

**[0073]** As seen in Table 2-1 and Table 2-2, in A1 to A32 of Examples of the present invention, hardness of the wire rod after rolling was low, and it could be expected that the service life of a die for cold forging would be improved. Therefore, they were excellent in cold forgeability. Even if the test pieces were heated to a temperature exceeding 900°C at the time of heating for quenching after cold working, coarse grains were not generated. Moreover, no cracking was generated on the surface of the cast slabs at the time of continuous casting, so that the scrappage rate of the cast slabs was low. Accordingly, it was obvious that the test pieces were excellent in manufacturability. All of A1 to A32 of Examples of the present invention after being subjected to heat treatment for measuring the grain size of prior austenite described above had tensile strength of 800 MPa or more.

**[0074]** In contrast, in the cases of comparative examples, any of cold forgeability, properties of preventing coarse grains, and manufacturability was poor. That is, in B 1 to B4, hot ductility was degraded due to the excessive addition amount of Bi, and manufacturability was poor. In B5 to B7, Bi was not added or the addition amount was excessively small, so that properties of preventing coarse grains were poor. In B8 and B9, the Ti-Nb-based precipitate generation index  $I_p$  exceeded due to the excessive addition amount of Ti or the small amount of the N content with respect to the addition amount of Ti, so that hardness of the wire rod after rolling was high and cold forgeability was poor.

[Table 1-1]

Symbols	Chemical composition (unit: mass%, remainder: Fe and impurities)															I <sub>FN</sub>	I <sub>P</sub>
	C	Si	Mn	Cr	S	Al	Ti	B	Bi	P	N	Mo	Cu	Ni	Nb		
Examples of Invention	A1	0.25	0.24	0.81	0.31	0.005	0.029	0.025	0.0025	0.0011	0.010	0.0040				0.0110	0.0035
	A2	0.24	0.25	0.80	0.29	0.006	0.030	0.024	0.0023	0.0030	0.009	0.0038				0.0107	0.0034
	A3	0.25	0.25	0.78	0.30	0.004	0.031	0.026	0.0025	0.0099	0.011	0.0042				0.0113	0.0036
	A4	0.25	0.26	0.80	0.28	0.005	0.028	0.025	0.0022	0.0100	0.010	0.0037				0.0121	0.0038
	A5	0.24	0.22	0.82	0.34	0.006	0.030	0.005	0.0025	0.0028	0.009	0.0001				0.0047	0.0014
	A6	0.26	0.20	0.86	0.31	0.008	0.036	0.014	0.0021	0.0029	0.008	0.0036				0.0014	0.0006
	A7	0.25	0.21	0.79	0.29	0.004	0.042	0.020	0.0024	0.0018	0.012	0.0041				0.0057	0.0019
	A8	0.24	0.26	0.83	0.27	0.006	0.031	0.026	0.0029	0.0030	0.013	0.0070				0.0015	0.0008
	A9	0.25	0.25	0.79	0.31	0.006	0.041	0.026	0.0021	0.0025	0.011	0.0026				0.0169	0.0052
	A10	0.25	0.27	0.75	0.33	0.004	0.001	0.050	0.0019	0.0038	0.009	0.0075				0.0238	0.0075
	A11	0.23	0.28	0.84	0.27	0.005	0.026	0.035	0.0034	0.0041	0.008	0.0051				0.0172	0.0054
	A12	0.26	0.24	0.77	0.32	0.007	0.028	0.029	0.0023	0.0052	0.012	0.0044			0.010	0.0136	0.0058
A13	0.22	0.26	0.86	0.33	0.003	0.031	0.024	0.0026	0.0033	0.009	0.0052			0.006	0.0058	0.0029	
A14	0.20	0.25	1.00		0.002	0.029	0.027	0.0021	0.0026	0.010	0.0044				0.0116	0.0037	
A15	0.20	0.24	0.20	1.26	0.005	0.041	0.022	0.0019	0.0043	0.001	0.0058				0.0017	0.0008	
A16	0.35	0.01	0.81	0.32	0.007	0.034	0.028	0.0025	0.0035	0.011	0.0062				0.0063	0.0022	
A17	0.26	0.05	0.35	0.70	0.005	0.030	0.025	0.0025	0.0030	0.010	0.0040				0.0110	0.0035	
A18	0.25	0.26	0.78	0.32	0.004	0.040	0.022	0.0021	0.0044	0.011	0.0052	0.05			0.0038	0.0014	
A19	0.25	0.24	0.81	0.36	0.007	0.036	0.021	0.0029	0.0029	0.013	0.0038		0.02	0.02	0.0077	0.0025	
A20	0.34	0.21	0.75	0.42	0.004	0.020	0.022	0.0018	0.0035	0.009	0.0045				0.0063	0.0021	

Examples of Invention

[Table 1-2]

Symbols	Chemical composition (unit: mass%, remainder: Fe and impurities)															I <sub>FN</sub>	I <sub>P</sub>	
	C	Si	Mn	Cr	S	Al	Ti	B	Bi	P	N	Mo	Cu	Ni	Nb			
Examples of Invention	A21	0.28	0.22	0.50	0.20	0.006	0.025	0.026	0.0028	0.0041	0.012	0.0052	0.13				0.0078	0.0026
	A22	0.31	0.15	0.80	0.35	0.005	0.015	0.027	0.0037	0.0050	0.008	0.0056		0.10	0.12		0.0074	0.0025
	A23	0.25	0.24	0.78	0.28	0.004	0.028	0.024	0.0020	0.0031	0.009	0.0038					0.0107	0.0034
	A24	0.24	0.28	0.72	0.29	0.006	0.031	0.026	0.0022	0.0037	0.011	0.0045					0.0103	0.0033
	A25	0.26	0.29	0.75	0.34	0.004	0.018	0.024	0.0021	0.0044	0.012	0.0043					0.0090	0.0029
	A26	0.25	0.22	0.85	0.33	0.006	0.029	0.031	0.0033	0.0050	0.008	0.0060					0.0100	0.0033
	A27	0.24	0.29	0.69	0.38	0.004	0.019	0.028	0.0027	0.0080	0.013	0.0059					0.0074	0.0025
	A28	0.26	0.23	0.75	0.31	0.006	0.001	0.028	0.0024	0.0028	0.008	0.0042					0.0133	0.0042
Comparative Examples	A29	0.25	0.29	0.68	0.35	0.005	0.010	0.015	0.0026	0.0031	0.012	0.0036					0.0024	0.0009
	A30	0.25	0.23	0.85	0.29	0.007	0.026	0.022	0.0016	0.0035	0.009	0.0049					0.0049	0.0017
	A31	0.26	0.19	0.91	0.34	0.004	0.018	0.026	0.0038	0.0041	0.008	0.0052					0.0078	0.0026
	A32	0.25	0.29	0.73	0.41	0.006	0.029	0.024	0.0048	0.0043	0.011	0.0046					0.0079	0.0026
	B1	0.25	0.24	0.81	0.31	0.006	0.028	0.024	0.0023	<u>0.0105</u>	0.009	0.0039					0.0104	0.0033
	B2	0.24	0.26	0.78	0.33	0.005	0.031	0.022	0.0018	<u>0.0400</u>	0.011	0.0042					0.0073	0.0024
	B3	0.25	0.22	0.76	0.36	0.006	0.025	0.026	0.0031	<u>0.1100</u>	0.009	0.0048					0.0092	0.0030
	B4	0.23	0.27	0.78	0.29	0.004	0.033	0.024	0.0023	<u>0.1300</u>	0.011	0.0051					0.0062	0.0021
	B5	0.26	0.19	0.84	0.31	0.007	0.032	0.023	0.0019	<u>0.0005</u>	0.012	0.0043					0.0080	0.0026
	B6	0.24	0.23	0.91	0.27	0.008	0.029	0.022	0.0018	<u>0.0000</u>	0.010	0.0050					0.0045	0.0016
	B7	0.25	0.24	0.87	0.29	0.006	0.033	0.035	0.0022	<u>0.0000</u>	0.011	0.0062					0.0133	0.0043
	B8	0.26	0.24	0.81	0.33	0.006	0.027	<u>0.055</u>	0.0019	<u>0.0033</u>	0.010	0.0042					0.0403	0.0123
	B9	0.24	0.25	0.79	0.29	0.007	0.031	0.035	0.0023	0.0035	0.009	0.0001					0.0347	0.0104

[Table 2-1]

Examples of Invention	Symbols	Hardness after rolling (HV)	Coarse grain generation temperature (°C)	Cracking on surface of cast slab at time of continuous casting	Cold forgeability	Grain coarsening resistance	Manufacturability
	A1	163	960	None	Passed	Passed	Passed
	A2	160	960	None	Passed	Passed	Passed
	A3	162	960	None	Passed	Passed	Passed
	A4	164	960	None	Passed	Passed	Passed
	A5	153	940	None	Passed	Passed	Passed
	A6	154	940	None	Passed	Passed	Passed
	A7	154	950	None	Passed	Passed	Passed
	A8	150	940	None	Passed	Passed	Passed
	A9	169	980	None	Passed	Passed	Passed
	A10	178	1000	None	Passed	Passed	Passed
	A11	168	980	None	Passed	Passed	Passed
	A12	173	980	None	Passed	Passed	Passed
	A13	157	960	None	Passed	Passed	Passed
	A14	154	960	None	Passed	Passed	Passed
	A15	138	940	None	Passed	Passed	Passed
	A16	173	950	None	Passed	Passed	Passed
	A17	149	960	None	Passed	Passed	Passed
	A18	165	940	None	Passed	Passed	Passed
	A19	160	950	None	Passed	Passed	Passed
	A20	176	950	None	Passed	Passed	Passed

[Table 2-2]

		Hardness after rolling (HV)	Coarse grain generation temperature (°C)	Cracking on surface of cast slab at time of continuous casting	Cold forgeability	Grain coarsening resistance	Manufacturability
Examples of Invention	A21	177	955	None	Passed	Passed	Passed
	A22	170	965	None	Passed	Passed	Passed
	A23	160	960	None	Passed	Passed	Passed
	A24	157	965	None	Passed	Passed	Passed
	A25	162	970	None	Passed	Passed	Passed
	A26	164	960	None	Passed	Passed	Passed
	A27	155	950	None	Passed	Passed	Passed
	A28	165	960	None	Passed	Passed	Passed
	A29	149	950	None	Passed	Passed	Passed
	A30	157	960	None	Passed	Passed	Passed
	A31	165	965	None	Passed	Passed	Passed
	A32	160	960	None	Passed	Passed	Passed
Comparative Examples	B1	162	960	Present	Passed	Passed	Failed
	B2	156	950	Present	Passed	Passed	Failed
	B3	159	960	Present	Passed	Passed	Failed
	B4	152	950	Present	Passed	Passed	Failed
	B5	161	860	None	Passed	Failed	Passed
	B6	156	850	None	Passed	Failed	Passed
	B7	168	880	None	Passed	Failed	Passed
	B8	201	1040	None	Failed	Passed	Passed
	B9	187	1020	None	Failed	Passed	Passed

## [Industrial Applicability]

**[0075]** According to the present invention, it is possible to provide a steel in which both softening at the time of cold forging and suppressing of generation of coarse grains at the time of quenching after cold forging can be achieved. In addition, the steel according to the present invention can be manufactured under conditions within a range in which cracking is not generated at the time of casting, at the time of rolling, and the like and no load is applied to manufacturing equipment, thereby being excellent in manufacturability. When the steel according to the present invention is applied to cold-forged components, wear on a die at the time of cold forging is suppressed and the service life of a die can be improved. In addition, when the steel according to the present invention is applied to cold-forged components, the cost of expensive dies can be reduced. Therefore, it is possible to particularly contribute to reduction of the manufacturing cost of high strength bolts of which tensile strength is 800 MPa or more. Moreover, the steel according to the present invention also has excellent machinability. Therefore, industrial contribution of the present invention is extremely significant.

## Claims

1. A steel comprising, as a chemical composition, by unit mass%,  
C: 0.15% to 0.40%,  
Mn: 0.10% to 1.50%,

S: 0.002% to 0.020%,  
 Ti: 0.005% to 0.050%,  
 B: 0.0005% to 0.0050%,  
 Bi: 0.0010% to 0.0100%,  
 P: 0.020% or less,  
 N: 0.0100% or less,  
 Si: 0% or more and less than 0.30%,  
 Cr: 0% to 1.50%,  
 Al: 0% to 0.050%,  
 Mo: 0% to 0.20%,  
 Cu: 0% to 0.20%,  
 Ni: 0% to 0.20%,  
 Nb: 0% to 0.030%, and  
 a remainder including Fe and impurities.

2. The steel according to claim 1, comprising, as the chemical composition, by unit mass%, one or more selected from the group consisting of  
 Si: 0.01% or more and less than 0.30%,  
 Cr: 0.01% to 1.50%, and  
 Al: 0.001% to 0.050%.

3. The steel according to claim 1 or 2, comprising, as the chemical composition, by unit mass%, one or more selected from the group consisting of  
 Mo: 0.02% to 0.20%,  
 Cu: 0.02% to 0.20%,  
 Ni: 0.02% to 0.20%, and  
 Nb: 0.002% to 0.030%.

4. The steel according to any one of claims 1 to 3,  
 wherein a N fixation index  $I_{FN}$  defined by the following Expression 1 is 0 or more,

$$I_{FN}=[Ti]-3.5\times[N] \dots (\text{Expression 1})$$

where, [Ti] indicates a Ti content by unit mass% and [N] indicates a N content by unit mass%.

5. The steel according to any one of claims 1 to 4,  
 wherein a Ti-Nb-based precipitate generation index  $I_p$  defined by the following Expression 2 is 0.0100 or less,

$$I_p=0.3\times[Ti]+0.15\times[Nb]-[N] \dots (\text{Expression 2})$$

where, [Ti] indicates a Ti content by unit mass%, [Nb] indicates a Nb content by unit mass%, and [N] indicates a N content by unit mass%.



## INTERNATIONAL SEARCH REPORT

International application No.

PCT/JP2016/078558

## A. CLASSIFICATION OF SUBJECT MATTER

C22C38/00(2006.01)i, C22C38/54(2006.01)i

According to International Patent Classification (IPC) or to both national classification and IPC

## B. FIELDS SEARCHED

Minimum documentation searched (classification system followed by classification symbols)

C22C38/00-38/60

Documentation searched other than minimum documentation to the extent that such documents are included in the fields searched

Jitsuyo Shinan Koho 1922-1996 Jitsuyo Shinan Toroku Koho 1996-2016

Kokai Jitsuyo Shinan Koho 1971-2016 Toroku Jitsuyo Shinan Koho 1994-2016

Electronic data base consulted during the international search (name of data base and, where practicable, search terms used)

## C. DOCUMENTS CONSIDERED TO BE RELEVANT

Category*	Citation of document, with indication, where appropriate, of the relevant passages	Relevant to claim No.
A	JP 2014-19904 A (Nippon Steel & Sumitomo Metal Corp.), 03 February 2014 (03.02.2014), (Family: none)	1-5
A	JP 2007-239028 A (Honda Motor Co., Ltd.), 20 September 2007 (20.09.2007), (Family: none)	1-5

☐ Further documents are listed in the continuation of Box C.☐ See patent family annex.

\* Special categories of cited documents:

"A" document defining the general state of the art which is not considered to be of particular relevance

"E" earlier application or patent but published on or after the international filing date

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"O" document referring to an oral disclosure, use, exhibition or other means

"P" document published prior to the international filing date but later than the priority date claimed

"T" later document published after the international filing date or priority date and not in conflict with the application but cited to understand the principle or theory underlying the invention

"X" document of particular relevance; the claimed invention cannot be considered novel or cannot be considered to involve an inventive step when the document is taken alone

"Y" document of particular relevance; the claimed invention cannot be considered to involve an inventive step when the document is combined with one or more other such documents, such combination being obvious to a person skilled in the art

"&amp;" document member of the same patent family

Date of the actual completion of the international search  
08 December 2016 (08.12.16)Date of mailing of the international search report  
20 December 2016 (20.12.16)Name and mailing address of the ISA/  
Japan Patent Office  
3-4-3, Kasumigaseki, Chiyoda-ku,  
Tokyo 100-8915, Japan

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**REFERENCES CITED IN THE DESCRIPTION**

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