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(54) CASE-HARDENED STEEL, CARBURIZED COMPONENT, AND PROCESS FOR PRODUCING CASE-HARDENED STEEL

(57) A case hardening steel includes, as a chemical composition, by mass%, C: 0.10% to 0.30%, Si: 0.02% to 1.50%, Mn: 0.30% to 1.80%, S: 0.003% to 0.020%, Cr: 0.40% to 2.00%, Al: 0.005% to 0.050%, Ti: 0.06% to 0.20%, Bi: 0.0001% to 0.0050%, Mo: 0% to 1.50%, Ni: 0% to 3.50%, V: 0% to 0.50%, B: 0% to 0.0050%, Nb: 0% or more and less than 0.040%, P: limited to 0.050% or less, N: limited to 0.0060% or less, O: limited to 0.0025% or less, and a remainder including an iron and impurities, and satisfies $Ti/S \geq 6.0$, in which, in a longitudinal section, a maximum diameter of Ti-based precipitates predicted by extreme value statistics under a condition that an inspection standard area is 100 mm^2, a number of inspections is 16 visual fields, and an area where prediction is performed is 30,000 mm^2, is 40 μm or less.

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Description

[Technical Field of the Invention]

[0001] The present invention relates to a case hardening steel, a carburized component, and a manufacturing method of a case hardening steel, and particularly to a case hardening steel which is excellent in coarse grain prevention properties and fatigue properties during carburizing, a manufacturing method thereof, and a carburized component which is obtained from the case hardening steel.

[0002] Priority is claimed on Japanese Patent Application No. 2015-256254, filed on December 28, 2015, the content of which is incorporated herein by reference.

[Related Art]

[0003] Carburized components such as gears, bearing components, rolling components, shafts, and constant-velocity joint components are typically manufactured by a method of forging medium carbon alloy steel for machine structural use specified in, for example, JIS G 4052, JIS G 4104, JIS G 4105, and JIS G 4106, machining the forged steel into a predetermined shape, and performing carburizing and quenching thereon.

[0004] During the forging performed when the carburized component is manufactured, cold forging (including rolling) or hot forging is performed. In cold forging, the surface skin and dimensional accuracy of a product are good, manufacturing costs are lower than in hot forging, and good yield is achieved. For this reason, in recent years, as the forging in the manufacturing of a carburized component, cold forging has been increasingly performed instead of hot forging. As a result, the number of carburized components manufactured by performing carburizing and quenching after cold forging has been remarkably increased in recent years.

[0005] However, as a main object of carburized components manufactured by performing carburizing and quenching after cold forging is to reduce heat treatment strain. For example, when a shaft is bent due to heat treatment strain, the function of the shaft is impaired. In addition, high heat treatment strain in gears and constant-velocity joint component causes noise and vibration.

[0006] The greatest cause of heat treatment strain in carburized components is coarse grains generated during carburizing. In the related art, in order to suppress the generation of coarse grains, annealing is performed before carburizing and quenching and after cold forging. However, in recent years, from the viewpoint of cost reduction, there is a strong demand for the omission of annealing. Therefore, there is a strong demand for steel in which coarse grains are not formed during carburizing even when annealing is omitted before carburizing and quenching.

[0007] On the other hand, among gears, bearing components, and rolling components, an object of the bearing components and rolling components that receive high contact pressure is to sufficiently ensure fatigue properties such as rolling fatigue properties. In such components, internal origin type damage is more frequently seen than surface origin type damage, and thus high depth carburizing is performed in order to suppress damage slightly inward of the surface where shear stress is likely to be maximized. However, since this high depth carburizing typically takes a long period of time of a dozen or so hours to several tens of hours, from the viewpoint of energy saving, a reduction in the carburizing time is required. In order to reduce the carburizing time, an increase in the carburizing temperature is effective. That is, the carburizing time can be reduced by setting the carburizing temperature, which is about 930°C in typical carburizing, to be in a temperature range of 990°C to 1090°C. However, when high temperature carburizing is performed in a temperature range of 990°C to 1090°C in order to reduce the carburizing time, coarse grains are generated, and there may be cases where fatigue properties such as rolling fatigue properties needed for carburized components are not sufficiently obtained. Therefore, there is a demand for case hardening steel which does not cause the generation of coarse grains even when high temperature carburizing is performed and is thus suitable for high temperature carburizing.

[0008] In addition, gears, bearing components, and rolling components which receive high contact pressure are generally large components. Such large components are typically manufactured by hot forging a steel bar, performing a heat treatment such as normalizing as necessary, and performing machining, carburizing and quenching, tempering, and polishing as necessary. In order to suppress the generation of coarse grains during carburizing, a hot forged member after the hot forging needs to be an appropriate material capable of suppressing coarse grains during carburizing. For this, it is necessary to use steel capable of suppressing the generation of coarse grains during carburizing as the material of the steel bar.

[0009] Patent Document 1 discloses a case hardening steel which contains Ti: 0.05% to 0.2%, S: 0.001% to 0.15%, and N: limited to less than 0.0051%, in which the amount of AlN precipitated after hot rolling is limited to 0.01% or less and coarse grain prevention properties and fatigue properties during carburizing are excellent.

[0010] In addition, Patent Document 2 discloses a case hardening steel which contains Ti: 0.03% to 0.30%, S: 0.010% to 0.10%, N: limited to 0.020% or less, in which the number density of Ti-based sulfides is specified.

[0011] However, in the case hardening steels disclosed in Patent Documents 1 and 2, a reduction in pinning force due to coarsening of precipitates used for pinning of grains and the balance between Ti and S contents are not considered, and there is a possibility that Ti-based precipitates used for the pinning effect may be insufficient. As a result, in the case hardening steels disclosed in Patent Documents 1 and 2, in a case where switching from hot forging to cold forging, annealing after the cold forging is omitted, or high temperature carburizing and the like are performed, there is a possibility that the coarse grain prevention properties may be insufficient. In addition, gears and shafts, which are main applications of case hardening steel, require not only a reduction in thermal strain but also a further improvement of fatigue strength properties.

[Prior Art Document]

[Patent Document]

[0012]

[Patent Document 1] Japanese Patent No. 4448456
[Patent Document 2] Japanese Unexamined Patent Application, First Publication No. 2007-31787

[Disclosure of the Invention]

[Problems to be Solved by the Invention]

[0013] The present invention has been made taking the foregoing circumstances into consideration, and an object thereof is to provide a case hardening steel having excellent coarse grain prevention properties during carburizing, a carburized component, and a manufacturing method of a case hardening steel. In a case of excellent coarse grain prevention properties during carburizing, heat treatment strain due to carburizing and quenching can be suppressed even when annealing before the carburizing is omitted, and excellent fatigue properties can be obtained after the carburizing and quenching.

[Means for Solving the Problem]

[0014]

(1) According to an aspect of the present invention, a case hardening steel includes, as a chemical composition, by mass%: C: 0.10% to 0.30%, Si: 0.02% to 1.50%, Mn: 0.30% to 1.80%, S: 0.003% to 0.020%, Cr: 0.40% to 2.00%, Al: 0.005% to 0.050%, Ti: 0.06% to 0.20%, Bi: 0.0001% to 0.0050%, Mo: 0% to 1.50%, Ni: 0% to 3.50%, V: 0% to 0.50%, B: 0% to 0.0050%, Nb: 0% or more and less than 0.040%, P: limited to 0.050% or less, N: limited to 0.0060% or less, O: limited to 0.0025% or less, and a remainder including an iron and impurities, in which an Expression (a) is satisfied, and in a longitudinal section, a maximum diameter of Ti-based precipitates predicted by extreme value statistics under a condition that an inspection standard area is 100 mm^2, a number of inspections is 16 visual fields, and an area where prediction is performed is 30,000 mm^2, is 40 μm or less,

$$\mathrm{Ti/S} \geq 6.0 \qquad \ldots\text{Expression (a)}$$

where Ti in the Expression (a) represents a Ti content by mass%, and S in the Expression (a) represents a S content by mass%.

(2) The case hardening steel according to (1) may include, as the chemical composition, by mass%, one or more selected from the group consisting of: Mo: 0.02% to 1.50%, Ni: 0.10% to 3.50%, V: 0.02% to 0.50%, B: 0.0002% to 0.0050%, and Nb: more than 0% and less than 0.040%.

(3) In the case hardening steel according to (1) or (2), a metallographic structure may contain a bainite, and a structure fraction of the bainite may be 30% or less.

(4) In the case hardening steel according to any one of (1) to (3), the metallographic structure may contain a ferrite, and a grain size number of the ferrite may be No. 8 to No. 11 specified in JIS G 0552.

(5) According to another aspect of the present invention, a carburized component includes: the case hardening steel according to any one of (1) to (4).

(6) According to still another aspect of the present invention, a manufacturing method of a case hardening steel includes: a heating process of heating steel including, as a chemical composition, by mass, C: 0.10% to 0.30%, Si:

0.02% to 1.50%, Mn: 0.30% to 1.80%, S: 0.003% to 0.020%, Cr: 0.40% to 2.00%, Al: 0.005% to 0.050%, Ti: 0.06% to 0.20%, Bi: 0.0001% to 0.0050%, Mo: 0% to 1.50%, Ni: 0% to 3.50%, V: 0% to 0.50%, B: 0% to 0.0050%, Nb: 0% or more and less than 0.040%, P: limited to 0.050% or less, N: limited to 0.0060% or less, O: limited to 0.0025% or less, and a remainder including an iron and impurities, and satisfies an Expression (b), at a temperature of 1150°C or higher for a holding time of ten minutes or longer; and a hot rolling process of hot rolling the steel into a wire rod or a steel bar,

$$\mathrm{Ti/S} \geq 6.0 \qquad \ldots\text{Expression (b)}$$

where Ti in the Expression (b) represents a Ti content (mass%), and S in the Expression (b) represents a S content (mass%).

(7) The manufacturing method of a case hardening steel according to (6) may further include: a cooling process of slow cooling the wire rod or the steel bar at a cooling rate of 1.00 °C/s or less in a temperature range of 800°C to 500°C after the hot rolling process.

(8) In the manufacturing method of a case hardening steel according to (6) or (7), in the hot rolling process, a finish temperature may be set to 840°C to 1000°C.

[Effects of the Invention]

[0015] The case hardening steel according to the aspect of the present invention has a predetermined chemical composition, and the maximum diameter of the Ti-based precipitates is controlled to be in a predetermined range, thereby achieving excellent coarse grain prevention properties during carburizing. Therefore, with the case hardening steel according to the aspect of the present invention, heat treatment strain due to carburizing and quenching can be suppressed, and excellent fatigue properties are obtained after the carburizing and quenching. In addition, the carburized component according to the aspect of the present invention has less heat treatment strain and has excellent fatigue properties.

[0016] With the manufacturing method of a case hardening steel according to the aspect of the present invention, the case hardening steel which has excellent coarse grain prevention properties during carburizing can be manufactured. The case hardening steel obtained by the manufacturing method can suppress heat treatment strain due to carburizing and quenching and achieve excellent fatigue properties after the carburizing and quenching.

[Embodiments of the Invention]

[0017] In order to achieve the above objects, the inventors conducted intensive studies. As a result, the following findings (i) to (v) were obtained.

(i) In the related art, there may be cases where crystallization of MnS, which is the origin of bending fatigue fracture, occurs during carburizing performed on case hardening steel, and sufficient fatigue properties cannot be obtained after carburizing and quenching performed on the case hardening steel. Contrary to this, by optimizing the relationship between the S content and the Ti content in case hardening steel (Ti/S ≥ 6.0), which is not considered in the related art, fine Ti-based carbosulfide can be formed instead of MnS, which is stretched in a rolling direction in which bending fatigue properties are deteriorated and is coarsened during carburizing performed on case hardening steel. Since fine Ti-based carbosulfide is formed instead of coarse MnS, excellent fatigue properties are obtained after carburizing and quenching performed on the case hardening steel.

(ii) In order to prevent coarsening of grains during carburizing performed on case hardening steel, it is effective to use Ti-based precipitates mainly containing TiC and TiCS finely precipitated during carburizing, instead of using AlN and NbN as pinning particles. Furthermore, by including a small amount of Bi in the case hardening steel, growth and coarsening of the Ti-based precipitates during the carburizing are suppressed, and thus coarse grain prevention properties are further improved.

[0018] In order to stably exhibit the pinning effect of the Ti-based precipitates during the carburizing performed on the case hardening steel, it is necessary that Ti-based precipitates finely precipitate in steel which is cooled after being hot rolled in a manufacturing process of the case hardening steel. For this, it is necessary to cause Ti-based precipitates to precipitate at phase interfaces during diffusion transformation from austenite in the cooling process after the hot rolling. When bainite is formed in the hot rolled structure, it is difficult for the Ti-based precipitates to precipitate at phase interfaces. Therefore, it is preferable to suppress the formation of bainite in case hardening steel as much as possible.

[0019] In order to cause Ti-based precipitates to finely precipitate in steel after being hot rolled and cooled, it is effective to optimize the hot rolling conditions. That is, it is preferable that Ti-based precipitates are first solid-solubilized in a matrix by setting the heating temperature during the hot rolling to a high temperature and the precipitation temperature range of the Ti-based precipitates after the hot rolling is set such that slow cooling is achieved. Through the heating, rolling, and cooling, the formation of bainite can be suppressed, and a large amount of Ti-based precipitates can be formed and finely dispersed.

(iii) By causing Nb carbonitride mainly containing NbC to finely precipitate during the carburizing performed on the case hardening steel in combination with the Ti-based precipitates, the coarse grain prevention properties are further improved. In order to stably exhibit the pinning effect of the Nb carbonitride during the carburizing performed on the case hardening steel, it is necessary that the Nb carbonitride finely precipitates in steel after being hot rolled and cooled in the manufacturing process of the case hardening steel. For this, similarly to the Ti-based precipitates, it is necessary to cause the Nb carbonitride to precipitate at phase interfaces during diffusion transformation from austenite in the cooling process after the hot rolling.

[0020] In order to cause the Nb carbonitride to finely precipitate in the steel after being hot rolled and cooled, it is preferable that Nb carbonitride is first solid-solubilized in a matrix by setting the heating temperature during the hot rolling to a high temperature and the precipitation temperature range of the Nb carbonitride is set such that slow cooling is achieved. Through the heating, rolling, and cooling, a large amount of Nb carbonitride can be finely dispersed. When bainite is formed in the hot rolled structure, it is difficult for the Nb carbonitride to precipitate at phase interfaces. Therefore, it is preferable to suppress the formation of bainite as much as possible.

(iv) When ferrite grains contained in the steel after being hot rolled and cooled are excessively fine, coarse grains tend to be generated during the carburizing performed on the case hardening steel. The grain size of the ferrite grains in the steel after being hot rolled and cooled can be optimized by controlling the finish rolling temperature.
(v) In a carburized component manufactured by performing carburizing and quenching on case hardening steel containing Ti, Ti-based precipitates become the origin of fatigue fracture, so that fatigue properties, particularly rolling fatigue properties tend to be insufficient. It is possible to improve the fatigue properties by reducing the N content in the chemical composition of the case hardening steel, setting the heating temperature during hot rolling to a high temperature, and reducing the maximum size of Ti precipitates.

[0021] The present invention has been made based on the above-described novel findings.

[0022] Hereinafter, case hardening steel according to an embodiment of the present invention (case hardening steel according to this embodiment), a carburized component according to the embodiment of the present invention (a carburized component according to this embodiment), and a manufacturing method thereof will be described in detail.

[0023] First, the chemical composition of the case hardening steel according to this embodiment will be described. Unless otherwise specified, "%" of the amount of each element means "mass%".

(C: 0.10% to 0.30%)

[0024] C is an element effective in improving the strength of steel. When the C content is less than 0.10%, a preferable tensile strength (for example, about 900 MPa) cannot be secured after carburizing, quenching and tempering. On the other hand, when the C content exceeds 0.30%, the steel becomes hard, the cold workability deteriorates, and the toughness of a core portion after the carburizing and quenching deteriorates. Therefore, it is necessary to set the C content to be in a range of 0.10% to 0.30%.

(Si: 0.02% to 1.50%)

[0025] Si is an element effective in deoxidizing steel. In addition, Si is an element effective in imparting necessary strength and hardenability to steel and improving the temper softening resistance of the steel. When the Si content is less than 0.02%, the effect is not sufficiently obtained. On the other hand, when the Si content exceeds 1.50%, the hardness of the steel increases and the cold forgeability deteriorates. For the above reasons, it is necessary to set the Si content to be in a range of 0.02% to 1.50%.

[0026] In a case where the case hardening steel is subjected to cold working, a suitable range of the Si content is 0.02% to 0.30%. In particular, in a case where cold forgeability is considered as being important, it is more desirable that the Si content is set to be in a range of 0.02% to 0.15%.

[0027] In addition, Si is an element effective in increasing the grain boundary strength, and in a case where case hardening steel is used as the material of a carburized component such as a bearing component and a rolling component, Si is an element effective in increasing the service life by suppressing the microstructural change and deterioration of the material in a rolling fatigue process of the carburized components. In a case where there is a demand for high-strengthening by including Si, a suitable range of the Si content is 0.20% to 1.50%. In particular, in a case where case hardening steel is used as the material of a carburized component having a high level of rolling fatigue strength, it is more desirable that the Si content is set to be in a range of 0.40% to 1.50%.

[0028] The effect of the inclusion of Si on the suppression of the microstructural change and deterioration of the material in the rolling fatigue process of the bearing component or the rolling component is particularly significant when the amount of retained austenite (so-called retained γ amount) in the structure after being subjected to carburizing and quenching is 30% to 40%. In order to control the retained γ amount to be in this range, it is effective to perform nitriding in a diffusion treatment after carburizing (so-called carbonitriding treatment). The nitriding treatment after the carburizing is appropriately performed under the condition that the nitrogen concentration on the surface is in a range of 0.2% to 0.6%. It is desirable that the carbon potential during carburizing in this case is set to be in a range of 0.9% to 1.3%.

(Mn: 0.30% to 1.80%)

[0029] Mn is an element effective in deoxidizing steel. In addition, Mn is an element effective in imparting necessary strength and hardenability to steel. When the Mn content is less than 0.30%, the effect is not sufficiently obtained. Therefore, the Mn content is set to 0.30% or more, and is desirably 0.50% or more. On the other hand, when the Mn content exceeds 1.80%, not only is the effect saturated, but also the cold forgeability deteriorates due to an increase in the hardness of the steel. Therefore, it is necessary to set the Mn content to 1.80% or less, and desirably 1.20% or less. In a case where the cold forgeability of the steel is considered as being important, it is desirable to set the Mn content to be in a range of 0.50% to 0.75%.

(P: 0.050% or Less)

[0030] P is an element which causes deterioration of cold forgeability by increasing deformation resistance during cold forging and thus causing deterioration of toughness. In addition, P is an element which causes deterioration of fatigue strength by embrittling grain boundaries of a component after quenching and tempering. Therefore, it is desirable to reduce the P content as much as possible. However, when the P content exceeds 0.050%, deterioration of the cold forgeability and fatigue strength becomes significant, so that the P content is limited to 0.050% or less. A suitable range of the P content is 0.015% or less. The P content may also be 0%.

(S: 0.003% to 0.020%)

[0031] S is an element that forms MnS in steel. Since MnS can be the origin of bending fatigue fracture of a carburized component, it is necessary to prevent the formation of MnS. Therefore, the S content is set to 0.020% or less, and the relationship between the S content and the Ti content is set to be in a range satisfying Expression (1). When the S content is in the above range, S in steel is present as Ti-based carbosulfide, and thus excellent fatigue properties are obtained after carburizing and quenching. The S content is more preferably 0.015% or less. On the other hand, the Ti-based carbosulfide has the pinning effect that contributes to the prevention of the generation of coarse grain. In order to exhibit the effect, it is necessary to set the S content to 0.003% or more. The S content is preferably 0.005% or more.

$$\mathrm{Ti/S} \geq 6.0 \qquad \ldots\text{Expression (1)}$$

(Ti in Expression (1) is the Ti content (mass%), and S in the Expression (1) is the S content (mass%))

(Cr: 0.40% to 2.00%)

[0032] Cr is an element effective in improving the strength and hardenability of steel. Furthermore, in a case where case hardening steel is used as the material of a carburized component such as a bearing component and a rolling component, Cr increases the retained γ amount after carburizing and quenching and suppresses the microstructural change and deterioration of the material in a rolling fatigue process. Therefore, Cr is an element that contributes to an increase in the fatigue life of the carburized component. When the Cr content is less than 0.40%, the effect is insufficient. Therefore, it is necessary to set the Cr content to 0.40% or more. The Cr content is preferably 0.70% or more. On the other hand, when the Cr content exceeds 2.00%, the cold forgeability deteriorates due to an increase in the hardness of the steel. Therefore, it is necessary to set the Cr content to 2.00% or less. The Cr content is preferably 1.60% or less.

[0033] The effect of the inclusion of Cr on the suppression of the microstructural change and deterioration of the material in the rolling fatigue process of the bearing component or the rolling component is particularly significant when the retained γ amount in the structure after being subjected to carburizing and quenching is 30% to 40%. In order to control the retained γ amount to be in this range, it is effective to perform a nitriding treatment after the carburizing under the condition that the nitrogen concentration on the surface is in a range of 0.2% to 0.6%.

(Al: 0.005% to 0.050%)

[0034] Al is an element effective as a deoxidizing agent. When the Al content is less than 0.005%, the effect is insufficient. Therefore, the Al content is set to 0.005% or more. The Al content is preferably 0.025% or more. On the other hand, when the Al content exceeds 0.050%, a portion of AlN remains unsolubilized during heating before hot rolling performed at the time of manufacturing of case hardening steel and becomes a precipitation site of precipitates of Ti (Ti and Nb in a case where Nb is contained). In this case, fine dispersion of the Ti-based precipitates (Ti-based precipitates and Nb carbonitride in the case where Nb is contained) is inhibited such that grains coarsen during carburizing. Therefore, it is necessary to set the Al content to 0.050% or less. The Al content is preferably 0.040% or less.

(Ti: 0.06% to 0.20%)

[0035] Ti is an element that forms fine Ti-based carbide and Ti-based carbosulfide such as TiC, TiCS, and $Ti_4C_2S_2$ in steel and is an element effective for achieving γ grain refinement during carburizing. When the Ti content is less than 0.06%, the effect is insufficient, and thus the Ti content is set to 0.06% or more. On the other hand, when the Ti content exceeds 0.20%, precipitation hardening by TiC significant occurs and the cold workability deteriorates significantly. In addition, the formation of precipitates mainly containing TiN significantly occurs, and the rolling fatigue properties after carburizing and quenching deteriorate. Therefore, it is necessary to set the Ti content to 0.20% or less. The Ti content is preferably less than 0.15%.

[0036] When the case hardening steel according to this embodiment or a forged member obtained by forging the case hardening steel is subjected to carburizing and quenching, a solid-soluted Ti reacts with carbon and nitrogen infiltrating during the carburizing such that a large amount of fine TiC and TiN (hereinafter, sometimes referred to as "Ti(C, N)") precipitates on the carburized layer. The Ti(C, N) contributes to the improvement of rolling fatigue life in a carburized component such as a bearing component or a rolling component obtained by performing carburizing and quenching on the case hardening steel. Therefore, in a case where the bearing component or the rolling component, which demands a high level of rolling fatigue life, is manufactured, it is effective to promote the precipitation of the Ti(C, N) by setting the carbon potential during the carburizing to a higher level in a range of 0.9% to 1.3% or by performing a so-called carbonitriding treatment. The carbonitriding treatment is a treatment in which the above-described carburizing and nitriding in a diffusion treatment after the carburizing are performed, and in the nitriding treatment, the condition that the nitrogen concentration on the surface is in a range of 0.2% to 0.6% is appropriate.

(Bi: 0.0001% to 0.0050%)

[0037] Bi is an important element in the case hardening steel according to this embodiment. When a small amount of Bi is contained in the steel, sulfide is finely dispersed as the solidification structure of the steel (mainly dendrite structure) becomes refined. Furthermore, by including a small amount of Bi in the steel, during the carburizing, growth and coarsening of precipitates such as Ti-based precipitates that suppress coarsening of grains can be suppressed. In order to obtain the above effect, it is necessary to set the Bi content to 0.0001% or more. The Bi content is preferably 0.0010% or more. On the other hand, when the Bi content exceeds 0.0050%, the effect of refining the solidified structure is saturated and the hot workability of the steel deteriorates. This makes it difficult to perform the hot rolling at the time of manufacturing of the case hardening steel. For these reasons, the Bi content is set to 0.0050% or less. The Bi content is preferably 0.0040% or less.

(N: 0.0060% or Less)

[0038] When N is bonded to Ti in steel, coarse TiN which hardly contributes to the prevention of coarsening of grains is formed. TiN becomes a precipitation site of Ti-based precipitates mainly containing TiC and TiCS, and NbC and NbN mainly containing NbC (hereinafter, sometimes referred to as "Nb(C, N)") and inhibits fine precipitation of Ti-based precipitates and Nb(C, N). In this case, the generation of coarse grains cannot be sufficiently suppressed. This adverse effect is particularly significant in a case where the N content exceeds 0.0060%. For the above reasons, it is necessary to set the N content to 0.0060% or less. The N content is preferably less than 0.0051 %. The N content may also be 0%.

(O: 0.0025% or Less)

[0039] In high titanium steel (steel containing a large amount of Ti) such as the case hardening steel according to this embodiment, O in the steel forms Ti-based oxide inclusions. Since the Ti-based oxide inclusions become a precipitation site of TiC, when a large amount of the Ti-based oxide inclusions are present in the steel, TiC coarsely precipitates during the hot rolling in the manufacturing of the case hardening steel. In this case, coarsening of grains during the carburizing cannot be suppressed. Therefore, it is desirable to reduce the O content as much as possible. When the O content exceeds 0.0025%, the adverse effect becomes significant. Therefore, it is necessary to limit the O content to 0.0025% or less. A suitable range of the O content is 0.0020% or less. In a carburized component such as a bearing component or a rolling component, the oxide inclusions become the origin of rolling fatigue fracture. Therefore, the lower the O content of the case hardening steel, the longer the rolling life of the carburized component. Therefore, in a case where the case hardening steel is used as the material of the carburized component such as the bearing component or the rolling component, it is desirable to limit the O content to 0.0012% or less. The O content may also be 0%.

[0040] The case hardening steel according to this embodiment basically includes the above-described elements and the remainder consisting of Fe and impurities. However, instead of a portion of Fe, one or more elements selected from the group consisting of Mo, Ni, V, B, and Nb may be contained in the following ranges in addition to the above-mentioned elements. However, these elements are not necessarily contained. Therefore, the lower limits thereof are 0%. In addition, even when these elements are contained in amounts less than the following ranges, the elements do not impair the properties of the case hardening steel and are thus allowed.

[0041] In addition, the impurities are components incorporated from raw materials such as ore or scrap or from various environments in a manufacturing process when steel is industrially manufactured and are allowed in a range in which the steel is not adversely affected.

[0042] The chemical composition of the case hardening steel according to this embodiment may further include one or more of Mo, Ni, V, B, and Nb as necessary in the following ranges.

(Mo: 0.02% to 1.50%)

[0043] Mo is an element effective in improving the strength and hardenability of steel. Furthermore, Mo is an element effective in an increase in the fatigue life by increasing the retained γ amount in a bearing component or a rolling component obtained after carburizing and suppressing the microstructural change and deterioration of the material in a rolling fatigue process. In a case of obtaining these effects, it is preferable to set the Mo content to 0.02% or more. The Mo content is more preferably 0.05% or more. However, when the Mo content exceeds 1.50%, machinability and cold forgeability deteriorate due to an increase in hardness. For the above reasons, even in the case where Mo is contained, the Mo content is set to be in a range of 1.50% or less. The Mo content is preferably 0.50% or less.

[0044] The effect of the inclusion of Mo on the suppression of the microstructural change and deterioration of the material in the rolling fatigue process of the bearing component or the rolling component is particularly significant when the retained γ amount in the structure after being subjected to carburizing and quenching is 30% to 40%, like the above-described effect of Cr.

(Ni: 0.10% to 3.50%)

[0045] Ni is an element effective in improving the strength and hardenability of steel. In a case of obtaining the effect, it is preferable to set the Ni content to 0.10% or more. The Ni content is more preferably 0.20% or more. On the other hand, when the Ni content exceeds 3.50%, machinability and cold forgeability deteriorate due to an increase in hardness. Therefore, even in a case where Ni is contained, the Ni content is set to be in a range of 3.50% or less. The Ni content is preferably 2.00% or less.

(V: 0.02% to 0.50%)

[0046] V is an element effective in improving the strength and hardenability of steel. In a case of obtaining the effect, it is preferable to set the V content to 0.02% or more. However, when the V content exceeds 0.50%, machinability and cold forgeability deteriorate due to an increase in hardness. Therefore, even in a case where V is contained, the V content is set to be in a range of 0.50% or less. The V content is preferably 0.20% or less.

(B: 0.0002% to 0.0050%)

[0047] B is an element effective in improving the strength and hardenability of steel. In addition, B forms boron iron carbide in a steel bar or a wire rod in a cooling process after rolling and thus increases the growth rate of ferrite, thereby providing an effect of softening the rolled steel. Furthermore, B improves the grain boundary strength of a carburized material and also has an effect of improving fatigue strength and impact strength as a carburized component. In a case of obtaining these effects, it is preferable to set the B content to 0.0002% or more. The B content is more preferably 0.0005% or more. However, when the B content exceeds 0.0050%, the above-mentioned effects are saturated, and adverse effects such as deterioration of the impact strength are concerned. Therefore, even in a case where B is contained, the B content is set to be in a range of 0.0050% or less. The B content is preferably 0.0030% or less.

(Nb: More Than 0% and Less Than 0.040%)

[0048] Nb is an element that is bonded to C and N in steel during carburizing to form Nb(C, N) and is thus effective in suppressing coarsening of grains. By including Nb, the effect of preventing coarse grains due to Ti-based precipitates is further increased. This is because Nb is solid-solubilized in the Ti-based precipitates and thus suppresses coarsening of the Ti-based precipitates. The effect of the inclusion of Nb increases as the Nb content is increased. On the other hand, Nb causes deterioration of machinability and cold forgeability, and deterioration of carburizing properties. In particular, when the Nb content is 0.040% or more, the hardness of the material increases and the machinability and the cold forgeability deteriorate. In addition, it is difficult for Nb carbonitride to be solid-solubilized by heating during hot rolling of the rolled material, and the number of grains of the Nb carbonitride finely precipitated decreases, leading to degradation of the coarse grain prevention properties. Therefore, even in a case where Nb is contained, the Nb content is set to be less than 0.040%. In a case where workability such as machinability and cold forgeability is considered as being important, an appropriate range of the Nb content is less than 0.030%. In a case where carburizing properties are considered as being important in addition to workability, an appropriate range of the Nb content is less than 0.020%. Furthermore, in a case where carburizing properties are considered as being particularly important, an appropriate range of the Nb content is less than 0.010%.

[0049] Even when the Nb content is a small amount such as less than 0.030%, less than 0.020%, or less than 0.010%, Nb significantly improves the coarse grain prevention properties compared to a case where Nb is not contained. Therefore, in a case where it is desirable to obtain the above-described effect, the Nb content may be more than 0%.

[0050] In order to achieve both the coarse grain prevention properties and the workability, it is preferable to adjust the Nb content according to the Ti content. Specifically, it is preferable to set the total content (Ti + Nb) of the Nb content and the Ti content to 0.07% to 0.20%. In particular, in a case where the case hardening steel is subjected to carburizing at a high temperature or cold forging, a desirable range of the total content of the Nb content and the Ti content is more than 0.091% and less than 0.17%.

[0051] Next, the structure (metallographic structure) of the case hardening steel according to this embodiment will be described.

(Maximum Diameter of Ti-Based Precipitates Predicted by Extreme Value Statistics: 40 μm or Less)

[0052] In the case hardening steel according to this embodiment, in a longitudinal section, the maximum diameter of Ti-based precipitates predicted by extreme value statistics under the condition that an inspection standard area is 100 mm^2, a number of inspections is 16 visual fields, and an area where prediction is performed is 30,000 mm^2 is set to 40 μm or less.

[0053] One of the required properties of the carburized component obtained from the case hardening steel which is an object of this embodiment is the improvement of contact fatigue strength such as rolling fatigue properties and surface fatigue strength. When coarse Ti-based precipitates are present in the case hardening steel, the Ti-based precipitates become the origin of contact fatigue fracture in the carburized component manufactured by performing carburizing and quenching the case hardening steel, resulting in deterioration of the fatigue properties.

[0054] When the maximum diameter of Ti-based precipitates predicted by extreme value statistics under the condition that the inspection standard area is 100 mm^2, the number of inspections is 16 visual fields, and the area where prediction is performed is 30,000 mm^2 exceeds 40 μm, an adverse effect of the Ti-based precipitates on the contact fatigue properties becomes particularly significant. For the above reason, the maximum diameter of the Ti-based precipitates predicted by extreme value statistics under the above condition is set to be 40 μm or less, and is preferably 30 μm or less.

[0055] A method of measuring and predicting the maximum diameter of precipitates using extreme value statistics is based on the method described in pp. 233 to 239 of "Metal Fatigue: Effects of Small Defects and Nonmetallic Inclusions" published by YOKENDO Ltd., on March 8, 1993. In this embodiment, the two-dimensional inspection method in which maximum precipitates observed in a predetermined area (an area where prediction is performed: 30,000 mm^2) by two-dimensional inspection are estimated is used. Detailed measurement procedures will be described in Examples. The area where prediction is performed is set in consideration of a risk volume of a general component.

(Structure Fraction of Bainite: 30% or Less)

[0056] In the case hardening steel according to this embodiment, it is preferable that the structure fraction (area ratio) of bainite is 30% or less. When a bainitic structure is incorporated in the case hardening steel, phase interface precipitation of Ti-based precipitates is difficult, this causes the generation of coarse grains during carburizing. In addition, it is desirable that the bainitic structure in the case hardening steel is small from the viewpoint of improving cold workability. An adverse effect of the bainitic structure in the case hardening steel is particularly significant when the structure fraction of the bainite exceeds 30%. For the above reasons, it is preferable to limit the structure fraction of the bainite to 30% or less. In a case where carburizing conditions regarding the prevention of coarse grains during carburizing are severe, such as in a case where the case hardening steel is subjected to high temperature carburizing, a suitable range of the structure fraction of the bainite is 20% or less. In addition, in a case where carburizing conditions regarding the prevention of coarse grains during carburizing are more severe, such as in a case where the case hardening steel is subjected to cold forging, a suitable range of the structure fraction of the bainite is 10% or less. The bainitic structure may also be 0%. A structure other than the bainite is preferably a structure mainly containing ferrite and pearlite.

(Ferrite Grain Size: No. 8 to No. 11)

[0057] In the case hardening steel according to this embodiment, the grain size number of ferrite contained in the metallographic structure is preferably No. 8 to No. 11 specified in JIS G 0552. When ferrite grains of the case hardening steel are excessively fine, austenite grains are excessively refined during carburizing. When the austenite grains become excessively fine, the driving force for grain growth increases and coarse grains tend to be formed. In particular, when the ferrite grain size exceeds No. 11 specified in JIS G 0552, this tendency becomes significant. On the other hand, when the ferrite grain size is less than No. 8 specified in JIS G 0552, the ferrite is coarse, the ductility deteriorates, and thus cold forgeability also deteriorates. For the above reasons, it is preferable to set the ferrite grain size number to be in a range of No. 8 to No. 11 specified in JIS G 0552.

[0058] Since the case hardening steel according to this embodiment has excellent coarse grain prevention properties during carburizing, heat treatment strain due to carburizing and quenching can be suppressed. In addition, when the carburizing and quenching is performed, a carburized component having excellent fatigue properties is obtained. Even when the case hardening steel according to this embodiment is subjected to high temperature carburizing, the generation of coarse grains during the carburizing can be suppressed. Therefore, by performing the high temperature carburizing after forging, the carburizing time can be reduced. Furthermore, even for a carburized component in the related art in which switching from hot forging to cold forging is not performed due to deterioration of dimensional accuracy caused by heat treatment strain, switching to cold forging can also be performed. Moreover, annealing performed in the related art to suppress heat treatment strain after cold forging can be omitted.

[0059] A carburized component according to this embodiment includes the case hardening steel according to this embodiment. The carburized component according to this embodiment is manufactured, for example, by a method of forging the case hardening steel according to this embodiment, machining the forged steel into a predetermined shape, and performing carburizing, quenching and tempering thereon. During the forging, machining, carburizing and quenching, the chemical composition and the maximum diameter of the Ti-based precipitates do not change. Therefore, the carburized component according to this embodiment has the same chemical composition and Ti-based precipitates as those of the case hardening steel according to this embodiment. However, since the carburized component according to this embodiment is obtained through carburizing and quenching, the carburized component according to this embodiment is different from the case hardening steel in that a carburizing and quenching layer is formed on the surface.

[0060] Next, a preferable manufacturing method of the case hardening steel according to this embodiment will be described in detail.

[0061] The manufacturing method described below is merely an example, and as long as a case hardening steel satisfying the scope of this embodiment can be obtained, the manufacturing method of the case hardening steel according to this embodiment is not limited to the following manufacturing conditions.

<Melting Process, Casting Process, and Blooming Process>

[0062] Steel having the above-mentioned chemical composition is melted (melting process) according to a typical method such as a converter or an electric furnace, and was cast into a bloom having the above-mentioned chemical composition (casting process). Thereafter, blooming is performed as necessary (blooming process), thereby obtaining a rolled material to be hot rolled into a wire rod or a steel bar. The size of the bloom, the cooling rate during solidification, and the blooming conditions are not particularly limited.

<Heating Process, Hot Rolling Process, and Cooling Process>

[0063] Next, the rolled material having the above-described chemical composition is heated under the following conditions, hot rolled into a wire rod or a steel bar, and cooled, thereby obtaining case hardening steel.

[0064] In the manufacturing of the case hardening steel according to this embodiment, the rolled material having the above-mentioned chemical composition is heated at a temperature of 1 150°C or higher for a holding time of 10 minutes or longer (heating process), and the heated rolled material is hot rolled into a wire rod or a steel bar (hot rolling process). During the hot rolling, when the heating temperature is 1150°C or higher and the holding time is 10 minutes or longer, the Ti-based precipitates can be sufficiently solid-solubilized in a matrix.

[0065] When the heating temperature before the hot rolling is lower than 1150°C and/or the holding time is shorter than 10 minutes, the Ti-based precipitates and AlN (in a case where Nb is contained, Ti-based precipitates, Nb precipitates, and AlN) cannot be sufficiently solid-solubilized in the matrix. As a result, coarse Ti-based precipitates once formed in the casting process remain unsolubilized in the steel after being hot rolled and cooled, and the Ti-based precipitates (Ti-based precipitates and Nb-based precipitates in the case where Nb is contained) cannot be finely precipitated. Furthermore, coarsening of the Ti-based precipitates remaining unsolubilized in the heating process before the hot rolling proceeds due to Ostwald growth. As a result, coarse Ti-based precipitates and AlN (in the case where Nb is contained, coarse Ti-based precipitates, Nb-based precipitates, and AlN) are present in the steel after being hot rolled and cooled. In this case, the generation of coarse grains during carburizing cannot be suppressed. Therefore, during hot rolling, it is preferable to perform heating at a temperature of 1150°C or higher for a holding time of 10 minutes or longer. A suitable range of the heating condition in the hot rolling is a holding time of 10 minutes or longer at a temperature of 1180°C or higher. It is not necessary to limit the upper limits of the heating temperature and the holding time. However, the upper limit of the heating temperature may be set to 1300°C and the upper limit of the holding time may be set to 60 minutes in consideration of facility restrictions and productivity.

(Finish Temperature)

[0066] The finish temperature (finish rolling temperature) of the hot rolling is preferably set to 840°C to 1000°C. By setting the finish temperature of the hot rolling to be the above range, steel having a ferrite grain size number of No. 8 to No. 11 specified in JIS G 0552 can be obtained.

[0067] When the finish temperature is lower than 840°C, the ferrite grain size becomes too fine, and coarse grains are likely to be generated during carburizing. On the other hand, when the finish temperature exceeds 1000°C, the ferrite grains becomes coarse, and the hardness of the steel after being hot rolled and cooled increases, resulting in deterioration of cold forgeability. For the above reasons, it is preferable to set the finish temperature of the hot rolling to 840°C to 1000°C. In order to soften the steel, the finish temperature is preferably 920°C to 1000°C. On the other hand, in a case where the case hardening steel is cold-forged and is subjected to annealing after the cold forging and before carburizing and quenching, the finish temperature is preferably 840°C to 920°C.

(Cooling Rate)

[0068] After the hot rolling, the steel is cooled (cooling process). In the manufacturing of the case hardening steel according to this embodiment, it is preferable to perform slow cooling at a cooling rate (average cooling rate) of 1.00 °C/s or less in a temperature range of 800°C to 500°C after the hot rolling. By performing the cooling under the above cooling conditions after the hot rolling, the time for which the Ti-based precipitates pass a precipitation temperature range is sufficiently secured, and dispersion of fine Ti-based precipitates is promoted. In addition, by performing the cooling under the above cooling conditions, the structure fraction of bainite can be suppressed. As a result, steel in which the structure fraction of bainite is 30% or less and excellent coarse grain prevention properties are achieved during carburizing is obtained. When the cooling rate exceeds 1.00 °C/s in the above temperature range, there is concern that the structure fraction of the bainite may increase to more than 30%. In addition, at a high cooling rate in the above temperature range, the hardness of the steel after being hot rolled and cooled increases, resulting in deterioration of cold forgeability. Therefore, it is desirable to set the cooling rate in the above temperature range to be as low as possible. A suitable range of the cooling rate in the above temperature range is 0.70 °C/s or less.

[0069] In a case where air cooling is performed after the hot rolling, although depending on the size of the case hardening steel, there is concern that the cooling rate in a range of 800°C to 500°C may exceed 1.00 °C/s. Therefore, it is preferable to control the cooling rate to decrease. As a method of reducing the cooling rate, for example, a method in which a heat insulation cover or a heat insulation cover with a heat source is installed in a rear stage of a hot rolling line and slow cooling of the steel is performed after the hot rolling by the heat insulation cover may be employed.

<Spheroidizing Annealing Process>

[0070] Spheroidizing annealing may be performed on the steel ((wire rod or steel bar): case hardening steel) after the cooling process as necessary.

[0071] By performing the spheroidizing annealing, the steel is softened, and thus the load during cold forging can be reduced.

[0072] According to the above manufacturing method, the case hardening steel according to this embodiment is obtained. This case hardening steel is suitable as the material of a carburized component.

[0073] The carburized component according to this embodiment can be manufactured by the method in which the case hardening steel according to this embodiment is forged, is machined into a predetermined shape, and is subjected to carburizing and quenching. In a case of manufacturing a carburized component using the case hardening steel according to this embodiment, the carburizing and quenching may be performed after hot forging, or the carburizing and quenching may be performed after cold forging.

[0074] In a case of manufacturing a carburized component by performing carburizing and quenching after hot forging the case hardening steel, for example, the carburized component can be manufactured by hot forging the case hardening steel (wire rod or steel bar), performing a heat treatment such as normalizing as necessary, and performing machining, carburizing and quenching, tempering, and polishing as necessary.

[0075] Specifically, for example, the hot forging can be performed at a heating temperature of 1150°C or higher.

[0076] In addition, although conditions during the carburizing and quenching are not particularly limited, for example, high temperature carburizing may be performed such that the carburizing temperature is in a temperature range of 950°C to 1090°C. In order to improve the rolling fatigue life of the carburized component, the carbon potential during the carburizing may be set to be high in a range of 0.9% to 1.3%. In addition, a carbonitriding treatment in which nitriding is performed in a diffusion treatment after the carburizing may be performed. In the nitriding treatment after the carburizing, the condition that the nitrogen concentration on the surface is in a range of 0.2% to 0.6% is appropriate for improving the rolling fatigue life.

[Examples]

[0077] Hereinafter, the present invention will be specifically described in detail with reference to Examples.

[0078] Steel having the composition shown in Table 1 was melted in a converter, was continuously cast into a bloom, and was subjected to blooming as necessary, thereby obtaining a 162 mm^2(162 mm x 162 mm in a cross section) rolled material (billet).

[0079] Subsequently, the billet was heated at a heating temperature shown in Table 2 for a holding time of 10 minutes or longer, was hot rolled at a finish temperature for hot rolling shown in Table 2, and was cooled at a cooling rate shown in Table 2 in a temperature of 800°C to 500°C after the hot rolling, thereby manufacturing a steel bar with a diameter of 24 to 30 mm.

[Table 1]

	No.	C (mass%)	Si (mass%)	Mn (mass%)	P (mass%)	S (mass%)	Al (mass%)	Ti (mass%)	Expression (1) (Ti/S)	Bi (mass%)	N (mass%)	Cr (mass%)	Mo (mass%)	Ni (mass%)	V (mass%)	B (mass%)	Nb (mass%)	O (mass%)
Steel of Present Invention	1	0.17	0.31	0.61	0.016	0.014	0.014	0.11	7.9	0.0018	0.0047	1.21	-	-	-	-	-	0.0018
Steel of Present Invention	2	0.22	0.16	0.95	0.019	0.013	0.037	0.15	11.6	0.0021	0.0041	1.25	-	-	-	-	0.030	0.0017
Steel of Present Invention	3	0.17	0.29	0.52	0.011	0.012	0.021	0.12	10.0	0.0016	0.0046	1.06	-	-	-	-	-	0.0010
Steel of Present Invention	4	0.16	0.17	0.75	0.014	0.011	0.017	0.08	7.3	0.0014	0.0051	1.14	0.16	-	-	-	-	0.0018
Steel of Present Invention	5	0.15	0.16	0.92	0.013	0.009	0.020	0.09	10.0	0.0018	0.0047	1.20	0.18	-	-	-	-	0.0010
Steel of Present Invention	6	0.17	0.38	0.96	0.012	0.014	0.033	0.13	9.3	0.0017	0.0046	0.98	0.17	-	-	-	-	0.0016
Steel of Present Invention	7	0.23	0.38	0.94	0.017	0.016	0.032	0.15	9.4	0.0019	0.0056	1.10	0.15	-	-	-	-	0.0011
Steel of Present Invention	8	0.17	0.43	0.92	0.016	0.015	0.033	0.14	9.3	0.0017	0.0049	1.25	-	-	-	0.0018	-	0.0014
Steel of Present Invention	9	0.19	0.20	0.76	0.014	0.012	0.034	0.18	15.0	0.0018	0.0044	1.01	-	-	0.16	-	-	0.0014
Steel of Present Invention	10	0.25	0.16	0.51	0.017	0.008	0.040	0.17	21.0	0.0019	0.0038	1.24	-	0.30	-	-	-	0.0011
Steel of Present Invention	11	0.24	0.19	0.79	0.011	0.011	0.034	0.18	15.9	0.0017	0.0047	1.17	-	-	-	0.0021	-	0.0014
Steel of Present Invention	12	0.16	0.29	0.62	0.019	0.012	0.016	0.19	15.9	0.0035	0.0042	1.24	-	-	-	-	-	0.0018
Comparative Steel	13	0.17	0.14	0.61	0.012	0.015	0.029	0.19	12.6	-	0.0051	1.08	-	-	-	-	-	0.0011
Comparative Steel	14	0.15	0.18	0.92	0.012	0.016	0.017	0.18	11.2	0.0200	0.0049	1.03	-	-	-	-	-	0.0012
Comparative Steel	15	0.17	0.10	0.67	0.016	0.051	0.026	0.08	1.6	0.0018	0.0041	0.99	-	-	-	-	-	0.0017
Comparative Steel	16	0.25	0.25	0.92	0.016	0.019	0.019	0.07	3.7	0.0021	0.0048	1.15	-	-	-	-	-	0.0012
Comparative Steel	17	0.21	0.15	0.84	0.011	0.012	0.021	-	0.2	-	0.0132	1.04	-	-	-	-	-	0.0013
Comparative Steel	18	0.21	0.18	0.90	0.012	0.012	0.019	0.08	6.7	0.0021	0.0178	1.07	-	-	-	-	-	0.0010
Comparative Steel	19	0.19	0.26	0.81	0.013	0.013	0.036	0.09	6.9	0.0016	0.0044	0.98	-	-	-	-	-	0.0010
Comparative Steel	20	0.23	0.16	0.79	0.015	0.012	0.033	0.11	9.2	0.0017	0.0072	1.02	-	-	-	-	-	0.0011
Comparative Steel	21	0.18	0.16	0.51	0.016	0.015	0.032	0.12	8.0	0.0014	0.0042	1.00	-	-	-	-	0.0500	0.0016
Steel of Present Invention	22	0.15	0.81	0.58	0.017	0.014	0.025	0.10	7.1	0.0018	0.0039	1.07	-	-	-	-	-	0.0010
Steel of Present Invention	23	0.15	1.10	0.97	0.018	0.013	0.031	0.13	10.0	0.0014	0.0041	1.19	-	-	-	-	-	0.0017

[Table 2]

Classification	Steel No.	Hot rolling conditions			Structure fraction of bainite (%)	Ferrite grain size number	Maximum diameter of Ti-based precipitates in extreme value statistics (μm)	Vickers hardness (HV)	Carburizing simulation	Quality of material of 1050°C carburized material		
		Heating temperature (°C)	Finish temperature (°C)	Cooling rate (°C/s)					Grain coarsening temperature (°C)	γ grain size number	Rolling fatigue life (relative value)	Rotating bending fatigue strength
Steel of Present Invention	1	1210	920	0.59	0	10	18	189	>1100	10	2.7	OK
Steel of Present Invention	2	1210	930	0.58	0	9	38	173	>1100	9	2.7	OK
Steel of Present Invention	3	1250	940	0.59	0	9	36	180	>1100	10	1.9	OK
Steel of Present Invention	4	1230	930	0.54	0	10	29	187	>1100	10	2.4	OK
Steel of Present Invention	5	1220	910	0.52	5	10	35	193	>1100	10	2.5	OK
Steel of Present Invention	6	1150	940	0.51	0	9	28	164	>1100	10	2.2	OK
Steel of Present Invention	7	1170	920	0.56	0	9	35	177	>1100	10	2.5	OK
Steel of Present Invention	8	1160	930	0.48	0	8	17	207	>1100	10	2.7	OK
Steel of Present Invention	9	1220	930	0.59	0	9	36	209	>1100	10	2.1	OK
Steel of Present Invention	10	1240	930	0.58	5	9	26	180	>1100	10	2.0	OK
Steel of Present Invention	11	1240	900	0.63	0	9	17	189	>1100	9	2.3	OK
Steel of Present Invention	12	1220	920	0.59	0	9	18	170	>1100	10	2.2	OK
Comparative Steel	13	1180	910	0.54	0	10	34	164	1050	9	2.9	OK
Comparative Steel	14	1190	930	0.52	0	9	31	172	>1100	10	0.2	NG
Comparative Steel	15	1200	940	0.71	5	8	34	147	1000	3	0.4	NG
Comparative Steel	16	1240	930	0.60	0	8	28	217	>1100	9	1.0	NG
Comparative Steel	17	1250	930	0.69	0	7	-	181	1050	3	1.0	NG
Comparative Steel	18	1220	930	0.52	0	9	59	184	1050	3	0.7	NG
Comparative Steel	19	1000	900	0.58	0	9	61	195	1000	3	0.9	NG
Comparative Steel	20	1220	920	0.59	0	10	45	201	>1100	9	0.8	OK
Comparative Steel	21	1220	930	0.55	0	10	62	161	1000	3	0.6	NG
Steel of Present Invention	22	1200	940	0.57	0	9	29	195	>1100	10	2.4	OK
Steel of Present Invention	23	1210	920	0.58	0	9	31	225	>1100	9	2.5	OK

[0080] The microstructure of each steel bar (case hardening steel) after being hot rolled and cooled was observed, the structure was identified by the following method, and the structure fraction of bainite was measured.

[0081] In addition, for each steel bar (case hardening steel), the ferrite grain size was measured according to the specification of JIS G 0552 and the grain size number was examined.

[0082] For each steel bar (case hardening steel), the maximum diameter of Ti-based precipitates was predicted using extreme value statistics by the following method.

[0083] For each steel bar (case hardening steel), the Vickers hardness was measured as an index of cold workability by the following method.

[0084] In order to evaluate the coarse grain prevention properties, a carburizing simulation was performed under the following conditions.

[0085] Furthermore, the γ grain size number, the rolling fatigue life, and the rotational bending fatigue strength were evaluated as the quality of the material after the carburizing by the following method.

[0086] The results are shown in Table 2.

"Structure Fraction of Bainite"

[0087] A sample was taken by cutting (traversing) each steel bar (case hardening steel) in a direction perpendicular to the axial direction. After burying the obtained sample in a resin, the cut surface (observed section) was polished. The observed section after being polished was corroded with Nital to expose and observe the microstructure, and a bainitic structure in the microstructure was identified. Furthermore, the area ratio of the bainitic structure in the observed section was obtained and was used as the structure fraction (%) of bainite.

[0088] A structure other than the bainite was ferrite, or ferrite and pearlite.

"Maximum Diameter of Ti-Based Precipitates"

[0089] Prediction of the maximum diameter of the Ti-based precipitates using extreme value statistics was performed by the following method. Whether or not the precipitates were based on Ti was determined by the difference in contrast in an optical microscope. The validity of the identification method based on the difference in contrast was confirmed in advance by a scanning electron microscope with an energy dispersive X-ray spectrometer.

[0090] A test piece was taken from each steel bar (case hardening steel), and a region of a 100 mm^2 inspection standard area (region of 10 mm x 10 mm) was prepared in advance for 16 visual fields in a longitudinal section of the steel bar. In addition, the maximum precipitate of Ti-based precipitates in each 100 mm^2 inspection standard area was detected and photographed with an optical microscope at a magnification of 1,000-fold. This was repeated 16 times (that is, the number of inspections was 16 visual fields) for the visual field of each 100 mm^2 inspection standard area. From the obtained photograph, the diameter of the maximum precipitate in each inspection standard area was measured. In a case where the precipitate is elliptical, the geometric mean of the major axis and the minor axis is obtained and taken as the diameter of the precipitate. 16 pieces of data of the 16 diameters of the maximum precipitates obtained were plotted on extreme value probability paper by the method described in pp. 233 to 239 of "Fatigue of Metals: Effects of Fine Defects and Inclusions" published by YOKENDO LTD. PUBLISHERS, a maximum precipitate distribution line (a linear function of maximum precipitate diameter and extreme value statistics standardized variable) was obtained, and the maximum precipitate distribution line was subjected to extrapolation such that the diameter of the maximum precipitates in an area of 30,000 mm^2 where prediction was performed was predicted.

"Vickers Hardness (HV)"

[0091] A sample was taken by cutting (traversing) each steel bar (case hardening steel) after rolling in the direction perpendicular to the axial direction. After burying the obtained sample in a resin, the cut surface (observed section) was polished. Regarding a portion at a diameter 1/4 depth from the surface of the observed section after being polished, the Vickers hardness was measured five times in total under a load of 10 kg based on "Vickers hardness test - Test method" in JIS Z 2244 (2009), and the average value was taken as the Vickers hardness. When the Vickers hardness is 230 HV or less, excellent cold forgeability was determined.

(Carburizing Simulation)

[0092] Each steel bar (case hardening steel) was subjected to spheroidizing annealing, and thereafter an upsetting test piece was prepared. After performing upsetting at a rolling reduction of 50%, a carburizing simulation was conducted under the following conditions.

[0093] In the carburizing simulation, the heating temperature was set to three temperatures, 1000°C, 1050°C, and 1100°C, and in a case of any of the heating temperatures, heating was performed for five hours, followed by water cooling. The cut surface of each test piece after the carburizing simulation was polished and then corroded, and prior austenite grain sizes were observed to obtain a grain coarsening temperature (coarse grains generation temperature). The measurement of the prior austenite grain sizes was performed according to JIS G 0551, about 10 visual fields were observed at a magnification of 400-fold, and the generation of coarse grains was determined when even a single coarse grain having a grain size number of No. 5 or less was present.

[0094] A grain coarsening temperature of higher than 1100°C was determined as good coarse grain prevention properties, and a grain coarsening temperature of 1100°C or lower was determined as inferior coarse grain prevention properties. The grain coarsening temperatures are shown in Table 2.

(Evaluation of Quality of Material After Carburizing)

[0095] Next, each steel bar (case hardening steel) was subjected to cold forging at a rolling reduction of 50%, a columnar rolling fatigue test piece having a diameter of 12.2 mm and an Ono type rotating bending test piece (with an R1.14 notch) having a parallel portion with a diameter of 9 mm were prepared, and carburizing was performed under the condition of five hours and a carbon potential of 0.8% at 1050°C. The temperature of a quenching oil was 130°C, and tempering was performed at 180°C for two hours.

[0096] For each of the obtained carburizing and quenching material, the γ (austenite) grain size of the carburized layer was investigated by the following method.

[0097] A samples was taken by cutting (traversing) the parallel portion subjected to the Ono type rotating bending after the carburizing, quenching and tempering, in the direction perpendicular to the axial direction. After burying the obtained sample in a resin, the cut surface (observed section) was polished. Corrosion was caused to expose austenite grains from the observed section after being polished, and the austenite grain size was measured in a visual field centered on a position at a depth of 200 μm from the surface according to the specification of JIS G 0551.

[0098] For each of the carburizing and quenching material, the rolling fatigue properties were evaluated using a point contact type rolling fatigue testing machine (Hertz maximum contact stress 5884 MPa). As a measure of fatigue life of the rolling fatigue properties, L10 life defined as "number of stress repetitions at which fatigue fracture occurs at a cumulative failure probability of 10% obtained by plotting test results on Weibull probability paper" was used. The rolling fatigue life represents a relative value of the L10 life of each material when the L10 life of Comparative Steel No. 17 as was set to 1.

[0099] For each of the carburizing and quenching material, the bending fatigue strength was evaluated using an Ono type rotating bending fatigue testing machine. Regarding the rotating bending fatigue strength, the material that withstood a stress of 550 MPa 10,000,000 times was evaluated as "OK", and the material that was fractured was evaluated as "NG".

[0100] These results are collectively shown in Table 2.

[0101] As shown in Table 2, the grain coarsening temperatures of steels of the present invention (Nos. 1 to 12, 22, and 23) were higher than 1100°C, the γ grain size of the carburized material heated at 1050°C was also No. 7 or higher in terms of grain size number, which means fine grains, and the rolling fatigue life and the result of the rotating bending fatigue test were good.

[0102] On the other hand, since Comparative Steel No. 13 did not contain Bi, the grain coarsening temperature was lower than those of the steels of the present invention.

[0103] In addition, in Comparative Steel No. 14, since the Bi content exceeded the upper limit specified in the present invention, initial cracks assumed to have occurred during hot rolling were present, and thus the rolling fatigue life and the result of the rotating bending fatigue test were inferior to those of the steels of the present invention.

[0104] Comparative Steel No. 15 had a large S content and did not satisfy Expression (1). Therefore, fatigue fracture originated from MnS had occurred, and the rolling fatigue life of the result of the rotating bending fatigue test were inferior to those of the steels of the present invention. In addition, in No. 15, precipitates of Ti-based carbonitride effective in preventing coarsening caused by the formation of a large amount of Ti-based sulfide could not be sufficiently obtained, and the grain coarsening temperature was lower than those of the steels of the present invention.

[0105] Since Comparative Steels Nos. 16 and 17 did not satisfy Expression (1), fatigue fracture originated from MnS occurred, and the rolling fatigue life and the result of the rotating bending fatigue test were inferior to those of the steels of the present invention.

[0106] In addition, in No. 17, precipitates of Ti-based carbonitride effective in preventing coarsening could not be sufficiently obtained, and the grain coarsening temperature was lower than those of the steels of the present invention.

[0107] In Comparative Steel No. 18, the N content was high, and coarse TiN was formed. Therefore, the rolling fatigue properties and the rotating bending fatigue properties were inferior to those of the steels of the present invention. Furthermore, in No. 18, the amount of precipitates of Ti-based carbonitride effective in preventing coarse grains was reduced due to the formation of coarse TiN. Therefore, the grain coarsening temperature was lower than those of the steels of the present invention.

[0108] In Comparative Steel No. 19, since the heating temperature before rolling was low, coarse Ti-based precipitates formed in the casting process remained unsolubilized and became coarsened in the heating process. Therefore, the rolling fatigue properties and the rotating bending fatigue properties were interior to those of the steels of the present invention. In addition, as a result of the reduction in the amount of precipitates of Ti-based carbonitride effective in preventing coarse grains, the grain coarsening temperature was inferior to those of the steels of the present invention.

[0109] Comparative Steel No. 20 had a large N content and thus had coarse TiN formed, so that the rolling fatigue properties were inferior to those of the steels of the present invention.

[0110] Since Comparative Steel No. 21 had a large Nb content, the carburizing properties were degraded, and a sufficient carbon concentration could not be obtained. As a result, the strength was insufficient, and the rolling fatigue life and the result of the rotating bending fatigue test were inferior to those of the steels of the present invention.

[Industrial Applicability]

[0111] The case hardening steel according to the present invention has a predetermined chemical composition, and the maximum diameter of Ti-based precipitates is controlled to be in a predetermined range, thereby achieving excellent coarse grain prevention properties during carburizing. Therefore, with the case hardening steel according to the present invention, heat treatment strain due to carburizing and quenching can be suppressed, and excellent fatigue properties are obtained after the carburizing and quenching. In addition, the carburized component manufactured by performing carburizing and quenching on the case hardening steel of the present invention has less heat treatment strain and has excellent fatigue properties.

[0112] With the manufacturing method of a case hardening steel according to the present invention, the case hardening steel which has excellent coarse grain prevention properties during carburizing can be manufactured. This case hardening steel can suppress heat treatment strain due to carburizing and quenching and achieves excellent fatigue properties after the carburizing and quenching.

[0113] Therefore, the industrial effect of the present invention is extremely remarkable.

Claims

1. A case hardening steel comprising, as a chemical composition, by mass%,
C: 0.10% to 0.30%,
Si: 0.02% to 1.50%,
Mn: 0.30% to 1.80%,

S: 0.003% to 0.020%,
Cr: 0.40% to 2.00%,
Al: 0.005% to 0.050%,
Ti: 0.06% to 0.20%,
Bi: 0.0001% to 0.0050%,
Mo: 0% to 1.50%,
Ni: 0% to 3.50%,
V: 0% to 0.50%,
B: 0% to 0.0050%,
Nb: 0% or more and less than 0.040%,
P: limited to 0.050% or less,
N: limited to 0.0060% or less,
O: limited to 0.0025% or less, and
a remainder including an iron and impurities,
wherein an Expression (1) is satisfied, and
in a longitudinal section, a maximum diameter of Ti-based precipitates predicted by extreme value statistics under a condition that an inspection standard area is 100 mm^2, a number of inspections is 16 visual fields, and an area where a prediction is performed is 30,000 mm^2, is 40 μm or less,

$$Ti/S \geq 6.0 \qquad \ldots\text{Expression (1)}$$

where Ti in the Expression (1) represents a Ti content by mass%, and S in the Expression (1) represents a S content by mass%.

2. The case hardening steel according to claim 1 comprising, as the chemical composition, by mass%, one or more selected from the group consisting of
Mo: 0.02% to 1.50%,
Ni: 0.10% to 3.50%,
V: 0.02% to 0.50%,
B: 0.0002% to 0.0050%, and
Nb: more than 0% and less than 0.040%.

3. The case hardening steel according to claim 1 or 2,
wherein a metallographic structure contains a bainite, and
a structure fraction of the bainite is 30% or less.

4. The case hardening steel according to any one of claims 1 to 3, wherein the metallographic structure contains a ferrite, and
a grain size number of the ferrite is No. 8 to No. 11 specified in JIS G 0552.

5. A carburized component comprising:

the case hardening steel according to any one of claims 1 to 4.

6. A manufacturing method of a case hardening steel comprising:

a heating process of heating steel including, as a chemical composition, by mass, C: 0.10% to 0.30%, Si: 0.02% to 1.50%, Mn: 0.30% to 1.80%, S: 0.003% to 0.020%, Cr: 0.40% to 2.00%, Al: 0.005% to 0.050%, Ti: 0.06% to 0.20%, Bi: 0.0001% to 0.0050%, Mo: 0% to 1.50%, Ni: 0% to 3.50%, V: 0% to 0.50%, B: 0% to 0.0050%, Nb: 0% or more and less than 0.040%, P: limited to 0.050% or less, N: limited to 0.0060% or less, O: limited to 0.0025% or less, and a remainder including an iron and impurities, and satisfies an Expression (2), at a temperature of 1150°C or higher for a holding time of ten minutes or longer; and
a hot rolling process of hot rolling the steel into a wire rod or a steel bar,

$$\mathrm{Ti/S} \geq 6.0 \qquad \ldots\text{Expression (2)}$$

where Ti in the Expression (2) represents a Ti content (mass%), and S in the Expression (2) represents a S content (mass%).

7. The manufacturing method of a case hardening steel according to claim 6 comprising:

a cooling process of slow cooling the wire rod or the steel bar at a cooling rate of 1.00 °C/s or less in a temperature range of 800°C to 500°C after the hot rolling process.

8. The manufacturing method of a case hardening steel according to claim 6 or 7,
wherein, in the hot rolling process, a finish temperature is set to 840°C to 1000°C.

INTERNATIONAL SEARCH REPORT

International application No.
PCT/JP2016/089086

A. CLASSIFICATION OF SUBJECT MATTER
C22C38/00(2006.01)i, *C21D8/06*(2006.01)i, *C22C38/58*(2006.01)i, *C21D1/06*(2006.01)n

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)
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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	1996-2017
Kokai Jitsuyo Shinan Koho	1971-2017	Toroku Jitsuyo Shinan Koho	1994-2017

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 2010-248630 A (JFE Steel Corp., JFE Bars & Shapes Corp.), 04 November 2010 (04.11.2010), (Family: none)	1-8
A	JP 2011-219854 A (JFE Steel Corp., JFE Bars & Shapes Corp.), 04 November 2011 (04.11.2011), (Family: none)	1-8
A	JP 2012-62536 A (JFE Steel Corp., JFE Bars & Shapes Corp.), 29 March 2012 (29.03.2012), (Family: none)	1-8

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

*	Special categories of cited documents:	"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
"A"	document defining the general state of the art which is not considered to be of particular relevance	"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
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"L"	document which may throw doubts on priority claim(s) or which is cited to establish the publication date of another citation or other special reason (as specified)	"&"	document member of the same patent family
"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		

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INTERNATIONAL SEARCH REPORT

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C (Continuation). DOCUMENTS CONSIDERED TO BE RELEVANT

Category*	Citation of document, with indication, where appropriate, of the relevant passages	Relevant to claim No.
A	JP 2007-31787 A (Kobe Steel, Ltd.), 08 February 2007 (08.02.2007), (Family: none)	1-8
A	JP 2005-240175 A (Nippon Steel Corp.), 08 September 2005 (08.09.2005), (Family: none)	1-8

Form PCT/ISA/210 (continuation of second sheet) (January 2015)

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- JP 4448456 B **[0012]**
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Non-patent literature cited in the description

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- Fatigue of Metals: Effects of Fine Defects and Inclusions. YOKENDO LTD. PUBLISHERS, 233-239 **[0090]**