(19) Europäisches Patentamt
European Patent Office
Office européen des brevets

(11) EP 3 312 299 A1

(12) EUROPEAN PATENT APPLICATION

published in accordance with Art. 153(4) EPC

(43) Date of publication:
25.04.2018 Bulletin 2018/17

(21) Application number: **16811762.0**

(22) Date of filing: **17.06.2016**

(51) Int Cl.:
C22C 38/00 *(2006.01)*
C21D 9/46 *(2006.01)*
C22C 38/06 *(2006.01)*
C22C 38/60 *(2006.01)*

(86) International application number:
PCT/JP2016/068169

(87) International publication number:
WO 2016/204288 (22.12.2016 Gazette 2016/51)

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

(30) Priority: **17.06.2015 JP 2015122260**

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(54) STEEL SHEET AND MANUFACTURING METHOD

(57) A steel plate improved in formability and wear resistance, having a predetermined chemical composition, having a metal structure of the steel plate having a ratio of number of carbides at ferrite grain boundaries with respect to a number of carbides in ferrite grains of over 1 and having a ferrite grain size of 5 μm to 50 μm, and having a Vickers hardness of steel plate of 100HV to 170HV.

Printed by Jouve, 75001 PARIS (FR)

Description

Technical Field

[0001] The present invention relates to steel plate and a method of production of the same.

Background Art

[0002] Gears, clutches, and other auto parts are produced through stamping, forging, press-forming, and other working processes. In the working processes, to improve and stabilize the product quality and reduce the manufacturing costs, improvement of the workability of the starting material carbon steel plate has been sought. Further, these parts are quenched and tempered, then used at a high strength, so excellent hardenability is demanded.

[0003] To secure the workability of carbon steel plate and secure hardenability, numerous proposals have been made in the past.

[0004] PLT 1 discloses high carbon steel plate excellent in workability, hardenability, and toughness after heat treatment containing, by mass%, C: 0.20 to 0.45%, Mn: 0.40 to 1.50%, P: 0.03% or less, S: 0.02% or less, P+S: 0.010% or more, Cr: 0.01 to 0.80%, Ti: 0.005 to 0.050%, and B: 0.0003 to 0.0050% and having a balance of Fe and unavoidable impurities, further, containing Sn: 0.05% or less and Te: 0.05% or less and containing a total of Sn+Te of 0.005% or more, and comprised of a mixed structure of ferrite and pearlite or a mixed structure of ferrite and cementite.

[0005] PLT 2 discloses a method of production of high hardenability high carbon hot rolled steel plate comprising hot rolling steel containing, by mass%, C: 0.2 to 0.7%, Si: 2% or less, Mn: 2% or less, P: 0.03% or less, S: 0.03% or less, sol. Al: 0.08% or less, and N: 0.01% or less and having a balance of iron and unavoidable impurities until the finishing temperature (Ar3 transformation point-20°C) or more, cooling it by a cooling rate of over 120°C/sec down to a cooling end temperature of 620°C or less, coiling it by a coiling temperature of 600°C or less to control it to a structure having over volume fraction 20% bainite phases, pickling it, and annealing it by an annealing temperature of 640°C to the Ac1 transformation point to obtain spheroidized structures.

Citation List

Patent Literature

[0006]

PLT 1: Japanese Patent No. 4319940
PLT 2; Japanese Patent No. 3879459

Summary of Invention

Technical Problem

[0007] However, the high carbon steel plate described in PLT 1 also uses the high hardness pearlite in the starting material structure and is not necessarily excellent in workability. PLT 2 does not describe a specific form of structure excellent in workability.

[0008] The present invention, in consideration of the current state of the prior art, has as its object the provision of steel plate improved in formability and wear resistance, in particular suitable for obtaining gears, clutches, and other parts by forming thick gauge plate, and a method of production of the same.

Solution to Problem

[0009] To solve the above problem and obtain steel plate suitable as a material for drive system parts etc., it can be understood that in steel plate containing the C necessary for raising the hardenability, it is sufficient to increase the grain size of ferrite, make the carbides (mainly cementite) spheroidal by suitable grain sizes, and reduce the pearlite structures. This is due to the following reasons.

[0010] Ferrite phases are low in hardness and high in ductility. Therefore, in a structure mainly comprised of ferrite, by making the grain size larger, it becomes possible to raise the formability of the material.

[0011] Carbides, by being made to suitably disperse in the metal structure, can maintain the formability of the material while imparting excellent wear resistance and rolling fatigue characteristics, so are structures essential for drive system parts. Further, carbides in steel plate are strong grains inhibiting slip. By making carbides present at the ferrite grain boundaries, it is possible to prevent the propagation of slip crossing crystal grain boundaries and suppress the formation of a shear zone, improve the cold forgeability, and simultaneously improve the formability of the steel plate.

[0012] However, cementite is a hard and brittle structure. If present in the state of a layered structure with ferrite, that is, pearlite, the steel becomes hard and brittle, so it must be made present in a spheroidal shape. If considering the cold forgeability and formation of cracks at the time of forging, the grain size has to be a suitable range.

[0013] However, no method of production for realizing the above structure has been disclosed up to now. Therefore, the inventors engaged in intensive research on a method of production for realizing the above structure.

[0014] As a result, they discovered that to make the metal structure of the steel plate after coiling after hot rolling a bainite structure comprised of small lamellar spacing fine pearlite or fine ferrite in which cementite is dispersed, the plate should be coiled up at a relatively low temperature (400°C to 550°C). By coiling at a relatively low temperature, the cementite dispersed in the ferrite also becomes easy to spheroidize. Next, as the first stage annealing, the cementite should be partially spheroidized by annealing at a temperature of right below the Ac1 point. Next, as the second stage annealing, part of the ferrite grains should be left while causing part to transform to austenite by annealing at a temperature between the Ac1 point and the Ac3 point (so-called dual phase region of ferrite and austenite). After that, the plate should be slowly cooled to cause the remaining ferrite grains to grow while using these as nuclei for transformation of austenite to ferrite to thereby obtain large ferrite phases while causing cementite to precipitate at the grain boundaries and realize the above structure.

[0015] That is, they discovered that a method of production of steel plate simultaneously satisfying hardenability and formability is difficult to realize even if adjusting the hot rolling conditions, annealing conditions, etc. separately and that it can be realized by achieving optimization in a so-called "integrated" process of hot rolling and annealing processes etc.

[0016] The present invention was made based on the above findings and has as its gist the following:

(1) A steel plate comprising, by mass%, C: 0.10 to 0.40%, Si: 0.01 to 0.30%, Mn: 1.00 to 2.00%, P: 0.020% or less, S: 0.010% or less, Al: 0.001 to 0.10%, N: 0.010% or less, O: 0.020% or less, Cr: 0.50% or less, Mo: 0.10% or less, Nb: 0.10% or less, V: 0.10% or less, Cu: 0.10% or less, W: 0.10% or less, Ta: 0.10% or less, Ni: 0.10% or less, Sn: 0.050% or less, Sb: 0.050% or less, As: 0.050% or less, Mg: 0.050% or less, Ca: 0.050% or less, Y: 0.050% or less, Zr: 0.050% or less, La: 0.050% or less, Ce: 0.050% or less and a balance of Fe and unavoidable impurities, wherein a metal structure of the steel plate has a ratio of a number of carbides at ferrite grain boundaries with respect to a number of carbides in ferrite grains of over 1, has a ferrite grain size of 5 μm to 50 μm, and has an area ratio of pearlite of 6% or less; and a Vickers hardness of the steel plate is 100HV to 170HV.

(2) The steel plate according to (1) containing one or both of Ti: 0.10% or less and B: 0.010% or less instead of part of the Fe.

(3) A method of production of the steel plate according to (1) or (2), the method of production of steel plate comprising hot rolling a steel slab of a chemical composition according to (1) or (2) during which completing finish hot rolling at a 750°C to 850°C temperature region to obtain hot rolled steel plate, coiling the hot rolled steel plate at 400°C to 550°C, pickling the coiled up hot rolled steel plate, holding the pickled hot rolled steel plate at a 650°C to 720°C temperature region for 3 hours to 60 hours as first stage annealing, then holding the hot rolled steel plate at a 725°C to 790°C temperature region for 3 hours to 50 hours as second stage annealing, and cooling the annealed hot rolled steel plate to 650°C by a cooling rate of 1°C/hour to 30°C/hour.

Advantageous Effects of Invention

[0017] According to the present invention, it is possible to provide steel plate excellent in formability and wear resistance, in particular suitable for obtaining gears, clutches, and other parts by forming of thick gauge plate, and a method of production of the same. Description of Embodiments

[0018] Below, the present invention will be explained in detail. First, the reasons for limitation of the chemical composition of the steel plate of the present invention will be explained. Below, the "%" of the components mean "mass%".

C: 0.10 to 0.40%

[0019] C is an element which forms carbides in the steel and is effective for strengthening of steel and refinement of ferrite grains. In order to suppress the formation of a matte surface of the steel plate at the time of cold forming and secure the beautiful appearance of the cold formed part, it is necessary to suppress coarsening of the ferrite grains. If less than 0.10%, the volume fraction of carbides is insufficient and coarsening of the carbides cannot be suppressed during box annealing, so C is made 0.10% or more. Preferably it is 0.12% or more.

[0020] On the other hand, if C is over 0.40%, the volume fraction of carbides increases, a large amount of cracks are formed and become starting points of fracture when applying a load for an instant, and the impact resistance characteristic deteriorates, so C is made 0.40% or less. Preferably, it is 0.38% or less.

Si: 0.01 to 0.30%

[0021] Si is an element which acts as a deoxidizing agent and has an effect on the form of the carbides. To obtain the deoxidizing effect, Si is made 0.01% or more. Preferably, it is 0.05% or more.

[0022] On the other hand, if over 0.30%, the ductility of the ferrite deteriorates, fractures more easily occur at the time of cold working, and the cold workability deteriorates, so Si is 0.30% or less. Preferably, it is 0.28% or less.

Mn: 1.00 to 2.00%

[0023] Mn is an element which raises the hardenability and contributes to the improvement of the strength. If less than 1.00%, securing the strength after hardening and the residual carbides after hardening becomes difficult, so Mn is made 1.00% or more. Preferably, it is 1.09% or more.

[0024] On the other hand, if over 2.00%, the Mn segregation becomes extremely band shaped and the workability remarkably deteriorates, so Mn is made 2.00% or less. Preferably, it is 1.91% or less.

Al: 0.001 to 0.10%

[0025] Al is an element acting as a deoxidizing agent of the steel and stabilizing ferrite. With less than 0.001%, the effect due to addition is not sufficiently obtained, so Al is made 0.001% or more. Preferably it is 0.004% or more.

[0026] On the other hand, if over 0.10%, a large amount of inclusions are formed and the cold workability deteriorates, so Al is made 0.10% or less. Preferably it is 0.08% or less.

[0027] The following elements are impurities and have to be controlled to certain amounts or less.

P: 0.0001 to 0.020%

[0028] P is an element which segregates at the ferrite grain boundaries and suppresses the formation of grain boundary carbides. The smaller the amount, the better, but in the refining process, if reducing P to less than 0.0001%, the refining costs greatly rise, so P is made 0.0001% or more. Preferably, it is 0.0013% or more.

[0029] On the other hand, if over 0.020%, the number ratio of the grain boundary carbides decreases and the cold workability deteriorates, so P is made 0.020% or less. Preferably, it is 0.018% or less.

S: 0.0001 to 0.010%

[0030] S is an impurity element forming MnS and other nonmetallic inclusions. Nonmetallic inclusions become starting points of fracture at the time of cold working, so S is preferably as small as possible, but if reducing it to less than 0.0001%, the refining costs greatly increase, so S is made 0.0001% or more. Preferably it is 0.0012% or more.

[0031] On the other hand, if over 0.010%, the cold workability deteriorates, so S is made 0.010% or less. Preferably, it is 0.007% or less.

N: 0.0001 to 0.010%

[0032] N is an element causing embrittlement of ferrite if contained in a large amount and is preferably as small as possible. The content of N may be made 0 as well, but if reducing it to less than 0.0001%, the refining costs greatly increase, so the substantive lower limit is 0.0001 to 0.0006%. On the other hand, if over 0.010%, the ferrite becomes brittle and the cold workability deteriorates, so N is made 0.010% or less. Preferably, it is 0.007% or less.

O: 0.0001 to 0.020%

[0033] O is an element forming coarse oxides in the steel if contained in a large amount and is preferably as small as possible. The content of O may also be 0%, but if reducing it to less than 0.0001%, the refining costs greatly increase, so the substantive lower limit is 0.0001 to 0.0011%. On the other hand, if over 0.020%, coarse oxides are formed in the steel and become starting points of fracture at the time of cold working, so O is made 0.02% or less. Preferably it is 0.017% or less.

Sn: 0.001 to 0.050%

[0034] Sn is an element which enters from the steel starting materials (scrap). It segregates at the grain boundaries and invites a drop in the number ratio of the grain boundary carbides, so is preferably as small as possible. The content of Sn may also be 0, but if reducing it to less than 0.001%, the refining costs greatly increase, so the substantive lower limit is 0.001 to 0.002%. On the other hand, if over 0.050%, the ferrite becomes brittle and the cold workability deteriorates, so Sn is made 0.050% or less. Preferably, it is 0.040% or less.

Sb: 0.001 to 0.050%

[0035] Sb, like Sn, is an element which enters from the steel starting materials (scrap). It segregates at the grain boundaries and invites a drop in the number ratio of the grain boundary carbides, so is preferably as small as possible. The content of Sb may also be 0, but if reducing it to less than 0.001%, the refining costs greatly increase, so the substantive lower limit is 0.001 to 0.002%. On the other hand, if over 0.050%, the ferrite becomes brittle and the cold workability deteriorates, so Sb is made 0.050% or less. Preferably, it is 0.040% or less.

As: 0.001 to 0.050%

[0036] As, like Sn and Sb, is an element which enters from the steel starting materials (scrap). It segregates at the grain boundaries and invites a drop in the number ratio of the grain boundary carbides, so is preferably as small as possible. The content of As may also be 0, but if reducing it to less than 0.001%, the refining costs greatly increase, so the substantive lower limit is 0.001 to 0.002%. On the other hand, if over 0.050%, the number ratio of the grain boundary carbides decreases and the cold workability deteriorates, so As is made 0.050% or less. Preferably, it is 0.040% or less.

[0037] The steel plate of the present invention has the above elements as basic components, but may further contain the following elements for the purpose of improving the cold forgeability of the steel plate. The following elements are not essential for obtaining the effects of the present invention, so their contents may be 0 as well.

Cr: 0.50% or less

[0038] Cr is an element which raises the hardenability and contributes to the improvement of the strength and, further, is an element which concentrates at the carbides and forms stable carbides even in the austenite phases. To obtain the effect of addition, Cr preferably is made 0.001% or more. More preferably, it is 0.007% or more. On the other hand, if over 0.50%, the carbides are liable to stabilize, the dissolution of the carbides at the time of hardening to become slow, and the required hardened strength to not be able to be achieved, so Cr is made 0.50% or less. Preferably, it is 0.45% or less.

Mo: 0.10% or less

[0039] Mo, like Mn, is an element effective for control of the form of carbides. To obtain the effect of addition, Mo preferably is made 0.001% or more. More preferably, it is 0.010% or more. On the other hand, if over 0.10%, the in-plane anisotropy of the "r" value deteriorates and the cold workability deteriorates, so Mo is made 0.10% or less. Preferably, it is 0.08% or less.

Nb: 0.10% or less

[0040] Nb is an element effective for control of the form of carbides. Further, it is an element refining the structure and contributing to improvement of the toughness. To obtain the effect of addition, Nb preferably is made 0.001% or more. More preferably, it is 0.002% or more. On the other hand, if over 0.10%, a large number of fine Nb carbides precipitate, the strength excessively rises, further, the number ratio of the grain boundary carbides decreases and the cold workability deteriorates, so Nb is made 0.10% or less. Preferably, it is 0.08% or less.

V: 0.10% or less

[0041] V also, like Nb, is an element effective for control of the form of carbides. Further, it is an element refining the structure and contributing to improvement of the toughness. To obtain the effect of addition, V preferably is made 0.001% or more. More preferably, it is 0.004% or more. On the other hand, if over 0.10%, a large number of fine V carbides precipitate, the strength excessively rises, further, the number ratio of the grain boundary carbides decreases and the cold workability deteriorates, so V is made 0.10% or less. Preferably, it is 0.08% or less.

Cu: 0.10% or less

[0042] Cu is an element which segregates at the crystal grain boundaries of the ferrite and forms fine precipitates to thereby contribute to the improvement of the strength. To obtain the effect of addition, Cu preferably is made 0.001% or more. More preferably, it is 0.005% or more. On the other hand, if over 0.10%, red shortness occurs and the productivity in the hot rolling deteriorates, so Cu is made 0.10% or less. Preferably, it is 0.08% or less.

W: 0.10% or less

[0043] W also, like Nb and V, is an element effective for control of the form of carbides. To obtain the effect of addition, W preferably is made 0.001% or more. More preferably, it is 0.003% or more. On the other hand, if over 0.10%, a large number of fine W carbides precipitate and the strength excessively rises. Further, since the number ratio of the grain boundary carbides decreases and the cold workability deteriorates, W is made 0.10% or less. Preferably, it is 0.08% or less.

Ta: 0.10% or less

[0044] Ta also, like Nb, V, and W, is an element effective for control of the form of carbides. To obtain the effect of addition, Ta preferably is made 0.001% or more. More preferably, it is 0.005% or more. On the other hand, if over 0.10%, a large number of fine Ta carbides precipitate and the strength excessively rises. Further, the number ratio of the grain boundary carbides decreases and the cold workability deteriorates, so Ta is 0.10% or less. Preferably, it is 0.08% or less.

Ni: 0.10% or less

[0045] Ni is an element effective for improving the toughness of a part. To obtain the effect of addition, Ni preferably is made 0.001% or more. More preferably, it is 0.004% or more. On the other hand, if over 0.10%, the number ratio of the grain boundary carbides decreases and the cold workability deteriorates, so Ni is made 0.10% or less. Preferably, it is 0.08% or less.

Mg: 0.050% or less

[0046] Mg is an element able to control the form of sulfides by addition of a trace amount. To obtain the effect of addition, Mg preferably is made 0.0001% or more. More preferably, it is 0.0008% or more. On the other hand, if over 0.050%, the ferrite becomes brittle and the cold workability deteriorates, so Mg is made 0.050% or less. Preferably, it is 0.040% or less.

Ca: 0.050% or less

[0047] Ca, like Mg, is an element able to control the form of sulfides by addition of a trace amount. To obtain the effect of addition, Ca preferably is made 0.001% or more. More preferably, it is 0.003% or more. On the other hand, if over 0.050%, coarse Ca oxides are formed and become starting points of fracture at the time of cold working, so Ca is made 0.050% or less. Preferably, it is 0.040% or less.

Y: 0.050% or less

[0048] Y, like Mg and Ca, is an element able to control the form of sulfides by addition of a trace amount. To obtain the effect of addition, Y preferably is made 0.001% or more. More preferably, it is 0.003% or more. On the other hand, if over 0.050%, coarse Y oxides are formed and become starting points of fracture at the time of cold working, so Y is made 0.050% or less. Preferably, it is 0.035% or less.

Zr: 0.050% or less

[0049] Zr, like Mg, Ca, and Y, is an element able to control the form of sulfides by addition of a trace amount. To obtain the effect of addition, Zr preferably is made 0.001% or more. More preferably, it is 0.004% or more. On the other hand, if over 0.050%, coarse Zr oxides are formed and become starting points of fracture at the time of cold working, so Zr is made 0.050% or less. Preferably, it is 0.045% or less.

La: 0.050% or less

[0050] La is an element effective for control of the form of sulfides by addition of a trace amount, but is also an element which segregates at the grain boundaries and invites a drop in the number ratio of the grain boundary carbides. To obtain the effect of addition, La preferably is made 0.001% or more. More preferably, it is 0.004% or more. On the other hand, if over 0.050%, the number ratio of the grain boundary carbides decreases and the cold workability deteriorates, so La is made 0.050% or less. Preferably, it is 0.045% or less.

Ce: 0.050% or less

[0051] Ce, like La, is an element able to control the form of sulfides by addition of a trace amount, but is also an element which segregates at the grain boundaries and invites a drop in the number ratio of the grain boundary carbides. To obtain the effect of addition, Ce preferably is made 0.001% or more. More preferably, it is 0.004% or more. On the other hand, if over 0.050%, the number ratio of the grain boundary carbides decreases and the cold workability deteriorates, so Ce is made 0.050% or less. Preferably, it is 0.046% or less.

[0052] The balance of the chemical composition of the steel plate of the present invention is comprised of Fe and unavoidable impurities.

[0053] Note that, part of the Fe may be replaced by one or both of Ti and B.

Ti: 0.10% or less

[0054] Ti is an element effective for control of the form of carbides. Further, it is also an element refining the structure and contributing to improvement of the toughness. To obtain the effect of addition, Ti preferably is made 0.001% or more. More preferably, it is 0.005% or more. On the other hand, if over 0.10%, coarse Ti oxides are formed and become starting points of fracture at the time of cold working, so Ti is 0.10% or less. Preferably, it is 0.08% or less.

B: 0.0001 to 0.010%

[0055] B is an element which raises the hardenability at the time of heat treatment of a part and makes the structure uniform and which contributes to improvement of the toughness. To obtain the effect of addition, B preferably is made 0.0001% or more. More preferably, it is 0.0006% or more. On the other hand, if over 0.010%, coarse B oxides are formed and become starting points of fracture at the time of cold working, so B is made 0.010% or less. Preferably, it is 0.009% or less.

[0056] Next, the structure of the steel plate of the present invention will be explained.

[0057] The structure of the steel plate of the present invention is substantially a structure comprised of ferrite and carbides. Carbides include not only the cementite (Fe_3C) of the compound of iron and carbon but also compounds where the Fe atoms in cementite are replaced by Mn, Cr, and other alloy elements and alloy carbides ($M_{23}C_6$, M_6C, MC, etc. [where M: Fe, and other metal elements added as alloys]).

[0058] When forming steel plate into a predetermined form, a shear zone is formed in the macrostructure of the steel plate and slip deformation occurs concentratedly near the shear zone. Slip deformation is accompanied with proliferation of dislocations. Near the shear zone, a region of high dislocation density is formed. Along with the increase of the amount of strain imparted to the steel plate, slip deformation is promoted and the dislocation density increases.

[0059] In cold forging, strong working is performed with an equivalent strain of over 1. For this reason, in conventional steel plate, it was not possible to prevent the formation of voids and/or cracks accompanying the increase in the dislocation density. In conventional steel plate, improvement of the cold forgeability was difficult. To solve this problem, it is effective to suppress the formation of a shear zone at the time of forming.

[0060] From the viewpoint of the microstructure, the formation of a shear zone is understood as the phenomenon of slip occurring at a certain single crystal grain crossing crystal grain boundaries and continuously propagating to the adjoining crystal grains. Accordingly, to suppress formation of a shear zone, it is necessary to prevent propagation of slip crossing crystal grain boundaries.

[0061] Carbides in steel plate are strong grains inhibiting slip. By forming carbides at the ferrite grain boundaries, propagation of slip crossing crystal grain boundaries can be prevented and formation of a shear zone can be suppressed so the cold formability can be improved. Simultaneously, the formability of the steel plate is improved.

[0062] The formability of steel plate is greatly dependent on the accumulation of strain inside the crystal grains (accumulation of dislocations). If propagation of strain to the adjoining crystal grains is inhibited at the crystal grain boundaries, the amount of strain inside the crystal grains will increase. As a result, the work hardening rate will increase and the formability will be improved.

[0063] Based on theory and principle, cold workability is considered to be strongly affected by the rate of coverage of the ferrite grain boundaries by carbides. High precision measurement of the coverage rate becomes necessary.

[0064] For measurement of the rate of coverage of ferrite grain boundaries by carbides in a three-dimensional space, serial sectioning SEM observation using an FIB to repeatedly cut and observe a sample in a scanning electron microscope or 3D EBSP observation becomes essential. A massive measurement time is required and technical knowhow has to be built up. This was clarified by the inventors. They concluded that this was not suitable as a general method of analysis.

[0065] For this reason, the inventors searched for a simple, high precision indicator for evaluation. As a result, the inventors discovered that if using the ratio of the number of carbides at the ferrite grain boundaries with respect to the number of carbides in the ferrite grains as an indicator, the cold workability can be evaluated and that if the ratio of the number of carbides at the ferrite grain boundaries with respect to the number of carbides in the ferrite grains is over 1, the cold workability is remarkably improved.

[0066] Note that, buckling, folding, and twisting of the steel plate occurring at the time of cold working all occur due to localization of strain accompanying the formation of a shear zone, so by forming carbides at the ferrite grain boundaries, formation of a shear zone and localization of strain are eased and occurrence of buckling, folding, and twisting is suppressed.

[0067] If the spheroidization rate of the carbides at the crystal grain boundaries is less than 80%, strain locally concentrates at the rod-shaped or plate-shaped carbides and voids and/or cracks easily are formed, so the spheroidization rate of the carbides at the crystal grain boundaries is preferably 80% or more, more preferably it is 90% or more.

[0068] If the average grain size of the carbides is less than 0.1 μm, the hardness of the steel plate remarkably increases and the workability deteriorates, so the average grain size of the carbides is preferably 0.1 μm or more. More preferably, it is 0.17 μm or more. On the other hand, if the average grain size of carbides is over 2.0 μm, at the time of cold working, coarse carbides become starting points of cracks and the cold workability deteriorates, so the average grain size of the carbides is preferably 2.0 μm or less. More preferably, it is 1.95 μm or less.

[0069] The carbides are observed by a scanning electron microscope. Before observation, the sample for observation of the structure is polished by wet polishing by Emery paper and a diamond abrasive having an average particle size of 1 μm, the observed surface is polished to a mirror finish, then a 3% nitric acid-alcohol solution is used to etch the structure. For the magnification of the observation, magnification enabling judgment of the structure of ferrite and carbides in 3000X is selected. Eight images of fields of 30 μm×40 μm at a plate thickness 1/4 layer are captured at random by the selected magnification.

[0070] The obtained structural images are analyzed by image analysis software such as made by Mitani Shoji K.K. (Win ROOF) so as to measure in detail the areas of the carbides contained in those regions. From the areas of the carbides, the circle equivalent diameter (=2×√(area/3.14)) is found. The average value is made the carbide grain size. Further, the spheroidization rate of the carbides was found by approximating the carbides by ovals of the equivalent area and equivalent inertia moment and calculating the ratio of the ones with ratios of the maximum lengths and maximum lengths in the direction perpendicular to the same becoming less than 3.

[0071] Note that to suppress the effect of measurement error due to noise, carbides with an area of 0.01 μm^2 or less are excluded from coverage by the evaluation. The number of carbides present at the ferrite grain boundaries is counted and the number of carbides at the ferrite grain boundaries is subtracted from the total number of carbides to calculate the number of carbides inside the ferrite grains. Based on the measured numbers, the number ratio of carbides at the ferrite grain boundaries to carbides inside the ferrite grains was found.

[0072] In the structure after annealing the cold rolled steel plate, it is possible to improve the cold workability by making the ferrite grain size 5.0 μm or more. If the ferrite grain size is less than 5 μm, the hardness increases and, at the time of cold working, fractures and cracks easily form, so the ferrite grain size is made 5 μm or more. Preferably, it is 7 μm or more.

[0073] On the other hand, if over 50 μm, the number of the carbides at the crystal grain boundaries suppressing propagation of slip decreases and the cold workability deteriorates, so the ferrite grain size is 50 μm or less. Preferably, it is 37 μm or less.

[0074] The ferrite grain size is measured by polishing the observed surface of the sample to a mirror finish by the above-mentioned polishing method, then etching the surface by a 3% nitric acid-alcohol solution and observing the structure of the observed surface by an optical microscope or scanning electron microscope and applying the line segment method to the captured images.

[0075] Further, the carbide of iron, that is, cementite, is a hard and brittle structure. If present as a layered structure with ferrite, that is, in the state of pearlite, the steel becomes hard and brittle. Therefore, pearlite has to be reduced as much as possible. In the steel plate of the present invention, it is made an area ratio of 6% or less.

[0076] Pearlite has a distinctive lamellar structure, so can be discerned by observation by an SEM or optical microscope. By calculating the regions of lamellar structures in any cross-section, it is possible to find the area ratio of pearlite.

[0077] Furthermore, by making the Vickers hardness of the steel plate 100HV to 170HV, it is possible to improve the cold workability. If the Vickers hardness is less than 100HV, buckling easily occurs during cold working, so the Vickers hardness is made 100HV or more. Preferably, it is 110HV or more.

[0078] On the other hand, if the Vickers hardness is over 170HV, the ductility deteriorates and internal fractures easily occur at the time of cold working, so the Vickers hardness is made 170HV or less. Preferably, it is 168HV or less.

[0079] Next, the method of production of the present invention will be explained.

[0080] The method of production of the present invention has as its basic idea to use a steel slab of the above-mentioned chemical composition, integrally manage the hot rolling conditions and annealing conditions, and control the structure of the steel plate.

[0081] To start, a steel slab obtained by continuously casting molten steel of each of the required chemical compositions was prepared for hot rolling. The continuously cast steel slab may be directly used for hot rolling or may be used for hot rolling after being cooled once, then heated.

[0082] If cooling once, then heating a steel slab to use it for hot rolling, the heating temperature is preferably 1000°C to 1250°C and the heating time is preferably 0.5 hour to 3 hours. When directly using the continuously cast steel slab for hot rolling, the temperature of the steel slab used for the hot rolling is preferably 1000°C to 1250°C.

[0083] If the steel slab temperature or steel slab heating temperature is over 1250°C or the steel slab heating time is over 3 hours, the decarburization of the surface layer of the steel slab becomes remarkable, at the time of heating before carburizing and quenching, the austenite grains at the surface layer of the steel plate abnormally grow, and the impact resistance deteriorates. For this reason, the steel slab temperature or steel slab heating temperature is preferably 1250°C or less and the heating time is preferably 3 hours or less. More preferably, it is 1200°C or less or 2.5 hours or less.

[0084] If the steel slab temperature or steel slab heating temperature is less than 1000°C or the heating time is less than 0.5 hour, the microsegregation or macrosegregation formed by the casting is not eliminated, regions remain inside the steel slab where Si, Mn, and other alloy elements locally concentrate, and the impact resistance deteriorates. For this reason, the steel slab temperature or steel slab heating temperature is preferably 1000°C or more and the heating time is preferably 0.5 hour or more. More preferably, it is 1050°C or more or 1 hour or more.

[0085] The finish rolling in the hot rolling is completed in the 750°C to 850°C temperature region. If the finish rolling temperature is less than 750°C, the deformation resistance of the steel plate increases and the rolling load remarkably rises. Further, the amount of roll wear increases and the productivity deteriorates. Along with this, the recrystallization required for improving the plasticity anisotropy does not sufficiently proceed. Therefore, the finish rolling temperature is made 750°C or more. In the point of promoting recrystallization, preferably it is 770°C or more.

[0086] If the finish rolling temperature is over 850°C, bulky scale is formed while the plate passes the run out table (ROT). Due to the scale, flaws are formed at the surface of the steel plate. When applying an impact load after cold forging and carburizing, quenching, and tempering, cracks easily form starting from the flaws, so the impact resistance of the steel plate deteriorates. For this reason, the finish rolling temperature is made 850°C or less. Preferably, it is 830°C or less.

[0087] When cooling hot rolled steel plate on the ROT after finish rolling, the cooling rate is preferably 10°C/sec to 100°C/sec. If the cooling rate is less than 10°C/sec, bulky scale is formed in the middle of cooling, formation of flaws due to the same cannot be suppressed, and the impact resistance deteriorates, so the cooling rate is preferably 10°C/sec or more. More preferably, it is 20°C/sec or more.

[0088] If cooling by a cooling rate over 100°C/sec from the surface layer to inside of the steel plate, the surface-most layer part is excessively cooled resulting in the formation of bainite, martensite, and other low temperature transformed structures. When paying out hot rolled steel plate coil cooled to 100°C to room temperature after coiling, fine cracks form in the low temperature transformed structure. These fine cracks are difficult to remove by pickling and cold rolling.

[0089] Further, if applying an impact load to the steel plate after cold forging and carburizing, quenching, and tempering, cracks grow starting from fine cracks, so the impact resistance deteriorates. For this reason, to keep bainite and martensite or other low temperature transformed structures from forming at the surface-most layer part of the steel plate, the cooling rate is preferably 100°C/sec or less. More preferably, it is 90°C/sec or less.

[0090] Note that, the cooling rate indicates the cooling ability received from the cooling facilities in each water spray section at the time when being cooled on the ROT down to the target temperature of coiling from the time when the hot rolled steel plate after finish rolling is water cooled at a water spray section after passing through a non-water spray section. It does not show the average cooling rate from the starting point of water spray to the temperature at which the steel plate is coiled up by the coiler.

[0091] The coiling temperature is made 400°C to 550°C. This is a temperature lower than the general coiling temperature. In particular, it is a condition not usually applied when the content of C is high. By coiling the hot rolled steel plate produced under the above-mentioned condition in this temperature range, the structure of the steel plate can be made a bainite structure comprised of fine ferrite in which carbides are dispersed.

[0092] If the coiling temperature is less than 400°C, the austenite, which had not yet been transformed before coiling, transforms to hard martensite. At the time of payout of the hot rolled steel plate coil, cracks form at the surface layer of the hot rolled steel plate and the impact resistance deteriorates.

[0093] Further, when recrystallizing ferrite from austenite, the driving force of recrystallization is small, the orientation of the recrystallized ferrite grains is strongly affected by the orientation of the austenite grains, and making the texture randomized becomes difficult. For this reason, the coiling temperature is made 400°C or more. Preferably, it is 430°C or more.

[0094] If the coiling temperature is over 550°C, large lamellar spacing pearlite is formed and high heat stability, bulky needle-shaped carbides are formed. These needle-shaped carbides remain even after two-stage annealing. At the time of cold forging and other forming of steel plate, the needle-shaped carbides form starting points for cracks.

[0095] Further, when recrystallizing ferrite from austenite, conversely, the driving force of recrystallization becomes too great. Even in such a case, the result becomes recrystallized ferrite grains strongly dependent on the orientation of the austenite grains. The texture is not randomized. For this reason, the coiling temperature is made 550°C or less. Preferably, it is 520°C or less.

[0096] The hot rolled steel plate coil is paid out and pickled, then is treated by two-stage step type annealing (two-stage annealing) holding it at two temperature regions. By treating the hot rolled steel plate by two-stage annealing, the stability of the carbides is controlled and the formation of carbides at the ferrite grain boundaries is promoted.

[0097] If cold rolling the pickled steel plate before the annealing treatment, the ferrite grains are refined, so the steel plate becomes harder to soften. For this reason, in this Description, it is not preferable to perform the cold rolling before the annealing. It is preferable to perform the annealing treatment without performing cold rolling after pickling.

[0098] The first stage annealing is performed at 650 to 720°C, preferably the Ac1 point or less temperature region. Due to this annealing, the carbides are made to coarsen and are made to partially spheroidize and the alloy elements are made to concentrate at the carbides to raise the thermal stability of the carbides.

[0099] In the first stage annealing, the heating rate up to the annealing temperature (below, referred to as the "first stage heating rate") is 30°C/hour to 150°C/hour. If the first stage heating rate is less than 30°C/hour, time is raised for raising the temperature and the productivity deteriorates, so the first stage heating rate is made 3°C/hour or more. Preferably, it is 10°C/hour or more.

[0100] On the other hand, if the first stage heating rate is over 150°C/hour, at the hot rolled steel plate coil, the temperature difference between the peripheral parts and the inside increases whereby scratches and seizing occur due to the difference in heat expansion and relief shapes are formed at the surface of the steel plate. At the time of cold forging and other forming, the relief shapes form starting points for cracks, the cold forgeability deteriorates, and the formability and the impact resistance after carburizing, quenching, and tempering deteriorates, so the first stage heating rate is made 150°C/hour or less. Preferably, it is 130°C/hour or less.

[0101] The annealing temperature in the first stage annealing (below, called the "first annealing temperature") is 650°C to 720°C. If the first annealing temperature is less than 650°C, the carbides are not sufficiently stabilized and at the time of the second stage annealing, it becomes difficult to make carbides remain in the austenite. For this reason, the first annealing temperature is made 650°C or more. Preferably, it is 670°C or more.

[0102] On the other hand, if the first stage annealing temperature is over 720°C, before the carbides rise in stability, austenite is formed and the above-mentioned control of changes in structures becomes difficult, so the first stage annealing temperature is made 720°C or less. Preferably, it is 700°C or less.

[0103] The annealing time in the first stage annealing (below, called the "first annealing time") is 3 hours to 60 hours. If the first annealing time is less than 3 hours, the carbides are not sufficiently stabilized and at the time of the second stage annealing, it becomes difficult to make carbides remain in the austenite. For this reason, the first annealing time is made 3 hours or more. Preferably, it is 5 hours or more.

[0104] On the other hand, if the first annealing time is over 60 hours, much greater stabilization of the carbides cannot be expected. Furthermore, the productivity deteriorates, so the first annealing time is made 60 hours or less. Preferably, it is 55 hours or less.

[0105] After that, the temperature is raised to 725 to 790°C, preferably, the Ac1 point to the A_3 point in temperature range, and austenite is made to form in the structure. At that time, the carbides in the fine ferrite grains dissolve in the austenite, but the carbides coarsened by the first stage annealing remain in the austenite.

[0106] If cooling is performed without this second stage annealing, the ferrite grain size cannot be enlarged and the ideal structure cannot be obtained.

[0107] The heating rate up to the annealing temperature of the second stage annealing (below, referred to as the "second stage heating rate") is 1°C/hour to 80°C/hour. At the time of the second stage annealing, austenite is formed and grows from the ferrite grain boundaries. At that time, by slowing the heating rate up to the annealing temperature, formation of nuclei of austenite is suppressed and, in the structure formed by slow cooling after annealing, the rate of coverage of grain boundaries by carbides can be raised.

[0108] For this reason, the second stage heating rate is preferably slower, but if less than 1°C/hour, time is required to raise the temperature and the productivity deteriorates, so the second stage heating rate is made 1°C/hour or more. Preferably, it is 10°C/hour or more.

[0109] If the second stage heating rate is over 80°C/hour, at the hot rolled steel plate coil, the temperature difference between the peripheral parts and the inside increases, and scratches and seizing occur due to the large difference in heat expansion, and as a result, deformation and relief shapes are formed at the surface of the steel plate. At the time of cold forging, the relief shapes form starting points for cracks, the cold forgeability and formability deteriorates, and the impact resistance after carburizing, quenching, and tempering also deteriorates, so the second stage heating rate is made 80°C/hour or less. Preferably, it is 70°C/hour or less.

[0110] The annealing temperature at the second stage annealing (below, called the "second stage annealing temperature") is made 725°C to 790°C. If the second stage annealing temperature is less than 725°C, the amount of formation of austenite becomes smaller and, after the cooling after the second stage annealing, the number of carbides at the ferrite grain boundaries decreases and the ferrite grain size becomes smaller. For this reason, the second stage annealing temperature is made 725°C or more. Preferably, it is 735°C or more.

[0111] On the other hand, if the second stage annealing temperature exceeds 790°C, it becomes difficult to make carbides remain at the austenite and control of changes in structure becomes difficult, so the second stage annealing temperature is made 790°C or less. Preferably, it is 770°C or less.

[0112] The annealing time in the second stage annealing (second stage annealing time) is made 3 hours to 50 hours. If the second stage annealing time is less than 3 hours, the amount of production of austenite becomes smaller and the carbides in the ferrite grains do not sufficiently dissolve so it becomes difficult to increase the number of carbides at the ferrite grain boundaries and, further, the ferrite grain size becomes smaller. For this reason, the second stage annealing time is made 3 hours or more. Preferably, it is 5 hours or more.

[0113] On the other hand, if the second stage annealing time is over 50 hours, it becomes difficult to make carbides remain in the austenite. Further, the manufacturing costs increase, so the second stage annealing time is made less than 50 hours. Preferably, it is 40 hours or less.

[0114] After the two-stage annealing, the steel plate is cooled by a 1°C/hour to 30°C/hour cooling rate down to 650°C.

[0115] By using slow cooling to slowly cool the austenite produced in the second stage annealing, the austenite is transformed to ferrite, carbon atoms are adsorbed at the carbides remaining in the austenite, the carbides and austenite cover the ferrite grain boundaries, and, finally, it is possible to obtain a structure in which a large number of carbides are present at the ferrite grain boundaries.

[0116] For this reason, the cooling rate is preferably slow, but if less than 1°C/hour, the time required for cooling increases and the productivity deteriorates, so the cooling rate is made 1°C/hour or more. Preferably, it is 10°C/hour or more.

[0117] On the other hand, if the cooling rate is over 30°C/hour, the austenite transforms to pearlite and the hardness of the steel plate increases, the cold forgeability deteriorates, and the impact resistance after carburizing, quenching, and tempering deteriorates, so the cooling rate is made 30°C/hour or less. Preferably, it is 20°C/hour or less.

[0118] Furthermore, the steel plate cooled down to 650°C is cooled down to room temperature. The cooling rate at this time is not limited.

[0119] The atmosphere in the two-stage annealing is not limited to any specific atmosphere. For example, it may also be any atmosphere of a 95% or more nitrogen atmosphere, 95% or more hydrogen atmosphere, and air atmosphere.

[0120] As explained above, according to the method of production of integrally managing the hot rolling conditions and annealing conditions of the present invention and controlling the structure of the steel plate, it is possible to produce steel plate excellent in formability at the time of cold forging combining drawing and thickening and further excellent in the hardenability required for improving the impact resistance after carburizing, quenching, and tempering.

Examples

[0121] Next, examples will be explained, but the levels in the examples are illustrations of conditions employed for confirming the workability and effects of the present invention. The present invention is not limited to these illustrations of conditions. The present invention can employ various conditions so long as not departing from the gist of the present invention and as achieving the object of the present invention.

[0122] The cold workability was evaluated by taking a JIS No. 5 tensile test piece from the plate thickness 3 mm material as annealed and conducting a tensile test. The total elongations in the direction of 0° from the rolling direction and the direction of 90° from the rolling direction were evaluated. In the case where, in both directions, they were 35% or more and the difference of the total elongations $|\Delta EL|$ in the two directions was 4% or less, it is judged that the cold workability was excellent.

[0123] The hardenability was evaluated by grinding a plate thickness 3 mm material as annealed to a plate thickness 1.5 mm, holding it in a vacuum atmosphere at 880°C×10 minutes, hardening it by a 30°C/sec cooling rate, and judging the hardenability was excellent if the fraction of martensite was 60% or more.

Example 1

[0124] Continuously cast slabs of the chemical compositions shown in Table 1 (steel ingots) were heated at 1240°C for 1.8 hours, then hot rolled. The finish hot rolling was completed at 890°C. After that, the plates were coiled up at 510°C to produce plate thickness 3 mm hot rolled coils. The hot rolled coils were pickled and loaded into a box type annealing furnace. The atmosphere was controlled so as to include 95% hydrogen-5% nitrogen. The coils were heated from room temperature to 705°C and held at 705°C for 36 hours to make the temperature distribution inside the hot rolled coils uniform, then were heated to 760°C and held at 760°C for 10 hours.

[0125] After that, the steel plates were cooled down to 650°C by a 10°C/hour cooling rate, then were furnace cooled down to room temperature to prepare samples for evaluation of characteristics. Note that, the structures of the samples were measured by the above-mentioned method.

Table 1

	C	Si	Mn	P	S	Al	N	O	Ti	Cr	Mo	B	Nb	V	Cu	W	Ta	Ni	Sn	Sb	As	Mg	Ca	Y	Zr	La	Ce	Remarks
A	0.34	0.23	1.47	0.0044	0.004																							Com.steel
B	0.34	0.19	1.53	0.0084	0.0064																							Inv.steel
C	0.16	0.13	1.15	0.0055	0.0052	0.062							0.071	0.033										0.048				Inv.steel
D	0.11	0.26	1.76	0.0076	0.0027		0.01		0.019																			Inv.steel
E	0.35	0.25	1.71	0.0046	0.001																							Inv.steel
F	0.32	0.19	1.04	0.0155	0.0043							0.0082		0.086		0.085						0.0059			0.045			Inv.steel
G	0.67	0.08	1.29	0.0065	0.0038																							Inv.steel
H	0.22	0.12	1.84	0.0018	0.0091						0.046												0.027			0.048		Inv.steel
I	0.16	0.06	1.15	0.004	0.0053	0.087				0.342		0.0091									0.009				0.011			Com.steel
J	0.21	0.12	1.35	0.0082	0.0078				0.02			0.003				0.037											0.002	Inv.steel
K	0.13	0.21	1.76	0.003	0.0004			0.0148						0.005						0.041				0.043				Com.steel
L	0.18	0.13	1.34	0.0053	0.0064		0.0002										0.009			0.008								Inv.steel
M	0.19	0.04	1.95	0.0087	0.0038																							Com.steel
N	0.21	0.16	1.86	0.02	0.0098																							Com.steel
O	0.37	0.3	1.68	0.0112	0.0044			0.0121		1.06	0.39					0.062							0.008			0.006		Com.steel
P	0.37	0.19	2	0.0074	0.0093																							Inv.steel
Q	0.12	1.18	1.09	0.0071	0.0074																							Inv.steel
R	0.38	0.09	1.11	0.0081	0.0016		0.0056						0.003					0.064				0.0416						Inv.steel
S	0.31	0.08	1.84	0.0104	0.0011									0.053			0.003	0.093				0.035						Inv.steel
T	0.19	0.19	1.91	0.0086	0.0057																							Inv.steel
U	0.21	0.02	1.02	0.0113	0.28																							Inv.steel
V	0.17	0.02	1.07	0.0194	0.0073																							Com.steel
W	0.11	0.01	1.41	0.0081	0.0034	0.014				0.043								0.023	0.026								0.004	Inv.steel
X	0.35	0.28	1.16	0.0013	0.0005						0.09								0.039		0.02							Inv.steel
Y	0.23	0.02	1.44	0.0006	0.0057																							Inv.steel
O	0.36	0.03	1.06	0.0132	0.002										0.093										0.018		0.037	Inv.steel
AA	0.3	0.27	0.71	0.0106	0.001																							Inv.steel
AB	0.31	0.2	1.52	0.0153	0.0075			0.0021								0.011										0.002		Inv.steel
AC	0.31	0.11	1.4	0.0045	0.0031			0.0182	0.003						0.052				0.033									Inv.steel
AD	0.14	0.26	1.34	0.011	0.0009	0.61							0.006				0.093				0.05			0.033				Inv.steel

[0126] Table 2 shows the results of measurement or evaluation of the Vickers hardness of the produced samples, the ratio of the number of carbides at the ferrite grain boundaries to the number of carbides inside the ferrite grains, the pearlite area ratio, the cold workability, and the hardenability.

Table 2

	Hot rolling conditions		Carbide size [μm]	Ferrite grain size [μm]	Pearlite area ratio [%]	Vickers hardness [HV]	No. of grain boundary carbides/ No. of grain carbides	Total elongation [%]		Elongation anisotropy	Martensite fraction [%]	Remarks
	Finish hot rolling temp. [°C]	Coiling temp. [°C]						0° direction	90° direction	$\lvert EL_{0°}-EL_{90°}\rvert$		
A-1	776	591	0.94	18-6	1.1	130	8.65	33.8	34.3	0.5	93	Comp. steel
B-1	815	490	1.09	22.1	0.3	123	7.94	39.0	39.7	0.7	95	Inv. steel
C-1	798	369	1.23	27.2	1.0	110	9.14	41.5	43.0	1.5	62	Comp. steel
D-1	921	500	1.08	32.0	1.6	118	6.94	40.0	40.7	0.7	72	Comp. steel
E-1	763	410	1.18	24.5	1.2	125	6.17	38.7	39.6	0.9	94	Inv. steel
F-1	824	442	1.15	21.6	0.6	121	8.07	39.4	40.5	1.1	74	Inv. steel
G-1	773	427	1.17	18.4	1.6	149	9.14	34.3	35.6	1.3	98	Comp. steel
H-1	820	485	1.09	26.1	1.5	114	6.88	40.7	41.6	0.9	96	Inv. steel
1-1	710	481	1.09	22.7	0.6	110	9.42	41.4	42.0	0.6	69	Comp. steel
J-1	810	497	1.08	23.8	1.8	113	9.52	40.8	42.3	1.5	75	Inv. steel
K-1	784	407	1.19	32.5	2.0	114	7.64	40.7	42.1	1.4	83	Inv. steel
L-1	759	543	1.02	23.1	0.3	114	9.00	40.6	41.1	0.5	72	Inv. steel
M-1	791	541	1.02	26.4	0.3	109	8.48	41.6	42.9	1.3	97	Inv. steel
N-1	806	499	1.08	25.9	2.1	116	8.17	40.3	41.6	1.3	95	Inv. steel
0-1	778	478	1.04	19.0	0.9	135	8.22	36.9	37.5	0.6	38	Comp. steel
P-1	807	453	1.13	25.2	1.2	121	8.47	39.4	40.8	1.4	100	Inv. steel
Q-1	832	542	1.02	18.9	8.3	178	5.99	28.8	29.6	0.8	7	Comp. steel

(continued)

	Hot rolling conditions		Carbide size [μm]	Ferrite grain size [μm]	Pearlite area ratio [%]	Vickers hardness [HV]	No. of grain boundary carbides/ No. of grain carbides	Total elongation [%]		Elongation anisotropy	Martensite fraction [%]	Remarks
	Finish hot rolling temp. [°C]	Coiling temp. [°C]						0° direction	90° direction	$\lvert EL_{0°}-EL_{90°} \rvert$		
R-1	758	511	1.07	20.5	1.9	119	7.94	39.8	40.6	0.8	82	Inv. steel
S-1	840	391	1.20	27.0	0.7	111	6.97	41.2	42.6	1.4	105	Comp. steel
T-1	756	538	1.02	25.4	1.9	119	6.18	39.8	40.7	0.9	95	Inv. steel
U-1	817	510	1.07	22.6	1.5	106	7.70	38.6	43.4	4.8	69	Comp. steel
V-1	788	633	0.87	19.4	1.3	110	6.35	32.8	33.8	1.0	68	Comp. steel
W-1	761	446	1.14	31.9	1.4	100	9.85	43.2	44.2	1.0	67	Inv. steel
X-1	831	455	1.14	20.9	0.7	129	8.74	38.5	39.4	0.9	81	Inv. steel
Y-1	818	440	1.15	25.6	1.9	106	7.94	42.1	42.9	0.8	81	Inv. steel
Z-1	763	456	1.14	21.9	0.6	112	8.14	41.0	41.5	0.5	78	Inv. steel
AA-1	824	414	1.19	21.8	0.7	123	6.09	39.0	40.3	1.3	51	Comp. steel
AB-1	843	454	1.13	23.1	1.5	122	8.94	39.3	39.8	0.5	92	Inv. steel
AC-1	834	508	1.07	21.8	0.9	117	9.51	40.1	41.6	1.5	87	Inv. steel
AD-1	791	460	1.13	18.0	1.8	138	7.13	34.8	35.2	0.4	5	Comp. steel

[0127] As shown in Table 2, Invention Steels B-1, E-1, F-1, H-1, J-1, K-1, L-1, M-1, N-1, P-1, R-1, T-1, W-1, X-1, Y-1, Z-1, AB-1, and AC-1 all have a ratio of the number of carbides at the ferrite grain boundaries with respect to the number of carbides inside the ferrite grains of over 1, a Vickers hardness of 170HV or less, and excellent cold workability and hardenability.

[0128] As opposed to this, Comparative Steel G-1 is high in amount of C and deteriorates in cold workability. Comparative Steel O-1 is high in amount of Mo and amount of Cr and is high in stability of carbides, so the carbides do not dissolve at the time of hardening, the amount of formation of austenite is small, and the hardenability is inferior.

[0129] Comparative Steels Q-1 and AD-1 are high in amounts of Si and Al and high in A3 point, so the amount of formation of austenite at the time of hardening is small and the hardenability is inferior. Comparative Example U-1 is high in amount of S, has coarse MnS formed in the steel, and is low in cold workability. Comparative Example AA-1 is low in amount of Mn and inferior in hardenability.

[0130] Comparative Example I-1 is low in finishing temperature of hot rolling and deteriorates in productivity. Comparative Example D-1 is high in finishing temperature of hot rolling and has scale flaws formed at the steel plate surface. Comparative Examples C-1 and S-1 are low in coiling temperature of hot rolling, are increased in number of bainite, martensite, and other low temperature transformed structures, become brittle resulting in frequent fracture at the time of pay out of the hot rolled coil, and deteriorates in productivity.

[0131] Comparative Examples A-1 and V -1 are high in coiling temperature of hot rolling and have hot rolled structures formed with large lamellar spacing bulky pearlite and high heat stability needle-shaped coarse carbides. The carbides remain in the steel plate even after two-stage step type annealing and the cold workability deteriorates.

Example 2

[0132] To investigate the effects of the annealing conditions, steel slabs of the chemical compositions shown in Table 1 were heated at 1240°C for 1.8 hours, then used for hot rolling. The finish hot rolling was ended at 820°C, then the plates were cooled on the ROT by a 45°C/sec cooling rate down to 520°C and coiled at 510°C to produce plate thickness 3.0 mm hot rolled coils. These were annealed by two-stage step type box annealing under the annealing conditions shown in Table 3 to prepare plate thickness 3.0 mm samples.

[0133] Table 3 shows the results of measurement or evaluation of the carbide size, ferrite grain size, Vickers hardness, ratio of the number of carbides at the ferrite grain boundaries to the number of carbides in the ferrite grains, pearlite area ratio, cold workability, and hardenability of the produced samples.

Table 3

	1st stage annealing		2nd stage			Carbide size [μm]	Ferrite grain size [μm]	Pearlite area ratio [%]	Vickers hardness [HV]	No. of grain boundary carbides/ No. of grain carbides	Total elongation [%]		Elongation anisotropy	Martensite fraction [%]	Remarks
	Holding temp. [°C]	Holding time [hr]	Holding temp. [°C]	Holding time [hr]	Cooling rate [°C/sec]						0° direction	90° direction	$\|EL_{0°}-EL_{90°}\|$		
A-2	748	25	787	38	18	1.23	26.3	7.5	176	1.9	29.2	30.4	1.2	93	Comp. steel
B-2	659	60	789	9	8	1.51	32.1	0.8	134	8.9	37.0	37.5	0.5	95	Inv. steel
C-2	657	39	755	38	17	0.96	21.6	1.2	117	5.3	40.0	40.9	0.9	69	Inv. steel
D-2	687	50	763	15	36	0.87	23.3	0.4	154	4.7	33.3	34.0	0.7	81	Comp. steel
E-2	667	48	766	44	17	1.10	21.1	0.3	135	5.4	36.8	38.0	1.2	91	Inv. steel
F-2	701	47	739	47	11	2.23	26.0	1.9	131	8.3	37.5	38.3	0.8	74	Inv. steel
G-2	692	26	734	27	13	0.92	12.6	0.0	158	7.8	32.5	33.5	1.0	98	Comp. steel
H-2	710	40	733	23	19	1.02	15.2	0.8	143	4.1	35.3	35.8	0.5	96	Inv. steel
1-2	677	49	783	8	9	1.33	31.7	1.9	117	7.1	40.1	41.2	1.1	62	Inv. steel
J-2	665	23	773	24	11	1.28	28.6	0.3	118	8.2	40.0	41.0	1.0	75	Inv. steel
K-2	654	22	785	20	21	1.01	30.6	0.8	120	4.7	39.6	40.3	0.7	83	Inv. steel
L-2	705	1	739	21	8	1.17	17.4	0.9	165	7.4	31.3	32.0	0.7	72	Comp. steel
M-2	658	9	776	50	27	1.01	26.7	1.9	110	3.8	41.4	42.4	1.0	97	Inv. steel
N-2	715	15	774	19	10	1.37	30.7	0.5	128	9.6	38.1	39.6	1.5	95	Inv. steel
0-2	680	46	760	13	18	0.92	15.5	8.3	175	7.5	29.4	29.9	0.5	35	Comp. steel
P-2	674	6	731	1	18	0.93	10.9	13.2	213	0.4	22.3	23.8	1.5	98	Comp. steel

(continued)

	1st stage annealing		2nd stage			Carbide size [μm]	Ferrite grain size [μm]	Pearlite area ratio [%]	Vickers hardness [HV]	No. of grain boundary carbides/ No. of grain carbides	Total elongation [%]		Elongation anisotropy	Martensite fraction [%]	Remarks
	Holding temp. [°C]	Holding time [hr]	Holding temp. [°C]	Holding time [hr]	Cooling rate [°C/sec]						0° direction	90° direction	$\lvert EL_{0°}-EL_{90°}\rvert$		
Q-2	673	15	786	44	13	1.16	35.4	0.5	1.62	3.1	31.8	32.3	0.5	7	Comp. steel
R-2	680	11	769	29	12	1.17	22.3	1.9	122	6.0	39.2	40.2	1.0	82	Inv. steel
S-2	692	45	749	9	9	1.30	20.7	1.9	150	7.8	39.4	40.5	1.1	96	Inv. steel
T-2	618	25	773	35	5	1.77	18.6	1.6	161	7.7	31.9	33.0	1.1	95	Comp. steel
U-2	705	22	779	27	18	1.07	27.4	0.2	103	3.3	38.1	43.8	5.7	70	Comp. steel
V-2	669	25	787	42	26	0.98	30.7	0.3	106	2.7	44.2	45.0	0.8	72	Inv. steel
W-2	677	89	772	42	28	0.93	29.5	0.7	102	3.2	43.0	44.2	1.2	67	Comp. steel
X-2	684	34	710	34	5	0.74	7.4	1.1	160	0.5	38.1	38.5	0.4	61	Comp. steel
Y-2	652	48	761	54	13	1.14	24.5	8.0	173	5.7	29.7	30.3	0.6	81	Comp. steel
Z-2	677	38	730	7	17	0.79	9.4	1.3	1.49	9.4	36.8	37.8	1.0	78	Inv. steel
AA-2	668	23	732	39	25	0.56	13.2	1.8	134	3.1	36.9	37.4	0.5	49	Comp. steel
AB-2	669	42	811	35	24	1.12	25.7	10.2	188	2.5	27.0	27.5	0.5	92	Comp. steel
AC-2	698	5	748	37	2	1.87	27.6	1.5	131	7.9	37.5	38.5	1.0	87	Inv. steel
AD-2	679	59	777	21	11	1.19	24.7	1.1	134	5.7	37.0	37.5	0.5	4	Comp. steel

[0134] As shown in Table 3, the Invention Steels B-2, C-2, E-2, F-2, H-2, 1-2, J-2, K-2, M-2, N-2, R-2, S-2, V-2, Z-2, and AC-2 all have a ratio of the number of carbides at the ferrite grain boundaries with respect to the number of carbides inside the ferrite grains of over 1 and a Vickers hardness of 170HV or less and are excellent in cold workability and hardenability.

[0135] As opposed to this, Comparative Steel G-1 is high in amount of C and deteriorates in cold workability. Comparative Steel O-1 is high in amount of Mo and amount of Cr and deteriorates in cold workability. Further, the carbides are high in stability, so at the time of hardening, the carbides will not dissolve, the amount of production of austenite is small, and the hardenability is inferior.

[0136] Comparative Steel Q-1 is high in amount of Si and high in hardness of ferrite, so deteriorates in workability. Further, it is high in the A3 point, so the amount of production of austenite at the time of hardening is small and the hardenability is inferior. Comparative Steel AD-1 is high in amount of Al and high in A3 point, so the amount of production of austenite at the time of hardening is small and the hardenability is inferior. Comparative Steel U-1 is high in amount of S and is formed with coarse MnS in the steel, so deteriorates in cold workability. Comparative Steel AA-1 is low in amount of Mn and inferior in hardenability.

[0137] Comparative Steel T-2 is low in holding temperature at the time of the first stage annealing of the two-stage step type box annealing, is insufficient in coarsening treatment of carbides at the Ac1 temperature or less, and is insufficient in thermal stability of the carbides, so is reduced in carbides remaining at the time of second stage annealing, cannot be suppressed in pearlite transformation in the structure after gradual cooling, and deteriorates in cold workability.

[0138] Comparative Steel A-2 has a high holding temperature at the time of the first stage annealing of the two-stage step type box annealing, is formed with austenite during the annealing, cannot be raised in stability of carbides, is decreased in carbides remaining at the time of second stage annealing, cannot be suppressed in pearlite transformation in the structure after gradual cooling, and deteriorates in cold workability.

[0139] Comparative Steel L-2 is short in holding time at the time of the first stage annealing of the two-stage step type annealing, is insufficient in the coarsening treatment of the carbides at the Ac1 temperature or less, and is insufficient in the thermal stability of the carbides, so is decreased in the carbides remaining at the time of the second stage annealing, cannot suppress pearlite transformation in the structure after gradual cooling, and deteriorates in cold workability.

[0140] Comparative Steel W-2 is long in holding time at the time of the first stage annealing of the two-stage step type annealing and deteriorates in productivity. Comparative Steel X-2 is low in holding temperature at the time of the second stage annealing at the time of two-stage step annealing, is small in amount of production of austenite, cannot be increased in number ratio of carbides at the grain boundaries, and deteriorates in cold workability.

[0141] Comparative Steel AB-2 is high in holding temperature at the time of second stage annealing in the two-stage step type box annealing and is promoted in dissolution of the carbides, so decreases the residual carbides, cannot suppress pearlite transformation in the structure after gradual cooling, and deteriorates in cold forgeability.

[0142] Comparative Steel P-2 is low in holding temperature at the time of second stage annealing of the two-stage step type box annealing, has little formation of austenite, cannot be increased in the number ratio of carbides at the ferrite grain boundaries, and deteriorates in cold workability Comparative Steel Y-2 is long in holding time at the time of second stage annealing of the two-stage step type box annealing and is promoted in dissolution of carbides, so is decreased in remaining carbides, cannot suppress pearlite transformation in the structure after gradual cooling, and deteriorates in cold forgeability.

[0143] Comparative Steel D-2 is large in cooling rate from the end of the second stage annealing of the two-stage step type box annealing down to 650°C, experiences pearlite transformation at the time of cooling, and deteriorates in cold workability.

Industrial Applicability

[0144] As explained above, according to the present invention, it is possible to produce and provide steel plate excellent in formability and wear resistance. The steel plate of the present invention is steel plate suitable as a material for auto parts, edged tools, and other machine parts produced through stamping, bending, press-forming, and other working processes, so the present invention is high in industrial applicability.

Claims

1. A steel plate comprising, by mass%,

C: 0.10 to 0.40%,
Si: 0.01 to 0.30%,
Mn: 1.00 to 2.00%,

P: 0.020% or less,
S: 0.010% or less,
Al: 0.001 to 0.10%,
N: 0.010% or less,
O: 0.020% or less,
Cr: 0.50% or less,
Mo: 0.10% or less,
Nb: 0.10% or less,
V: 0.10% or less,
Cu: 0.10% or less,
W: 0.10% or less,
Ta: 0.10% or less,
Ni: 0.10% or less,
Sn: 0.050% or less,
Sb: 0.050% or less,
As: 0.050% or less,
Mg: 0.050% or less,
Ca: 0.050% or less,
Y: 0.050% or less,
Zr: 0.050% or less,
La: 0.050% or less,
Ce: 0.050% or less and
a balance of Fe and unavoidable impurities,
wherein a metal structure of the steel plate
has a ratio of a number of carbides at ferrite grain boundaries with respect to a number of carbides in ferrite grains of over 1,
has a ferrite grain size of 5 μm to 50 μm, and
has an area ratio of pearlite of 6% or less; and
a Vickers hardness of the steel plate is 100HV to 170HV.

2. The steel plate according to claim 1 containing one or both of

Ti: 0.10% or less and
B: 0.010% or less
instead of part of the Fe.

3. A method of production of the steel plate according to claim 1 or 2, the method of production of steel plate comprising hot rolling a steel slab of a chemical composition according to claim 1 or 2 during which completing finish hot rolling at a 750°C to 850°C temperature region to obtain hot rolled steel plate,
coiling the hot rolled steel plate at 400°C to 550°C,
pickling the coiled up hot rolled steel plate, holding the pickled hot rolled steel plate at a 650°C to 720°C temperature region for 3 hours to 60 hours as first stage annealing, then
holding the hot rolled steel plate at a 725°C to 790°C temperature region for 3 hours to 50 hours as second stage annealing, and
cooling the annealed hot rolled steel plate to 650°C by a cooling rate of 1°C/hour to 30°C/hour.

INTERNATIONAL SEARCH REPORT

International application No.
PCT/JP2016/068169

A. CLASSIFICATION OF SUBJECT MATTER
C22C38/00(2006.01)i, *C21D9/46*(2006.01)i, *C22C38/06*(2006.01)i, *C22C38/60*(2006.01)i

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

B. FIELDS SEARCHED

Minimum documentation searched (classification system followed by classification symbols)
C22C38/00-38/60, C21D9/46

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

Jitsuyo Shinan Koho	1922-1996	Jitsuyo Shinan Toroku Koho	1996-2016
Kokai Jitsuyo Shinan Koho	1971-2016	Toroku Jitsuyo Shinan Koho	1994-2016

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

C. DOCUMENTS CONSIDERED TO BE RELEVANT

Category*	Citation of document, with indication, where appropriate, of the relevant passages	Relevant to claim No.
A	JP 2007-270330 A (JFE Steel Corp.), 18 October 2007 (18.10.2007), claims; paragraphs [0036] to [0060]; examples; tables 1 to 4 (Family: none)	1-3
A	JP 11-80884 A (Nisshin Steel Co., Ltd.), 26 March 1999 (26.03.1999), claims; paragraphs [0016] to [0020], [0028] to [0029]; examples; tables 1, 2 (Family: none)	1-3
A	JP 11-269552 A (Nisshin Steel Co., Ltd.), 05 October 1999 (05.10.1999), claims; paragraphs [0023] to [0032], [0036] to [0042]; examples; tables 1 to 3 (Family: none)	1-3

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

* Special categories of cited documents:
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Date of the actual completion of the international search 12 July 2016 (12.07.16)	Date of mailing of the international search report 26 July 2016 (26.07.16)
Name and mailing address of the ISA/ Japan Patent Office 3-4-3,Kasumigaseki,Chiyoda-ku, Tokyo 100-8915,Japan	Authorized officer Telephone No.

INTERNATIONAL SEARCH REPORT

International application No.
PCT/JP2016/068169

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 2012-62496 A (Nippon Steel Corp.), 29 March 2012 (29.03.2012), claims; paragraphs [0016] to [0040], [0048] to [0052]; examples; tables 1, 2 (Family: none)	1-3
A	JP 11-61272 A (Sumitomo Metal Industries, Ltd.), 05 March 1999 (05.03.1999), claims; paragraphs [0018] to [0028]; examples; tables 1, 2 (Family: none)	1-3
A	WO 2007/088985 A1 (JFE Steel Corp.), 09 August 2007 (09.08.2007), claims; page 11, lines 9 to 24; page 13, lines 12 to 22 & US 2009/0173415 A1 claims; paragraphs [0065], [0076] to [0077] & EP 1980635 A1 & CN 101379208 A & KR 10-2008-0077254 A & JP 2007-231416 A	1-3
A	WO 2007/116599 A1 (JFE Steel Corp.), 18 October 2007 (18.10.2007), claims & US 2009/0308504 A1 claims & EP 2003220 A1 & CN 101379207 A & KR 10-2008-0081955 A & JP 2007-270325 A	1-3
P,A	JP 2015-117406 A (Nippon Steel & Sumitomo Metal Corp.), 25 June 2015 (25.06.2015), claims; examples; tables 1-1 to 6-2 (Family: none)	1-3

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- JP 3879459 B **[0006]**