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(71) Applicant: **Nippon Steel Corporation**
Tokyo 100-8071 (JP)

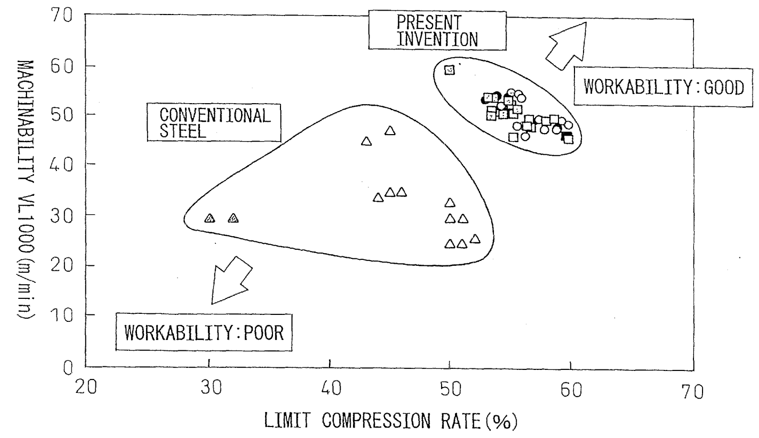
(72) Inventors:
- **HASHIMURA, Masayuki**
Tokyo 100-8071 (JP)
- **MIYANISHI, Kei**
Tokyo 100-8071 (JP)
- **KOZAWA, Shuji**
Tokyo 100-8071 (JP)
- **KUBOTA, Manabu**
Tokyo 100-8071 (JP)
- **OCHI, Tatsuro**
Tokyo 100-8071 (JP)

(74) Representative: **Vossius & Partner**
Siebertstrasse 4
81675 München (DE)

(54) STEEL FOR CASE HARDENING WHICH HAS EXCELLENT COLD WORKABILITY AND MACHINABILITY AND WHICH EXHIBITS EXCELLENT FATIGUE CHARACTERISTICS AFTER CARBURIZING AND QUENCHING, AND PROCESS FOR PRODUCTION OF SAME

(57) Cold worked, machined, and carburized quenched case-hardened steel prevented from formation of coarse grains, that is, case-hardened steel superior in cold workability, machinability, and fatigue characteristics after carburized quenching **characterized by** limiting, by mass%, S: 0.001 to 0.15%, Ti: 0.05 to 0.2%, Al: 0.04% or less, and N: 0.0050% or less, containing other specific ingredients in specific ranges, furthermore containing one or more of Mg: 0.003% or less, Zr: 0.01% or less, and Ca: 0.005% or less, limiting the amount of precipitation of AlN to 0.01% or less, and having a density d (/mm^2) of sulfides with a equivalent circle diameter of over 20 μm and an aspect ratio of over 3 and a content of S [S] (mass%) satisfying $d \leq 1700[S]+20$.

Fig.1



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Description

Technical Field

[0001] The present invention relates to case-hardened steel produced by hot rolling, hot forging, or other hot working, then cold forged, rolled, or otherwise cold worked, cut, etc., then treated by carburized quenching and a method of production of the same.

Background Art

[0002] Gears, bearings, and other rolling parts and constant velocity joints, shafts, and other rotation transmission parts require surface hardness, so are treated by carburized quenching. These carburized parts are, for example, produced by the process of using medium carbon alloy steel for machine structures prescribed by JIS G 4052, JIS G 4104, JIS G 4105, JIS G 4106, etc. and hot forging, warm forging, cold forging, rolling, or otherwise plastic working it or cutting it to obtain a predetermined shape, then treating it by carburized quenching.

[0003] When producing carburized parts, the heat treatment strain arising due to the carburized quenching sometimes causes the shape precision of the parts to degrade. In particular, with gears, constant velocity joints, or other parts, the heat treatment strain becomes a cause of noise or vibration. Furthermore, it sometimes causes a deterioration of fatigue characteristics at the contact surface.

[0004] Further, with a shaft etc., if the distortion due to heat treatment strain becomes large, the efficiency of transmission of power or the fatigue characteristics are impaired. The biggest reason for this heat treatment strain is the coarse grains formed unevenly due to the heating at the time of carburized quenching.

[0005] In the past, annealing was performed after forging and before carburized quenching so as to suppress the formation of coarse grains. However, if annealing, the increase in production costs becomes an issue.

[0006] Further, gears, bearings, and other rolling parts are subjected to high surface pressures, so are treated by deep carburization. With deep carburization, to shorten the carburization time, usually the 930°C or so carburization temperature is raised to a 990 to 1090°C temperature region. For this reason, with deep carburization, coarse grains easily form.

[0007] To suppress the formation of coarse grains at the time of carburized quenching, the quality of the case-hardened steel, that is, the material before plastic working, is important.

[0008] To suppress coarsening of the crystal grains at a high temperature, fine precipitates are effective. Case-hardened steel utilizing Nb and Ti precipitates, AlN, etc. have been proposed (for example, Patent Literatures 1 to 5).

Citation List

Patent Literature

[0009]

PTL 1: Japanese Patent Publication (A) No. 11-335777
PTL 2: Japanese Patent Publication (A) No. 2001-303174
PTL 3: Japanese Patent Publication (A) No. 2004-183064
PTL 4: Japanese Patent Publication (A) No. 2004-204263
PTL 5: Japanese Patent Publication (A) No. 2005-240175

Summary of Invention

Technical Problem

[0010] However, if utilizing fine precipitates to suppress the formation of coarse grains, precipitation strengthening will cause the case-hardened steel to harden. Further, the addition of alloy elements for forming precipitates will also cause the case-hardened steel to harden. For this reason, with steel prevented from forming coarse grains at a high temperature, the deterioration of cold forgeability, cutting, and other cold workability became a new issue.

[0011] In particular, cutting is working requiring a high precision close to the final shape. A slight rise in hardness has a great effect on the precision. Therefore, when using case-hardened steel, it is extremely important not only to prevent the formation of coarse grains, but to also consider the machineability (ease of cutting of material).

[0012] In the past, to improve the machineability, it has been known to be effective to add Pb, S, and other elements improving the machineability.

[0013] However, Pb is a substance having an environmental load. Due to the importance of environmentally friendly technology, addition of Pb to steel materials is being limited.

[0014] Further, S forms MnS etc. in the steel to improve the machineability, but the coarse MnS inclusions elongated by the hot working become origin of fracture. For this reason, addition of a large amount of S can easily become a cause of a deterioration of cold forgeability or rolling contact fatigue or other mechanical properties.

[0015] The present invention, in view of this situation, prevents the formation of coarse grains in case-hardened steel which is forged, rolled, or otherwise cold worked, cut, and treated by carburized quenching such as in carburized parts in which fatigue characteristics are demanded, in particular bearing parts, rolling parts, etc. in which rolling contact fatigue characteristics are demanded, and provides case-hardened steel superior in cold workability, machinability, and fatigue characteristics after carburized quenching and a method of production of the same.

Solution to Problem

[0016] If treating steel to which Ti has been added by carburized quenching, Ti precipitates will form origin of fatigue fracture and the fatigue characteristics, in particular the rolling contact fatigue characteristic, will easily be degraded. However, if limiting the content of N and raising the hot rolling temperature etc. so as to cause the Ti precipitates to finely disperse, achievement of both prevention of coarse grains and good fatigue characteristics is possible. Furthermore, for improvement of the machineability, it is important to add S and add one or more of Mg, Zr, and Ca to control the size and shape of the sulfides.

[0017] The gist of the present invention is as follows.

(1) Case-hardened steel superior in cold workability, machinability, and fatigue characteristics after carburized quenching characterized by containing, by mass%,
C: 0.1 to 0.5%,
Si: 0.01 to 1.5%,
Mn: 0.3 to 1.8%,
S: 0.001 to 0.15%,
Cr: 0.4 to 2.0%, and
Ti: 0.05 to 0.2%,
limiting
Al: 0.04% or less,
N: 0.0050% or less,
P: 0.025% or less,
O: 0.0025% or less,
further having one or more of
Mg: 0.003% or less,
Zr: 0.01% or less, and
Ca: 0.005% or less,
having a balance of iron and unavoidable impurities,
limiting an amount of precipitation of AlN to 0.01% or less, and
having a density d (/mm^2) of sulfides of a equivalent circle diameter of over 20 μm and an aspect ratio of over 3 and a content of S [S] (mass%) satisfying

$$d \leq 1700[S]+20.$$

(2) Case-hardened steel superior in cold workability, machinability, and fatigue characteristics after carburized quenching as set forth in the above (1), characterized by further containing, by mass%,
Nb: less than 0.04%.

(3) Case-hardened steel superior in cold workability, machinability, and fatigue characteristics after carburized quenching as set forth in the above (1) or (2), characterized by further containing, by mass%, one or more of
Mo: 1.5% or less,
Ni: 3.5% or less,
V: 0.5% or less, and
B: 0.005% or less.

(4) Case-hardened steel superior in cold workability, machinability, and fatigue characteristics after carburized quenching as set forth in any one of the above (1) to (3), characterized by limiting a structural fraction of bainite to

30% or less.

(5) Case-hardened steel superior in cold workability, machinability, and fatigue characteristics after carburized quenching as set forth in any one of the above (1) to (4), **characterized in that** a grain size number of ferrite is 8 to 11 as defined by JIS G 0551.

(6) Case-hardened steel superior in cold workability, machinability, and fatigue characteristics after carburized quenching as set forth in any one of the above (1) to (5), **characterized in that** a maximum size of Ti precipitates is 40 μm or less.

(7) A method of production of case-hardened steel superior in cold workability, machinability, and fatigue characteristics after carburized quenching characterized by heating a steel material comprised of the ingredients of any of the above (1) to (3) to 1150°C or more, hot working it at a finishing temperature of 840 to 1000°C, and cooling it in a 800 to 500°C temperature range by 1°C/s or less.

Advantageous Effects of Invention

[0018] The case-hardened steel of the present invention is superior in forgeability, machineability, and other workability. Even when producing parts by the cold forging process, coarsening of the crystal grains due to heating at the time of carburized quenching is suppressed. Deterioration of the dimensional precision due to quenching strain is much smaller than the past.

[0019] Further, according to the case-hardened steel of the present invention, the problem of the deterioration of machinability due to the prevention of formation of coarse grains in the past is solved. Further, higher precision of part shapes is achieved. Furthermore, the tool life also becomes longer.

[0020] Further, parts made of the case-hardened steel of the present invention are kept from forming coarse grains even in high temperature carburization, sufficient strength characteristics such as rolling contact fatigue characteristics can be obtained, etc. The contribution to industry is extremely remarkable.

Brief Description of Drawings

[0021]

FIG. 1 is a view for explaining a balance of machineability and cold workability of the present invention.
FIG. 2 is a view showing a position for measuring a cooling rate at the time of solidification.
FIG. 3 is a view showing a test piece used for an upset test.

Description of Embodiments

[0022] Coarsening of crystal grains due to carburized quenching is prevented by using precipitates as pinning particles to suppress grain growth. In particular, making Ti precipitates mainly comprised of TiC and TiCS precipitate finely at the time of cooling after hot working is extremely effective for preventing the formation of coarse grains. Furthermore, to prevent the formation of coarse grains, it is preferable to make NbC and other Nb precipitates finely precipitate in the case-hardened steel.

[0023] However, if the amount of N contained in the steel is great, the coarse TiN formed at the time of casting will not be solubilized by the heating of the hot rolling or hot forging and will sometimes remain in large amounts. If coarse TiN remains, at the time of carburized quenching, the TiN will act as precipitation nuclei resulting in TiC, TiCS, and furthermore NbC precipitating and fine dispersion of the precipitates being inhibited. Therefore, to enable fine Ti precipitates and Nb precipitates to prevent formation of coarse grains at the time of carburized quenching, it is important to reduce the amount of N and solubilize the Ti precipitates and Nb precipitates at the time of heating in hot working.

[0024] Further, if coarse AlN remains at the time of heating in hot working, in the same way as TiN, formation of fine precipitates acting as pinning particles is inhibited.

[0025] However, the temperature at which AlN forms a solid solution is lower than that of TiN, so compared with TiN, it is easier to solubilize at the time of heating in hot rolling. Furthermore, during the hot working and at the time of cooling after that, AlN precipitates and grows slower than Ti precipitates and Nb precipitates. Therefore, by preventing AlN from remaining at the time of heating in hot working, it is possible to limit the amount of precipitation of the AlN contained in the case-hardened steel.

[0026] Therefore, according to the case-hardened steel of the present invention limited in amount of precipitation of AlN, it is possible to utilize fine Ti precipitates and Nb precipitates to prevent the formation of coarse grains at the time of carburized quenching.

[0027] Furthermore, to enable the pinning effect of Ti precipitates and Nb precipitates to be stably exhibited, it is effective to cause Ti precipitates and Nb precipitates to precipitate by interphase boundary precipitation in the process of cooling after hot working and the diffusion and transformation from austenite. However, if bainite forms in the cooling process after hot rolling, interphase boundary precipitation of precipitates will become difficult.

[0028] Therefore, it is preferable to control the structure of the steel after hot rolling and suppress the formation of bainite and is more preferable to obtain a structure substantially not containing any bainite.

[0029] In the method of production, first, it is necessary to heat the steel material so that the Al, Ti, and Nb precipitates solute. In particular, it is important to raise the heating temperature of hot rolling, hot forging, or other hot working and cause the Ti precipitates and Nb precipitates to solute.

[0030] Next, after hot working, that is, after hot rolling or after hot forging, it is necessary to slow the cooling in the temperature region of precipitation of Ti precipitates and Nb precipitates. As a result, it is possible to make the Ti precipitates and Nb precipitates finely disperse in the case-hardened steel.

[0031] Further, if the ferrite grains of the steel material before carburized quenching are excessively fine, at the time of heating for carburization, coarse grains will easily form. For that reason, it is necessary to control the finishing temperature of the hot rolling or hot forging to prevent formation of fine ferrite.

[0032] Further, when working the case-hardened steel of the present invention into a gear etc., the teeth are formed by forging and gear cutting before carburized quenching. At that time, MnS and other sulfides cause the cold forgeability to drop, but are extremely effective for gear cutting. That is, sulfides exhibit the effect of suppressing changes in tool shape due to wear of the cutting tools and extending so-called tool life.

[0033] In particular, in the case of precision shapes such as gears, if the cutting tool life is short, stable formation of gear shapes is not possible. For this reason, the cutting tool life has an effect not simply on the production efficiency or cost, but also the shape precision of the parts.

[0034] Therefore, to improve the machinability, it is desirable to cause formation of sulfides in the steel.

[0035] On the other hand, in hot rolling or hot forging, in particular the coarse MnS or other sulfides are often elongated. Furthermore, if the sulfides increase in length, the probability of their appearing as defects in the parts also becomes higher and the performance of the parts is lowered. Therefore, not only the size of the sulfides, but also control of the shape so as not to elongate is important.

[0036] Note that, to suppress coarsening of the sulfides, it is preferable to control the solidification speed at the time of casting.

[0037] To reduce the MnS and other soft sulfides, it is also effective to add Ti and cause the formation of TiCS and other Ti sulfides. However, if the soft MnS is reduced, the added S will no longer contribute to the improvement of the machineability.

[0038] Therefore, to improve the machineability, it is important to not only add S, but also control the soft sulfides in the molten steel to which Ti is added.

[0039] Therefore, it is preferable to control the shape of sulfides by control of the AlN required for suppressing coarse grains, addition of Ti, control of the amount of S, and, furthermore, addition of Zr, Mg, and Ca.

[0040] The machineability and cold workability will be further explained.

[0041] At the time of cold working, the sulfides mainly comprised of MnS deform and become origin of fracture. In particular, the coarse MnS lowers the limit compression rate and other aspects of cold forgeability. Further, if the MnS in the steel is coarse, anisotropy of the material characteristics will occur due to the shape of the MnS.

[0042] To apply case-hardened steel to various complicated parts, stable mechanical properties are demanded in all directions. For this reason, in the case-hardened steel of the present invention, it is preferable to make the sulfides mainly comprised of MnS finer and make their shapes substantially spherical. Further, it is more preferable that the change in shape be small even after forging and other cold working.

[0043] Addition of Zr, Mg, and Ca is effective for causing dispersion of fine sulfides. Furthermore, if Zr, Mg, Ca, etc. solute in the MnS, the resistance to deformation becomes higher and the sulfides no longer easily deform. Therefore, the addition of Zr, Mg, and Ca is extremely effective for suppression of elongating.

[0044] On the other hand, from the viewpoint of the machineability, increase of the amount of S is important. Due to the addition of S, the tool life at the time of cutting is improved. This effect is determined by the total amount of S. The effect of the shape of the sulfides is small. For this reason, by increasing the amount of addition of S and controlling the shape of the sulfides, it is possible to achieve both cold forgeability and machineability (tool life).

[0045] In case-hardened steel, not only the prevention of formation of coarse grains at the time of carburized quenching, but also securing cold workability and machineability is important. If increasing the amount of S, the machineability is improved, but a deterioration of cold workability is invited. Therefore, it is also important to secure a good cold workability when compared by the same amount of S.

[0046] FIG. 1 compares the relationship of machineability and cold workability for case-hardened steel with a good coarse grain characteristic suppressed in formation of coarse grains at the time of carburized quenching. In the present invention, it is possible to maintain a good coarse grain characteristic (coarse grain formation temperature > 970°C) while achieving both cold workability (limit compression rate) and machineability (drillability VL1000). In FIG. 1, the further to the top right, the better the balance of machineability and cold workability of the material.

[0047] Below, the present invention will be explained in detail.

[0048] First, the composition of ingredients will be explained. Below, "mass%" will be simply described as "%".

[0049] C is an element raising the strength of steel. In the present invention, to secure the tensile strength, 0.1% or more of C is added. An amount of C of 0.15% or more is preferable. On the other hand, if the content of C exceeds 0.5%, the steel remarkably hardens and the cold workability is degraded, so the upper limit is made 0.5%. Further, to secure toughness of the core part after carburization, the amount of C is preferably made 0.4% or less. An amount of C of 0.3% or less is more preferable.

[0050] Si is an element effective for deoxidation of steel. In the present invention, 0.01% or more is added. Further, Si is an element strengthening steel and improving the quenchability. Addition of 0.02% or more is preferable. Furthermore, Si is an element effective for increasing the grain boundary strength. Furthermore, in bearing parts and rolling parts, it is an element effective for extending lifetime by suppressing structural changes and deterioration of quality in the process of rolling contact fatigue. For this reason, when aiming at increasing the strength, addition of 0.1% or more is more preferable. In particular, to raise the rolling contact fatigue strength, addition of 0.2% or more of Si is preferable.

[0051] On the other hand, if the amount of Si exceeds 1.5%, the hardening causes the cold forging and other cold workability to deteriorate, so the upper limit is made 1.5%. Further, to raise the cold workability, it is preferable to make the amount of Si 0.5% or less. In particular, when stressing cold forgeability, the amount of Si is preferably 0.25% or less.

[0052] Mn is effective for deoxidation of steel. Furthermore, it is an element improving the strength and quenchability of steel. In the present invention, 0.3% or more is added. On the other hand, if the amount of Mn exceeds 1.8%, the rise in hardness causes the cold forgeability to be degraded, so 1.8% is made the upper limit. The preferable range of the amount of Mn is 0.5 to 1.2%. Note that, when stressing the cold forgeability, it is preferable to make the upper limit of the amount of Mn 0.75%.

[0053] S is an element forming MnS in steel and improving the machineability. In the present invention, to improve the machineability, the content of S is made 0.001% or more. The preferable lower limit of the amount of S is 0.1%. On the other hand, if the amount of S is over 0.15%, grain boundary segregation causes grain boundary embrittlement to be invited, so the upper limit is made 0.15%. Further, if considering the fact that the parts require high strength, the amount of S is preferably 0.05% or less. Furthermore, when considering the strength or cold workability and, furthermore, the stability of the same, the amount of S is preferably made 0.03% or less.

[0054] Note that, in the past, in bearing parts and rolling parts, it was considered necessary to reduce the S since MnS caused deterioration of the rolling fatigue life. However, the inventors etc. discovered that for improvement of the machinability, the content of S has a large effect, while for improvement of the cold workability, the shape of the sulfides has a large effect. In the present invention, one or more of Mg, Zr, and Ca are added to control the shape of the sulfides, so it is possible to make the amount of S 0.01% or more. When stressing the machineability, the amount of S is preferably made 0.02% or more.

[0055] Cr is an element effective for improving the strength and quenchability of steel. In the present invention, 0.4% or more is added. Furthermore, in bearing parts and rolling parts, it is effective for increasing the residual amount of γ of the surface layer after carburization and increasing lifetime by suppressing changes in structure and degradation of quality in the process of rolling contact fatigue, so addition of 0.7% or more is preferable. The more preferable amount of Cr is 1.0% or more. On the other hand, if adding Cr over 2.0%, the rise in hardness causes the cold workability to be degraded, so the upper limit is made 2.0%. To improve the cold forgeability, the amount of Cr is preferably made 1.5% or less.

[0056] Ti is an element forming carbides, carbosulfides, nitrides, and other precipitates in the steel. In the present invention, to utilize the fine TiC and TiCS to prevent the formation of coarse grains at the time of carburized quenching, 0.05% or more of Ti is added. The preferable lower limit of the amount of Ti is 0.1%. On the other hand, if adding over 0.2% of Ti, precipitation hardening causes the cold workability to remarkably degrade, so the upper limit of the amount of Ti is made 0.2%. Further, to suppress precipitation of TiN and improve the rolling contact fatigue characteristic, it is preferable to make the amount of Ti 0.15% or less.

[0057] Al is a deoxidizing agent. Addition of 0.005% or more is preferable, but the invention is not limited to this. On the other hand, if the amount of Al exceeds 0.04%, the AlN will remain without being solubilized by the heating of the hot working. For this reason, the coarse AlN will form precipitation nuclei for precipitates of Ti and Nb and formation of fine precipitates will be inhibited. Therefore, to prevent coarsening of the crystal grains at the time of carburized quenching, the amount of Al has to be made 0.04% or less.

[0058] N is an element forming nitrides. In the present invention, to suppress the formation of coarse TiN and AlN, the upper limit of the amount of N is made 0.0050%. This is because coarse TiN and AlN form precipitation nuclei for Ti precipitates mainly comprised of TiC and TiCS and Nb carbonitrides mainly comprised of NbC etc. and inhibit the dispersion of fine precipitates.

[0059] P is an impurity. It is an element which raises the resistance to deformation at the time of cold working and degrades the toughness. If excessively included,'the cold forgeability is degraded, so the content of P has to be limited to 0.025% or less. Further, to suppress embrittlement of the crystal grain boundaries and improve the fatigue strength, the content of P is preferably made 0.015% or less.

[0060] O is an impurity. It forms oxide inclusions in the steel and impairs the workability, so the content is limited to 0.0025% or less. Further, the case-hardened steel of the present invention includes Ti, so oxide inclusions including Ti are formed and act as precipitation nuclei causing TiC to precipitate. If the oxide inclusions increase, the formation of fine TiC is sometimes suppressed at the time of hot working.

[0061] Therefore, to make the Ti precipitates mainly comprised of TiC and TiCS finely disperse and suppress the coarsening of crystal grains at the time of carburized quenching, the upper limit of the amount of O is preferably made 0.0020%.

[0062] Furthermore, in bearing parts and rolling parts, the oxide inclusions sometimes serve as origin of rolling contact fatigue fracture. For this reason, when used for bearing parts and rolling parts, to improve the rolling life, the O content is preferably limited to 0.0012% or less.

[0063] Furthermore, in the case-hardened steel of the present invention, to control the form of the sulfides, it is necessary to add one or more of Mg, Zr, and Ca. Mg, Zr, and Ca form roughly spherical sulfides and further raise the deformation ability of MnS to suppress elongating due to hot working. In particular, Mg and Zr exhibit remarkable effects even when included in very small amounts, so care is preferably exercised in secondary materials etc. Furthermore, to stabilize the amounts of addition of Mg and Zr, it is preferable to use refractories containing Mg and Zr to control the content.

[0064] Mg is an element forming oxides and sulfides. Due to the inclusion of Mg, composite sulfides (Mn,Mg)S with MgS or MnS etc. are formed, so it is possible to suppress elongating of MnS. A very small amount of Mg is effective for control of the form of the MnS. To improve the workability, addition of 0.0002% or more of Mg is preferable.

[0065] Further, oxides of Mg finely disperse and form the nuclei for formation of MnS and other sulfides. To utilize oxides of Mg to suppress the formation of coarse sulfides, addition of 0.0003% or more of Mg is preferable. Furthermore, if adding Mg, the sulfides become somewhat hard and become harder to elongate due to hot working.

[0066] For control of the shape of the sulfides to contribute to improvement of the machinability and prevent the cold workability from being detracted from, addition of 0.0005% or more of Mg is preferable. Note that, hot forging has the effect of causing fine sulfides to uniform disperse and is effective for improvement of the cold workability.

[0067] On the other hand, oxides of Mg easily float up in molten steel, so the yield is low. From the viewpoint of the production costs, the upper limit of the content of Mg is preferably 0.003%. Further, if excessively adding Mg, large amounts of oxides are formed in the molten steel and deposition on the refractories, clogging of nozzles, and other trouble in steelmaking are sometimes caused. Therefore, the amount of addition of Mg is more preferably made 0.001% or less.

[0068] Zr is an element forming oxides, sulfides, and nitrides. If adding a very small amount of Zr, it combines with the Ti in the molten steel to form fine oxides, sulfides, and nitrides. Therefore, in the present invention, the addition of Zr is extremely effective for the control of inclusions and precipitates. To control the form of the inclusions and improve the workability, addition of 0.0002% or more of Zr is preferable, but the invention is not limited to this.

[0069] Oxides, sulfides, and nitrides including Zr and Ti form precipitation nuclei for MnS at the time of solidification. The Zr and Ti dissolve into the MnS precipitated around these oxides, sulfides, and nitrides including Zr and Ti resulting in a deterioration of the deformation ability. Therefore, to suppress the deformation of MnS and prevent elongating due to hot working, addition of 0.0003% or more of Zr is preferable.

[0070] On the other hand, Zr is an expensive element, so from the viewpoint of the production costs, the upper limit of the amount of Zr is preferably made 0.01%. The more preferable amount of Zr is 0.005% or less, still more preferably 0.003% or less.

[0071] Ca is an element forming oxides and sulfides. To control the form of the inclusions and improve the workability, 0.0002% or more of Ca is preferably added. The CaS and (Mn,Ca)S and the composite sulfides with Ti formed by the addition of Ca act as precipitation nuclei for MnS at the time of solidification.

[0072] In particular, the Ca and Ti dissolve in the MnS precipitated around the oxides and sulfides containing Ca and Ti resulting in a deterioration of the deformation ability. Therefore, to suppress deformation of MnS and prevent elongating due to hot working, addition of 0.0003% or more of Ca is preferable.

[0073] On the other hand, in the same way as Mg, if excessively adding Ca, deposition of the oxides on the refractories, clogging of nozzles, and other trouble in steelmaking are sometimes caused. Therefore, the amount of Ca is preferably made 0.005% or less.

[0074] Further, addition of two or more of Mg, Zr, and Ca is more preferable. It is possible to make roughly spherical sulfides finely disperse. When adding two or more of Mg, Zr, and Ca, it is preferable to make the total content 0.0005% or more. Further, to prevent deposition on the refractories etc. even when adding two or more of Mg, Zr, and Ca, it is preferable to make the total content 0.006% or less, more preferable to make it 0.003% or less.

[0075] Furthermore, to suppress the formation of coarse grains at the time of carburized quenching, in the same way as Ti, addition of Nb forming carbonitrides is preferable. Nb, in the same way as Ti, is an element bonding with C and N in the steel to form carbonitrides. Due to the addition of Nb, the effect of suppression of formation of coarse grains due to the Ti precipitates becomes more remarkable. Even if the amount of Nb added is very small, compared with the case of not adding Nb, the addition is extremely effective for prevention of coarse grains.

[0076] This is because the Nb forms a solid solution in the Ti precipitates and suppresses coarsening of the Ti precipitates. To suppress the formation of coarse grains at the time of heating in carburized quenching, addition of 0.01% or more of Nb is preferable, but the invention is not limited to this. On the other hand, if adding Nb in an amount of 0.04% or more, the steel hardens and the cold workability, in particular the cold forgeability and machinability, and, furthermore, the carburization characteristics are sometimes degraded. Therefore, the amount of addition of Nb is preferably made less than 0.04%. When stressing the cold forgeability or other cold workability and machinability, the preferable upper limit of the amount of Nb is less than 0.03%. Further, when stressing the carburization ability in addition to the workability, the preferable upper limit of the amount of Nb is less than 0.02%.

[0077] Further, to achieve both prevention of coarse grains and workability, it is preferable to adjust the total of the amount of addition of Nb and the amount of addition of Ti. The preferable range of Ti+Nb is 0.07% to less than 0.17%. In particular, in high temperature carburization or cold forged parts, the preferable range of Ti+Nb is over 0.09% to less than 0.17%.

[0078] Furthermore, to improve the strength and quenchability of the steel, one or more of Mo, Ni, V, B, and Nb may be added.

[0079] Mo is an element improving the strength and quenchability of steel. In the present invention, it is effective for increasing the amount of residual γ at the surface layer of carburized parts and further to increase the lifetime by suppression of structural changes and quality changes in the process of rolling contact fatigue. However, if adding over 1.5% of Mo, the rise in hardness causes the machinability and cold forgeability, to be degraded in some cases.

[0080] Therefore, making the content of Mo 1.5% or less is preferable. Mo is an expensive element. From the viewpoint of the production costs, making the amount 0.5% or less is more preferable.

[0081] Ni, in the same way as Mo, is an element effective for improving the strength and quenchability of the steel. However, if adding Ni over 3.5%, the rise in the hardness causes the cuttability and cold forgeability to deteriorate in some cases, so making the content of Ni 3.5% or less is preferable. Ni is also an expensive element. From the viewpoint of the production costs, the preferable upper limit is 2.0%. The further preferable upper limit of the amount of Ni is 1.0%.

[0082] V is an element improving the strength and quenchability if forming a solid solution in the steel. If the amount of V is over 0.5%, the rise in the hardness causes the machinability and cold forgeability to deteriorate in some cases, so making the upper limit of content 0.5% is preferable. The preferable upper limit of the amount of V is 0.2%.

[0083] B is an element effective for raising the quenchability of steel with addition in a very fine amount. Further, B forms boron-iron carbides in the cooling process after hot rolling, increases the growth rate of ferrite, and promotes softening. Furthermore, it is also effective for improving the grain boundary strength of carburized parts and for improving the fatigue strength and impact strength. However, if adding B in over 0.005%, the effect becomes saturated and the impact strength is degraded, so the upper limit of the content is preferably 0.005%. The preferable upper limit of the amount of B is 0.003%.

[0084] Note that, the effect of the addition of Si and Cu and, furthermore, the addition of Mo in suppressing structural changes and quality changes in bearing parts and rolling parts in the process of rolling contact fatigue is particularly large when the residual austenite (residual γ) at the surface layer after carburization is 30 to 40%. To control the residual amount of y of the surface layer to 30 to 40% in range, carbonitridation treatment is effective. Carbonitridation treatment is treatment for carburization, then nitridation in the process of diffusion treatment.

[0085] To make the residual amount of γ of the surface layer 30 to 40%, it is preferable to perform carbonitridation so that the nitrogen concentration of the surface layer becomes 0.2 to 0.6% in range. Note that, in this case, it is preferable to make the carbon potential at the time of carburization 0.9 to 1.3% in range.

[0086] Further, in the case-hardened steel of the present invention, the carbon and nitrogen penetrating the surface layer at the time of carburized quenching and the solute Ti react and fine Ti(C,N) precipitate in large amounts at the carburized layer. In particular, at the bearing parts and rolling parts, the Ti(C,N) at the surface layer causes the rolling fatigue life to be improved.

[0087] Therefore, to improve the rolling fatigue life, it is preferable to set the carbon potential at the time of carburization to 0.9 to 1.3%. Further, with carburization, then nitridation in the process of diffusion treatment, that is, carbonitridation treatment, it is preferable to set the conditions so that the nitrogen concentration of the surface becomes 0.2 to 0.6% in range.

[0088] Next, among the precipitates included in the case-hardened steel of the present invention, AlN and sulfides will be explained.

[0089] AlN forms the precipitation nuclei for Ti precipitates and Nb precipitates and inhibits the formation of fine precipitates. Therefore, in the present invention, it is necessary to limit the amount of precipitation of AlN included in the case-hardened steel. If the amount of precipitation of AlN is excessive, coarse grains are liable to be formed at the time of carburized quenching, so the amount of precipitation of AlN in the case-hardened steel is limited to 0.01% or less. The preferable upper limit of the amount of precipitation of AlN is 0.005%.

[0090] To suppress the amount of precipitation of AlN of the case-hardened steel, it is necessary to raise the hot working heating temperature and promote solubilization. The case-hardened steel of the present invention is limited in amount of N, so if heating it to a temperature where AlN is solubilized, the Ti precipitates and Nb precipitates can also be solubilized.

[0091] Note that, the amount of precipitation of AlN can be measured by chemical analysis of the extraction residue. The extraction residue is obtained by etching the steel by a bromine methanol solution and filtering by a 0.2 μm filter. Note that, even if using a 0.2 μm filter, in the process of filtration, the precipitates cause the filter to clog, so extraction of 0.2 μm or smaller fine precipitates is also possible.

[0092] MnS is useful for the improvement of the machinability, so it is necessary to secure the density. On the other hand, elongated coarse MnS impairs the cold workability, so the size and form have to be controlled.

[0093] The inventors etc. studied the relationship between the content of S, the size and shape of MnS inclusions, and the machinability and cold workability.

[0094] As a result, it was learned that when MnS inclusions observed under an optical microscope have a equivalent circle diameter of over 20 μm and an aspect ratio of over 3, they become origin of fracture at the time of cold working.

[0095] The equivalent circle diameter of an MnS inclusion is the diameter of a circle having an area equal to the area of the MnS inclusion and can be found by image analysis. The aspect ratio is the ratio of the length of the MnS inclusion divided by the thickness of the MnS.

[0096] Next, the inventors etc. studied the effects of the distribution of sulfides. The MnS inclusions of a hot rolled material of a diameter of 30 mm were observed under a scanning electron microscope and analyzed for the relationship of size, aspect ratio and density, and cold workability and machinability. The MnS inclusions are examined at a part of 1/2 radius from the surface of the cross-section parallel to the rolling direction. Ten fields of 1 mm×1 mm area were examined and the equivalent circle diameters, aspect ratios, and numbers of the sulfide inclusions present were found. Note that, the fact that the inclusions are sulfides was confirmed by an energy dispersive X-ray spectrometer attached to a scanning electron microscope.

[0097] The number of MnS inclusions with a equivalent circle diameter over 20 μm and an aspect ratio over 3 was counted and divided by the area to find the density d. It was learned that the density d of sulfides is influenced by the amount of S, so to achieve both machinability and cold workability, the following relation must be satisfied:

$$d \leq 1700[S]+20 \ (/\text{mm}^2)$$

[0098] Here, [S] indicates the content (mass%) of S. Furthermore, if coarse Ti precipitates are present in the steel, they become origin of contact fatigue fracture and the fatigue characteristics deteriorate in some cases.

[0099] The contact fatigue strength is a required characteristic of a carburized part and is the rolling contact fatigue characteristic or surface fatigue strength. To raise the contact fatigue strength, making the maximum size of the Ti precipitates less than 40 μm is preferable.

[0100] The maximum size of the Ti precipitates is found by statistics of extremes measured in the cross-section of the longitudinal direction of the case-hardened steel using a standard inspection area of 100 mm^2, inspection of 16 fields, and a prediction area of 30000 mm^2.

[0101] The method of measurement of the maximum size of precipitates using statistics of extremes is, for example, as described in Yukitaka Murakami, "Metal Fatigue - Effects of Small Defects and Nonmetallic Inclusions", Yokendo, pp. 233 to 239 (1993), a two-dimensional test method of estimating the largest precipitates obtained in a fixed area, that is, a prediction area (30000 mm^2).

[0102] The values are plotted on an extreme probability paper, the primary function of the maximum precipitate size and statistics of extremes standardized variable is found, and the maximum precipitate distribution line is extrapolated to predict the size of the largest precipitate in the prediction area.

[0103] Next, the structure of the case-hardened steel of the present invention will be explained.

[0104] The structural fraction of bainite in the case-hardened steel is preferably limited to 30% or less. This is because to prevent the formation of coarse grains at the time of carburized quenching, it is preferable to form fine precipitates at the grain boundary. That is, if the structural fraction of bainite formed at the time of cooling after hot working exceeds 30%, it becomes harder for the Ti precipitates and the Nb precipitates to be made to precipitate by interphase boundary precipitation.

[0105] Suppressing the structural fraction of bainite to 30% or less is also effective for improving the cold workability.

[0106] In the case of high temperature carburization or otherwise when the conditions for prevention of coarse grains are severe, the upper limit of the structural fraction of bainite is preferably made 20%, more preferably 10% or less. Furthermore, when cold forging, then performing high temperature carburization etc., the upper limit of the structural fraction of bainite is preferably made 5% or less.

[0107] If the ferrite grains of the case-hardened steel of the present invention are excessively fine, coarse grains easily form. This is because at the time of carburized quenching, the austenite grains become excessively fine. In particular, if the grain size number of the ferrite exceeds 11 as defined by JIS G 0551, coarse grains easily are formed. On the other hand, if the grain size number of ferrite of the case-hardened steel becomes less than 8 as defined by JIS G 0551, the ductility falls and the cold workability is impaired in some cases. Therefore, the grain size number of ferrite of the case-hardened steel is preferably 8 to 11 in range as defined by JIS G 0551.

[0108] Next, the method of production of case-hardened steel of the present invention will be explained.

[0109] Steel is produced by a converter, electric furnace, or other usual method, adjusted in ingredients, and passed through a casting process and, if necessary, a blooming process, to obtain a steel material. The steel material is hot worked, that is, hot rolled or hot forged, to produce steel rails or steel bars.

[0110] The sulfides of the steel material often precipitate in the molten steel or at the time of solidification. The size of the sulfides is greatly influenced by the cooling rate at the time of solidification. Therefore, to prevent the coarsening of the sulfides, it is important to control the cooling rate at the time of solidification.

[0111] The cooling rate at the time of solidification is defined as the cooling rate at the part of 1/2 of the distance from the surface to the centerline in the thickness direction on the centerline of the cast bloom width W in the cross-section of the cast bloom shown in FIG. 2 (position from the surface of T/4 from the surface with respect to the cast bloom thickness T). To suppress coarsening of the sulfides, the cooling rate at the time of solidification is preferably made 3°C/min or more. Preferably it is made 5°C/min or more, more preferably 10°C/min or more. Note that, the cooling rate at the time of solidification can be confirmed by the secondary dendrite arm spacing.

[0112] The cast bloom is reheated as it is and hot worked to produce case-hardened steel or the material obtained by a blooming process is reheated and hot worked to produce case-hardened steel. In general, a cast bloom is bloomed to form a billet, cooled to room temperature, then reheated to produce case-hardened steel. Furthermore, in the production of gears or other parts, hot forging is sometimes applied. At that time, in blooming, it is preferable to hold the steel at a 1150°C or more high temperature for 10 minutes or more and cause the Ti and Nb precipitates to solute.

[0113] To produce case-hardened steel, the steel material is heated. If the heating temperature is less than 1150°C, it is not possible to make the Ti precipitates, Nb precipitates, and AlN solute in the steel, and coarse Ti precipitates, Nb precipitates, and AlN will remain.

[0114] To cause the fine Ti precipitates or Nb precipitates to disperse in the case-hardened steel after hot working and suppress the formation of coarse grains at the time of carburized quenching, it is necessary to make the heating temperature 1150°C or more. The preferable lower limit of the heating temperature is 1180°C or more.

[0115] The upper limit of the heating temperature is not prescribed, but if considering the load of the heating furnace, 1300°C or less is preferable. To make the steel material uniform in temperature and cause the precipitates to solute, a holding time of 10 minutes or more is preferable. The holding time is preferably 60 minutes or less from the viewpoint of productivity.

[0116] If the finishing temperature of the hot working is less than 840°C, the ferrite crystal grains become fine and coarse grains easily form at the time of carburized quenching. On the other hand, if the finishing temperature exceeds 1000°C, hardening occurs and the cold workability deteriorates. Therefore, the finishing temperature of hot working is made 840 to 1000°C. Note that, the preferable range of the finishing temperature is 900 to 970°C, and the more preferable range is 920 to 950°C.

[0117] The cooling conditions after the hot working are important for causing the Ti precipitates and Nb precipitates to finely disperse. The temperature range at which precipitation of Ti precipitates and Nb precipitates is promoted is 500 to 800°C. Therefore, the cooling is performed slowly by 1°C/s or less from a 800°C to 500°C temperature range to promote the formation of Ti precipitates and Nb precipitates.

[0118] If the cooling rate exceeds 1°C/s, the time of passage through the region of the precipitation temperature of Ti precipitates and Nb precipitates becomes shorter and the formation of fine precipitates becomes insufficient. Further, if the cooling rate becomes faster, the structural fraction of bainite becomes larger. Further, if the cooling rate is large, the case-hardened steel hardens and the cold workability deteriorates, so the cooling rate is preferably 0.7°C/s or less.

[0119] Note that, as the method for reducing the cooling rate, the method of setting a heat retaining cover or heat retaining cover with a heat source after the rolling line and thereby slowing the cooling may be mentioned.

[0120] The case-hardened steel of the present invention can be applied to parts produced by a cold forging process or parts produced by hot forging. The hot forging process, for example, may comprise hot forging of steel bar, normalization or other heat treatment if necessary, cutting, carburized quenching, and grinding or polishing if necessary.

[0121] By using the case-hardened steel of the present invention, hot forging it at for example a 1150°C or more heating temperature, then, as necessary, treating it by normalization, it is possible to suppress the formation of coarse grains even if applying high temperature carburization in a 950 to 1090°C temperature region. For example, in the case of bearing parts or rolling parts, even if treating them by high temperature carburization, superior rolling contact fatigue characteristics can be obtained.

[0122] The carburized quenching is not particularly limited, but when aiming at a high rolling fatigue life in bearing parts and rolling parts, it is preferable to set the carbon potential at 0.9 to 1.3%. Further, carburization, then nitridation in the process of diffusion treatment, that is, carbonitridation treatment, is also effective. Conditions whereby the nitrogen concentration of the surface becomes 0.2 to 0.6% in range are suitable. By selecting these conditions, fine Ti(C,N) precipitates in large amounts at the carburized layer and the rolling life is improved.

Example 1

[0123] Steels having the compositions of ingredients shown in Tables 1 to 3 were produced and cast at solidification cooling rates of 10 to 11°C/min. The blank fields in the ingredients of Tables 1 to 3 mean the elements are deliberately not added, while the underlines indicate the figures are outside the ranges of the present invention.

[0124] The solidification cooling rate was adjusted in advance based on data analyzing the relationship between the cooling conditions and solidification cooling rate when casting various sizes of cast blooms. The solidification cooling rate of some of the cast blooms was confirmed by secondary dendrite arm spacing to be 10 to 11°C/min in range. Some of the cast blooms were bloomed in accordance with need.

Table 1

No.	Chemical ingredients (mass%)																		Remarks
	C	Si	Mn	P	S	Cr	Ti	Al	N	O	Zr	Mg	Ca	Nb	Mo	Ni	V	B	
1	0.21	0.19	1.30	0.018	0.011	1.06	0.13	0.026	0.0030	0.0011	0.0024								Inv. ex.
2	0.20	0.20	0.38	0.022	0.014	1.10	0.14	0.024	0.0047	0.0014		0.0005							
3	0.21	0.19	0.98	0.014	0.015	1.20	0.06	0.035	0.0033	0.0014			0.0025						
4	0.19	0.18	0.84	0.014	0.014	1.28	0.08	0.027	0.0045	0.0012	0.0007		0.0006						
5	0.19	0.21	0.88	0.005	0.016	1.22	0.08	0.038	0.0026	0.0015	0.0013	0.0020							
6	0.20	0.19	0.58	0.014	0.013	1.13	0.06	0.018	0.0029	0.0014		0.0008	0.0014						
7	0.18	0.24	0.70	0.015	0.010	1.22	0.07	0.038	0.0029	0.0012	0.0025	0.0018	0.0013						
8	0.20	0.19	0.41	0.021	0.030	1.23	0.10	0.026	0.0045	0.0014									
9	0.21	0.21	1.23	0.011	0.026	1.10	0.12	0.037	0.0035	0.0015									
10	0.19	0.21	1.04	0.017	0.031	1.23	0.11	0.038	0.0028	0.0014	0.0005								
11	0.19	0.25	1.63	0.018	0.029	1.05	0.07	0.020	0.0031	0.0012		0.0015							
12	0.22	0.21	0.81	0.016	0.028	1.22	0.11	0.016	0.0032	0.0011			0.0012						
13	0.20	0.19	1.60	0.009	0.026	1.15	0.14	0.028	0.0026	0.0012	0.0016	0.0015	0.0014						
14	0.19	0.19	0.99	0.018	0.029	1.15	0.15	0.034	0.0027	0.0010	0.0018	0.0011							
15	0.32	0.22	0.38	0.018	0.048	1.22	0.06	0.030	0.0032	0.0010		0.0015	0.0013						
16	0.21	0.25	0.32	0.024	0.026	1.16	0.10	0.034	0.0026	0.0012	0.0018	0.0003	0.0019						
17	0.22	0.18	1.77	0.009	0.015	1.21	0.12	0.022	0.0028	0.0011				0.024					
18	0.21	0.20	0.54	0.025	0.013	1.21	0.12	0.014	0.0034	0.0014				0.021					
19	0.19	0.23	0.86	0.005	0.012	1.22	0.09	0.027	0.0035	0.0012	0.0004			0.012					
20	0.21	0.22	1.31	0.023	0.016	1.28	0.11	0.023	0.0034	0.0011		0.0012		0.019					
21	0.21	0.25	0.57	0.016	0.013	1.13	0.14	0.037	0.0047	0.0015			0.0006	0.013					
22	0.19	0.19	1.19	0.011	0.011	1.22	0.08	0.021	0.0041	0.0010	0.0008		0.0004	0.013					
23	0.22	0.19	0.57	0.013	0.013	1.13	0.05	0.019	0.0025	0.0014	0.0030	0.0015		0.016					

(continued)

No.	Chemical ingredients (mass%)																		Remarks
	C	Si	Mn	P	S	Cr	Ti	Al	N	O	Zr	Mg	Ca	Nb	Mo	Ni	V	B	
24	0.18	0.24	0.74	0.016	0.011	1.16	0.12	0.017	0.0032	0.0011		0.0014	0.0015	0.025					
25	0.21	0.23	1.15	0.019	0.015	1.18	0.05	0.018	0.0032	0.0014	0.0027	0.0017	0.0009	0.014			0.13		
26	0.22	0.21	0.48	0.013	0.013	1.27	0.07	0.025	0.0031	0.0014	0.0017	0.0007	0.0004	0.014		0.30			
27	0.20	0.20	0.45	0.015	0.010	1.15	0.09	0.037	0.0036	0.0010	0.0010	0.0005	0.0011	0.020					
28	0.20	0.22	1.11	0.022	0.017	1.12	0.13	0.024	0.0048	0.0015	0.0006	0.0016	0.0013	0.012				0.0015	
29	0.22	0.20	1.19	0.016	0.025	1.26	0.09	0.034	0.0029	0.0013				0.014					
30	0.21	0.24	1.08	0.008	0.025	1.08	0.15	0.036	0.0030	0.0011				0.010					
31	0.21	0.25	1.16	0.011	0.031	1.28	0.05	0.039	0.0028	0.0010	0.0022			0.022					
32	0.19	0.23	1.73	0.009	0.040	1.23	0.06	0.016	0.0041	0.0015	0.0014			0.014					
33	0.22	0.25	0.74	0.007	0.025	1.18	0.10	0.008	0.0026	0.0010		0.0011		0.016					
34	0.21	1.22	1.22	0.009	0.030	1.13	0.15	0.009	0.0038	0.0015			0.0008	0.023					
35	0.18	0.22	1.35	0.011	0.032	1.25	0.14	0.013	0.0039	0.0011	0.0020	0.0015	0.0009	0.024					

Table 2

No.	Chemical ingredients (mass%)																		Remarks
	C	Si	Mn	P	S	Cr	Ti	Al	N	O	Zr	Mg	Ca	Nb	Mo	Ni	V	B	
36	0.19	0.19	1.72	0.009	0.029	0.55	0.12	0.039	0.0049	0.0015	0.0015	0.0006		0.019					Inv. ex.
37	0.22	0.18	1.68	0.024	0.028	1.06	0.06	0.023	0.0030	0.0012		0.0018	0.0012	0.020					
38	0.21	0.20	0.32	0.010	0.028	1.08	0.10	0.032	0.0039	0.0013	0.0013	0.0006	0.0013	0.019					
39	0.20	0.21	1.02	0.018	0.030	1.05	0.09	0.010	0.0046	0.0011	0.0019	0.0012	0.0013	0.020			0.21		
40	0.19	0.20	0.33	0.025	0.035	0.62	0.12	0.022	0.0045	0.0015	0.0013	0.0004	0.0013	0.016		0.95			
41	0.19	0.20	1.16	0.013	0.028	1.20	0.09	0.032	0.0049	0.0015	0.0021	0.0010	0.0013	0.022				0.0016	
42	0.19	0.23	1.37	0.012	0.017	1.08	0.13	0.032	0.0035	0.0013	0.0017				0.14				
43	0.21	0.18	1.00	0.016	0.013	1.07	0.11	0.019	0.0044	0.0014		0.0004			0.16				
44	0.20	0.25	1.69	0.020	0.016	1.15	0.05	0.035	0.0031	0.0012			0.0010		0.14				
45	0.21	0.20	0.76	0.019	0.017	1.06	0.08	0.033	0.0031	0.0013	0.0012		0.0017		0.14				
46	0.20	0.22	1.52	0.015	0.015	1.30	0.10	0.018	0.0048	0.0013	0.0017	0.0007			0.12				
47	0.19	0.25	1.34	0.012	0.027	1.21	0.12	0.012	0.0041	0.0011				0.020	0.13				
48	0.22	0.22	0.64	0.014	0.027	1.11	0.13	0.032	0.0050	0.0014				0.011	0.16				
48	0.19	0.21	0.45	0.010	0.027	1.28	0.13	0.019	0.0026	0.0010	0.0010			0.022	0.16				
49	0.21	0.21	0.56	0.021	0.044	1.62	0.15	0.039	0.0033	0.0010	0.0020			0.019	0.13				
50	0.20	0.18	1.02	0.023	0.054	1.15	0.11	0.019	0.0033	0.0013		0.0003		0.011	0.15				
51	0.22	0.23	0.75	0.022	0.026	1.25	0.06	0.019	0.0047	0.0010			0.0005	0.014	0.16				
52	0.21	0.18	0.38	0.017	0.028	0.72	0.09	0.028	0.0031	0.0012	0.0018	0.0008	0.0012	0.013	0.92				
53	0.21	0.20	0.82	0.018	0.029	1.12	0.09	0.035	0.0040	0.0012	0.0025	0.0004		0.019	0.12				
54	0.21	0.23	0.56	0.011	0.031	1.08	0.09	0.013	0.0049	0.0013		0.0010	0.0017	0.014	0.15				

Table 3

No.	Chemical ingredients (mass%)																		Remarks
	C	Si	Mn	P	S	Cr	Ti	Al	N	O	Zr	Mg	Ca	Nb	Mo	Ni	V	B	
55	0.19	0.24	1.72	0.013	0.012	1.10		0.035	0.0126	0.0014									Comp. ex.
56	0.18	0.23	0.98	0.013	0.013	1.14	0.08	0.018	0.0031	0.0011									
57	0.20	0.19	1.15	0.024	0.013	1.26	0.12	0.034	0.0036	0.0012									
58	0.19	0.22	1.06	0.025	0.012	1.09	0.14	0.034	0.0036	0.0013									
59	0.19	0.21	1.57	0.017	0.030	1.10		0.017	0.0043	0.0011									
60	0.19	0.25	0.79	0.005	0.030	1.14		0.029	0.0030	0.0012									
61	0.20	0.24	1.72	0.008	0.012	1.28	0.14	0.034	0.0040	0.0012	0.0010		0.0005						
62	0.19	0.19	0.84	0.007	0.027	1.27	0.15	0.015	0.0048	0.0015	0.0014		0.0014						
63	0.20	0.19	0.31	0.009	0.014	1.17	0.12	0.022	0.0045	0.0013	0.0009								
64	0.20	0.22	0.75	0.017	0.030	1.07	0.13	0.016	0.0031	0.0010	0.0025								
65	0.20	0.24	0.71	0.023	0.030	1.13	0.14	0.020	0.0032	0.0013		0.0010							
66	0.19	0.20	1.52	0.022	0.011	1.25	0.13	0.037	0.0043	0.0015	0.0025		0.0017	0.017					
67	0.20	0.22	1.52	0.009	0.026	1.10	0.09	0.015	0.0049	0.0014	0.0024		0.0016	0.023					
68	0.18	0.19	0.42	0.025	0.015	1.23	0.13	0.029	0.0043	0.0011	0.0004			0.014					
69	0.20	0.21	1.78	0.013	0.027	1.26	0.13	0.011	0.0043	0.0012	0.0018			0.012					
70	0.20	0.20	1.11	0.019	0.031	1.24	0.10	0.033	0.0031	0.0014		0.0016		0.023					
71	0.20	0.24	1.02	0.022	0.017	1.09	0.12	0.013	0.0124	0.0012	0.0011								
72	0.21	0.22	0.87	0.018	0.017	1.25	0.11	0.014	0.0145	0.0012		0.0006							
73	0.19	0.21	1.02	0.019	0.013	1.26	0.10	0.018	0.0086	0.0011			0.0011						
74	0.18	0.20	0.34	0.015	0.026	1.15	0.12	0.031	0.0098	0.0015	0.0024								
75	0.19	0.22	0.33	0.008	0.030	1.16	0.06	0.030	0.0146	0.0012		0.0015							
76	0.20	0.19	1.74	0.009	0.028	1.25	0.12	0.006	0.0113	0.0010			0.0016						

(continued)

No.	Chemical ingredients (mass%)																		Remarks
	C	Si	Mn	P	S	Cr	Ti	Al	N	O	Zr	Mg	Ca	Nb	Mo	Ni	V	B	
77	0.20	0.24	1.57	0.009	0.011	1.25		0.020	0.0031	0.0013		0.0009							
78	0.20	0.24	0.32	0.020	0.013	1.07	0.30	0.040	0.0049	0.0010	0.0011								
79	0.20	0.22	1.60	0.015	0.027	1.05	0.14	0.034	0.0026	0.0011	0.0011			0.120					
80	0.20	0.24	0.82	0.021	0.032	1.05	0.14	0.012	0.0045	0.0031	0.0020								
81	0.18	0.24	1.45	0.006	0.032	1.16	0.10	0.008	0.0049	0.0015	0.0005								
82	0.21	0.20	0.77	0.023	0.031	1.10	0.05	0.007	0.0040	0.0015	0.0004								
83	0.21	0.20	0.77	0.023	0.031	1.10	0.05	0.007	0.0040	0.0015	0.0004								
84	0.21	0.22	1.31	0.007	0.016	1.25	0.15	0.016	0.0031	0.0012					0.14				
85	0.22	0.24	1.79	0.018	0.010	1.10	0.11	0.020	0.0038	0.0011					0.15				
86	0.20	0.22	0.79	0.018	0.014	1.23	0.10	0.009	0.0031	0.0014				0.020	0.13				
87	0.22	0.23	0.78	0.009	0.010	1.07		0.035	0.0126	0.0012					0.15				
88	0.20	0.19	0.94	0.017	0.026	1.25	0.06	0.034	0.0033	0.0014					0.14				
89	0.20	0.22	0.89	0.025	0.031	1.28	0.07	0.032	0.0030	0.0013				0.020	0.13				

[0125] Next, the steels were hot worked to produce steel bars of diameters of 24 to 30 mm. The steels were observed under a microscope, the bainite fractions were measured, and the ferrite grain size numbers were determined based on the provisions of JIS G 0551. The Vickers hardnesses were measured based on JIS Z 2244 and used as indicators of cold workability and machineability. The amounts of precipitation of AlN were found by chemical analysis.

[0126] Further, the statistics of extremes method was used to predict the maximum sizes of the Ti precipitates. Table 4 to 6 show the hot working heating temperatures, finishing temperatures, cooling rates, bainite fractions, ferrite grain size numbers, AlN precipitation, Ti precipitate maximum sizes, and Vickers hardnesses. Note that, the cooling rate is the cooling rate in the 500 to 800°C range. This was found from the time required for cooling from 800°C to 500°C.

[0127] The maximum sizes of the Ti precipitates were found as follows. An optical microscope was used to observe the metal structures and contrast was used to differentiate the precipitates. Note that, the contrast of the precipitates was confirmed using a scanning electron microscope and energy dispersive X-ray spectrometer.

[0128] In the longitudinal direction cross-section of each test piece, 16 fields of regions of standard inspection areas of 100 mm^2 (10 mm×10 mm region) were prepared in advance. The largest Ti precipitates in each 100 square mm standard inspection area was detected and photographed by an optical microscope by 1000X.

[0129] This was repeated 16 times for the 16 fields of the standard inspection areas of 100 mm^2**.** In this way, the test was conducted for 16 fields and the size of the largest precipitate in each standard inspection area was measured from the obtained photographs. Note that, in the case of an ellipse, the geometric mean of the long axis and short axis is found and used as the size of the precipitate.

[0130] The 16 sets of data of the obtained maximum precipitate sizes were plotted on an extreme probability paper by the method described in Yukitaka Murakami, "Metal Fatigue - Effects of Small Defects and Nonmetallic Inclusions", Yokendo, pp. 233 to 239 (1993), the largest precipitate distribution line, that is, the primary function of the maximum precipitate size and statistics of extremes standardized variable, was found, the largest precipitate distribution line was extrapolated, and the diameters of the largest precipitates in the prediction area (30000 mm^2) were found.

[0131] Further, to evaluate the cold workability by cold forging, the test piece was annealed, then subjected to an upset test. The grooved test piece shown in FIG. 3 was obtained and measured for the limit compression rate until fracture. The compression rate was changed and 10 test pieces were used to find the probability of fracture. The compression rate when the probability became 50% was made the limit compression rate.

[0132] The higher this limit compression rate, the better the forgeability evaluated. This test method is a method of evaluation close to cold forging, but has also been considered an indicator showing the effects of sulfides on forgeability in hot forging.

[0133] The machineability was evaluated by a test finding the lifetime until a drill broke. Note that, the drilling was performed using a high speed steel straight shank drill having a diameter of 3 mm at a feed of 0.25 mm, a hole depth of 9 mm, and a drill projection of 35 mm using a water soluble cutting fluid.

[0134] The speed of the drill was fixed at 10 to 70 m/min in range and the cumulative hole depth until breakage was measured while drilling. Here, the cumulative hole depth is the product of the depth of one hole and the number of drilled holes.

[0135] The speed of the drill was changed and similar measurements conducted. The maximum value of the speed of the drill where the cumulative hole depth exceeds 1000 mm was found as VL1000. The larger the VL1000, the better the tool life and the more superior the machineability the material is evaluated as.

[0136] Further, the coarse grain characteristic was evaluated by taking a test piece from a steel bar after spheroidal annealing, cold upset forging it by a reduction rate of 50%, then heat treating it simulating carburized quenching (referred to as "carburization simulation"), and measuring the old austenite grain size.

[0137] The carburization simulation comprised heat treatment heating a test piece to 910 to 1010°C, holding it there for 5 hours, then water cooling it. The old austenite grain size was measured in accordance with JIS G 0551.

[0138] The old austenite grain size was measured and the temperature at which coarse grains formed (coarsening temperature) was found. Note that, the old austenite grain size was measured by observation at 400X for about 10 fields. If even one coarse grain of a grain size number of 5 or less was present, it was judged that coarse grains were formed.

[0139] The heating temperature of the carburized quenching treatment is usually 930 to 950°C, so a test piece with a coarsening temperature of 950°C or less was judged to be inferior in crystal grain coarsening characteristic.

[0140] Next, the reduction rate was made 50%, the steel was cold forged, and a cylindrical rolling contact fatigue test piece of a diameter of 12.2 mm was obtained and treated by carburized quenching. The carburized quenching was performed by heating the steel in an atmosphere of a carbon potential of 0.8% to 950°C, holding it there fore 5 hours, and quenching it in oil of a temperature of 130°C. Furthermore, the steel was held at 180°C for 2 hours and tempered. These carburized quenched materials were investigated for the γ granularity (carburized layer austenite grain size number) of the carburized layers based on JIS G 0551.

[0141] Furthermore, a point contact type rolling contact fatigue test rig (Hertz maximum contact stress 5884 MPa) was used to evaluate the rolling contact fatigue characteristic. As a measure of the fatigue life, the L_{10} life, defined as "the number of cycles of stress to fatigue fracture at a probability of failure of 10% obtained by plotting the test results on a Weibull probability paper", was used. However, materials with frequent breakage at a reduction rate of 50% were not subjected to subsequent fatigue tests.

[0142] The results of these investigations are summarized in Tables 4 to 6. The rolling fatigue life shows the relative value of the L_{10} life of each material indexed to the L_{10} life of No. 55 (comparative example) as "1".

Table 4

No.	Hot working			Bainite fraction (%)	Ferrite AlN grain size number	precipitation (%)	Ti precipitate max. size μm	Sulfide density (/mm^2)	Vickers hardness (HV)	Coarsening temp. (°C)	Carburized Limit layer austenite grain size number	comp. rate (%)	Machine-ability VL1000 (m/min)	Fatigue life (rel. value)	Remarks
	Heating temp. (°C)	Finishing temp. (°C)	Cooling rate (°C/s)												
1	1270	930	0.50	0	9.8	0.003	21	16.0	180	>1050	9.8	58	48	3.5	Inv. ex.
2	1260	950	0.53	0	9.0	0.004	23	29.5	183	>1050	8.8	56	46	3.7	
3	1190	940	0.53	0	9.4	0.004	26	26.6	187	>1050	9.9	56	45	3.0	
4	1210	940	0.53	0	9.4	0.004	25	13.2	184	>1050	8.9	55	49	3.4	
5	1260	940	0.55	0	9.8	0.003	23	11.6	185	>1050	8.7	56	46	3.5	
6	1220	930	0.53	0	9.2	0.004	27	27.9	194	>1050	8.6	57	46	2.8	
7	1190	940	0.48	0	10.5	0.003	29	25.6	188	>1050	8.0	55	49	2.6	
8	1180	940	0.57	0	9.2	0.004	26	47.5	172	>1050	9.7	55	55	2.5	
9	1220	930	0.55	0	10.2	0.003	31	52.0	183	>1050	8.9	54	51	3.8	
10	1250	940	0.49	0	9.5	0.004	27	36.2	188	>1050	8.5	53	54	3.4	
11	1270	930	0.48	0	9.8	0.003	24	53.3	176	>1050	10.0	56	50	3.0	
12	1230	950	0.56	0	9.8	0.003	30	37.3	178	>1050	8.4	54	53	3.2	
13	1200	930	0.47	0	9.4	0.003	24	51.4	187	>1050	8.4	56	50	2.6	
14	1270	930	0.46	0	10.2	0.002	32	39.2	183	>1050	8.5	54	51	3.2	
15	1190	940	0.52	5	9.0	0.004	25	30.2	192	>1050	9.8	52	51	3.4	
16	1240	930	0.48	0	9.3	0.003	24	41.8	177	>1050	9.9	56	52	2.8	
17	1220	940	0.47	0	10.1	0.002	25	27.9	173	>1050	9.1	59	46	2.7	
18	1250	950	0.46	0	10.4	0.003	31	23.2	178	>1050	9.6	56	49	3.7	
19	1190	940	0.57	0	10.0	0.002	23	19.7	174	>1050	9.6	56	46	3.1	
20	1270	940	0.56	0	10.0	0.003	29	10.9	180	>1050	9.4	58	48	2.5	
21	1230	930	0.52	0	9.2	0.002	26	21.7	194	>1050	10.0	60	48	3.2	

(continued)

No.	Hot working			Bainite fraction (%)	Ferrite AlN grain size number	precipitation (%)	Ti precipitate max. size μm	Sulfide density (/mm^2)	Vickers hardness (HV)	Coarsening temp. (°C)	Carburized Limit layer austenite grain size number	comp. rate (%)	Machine-ability VL1000 (m/min)	Fatigue life (rel. value)	Remarks
	Heating temp. (°C)	Finishing temp. (°C)	Cooling rate (°C/s)												
22	1190	950	0.45	0	9.5	0.004	27	22.5	179	>1050	8.4	58	50	2.5	
23	1220	930	0.57	0	9.7	0.004	27	25.6	181	>1050	9.0	59	48	3.2	
24	1230	940	0.50	0	10.5	0.004	27	16.4	192	>1050	8.9	56	46	2.9	
25	1250	930	0.56	0	10.2	0.004	29	28.3	174	>1050	9.3	59	50	3.3	
26	1190	930	0.53	0	9.0	0.003	21	20.2	193	>1050	8.4	55	49	2.8	
27	1250	940	0.52	0	10.3	0.004	27	25.1	175	>1050	8.7	59	48	2.6	
28	1230	940	0.51	0	10.4	0.003	30	10.9	184	>1050	8.4	55	46	2.6	
29	1200	940	0.52	0	9.6	0.002	27	59.5	180	>1050	9.1	54	50	3.3	
30	1200	940	0.46	0	9.6	0.003	29	46.8	177	>1050	9.0	56	53	3.7	
31	1230	940	0.56	0	10.4,	0.003	22	57.1	175	>1050	9.5	49	58	3.6	
32	1270	930	0.48	0	9.8	0.003	25	60.6	189	>1050	8.6	54	53	3.5	
33	1200	950	0.56	0	9.3	0.004	29	53.3	189	>1050	8.6	55	51	3.3	
34	1200	940	0.45	0	9.8	0.003	28	50.0	191	>1050	9.7	54	50	3.7	
35	1280	940	0.49	0	9.0	0.002	23	38.1	176	>1050	9.5	54	53	3.2	

Table 5

No.	Hot working			Bainite fraction (%)	Ferrite grain size number	AlN precipitation (%)	Ti precipitate max. size μm	Sulfide density (/mm^2)	Vickers hardness (HV)	Coarsening temp. (°C)	Carburized layer austenite grain size number	Limit compression rate (%)	Machingability VL1000 (m/min)	Fatigue life (rel. value)	Remarks
	Heating No. temp. (°C)	Finishing temp. (°C)	Cooling rate (°C/s)												
36	1210	940	0.52	0	8.8	0.003	26	53.9	173	>1050	8.9	55	52	3.4	Inv. ex.
37	1270	950	0.48	0	10.4	0.003	27	41.6	178	>1050	8.9	54	53	3.0	
38	1190	950	0.46	0	9.7	0.003	23	45.0	173	>1050	8.6	53	53	3.2	
39	1260	940	0.56	0	8.9	0.003	27	36.9	194	>1050	9.2	55	52	3.3	
40	1240	950	0.47	0	9.5	0.003	24	59.7	187	>1050	8.5	56	52	3.7	
41	1200	940	0.46	0	9.6	0.004	30	40.3	174	>1050	9.8	54	52	2.6	
42	1200	930	0.49	4	9.5	0.003	20	15.2	201	>1050	8.8	58	43	3.1	
43	1280	950	0.50	4	10.2	0.003	29	15.2	193	>1050	9.5	54	42	3.9	
44	1260	940	0.45	4	9.9	0.004	27	29.3	185	>1050	9.1	55	41	3.1	
45	1260	940	0.57	5	9.5	0.003	22	15.1	188	>1050	9.4	57	41	3.3	
46	1200	950	0.50	7	9.5	0.002	28	28.9	188	>1050	8.2	56	43	3.3	
47	1240	950	0.47	6	9.1	0.002	23	47.9	202	>1050	9.7	53	47	3.0	
48	1250	950	0.56	5	9.7	0.004	26	32.8	184	>1050	8.4	51	47	3.3	
48	1280	940	0.49	5	9.1	0.002	32	44.6	196	>1050	8.4	52	48	3.1	
49	1190	930	0.02	16	10.1	0.003	28	70.0	189	>1050	8.1	51	52	3.1	
50	1280	950	0.51	3	10.1	0.003	26	64.7	198	>1050	8.7	52	50	3.5	
51	1220	940	0.55	5	10.0	0.003	22	32.0	197	>1050	9.2	52	47	3.5	
52	1200	940	0.47	14	9.9	0.004	24	48.8	205	>1050	8.2	53	45	3.8	
53	1280	940	0.55	3	9.9	0.004	21	57.5	186	>1050	8.1	51	48	3.2	
54	1200	950	0.50	4	9.0	0.002	27	40.5	194	>1050	9.0	53	49	3.5	

Table 6

No.	Hot working			Bainite fraction (%)	Ferrite grain size number	AlN precipiation (%)	Ti precipitate max. size μm	Sulfide density (/mm^2)	Vickers hardness (HV)	Coarsening temp. (°C)	Carburized layer austenite grain size number	Limit comp. rate (%)	Machine-ability VL1000 (m/min)	Fatigue life (rel. value)	Remarks
	Heating temp. (°C)	Finishing temp. (°C)	Cooling rate (°C/s)												
55	1210	900	0.47	0	10.3	0.003	-	70.5	165	950	3.7	58	40	1.0	Comp. ex.
56	1200	930	0.51	0	10.4	0.002	22	46.9	191	>1050	8.1	50	30	2.6	
57	1220	930	0.45	0	9.8	0.003	27	45.1	195	>1050	8.2	51	30	2.6	
58	1210	930	0.53	0	9.1	0.004	28	58.9	176	>1050	8.5	50	33	2.8	
59	1190	950	0.56	0	9.1	0.003	-	126.6	160	910	3.5	45	47		
60	1220	950	0.52	0	8.9	0.004	-	149.5	162	910	3.7	43	49		
61	1000	930	0.52	0	10.3	0.003	52	22.8	190	910	4.9	59	46	3.2	
62	980	940	0.46	0	9.7	0.003	54	57.7	181	920	3.4	56	53	3.3	
63	1000	940	0.56	0	9.2	0.003	52	12.9	183	910	3.0	59	48	3.1	
64	980	940	0.57	0	10.4	0.003	55	48.4	193	910	4.5	56	55	2.9	
65	980	940	0.49	0	10.4	0.003	49	48.1	183	920	4.1	56	52	2.6	
66	1000	950	0.46	0	9.3	0.003	52	21.2	177	910	4.3	58	47	3.7	
67	980	940	0.50	0	10.5	0.003	53	54.4	181	920	4.5	55	52	3.6	
68	1000	940	0.52	0	9.6	0.004	52	24.2	172	910	3.4	58	50	3.2	
69	980	950	0.51	0	9.2	0.004	56	41.6	180	910	4.9	54	55	3.6	
70	980	950	0.52	0	10.0	0.003	55	35.4	189	920	3.1	54	53	2.7	
71	1210	940	0.54	0	8.9	0.003	61	23.7	188	930	3.7	50	25	2.7	
72	1240	940	0.49	0	9.8	0.003	56	20.7	176	930	3.5	52	26	2.6	
73	1260	940	0.53	0	9.3	0.003	36	17.0	180	>1050	9.8	51	25	2.7	
74	1270	940	0.48	0	9.0	0.002	40	42.7	194	>1050	9.4	45	35		
75	1210	940	0.51	0	9.7	0.003	70	42.8	193	930	3.7	44	34		

(continued)

No.	Hot working			Bainite fraction (%)	Ferrite grain size number	AlN precipiation (%)	Ti precipitate max. size μm	Sulfide density (/mm^2)	Vickers hardness (HV)	Coarsening temp. (°C)	Carburized layer austenite grain size number	Limit comp. rate (%)	Machine-ability VL1000 (m/min)	Fatigue life (rel. value)	Remarks
	Heating temp. (°C)	Finishing temp. (°C)	Cooling rate (°C/s)												
76	1240	930	0.45	0	10.2	0.004	59	51.9	189	920	3.7	46	35		
77	1270	930	0.55	0	10.3	0.003	-	76.5	165	910	3.0	58	50	1.1	
78	1180	950	0.47	0	9.7	0.003	76	25.3	203	910	3.2	30	30		
79	1200	930	0.47	0	9.8	0.004	31	55.5	205	910	3.4	32	30		
80	1200	940	0.50	0	10.1	0.003	24	34.7	179	910	4.0	57	53	0.3	
81	1200	930	1.50	35	9.9	0.002	25	54.6	220	930	3.4	30	30		
82	1200	1030	0.56	0	7.0	0.002	23	40.1	184	910	3.5	53	54	1.2	
83	1200	850	0.56	0	12.0	0.002	23	40.1	184	910	3.5	53	54	1.3	
84	1190	930	0.54	0	8.9	0.003	24	48.2	194	>1050	8.6	47	25		
85	1280	940	0.56	0	10.0	0.003	23	56.9	191	>1050	8.1	46	28		
86	1230	930	0.46	0	10.0	0.004	28	54.5	205	>1050	9.0	45	25		
87	1200	900	0.46	0	10.5	0.003	-	75.4	175	910	3.7	50	35	1.2	
88	1230	940	0.52	0	9.9	0.003	23	132.5	200	>1050	9.1	41	43		
89	1250	940	0.56	0	9.9	0.003	24	116.2	201	>1050	8.5	41	43		

[0143] It is clear that the crystal grain coarsening temperature of the invention examples is 990°C or more, the γ grains of a 950°C carburized material are fine, regular grains, and the rolling contact fatigue characteristic is also superior. Regarding the cold forgeability and machineability as well, it is clear that they are superior compared with the comparative examples of similar amounts of S.

[0144] On the other hand, the comparative example of No. 55 corresponds to SCr420 prescribed by the JIS. It does not contain Ti, Mg, Zr, or Ca, so has a low coarsening temperature and coarse γ grains.

[0145] Further, Nos. 56 to 58 exhibit effects of prevention of coarse grains by Ti, but do not contain Ti, Mg, Zr, or Ca, so have inferior machineability and furthermore insufficient cold forgeability.

[0146] Nos. 59 and 60 are examples where the S is increased to try to improve the machineability, but do not contain Ti, Mg, Zr, or Ca, so have elongated sulfides and inferior cold forgeabilities.

[0147] Nos. 84 to 89 are examples where Mo and Nb are added and the quenchability is improved, while No. 87 corresponds to SCM420 prescribed by the JIS. However, No. 87 does not contain Ti, Mg, Zr, or Ca, so has a low coarsening temperature and coarse γ grains. Further, Nos. 84 to 86, 88, and 89 exhibit effects of prevention of coarse grains by Ti, but do not contain Ti, Mg, Zr, or Ca, so have inferior machineability and, furthermore, insufficient cold forgeability.

[0148] Nos. 71 to 76 have large contents of N, coarse Ti precipitates, and remarkable formation of coarse grains. Further, Nos. 71 to 73 have reduced rolling contact fatigue characteristics of carburized parts, while Nos. 74 to 76 are examples inferior in cold forgeability and not subjected to rolling contact fatigue tests.

[0149] No. 80 has a large O content, formation of coarse grains, and no good rolling contact fatigue characteristic as well.

[0150] No. 77 has a small Ti content and a small pinning effect of Ti, so has a reduced coarsening temperature.

[0151] No. 78 has a large Ti content, coarse Ti precipitates, reduced coarsening temperature, and degraded cold workability due to TiC precipitation hardening. Further, No. 78 has insufficient solubilization of Ti precipitates and reduced rolling contact fatigue characteristic of carburized parts.

[0152] No. 79 has a large Nb content, degraded cold workability due to precipitation hardening, and inferior prevention of coarse grains.

[0153] Nos. 61 to 70 have low heating temperatures, insufficient solid solutions of Ti precipitates and Nb precipitates, and inferior effects of prevention of coarse grains.

[0154] No. 81 has a fast cooling rate after hot rolling, increased bainite structural fraction after hot working, and formation of coarse grains.

[0155] No. 82 has a high finishing temperature in hot working, coarse ferrite crystal grain size, and degraded prevention of coarse grains.

[0156] No. 83 has a low finishing temperature in hot working, a fine ferrite crystal grain size, and inferior prevention of coarse grains.

Claims

1. Case-hardened steel superior in cold workability, machinability, and fatigue characteristics after carburized quenching **characterized by** containing, by mass%,
   C: 0.1 to 0.5%,
   Si: 0.01 to 1.5%,
   Mn: 0.3 to 1.8%,
   S: 0.001 to 0.15%,
   Cr: 0.4 to 2.0%, and
   Ti: 0.05 to 0.2%,
   limiting
   Al: 0.04% or less,
   N: 0.0050% or less,
   P: 0.025% or less,
   O: 0.0025% or less,
   further having one or more of
   Mg: 0.003% or less,
   Zr: 0.01% or less, and
   Ca: 0.005% or less,
   having a balance of iron and unavoidable impurities,
   limiting an amount of precipitation of AlN to 0.01% or less, and
   having a density d (/mm^2) of sulfides of a equivalent circle diameter of over 20 μm and an aspect ratio of over 3 and a content of S [S] (mass%) satisfying

$$d \leq 1700[S]+20.$$

2. Case-hardened steel superior in cold workability, machinability, and fatigue characteristics after carburized quenching as set forth in claim 1, **characterized by** further containing, by mass%,
Nb: less than 0.04%.

3. Case-hardened steel superior in cold workability, machinability, and fatigue characteristics after carburized quenching as set forth in claim 1 or 2, **characterized by** further containing, by mass%, one or more of
Mo: 1.5% or less,
Ni: 3.5% or less,
V: 0.5% or less, and
B: 0.005% or less.

4. Case-hardened steel superior in cold workability, machinability, and fatigue characteristics after carburized quenching as set forth in any one of claims 1 to 3, **characterized by** limiting a structural fraction of bainite to 30% or less.

5. Case-hardened steel superior in cold workability, machinability, and fatigue characteristics after carburized quenching as set forth in any one of claims 1 to 4, **characterized in that** a grain size number of ferrite is 8 to 11 as defined by JIS G 0551.

6. Case-hardened steel superior in cold workability, machinability, and fatigue characteristics after carburized quenching as set forth in any one of claims 1 to 5, **characterized in that** a maximum size of Ti precipitates is 40 μm or less.

7. A method of production of case-hardened steel superior in cold workability, machinability, and fatigue characteristics after carburized quenching **characterized by** heating a steel material comprised of the ingredients of any of claims 1 to 3 to 1150°C or more, hot working it at a finishing temperature of 840 to 1000°C, and cooling it in a 800 to 500°C temperature range by 1°C/s or less.

Fig.1

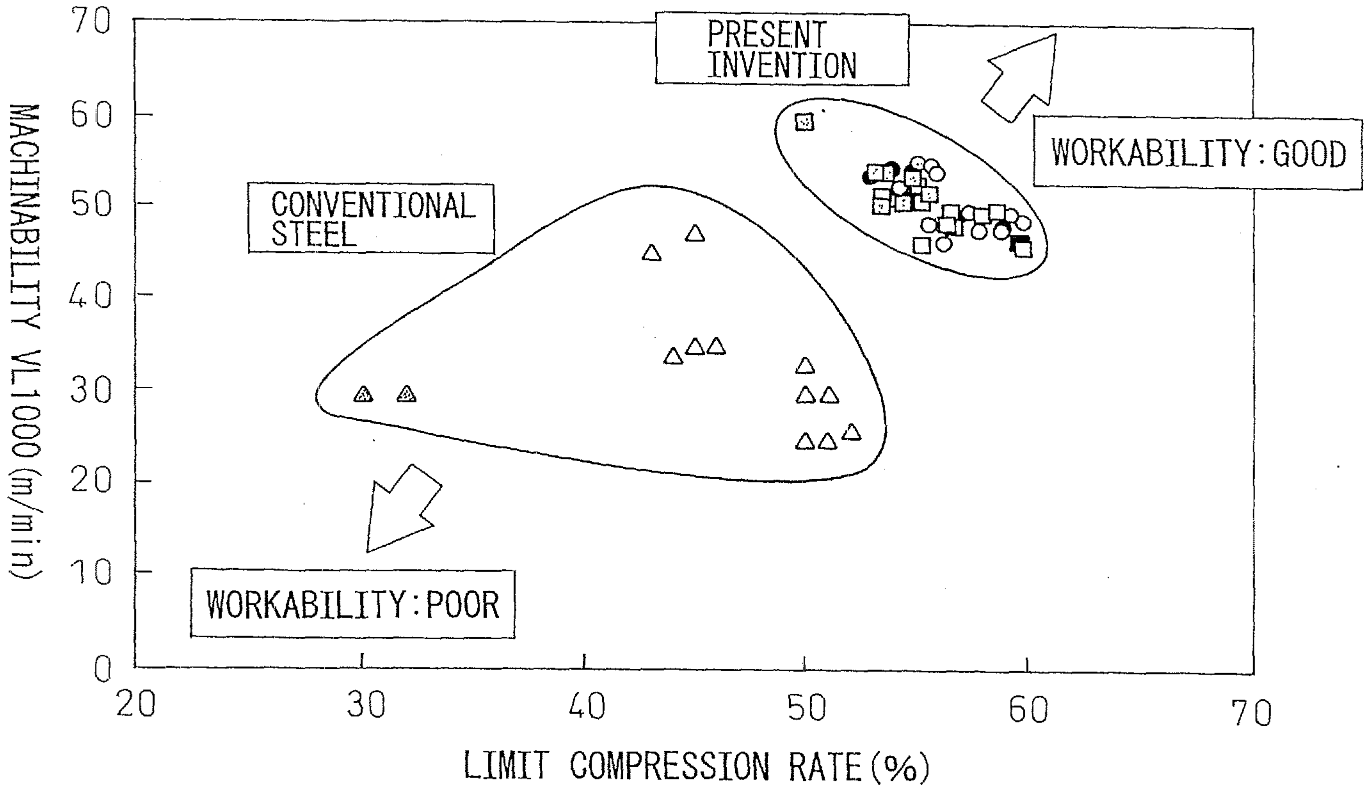

70
60
50
40
30
20
10
0
MACHINABILITY VL1000(m/min)
20
30
40
50
60
70
LIMIT COMPRESSION RATE(%)
PRESENT INVENTION
WORKABILITY:GOOD
CONVENTIONAL STEEL
WORKABILITY:POOR

Fig. 2

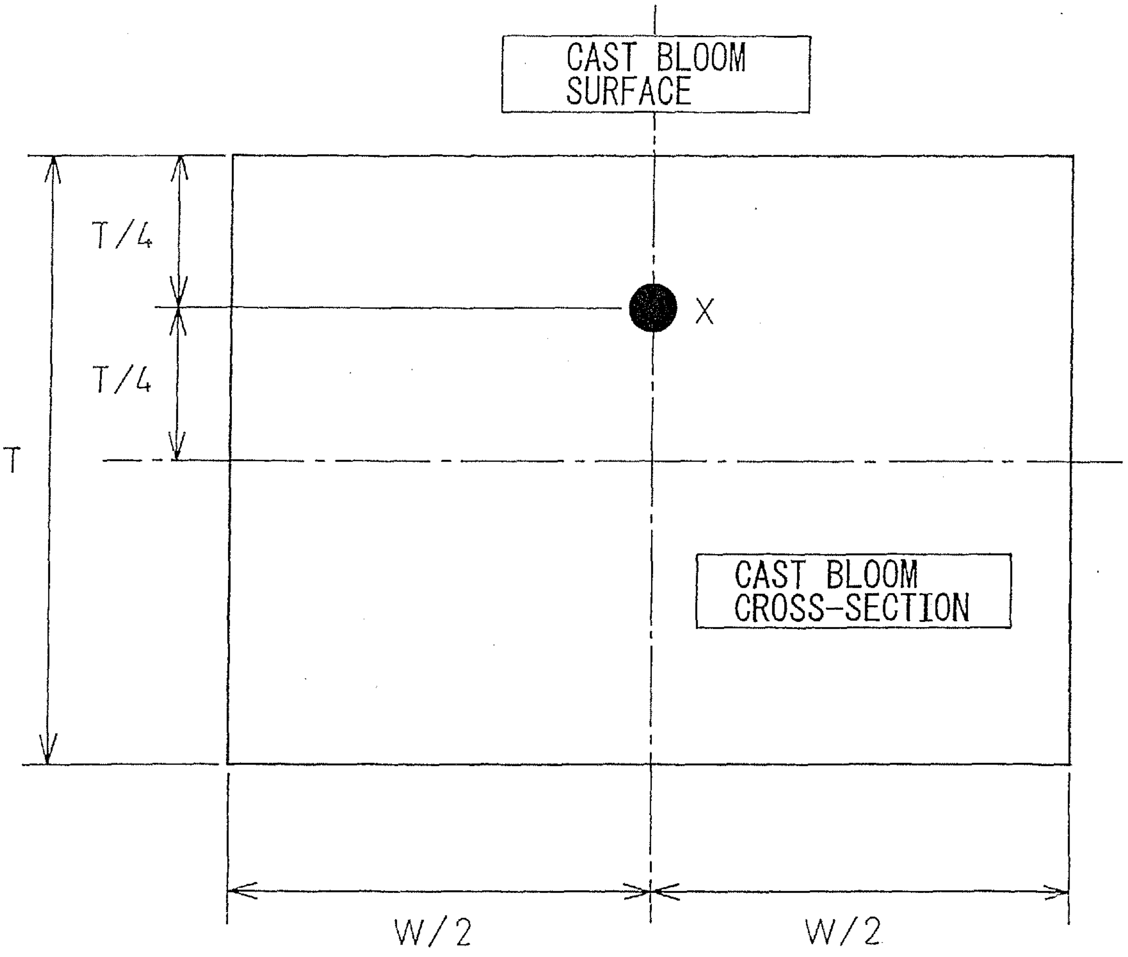
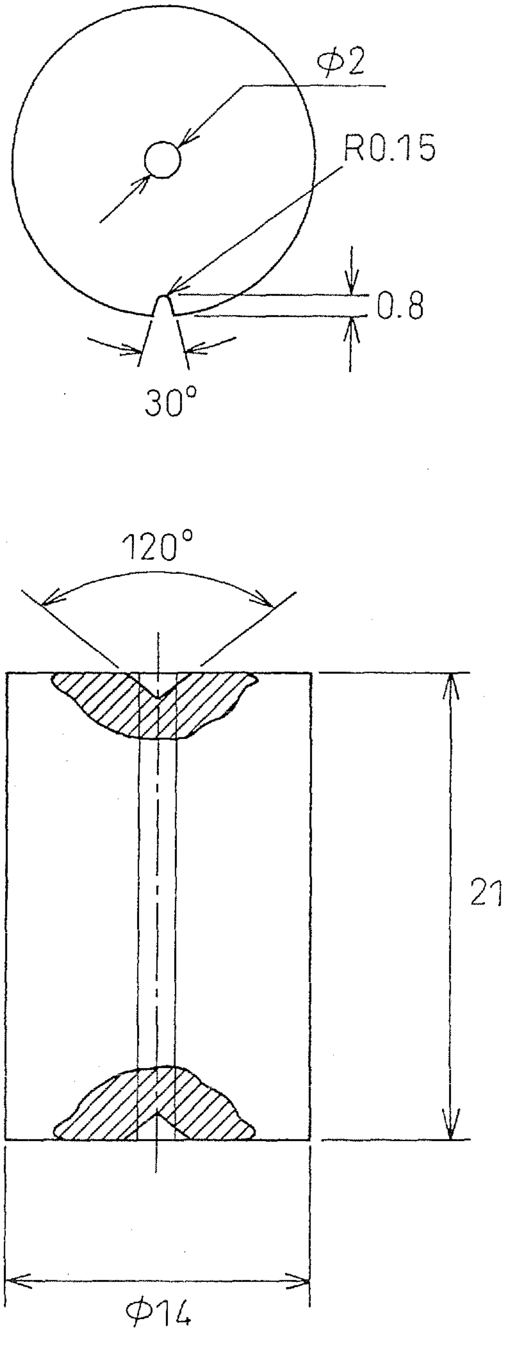


Fig.3



INTERNATIONAL SEARCH REPORT

International application No.
PCT/JP2009/068083

A. CLASSIFICATION OF SUBJECT MATTER
C22C38/00(2006.01)i, *C21D8/06*(2006.01)i, *C22C38/60*(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)
C22C38/00, C21D8/06, C22C38/60, C21D1/06

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	1996-2010
Kokai Jitsuyo Shinan Koho	1971-2010	Toroku Jitsuyo Shinan Koho	1994-2010

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 2005-240175 A (Nippon Steel Corp.), 08 September 2005 (08.09.2005), (Family: none)	1-7
A	JP 2004-183064 A (Nippon Steel Corp.), 02 July 2004 (02.07.2004), (Family: none)	1-7
A	JP 2005-220423 A (Kobe Steel, Ltd.), 18 August 2005 (18.08.2005), (Family: none)	1-7
A	JP 2007-31787 A (Kobe Steel, Ltd.), 08 February 2007 (08.02.2007), (Family: none)	1-7

☒ 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	"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
"E" earlier application or patent but published on or after the international filing date	"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
"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)	"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
"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	"&" document member of the same patent family

Date of the actual completion of the international search 08 January, 2010 (08.01.10)	Date of mailing of the international search report 19 January, 2010 (19.01.10)
Name and mailing address of the ISA/ Japanese Patent Office Facsimile No.	Authorized officer Telephone No.

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INTERNATIONAL SEARCH REPORT

International application No.
PCT/JP2009/068083

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 2008-106308 A (Sanyo Special Steel Co., Ltd.), 08 May 2008 (08.05.2008), (Family: none)	1-7

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- JP 2001303174 A **[0009]**
- JP 2004183064 A **[0009]**
- JP 2004204263 A **[0009]**
- JP 2005240175 A **[0009]**

Non-patent literature cited in the description

- **Yukitaka Murakami.** Metal Fatigue - Effects of Small Defects and Nonmetallic Inclusions. Yokendo, 1993, 233-239 **[0101] [0130]**