

A STEEL PLATE SUPERIOR IN FATIGUE CRACK PROPAGATION
RESISTANCE AND A METHOD OF PRODUCTION OF THE SAME

5 BACKGROUND OF THE INVENTION

1. Field of the Invention

The present invention relates to a steel plate superior in fatigue crack propagation resistance used for welded structural members of buildings, ships, bridges, construction machinery, offshore structures, etc. where
10 fatigue resistance is required and a method of production of the same.

2. Description of the Related Art

In general, buildings, ships, bridges, construction machinery, offshore structures, and other
15 welded structures are made using welded joints using diverse types of welding methods such as arc welding, plasma welding, laser welding, electron beam welding, etc.

These welded joints are subject to repeated load due to the wind, waves, mechanical vibration, etc., so improvement of the fatigue strength is extremely important. In general, as techniques for improving the fatigue strength, treatment after welding such as
20 grinding, TIG dressing, shot peening, and hammer peening are used. However, these have the following problems. Among these, grinding and TIG dressing improve the shape of the weld beads, but both are remarkably poor in work efficiency. Shot peening and hammer peening have the
25 effect of improving the fatigue strength, but shot peening requires massive machinery and requires various utilities.

Further, hammer peening has a large reaction force, so the results of the treatment are not stable and
35 sometimes conversely the press formability or fatigue strength end up dropping. Further, hammer peening gives too great a plastic deformation, so there was the defect

that use for thin plates was difficult. Further, grinding or hammer peening had the problem of applying machining of a low frequency of several Hz to the joint, so resulted in roughness at the machined surface,

5 concentration of stress at the concave parts, and, upon the joint receiving repeated load, cracks occurring from the stress concentration parts and therefore a drop in the fatigue strength of the joint as a whole.

Further, a weld zone generally suffers from
10 residual stress due to the heat input from the welding. This residual stress becomes a major factor in lowering the fatigue strength at the weld zone. Therefore, as another means for improving the fatigue strength, the method is known of causing the generation of compressive
15 residual stress at the welded joint or reducing the tensile residual stress occurring at the welded joint so as to improve the fatigue strength. For example, it is possible to impart compressive residual stress by shot peening the vicinity of the weld ending. This shot
20 peening is a technique imparting compressive residual stress by shooting steel balls of under 1 mm diameter at the part serving as the starting point of occurrence of fatigue cracks. Further, it is also known that heating the weld metal to remelt enables the shape of the welding
25 end to be improved or the tensile residual stress to be reduced. However, this shot peening requires steel shot. The disposal and cost of the steel shot become problems in some cases. Further, there is the problem that the amount of improvement of the fatigue strength becomes
30 uneven.

In this way, it is difficult to employ the technology for improvement of the fatigue strength by post-welding treatment to welded joints. Even if it can be employed, the amount of improvement of the fatigue
35 strength remains a low level. Therefore, technology not requiring post-welding treatment and enabling improvement of the fatigue strength of the welded joint as welded is

earnestly desired.

From this viewpoint, several steel plates suppressing the propagation of fatigue cracks so as to improve the fatigue strength of a welded joint as welded have been proposed. For example, Japanese Unexamined Patent Publication (Kokai) No. 06-271985 discloses that a general shipbuilding steel formed with a superfine grain structure at the surface by processing the ferrite during welding, known as SUF steel has the effect of lowering the speed of propagation of fatigue cracks. However, remarkably lowering the propagation speed by just making the ferrite finer is difficult. Further, the superfine grain structure formed on the surface ends up being lost in large part due to the heat affect during welding. According, sufficient improvement of the fatigue strength of a welded joint cannot be achieved.

Further, Japanese Unexamined Patent Publication (Kokai) No. 06-271985, Japanese Unexamined Patent Publication (Kokai) No. 07-090478, Japanese Unexamined Patent Publication (Kokai) No. 08-073980, Japanese Unexamined Patent Publication (Kokai) No. 10-168542, Japanese Unexamined Patent Publication (Kokai) No. 11-001742, Japanese Unexamined Patent Publication (Kokai) No. 2002-047531, and Japanese Unexamined Patent Publication (Kokai) No. 2003-003229 disclose a steel plate comprising a mixed structure of hard pearlite, bainite, or martensite as a second phase in a base phase of soft ferrite wherein the hard second phase inhibits the progression of cracks and thereby lowers the speed of propagation of fatigue cracks. However, these technologies do not enable suitable control of factors important for delaying crack progression such as the area fraction, the aspect ratio (long axis/short axis), and the hardness of the martensite, the hardness of the ferrite, and the distance between the ferrite and martensite, so the fatigue crack propagation resistance is not improved at all, the improvement is insufficient,

and the toughness of the steel becomes remarkably degraded.

For example, in Japanese Unexamined Patent Publication (Kokai) No. 06-271985, the martensite fraction is insufficient and a sufficient improvement in the fatigue crack propagation resistance cannot be obtained. In Japanese Unexamined Patent Publication (Kokai) No. 07-090478, when the martensite ratio exceeds 30%, a serious drop in toughness occurs and even if a hardness of the hard second phase of 30% or more of the ferrite is secured, if the hardness of the ferrite is not more than 150 or the hardness of the hard second phase is not more than 400, a sufficient effect of improvement of the fatigue crack propagation resistance cannot be obtained. In Japanese Unexamined Patent Publication (Kokai) No. 08-073980 as well, the martensite fraction is over 30% and the toughness of the steel ends up being seriously impaired. Japanese Unexamined Patent Publication (Kokai) No. 10-168542, Japanese Unexamined Patent Publication (Kokai) No. 11-001742, Japanese Unexamined Patent Publication (Kokai) No. 2002-047531, and Japanese Unexamined Patent Publication (Kokai) No. 2003-003229 do not suitably control the hardness and fraction of the ferrite and second phase and the distance between the same. When the second phase is bainite with a low hardness of 400 or less, even if the fraction is high, deterioration of the toughness is suppressed, but the effect of suppression of propagation is small. Further, when the second phase is martensite with a high hardness of 400 or more, serious deterioration of the toughness occurs with a fraction of 30% or more.

Further, Japanese Unexamined Patent Publication (Kokai) No. 08-225882 discloses a steel plate comprised of a two phase structure of ferrite and bainite wherein the fraction of the ferrite phase part, the hardness of the ferrite, the number of the phase boundaries of the ferrite and bainite, etc. are defined to specific ranges

so as to reduce the speed of progression of fatigue cracks. However, with the hardness level of bainite, the effect of improvement of the fatigue crack propagation resistance is insufficient. Even if the hardness of the ferrite is 150 or less, the effect is small.

Further, Japanese Unexamined Patent Publication (Kokai) No. 07-24992, Japanese Unexamined Patent Publication (Kokai) No. 08-199286, and Japanese Unexamined Patent Publication (Kokai) No. 09-095754 disclose a steel plate where, unlike the above concepts, a hard phase makes the matrix phase and a soft phase is dispersed in it as a second phase so as to reduce the fatigue crack propagation speed. These have the plastic deformation energy required for crack progression absorbed by the soft phase so as to promote crack closing behavior and suppress crack progression, but in welded joints with welding tensile residual stress, cracks tends to easily open, so with just the crack closing effect, a sufficient effect of improvement of the fatigue crack propagation resistance is not obtained.

Further, Japanese Unexamined Patent Publication (Kokai) No. 08-199286 and Japanese Unexamined Patent Publication (Kokai) No. 09-095754 disclose a steel plate securing a certain fraction of recovered or a recrystallized ferrite and promoting a specific texture structure so as to lower the speed of fatigue crack propagation. These aim to suppress the plastic deformation at the cracktip at the time of crack progression by specific textures. However, with just the texture of ferrite without the second phase structure defined, a sufficient fatigue crack propagation resistance cannot be obtained and, further, the plastic deformation at the cracktips can only be suppressed in the extremely low ΔK region, so the application is remarkably limited.

In this way, in the related art, suitable control of the structure for remarkably suppressing crack

progress is not possible. Development of steel plate able to stably reduce the fatigue crack propagation speed and steel plate able to contribute further to improvement of the fatigue life of welded joints is earnestly desired.

5 (See also 1998 Society of Materials Science Japan 24th Fatigue Symposium Papers, "Fatigue Properties of Surface Superfine Grain Steel Plate", p. 157 to 162.)

SUMMARY OF THE INVENTION

10 The object of the present invention is to solve the above problems of the related art and provide steel plate superior in fatigue crack propagation resistance used for welded structural members of buildings, ships, bridges, construction machinery, offshore structures, etc. and a method of production of the same. Specifically, the
15 present invention provides a steel plate having a fatigue crack propagation speed da/dN when a stress expansion coefficient range ΔK of a stress ratio of 0.1 is $20 \text{ MPa}\sqrt{\text{m}}$ of not more than 10^{-8} m/cycle and having a fatigue life at the time of a welded joint axial force fatigue test with
20 a stress ratio of 0.1 and a heat input of 10 to 30 kJ/ml of at least two times that of conventional steel and a method of production of the same.

The present invention was made as a result of intensive study for solving the above problems and has as
25 its gist the following:

(1) A steel plate superior in fatigue crack propagation resistance containing, by wt%,

C: 0.03 to 0.2%,

Si: 0.01 to 1.6%,

30 Mn: 0.5 to 2%,

P: 0.02% or less,

S: 0.02% or less,

Al: 0.001 to 0.1%, and

N: 0.001 to 0.008%

35 and a balance of Fe and unavoidable impurities, wherein a microstructure of the base material is a lamellar structure having ferrite with a Vickers hardness

of at least 150 as a matrix phase and having elongated martensite with a Vickers hardness of 400 to 900, an area fraction of 5 to 30%, and an aspect ratio (long axis/short axis) of 3 or more as a second phase, an average interlayer distance between the ferrite and martensite in the plate thickness direction is 3 to 50 μm , and a fatigue crack propagation speed da/dN when a stress intensity factor range ΔK of a stress ratio 0.1 is 20 $\text{MPa}\sqrt{\text{m}}$ is not more than 10^{-8} m/cycle.

(2) A steel plate superior in fatigue crack propagation resistance as set forth in (1) further containing, by wt%, one or more of

Cu: 0.1 to 2.5%,

Ni: 0.1 to 5%,

Cr: 0.01 to 1.5%,

Mo: 0.01 to 1.5%,

W: 0.01 to 1.5%,

Ti: 0.001 to 0.05%,

Nb: 0.005 to 0.2%,

Zr: 0.005 to 0.2%,

V: 0.005 to 0.2%, and

B: 0.0002 to 0.005%.

(3) A steel plate superior in fatigue crack propagation resistance as set forth in (1) or (2), further containing, by wt%, one or more of

Mg: 0.0005 to 0.01%,

Ca: 0.0005 to 0.01%, and

REM: 0.005 to 0.05%.

(4) A method of production of a steel plate superior in fatigue crack propagation resistance wherein a microstructure of a base material is a lamellar structure having ferrite with a Vickers hardness of at least 150 as a matrix phase and having elongated martensite with a Vickers hardness of 400 to 900, an area fraction of 5 to 30%, and an aspect ratio (long axis/short axis) of 3 or more as a second phase, an

average interlayer distance between the ferrite and martensite in the plate thickness direction is 3 to 50 μm , and a fatigue crack propagation speed da/dN when a stress intensity factor range ΔK of a stress ratio of 0.1 is 20 $\text{MPa}\sqrt{\text{m}}$ is not more than 10^{-8} m/cycle, comprising heating a slab containing the ingredients described in any one of (1) to (3) to a temperature of at least an A_{c3} transformation temperature to not more than 1350°C , then rolling by a cumulative reduction ratio of 10 to 80% in an austenite single phase region of the A_{r3} transformation temperature to 1250°C , then finishing rolling by a cumulative reduction ratio of 40 to 90% in a two phase region of austenite-ferrite of a rolling start temperature of not more than the A_{r3} transformation temperature and a rolling end temperature of not less than 600°C .

(5) A method of production of a steel plate superior in fatigue crack propagation resistance as set forth in (4), further comprising accelerated cooling by a cooling rate of 5 to 80°C/s to 20 to 400°C after said finishing rolling.

(6) A method of production of a steel plate superior in fatigue crack propagation resistance as set forth in (4) or (5), further comprising tempering in a temperature range of 300 to 500°C .

According to the present invention, it is possible to provide a steel plate superior in fatigue crack propagation resistance for use for welded structural members of buildings, ships, bridges, construction machinery, and offshore structures and a method for production of the same. Specifically, it exhibits the industrially useful remarkable effect of enabling a fatigue crack propagation speed da/dN when a stress intensity factor range ΔK is 20 $\text{MPa}\sqrt{\text{m}}$ of not more than 10^{-8} m/cycle, an improvement of the welded joint fatigue life of at least two times that of the past, and an

improvement of the reliability of the welded steel structure with respect to fatigue fracture.

These and other objects and features of the present invention will become clearer from the following description of the preferred embodiments given with
5 reference to the attached drawings.

BRIEF DESCRIPTION OF THE DRAWINGS

FIG. 1 is a view of a test piece used for the fatigue crack propagation test, and

10 FIG. 2 is a view of a test piece used for a welded joint fatigue test.

DESCRIPTION OF THE PREFERRED EMBODIMENTS

Preferred embodiments of the present invention will be described in detail below while referring to the
15 attached figures.

In general, it is known that the fatigue crack propagation speed does not depend on the microstructure or strength of steel. The inventors however engaged in intensive studies on this and as a result discovered that
20 by making ferrite the matrix phase, dispersing martensite in it as a second phase in layers, and further suitably controlling the hardness, area fraction, aspect ratio (long axis/short axis), and distance between layers in the thickness direction of the phases, the fatigue crack
25 propagation speed remarkably drops compared with the past. The mechanism behind the drop in the fatigue crack propagation speed is based on the change in the internal stress around the martensite which occur at the time of the martensite transformation during rolling and cooling
30 of the steel plate. This is effective for lowering the driving force behind crack progression. Due to this, a crack slows when it encounters martensite, cannot easily proceed in the martensite, and therefore detours or splits along the perimeter of the martensite. This delay
35 causes the slowing of the crack and the increase in the propagation distance due to the detouring and splitting of the crack and further the remarkably crack closing

behavior accompanying the crack detouring and splitting enable a large drop in the fatigue crack propagation speed.

5 The reasons for the range of limitation of the microstructure will be explained next. The factor having the greatest effect among the factors affecting the fatigue crack propagation speed is the area fraction of the martensite. The propagation speed drops sharply by at least 5%. This is caused by the increase in obstacles to
10 crack progression due to the increase in the martensite fraction. However, if over 30%, the toughness is seriously degraded, so the range was made 5 to 30%.

 To increase the internal stress and effectively lower the driving force behind crack progression, it is
15 necessary to lower the martensitic transformation start temperature. This is because if martensitic transformation occurs at a low temperature, the ferrite constraining the transformation is hard, so the internal stress increases due to its reaction force. The
20 martensitic transformation start temperature falls the greater the amount of concentration of the carbon in the austenite at the time of hot rolling. Further, the greater the amount of concentration of carbon, the greater the hardness of the martensite, so to make the
25 martensitic transformation start temperature 400°C or less, the hardness of the martensite has to be made at least 400. The reason why the martensitic transformation start temperature has to be made 400°C or less is that the internal stress is eased by the thermal shrinkage after
30 transformation if over 400°C and therefore the effect of delay of fatigue crack propagation becomes smaller. Further, if the hardness of the martensite is over 900, it is difficult to secure a martensite fraction of 5% or more and the martensite acts as an initiation point for
35 brittle fracture, so the hardness of the martensite should be controlled at 400 to 900Hv.

Further, as explained above, the harder the ferrite, the greater the constraint at the time of martensitic transformation, the greater the reaction force, and the higher the internal stress, so the hardness of the ferrite should be controlled at 150 Hv or more.

The greater the aspect ratio of the martensite, the greater the probability of encountering the martensite, which obstructs crack progression. Further, the detour and splitting distance increase, so this is effective for reducing the fatigue crack propagation speed. If the aspect ratio is smaller than 3, even if cracks meet the martensite, the detour and branching distance will be smaller, so the effect of improvement of the crack propagation resistance will be small. Therefore, the aspect ratio (long axis/short axis) of the martensite was made at least 3.

The ferrite phase and martensite phase have to be dispersed in layers. If the distance between layers is smaller than 3 μm , the internal stress introduced at the time of martensitic transformation will no longer act effectively and delay of the crack progression will become difficult. Further, if the interlayer distance is over 50 μm , the probability of the cracks encountering the martensite, that is, the effect of slowing, detouring, and branching the cracks will become smaller, so the range of the distance between layers should be controlled 3 to 50 μm .

Next, the reasons for limiting the ranges of the alloying elements will be explained. Note that below, % means wt%. C is one of the main elements of the composition of the present invention. It is included as an effective ingredient for controlling the fraction of the martensite and improving the strength of the steel. If less than 0.03%, it becomes difficult to secure a martensite fraction of 5% or more. If over 0.2%, the toughness and weld crack resistance of the base material

and the weld zone drop, so the range was limited to 0.03 to 0.2%.

Si is an element required as a deoxidation agent in addition to securing the strength. To obtain this effect, addition in an amount of at least 0.01% is required. Excess content of over 1.6% forms coarse oxides and invites a drop in ductility and toughness, so the amount was limited to 0.01 to 1.6%.

Mn is an essential element for increasing the strength, but if less than 0.5%, the strength of the base material cannot be secured. On the other hand, excess content of over 2% degrades the toughness of the base material, the toughness of the weld zone, the weld cracking, etc. due to grain boundary brittleness etc., so the amount was limited to 0.5 to 2%.

P is an element affecting the toughness of the steel. If over 0.02%, it seriously inhibits the toughness of not only the base material, but also the HAZ, so it is preferably as low as possible and the upper limit was made 0.02%.

S is preferably as low as possible like P. If over 0.02%, the precipitation of MnS becomes serious. This obstructs the HAZ toughness of the base material and lowers the ductility in the plate thickness direction, so the upper limit was made 0.02%.

Al is an element effective for deoxidation and refining austenite grain size. To obtain this effect, it must be contained in an amount of at least 0.001%. On the other hand, if contained in excess over 0.1%, it forms coarse oxides and seriously degrades the ductility, so the amount was made 0.001 to 0.1%.

N bonds with Al or Ti to act effectively to reduce the size of the austenite grains, so a small amount of N contributes to an improvement of the mechanical properties. Further, complete removal of the N from the steel is industrially impossible. Reduction more than necessary is also not preferable in that it places an

excess load on the production process. Therefore, as the range enabling control industrially and of an allowable load on the production process, the lower limit is made 0.001%. If contained in excess, the increased N is left
5 dissolved and the strain ageing property deteriorates, so the upper limit was made 0.008%.

The above summarizes the reasons for limitation of the basic ingredients of the present invention. In the present invention, however, for adjusting the strength
10 and toughness, it is also possible to include one or more of Cu, Ni, Cr, Mo, W, Ti, Nb, Zr, V, or B in accordance with need. The reasons for limitation of these elements are explained next.

Cu is an element effective for increasing the
15 strength without sacrificing the toughness. If less than 0.1%, there is no such effect, while if over 2.5%, cracks tends to occur at the time of heating the slab or welding. Therefore, the amount was limited to at 0.1 to 2.5%.

20 Ni is an element effective for improving the toughness and strength. To obtain this effect, it is necessary to add at least 0.1%, but excess addition over 5% results in saturation of the effect and the deterioration of the HAZ toughness or weldability.
25 Further, this is an expensive element, so considering economy as well, the amount was limited to at 0.1 to 5%.

Cr is preferred to be at least 0.01% in order to increase the quench hardenability and secure strength. On the other hand, an amount over 1.5% is not preferable for
30 the same reason as Ni. Therefore, the amount was limited to at 0.01 to 1.5%.

Mo is an element effective for improving the quench hardenability, improving the strength, improving the resistance to temper brittleness, and suppressing
35 recrystallization. To obtain these effects, addition of at least 0.01% is required, but if over 1.5%, the toughness and the weldability deteriorate. Therefore, the

amount was limited to at 0.01 to 1.5%.

W is an element necessary for increasing the quench hardenability and securing the strength, but the amount was made 0.01 to 1.5% as the range in which it can exhibit this effect and does not have a detrimental effect on other properties.

Ti is an element contributing to improvement of the strength of the base material through precipitation hardening and effective for refining the heated austenite grain size by formation of TiN stable even at a high temperature. To exhibit this effect, it must be included in an amount of at least 0.001%. On the other hand, if over 0.05%, coarse oxides are formed and the ductility is degraded seriously, so the amount was limited to at 0.001 to 0.05%.

Nb, Zr, and V contribute to the improvement of the strength of the base material through precipitation hardening, but if less than 0.005%, there is no such effect, while if added in excess over 0.2%, the ductility and toughness deteriorate. Therefore, the amounts of Nb, Zr, and V were limited to at 0.005 to 0.2% respectively.

B is an element which segregates at the austenite grain boundaries in the solid solution state and thereby enables the quench hardenability to be improved even in small amounts, but is also effective for suppression of recrystallization of the austenite in the state segregated at the grain boundaries. To exhibit the effects of improvement of the quench hardenability and suppression of recrystallization, addition of at least 0.0002% is required, but excess addition over 0.005% produces coarse precipitates and degrades the toughness. Therefore, the amount was limited to at 0.0002 to 0.005%.

Further, in the present invention, for improvement of the ductility and improvement of the joint toughness, it is also possible to add at one or more types of Mg, Ca, and REM in accordance with need. Mg, Ca, and REM are all effective for suppressing flattening of oxides during

hot rolling. They also effectively work to make the oxides finer and improve the joint toughness. To obtain these effects, the lower limit of content was 0.0005% for Mg, 0.0005% for Ca, and 0.005% for an REM. On the other hand, if contained in excess, the sulfides and oxides become coarser and a deterioration of the ductility and toughness is invited, so the upper limit of content was made 0.01% for Mg, 0.01% for Ca, and 0.05% for REM.

The above completes the explanation of the reasons for limitation of the microstructure and chemical composition of the basic requirements of the present invention. In addition, a suitable method of production for satisfying the microstructure requirements of the present invention will be proposed. However, the microstructure of the present invention exhibits its effect without regard as to the means for achieving it. The method of production of steel plate superior in fatigue crack propagation resistance described in claims 1 to 3 of the present invention are not limited to the methods shown in claims 4 to 6.

Before hot rolling, the steel slab has to be made 100% austenite. For this, the steel bloom has to be heated to a temperature of at least the Ac_3 transformation temperature. However, if the slab is heated over $1350^{\circ}C$, the austenite grains remarkably increase in size, so fine ferrite can no longer be obtained after rolling. Therefore, the upper limit of the heating temperature was made $1350^{\circ}C$.

The hot rolling after this was limited to the temperature region of the Ar_3 transformation point to $1250^{\circ}C$ because rolling in the austenite region enables the transformation temperature to be raised and the transformed structure to be made finer, so fine ferrite can be obtained in the two phase rolling. With a cumulative reduction ratio of less than 10%, this effect is small. Further, if over 80%, the reduction ratio in

the following two phase rolling can no longer be secured. Therefore, the upper limit was made 80%. In this case, it is preferable to control the rolling in the austenite region and make the austenite grains further finer before the two phase rolling.

In the present invention, it is necessary to cause the elongated, hard martensite to disperse in layers in the hard ferrite. Therefore, finishing rolling at the A_{r3} transformation temperature or less plays an extremely important role and is an essential step in the present invention. To improve the hardness of the ferrite, improve the hardness of the martensite, elongated martensite, and lower the transformation start temperature, finishing rolling at the A_{r3} transformation temperature or less is required. The lower the rolling temperature, the better, but the lower the temperature, the higher the deformation resistance, so the rolling load rises and rolling becomes difficult. Further, if less than 600°C, a fraction of martensite of 5% or more can no longer be secured. Therefore, the rolling end temperature was made at least 600°C.

With a cumulative reduction ratio of the finishing rolling of less than 40%, the hardness of the ferrite rises, the hardness of the martensite rises, the effect of elongation is small, and the interlayer distance between the ferrite and the martensite in the thickness direction ends up increasing, so the larger the cumulative reduction ratio, the better. Therefore, the cumulative reduction ratio of the finishing rolling was made 40 to 90%.

As the cooling method after the two phase rolling, to ensure the martensitic transformation, down to below the martensitic transformation start temperature, accelerated cooling by a cooling rate of 5 to 80°C/s down to 20 to 400°C is necessary. The cooling rate in the case of accelerated cooling was limited to 5 to 80°C/s because

if less than 5°C/s, martensitic transformation is difficult and no improvement in the strength or toughness of the base material can be expected, while if over 80°C/s, a large difference occurs in the structure and properties between the surface and interior.

Further, the accelerated cooling is stopped at 20 to 400°C in accordance with the desired strength and toughness level of the steel plate. Making the end temperature of the accelerated cooling less than 20°C has no effect at all in controlling the material properties and only invites a rise in manufacturing cost, so is meaningless. Conversely, if ending the accelerated cooling at over 400°C, martensitic transformation becomes difficult, the internal stress is eased, and no improvement in the fatigue crack propagation resistance can be expected.

The tempering performed in accordance with need after the rolling and cooling aims at improving the toughness of the base material structure through recovery, so the heating temperature has to be less than the temperature region A_{c1} where reverse transformation does not occur. Further, if over 500°C, the internal stress is eased, so the fatigue crack propagation resistance deteriorates, so the upper limit was made 500°C. In addition, recovery causes a reduction in the lattice defect density through disappearance and merger of the dislocations. To realize this, it is necessary to heat to 300°C or more, so the lower limit was made 300°C. Note that the tempered martensite produced by this tempering heat treatment is also defined as the martensite of the microstructure requirement of the present invention.

Examples

Below, the effects of the present invention will be explained in further detail by examples. The chemical compositions of the test steels used in the examples are

shown in Table 1. The individual test steels were cast into ingots and rolled to or continuously cast into slabs. Steel Nos. 1 to 20 of Table 20 satisfy the range of chemical composition of the present invention, while
5 Steel Nos. 21 to 25 do not satisfy the range of chemical composition of the present invention.

Slabs of the chemical compositions of Table 1 were rolled into steel plate by the conditions shown in Table 2. Test Steel Nos. A1 to A23 were produced by methods
10 relating to claims 4 to 6. Further, Test Steel Nos. B1 to B12 did not satisfy the production conditions of the present invention. The mechanical properties at room temperature are shown together in Table 2. Table 3 shows the results of examination of the microstructures and
15 results of fatigue tests of the steel plate comprised of the Steel Nos. 1 to 25 and Test Steel Nos. A1 to A23 and B1 to B12.

Table 2

Class	Test steel no.	Steel no.	Slab thick. (mm)	Heat. temp. (°C)	Rough rolling conditions		Finishing rolling conditions		Finished plate thick. (mm)	Cool. start temp. (°C)	Cooling rate (°C/sec)	Cooling end temp. (°C)	Mechanical properties			
					Start temp. (°C)	Cum. red. (%)	Start temp. (°C)	Ar ₃ trans. temp. (°C) *1					Yield stress (MPa)	Tens. strength (MPa)	Elong. (%)	vE-5°C (J)
P r e s e n t i n v e n t i o n	A1	1	240	1150	1050	75	700	771	67	710	25	<100	480	580	32	186
	A2	2	240	1150	1100	53	720	766	78	700	20	300	(*)	585	31	180
	A3	3	200	1200	1150	60	690	752	75	690	25	<100	-	565	30	168
	A4	4	200	1100	1050	56	725	742	71	720	20	<100	-	620	29	158
	A5	5	240	1150	1100	75	700	780	75	680	25	<100	-	630	28	145
	A6	6	240	1200	1150	58	730	754	80	715	20	<100	-	618	28	135
	A7	7	200	1150	1000	65	740	770	71	720	20	<100	-	605	27	128
	A8	8	240	1200	1150	50	700	772	67	680	10	<100	-	654	26	110
	A9	9	240	1150	1100	67	725	782	75	705	25	<100	-	660	25	122
	A10	10	240	1100	1000	58	730	784	75	710	25	<100	-	685	24	105
	A11	11	200	1200	1150	50	740	773	75	720	10	<100	-	673	25	102
	A12	12	240	1230	1170	56	735	783	67	710	20	250	-	625	28	149
	A13	13	240	1200	1150	50	720	777	75	710	25	<100	-	630	29	148
	A14	14	240	1150	1100	58	750	792	75	720	20	<100	350	590	32	175
	A15	15	200	1240	1190	70	700	766	75	705	15	<100	-	513	28	104
	A16	16	200	1200	1150	60	735	793	76	718	30	<100	-	625	28	116
	A17	17	240	1100	1050	58	730	787	78	715	25	<100	-	620	29	110
	A18	18	240	1150	1100	58	690	732	75	702	20	<100	-	495	30	165
	A19	19	200	1100	1050	68	720	780	75	710	15	<100	450	438	32	185
	A20	20	200	1100	1050	63	700	757	76	705	20	350	(*)	505	28	118
	A21	21	200	1200	1150	50	680	750	75	690	25	<100	-	385	36	205
	A22	22	240	1050	1000	48	740	752	80	720	20	<100	-	586	12	10
	A23	24	200	1050	1100	60	750	798	75	730	15	<100	-	343	34	199
C o m p o s i t i o n	B1	11	200	1150	1050	-	905	773	-	880	20	-	-	356	38	185
	B2	12	200	1100	1000	40	880	783	75	640	20	<100	-	425	29	163
	B3	13	200	1050	1100	81	715	777	33	700	25	<100	-	388	32	175
	B4	14	240	1150	1000	58	725	792	75	710	25	<100	600	512	33	180
	B5	15	200	1050	1100	63	780	766	67	760	20	<100	-	505	31	176
	B6	16	240	1150	1100	-	920	793	-	885	0.5(AC)	-	-	335	39	192
	B7	17	240	1150	1100	79	760	787	60	740	3	<100	-	388	30	180
	B8	18	240	1150	1100	67	850	732	75	830	5	<100	550	518	27	178
	B9	19	240	1150	1100	85	880	780	40	680	20	<100	-	438	26	138
	B10	20	200	1150	1000	50	740	757	75	720	25	550	(*)	395	28	156
	B11	23	200	1150	1100	60	800	787	75	780	10	550	(*)	492	15	13
	B12	25	240	1150	1100	75	950	641	75	930	24	-	-	825	8	8

*1: Estimated value from chemical composition, *2: Self tempering due to recalcrescence

Table 3

Class	Test steel no.	Steel no.	Matrix phase		Second phase		Interlayer		Fatigue properties			
			Phase and microstructure*	Hardness (Hv)	Phase and microstructure*	Hardness (Hv)	Area fraction (%)	Aspect ratio (-)	distance between matrix and second phase	Propagation speed when $\Delta K=20\text{MPa}\sqrt{\text{m}}$ (m/cycle)	Welded joint fatigue test	Relative life (x) life to BL
Present invention	A1	1	F	154	M(T)	420	20	5.5	20	9.5E-09	4.5E+05	2.0
	A2	2	F	158	M(T)	415	25	7.8	15	9.2E-09	4.6E+05	2.1
	A3	3	F	155	M	880	6	5.8	32	8.4E-09	4.7E+05	2.1
	A4	4	F	160	M	540	10	4.4	30	7.0E-09	5.0E+05	2.3
	A5	5	F	170	M	800	8	6.0	25	6.5E-09	5.1E+05	2.3
	A6	6	F	175	M	700	16	12.3	40	6.0E-09	5.2E+05	2.4
	A7	7	F	180	M	600	18	4.9	20	3.2E-09	6.4E+05	2.9
	A8	8	F	185	M	820	12	7.3	15	4.5E-09	5.7E+05	2.6
	A9	9	F	185	M	650	20	7.8	28	3.5E-09	6.2E+05	2.8
	A10	10	F	190	M	650	28	10.2	15	3.1E-09	6.5E+05	2.9
	A11	11	F	173	M	640	25	6.2	15	3.5E-09	6.2E+05	2.8
	A12	12	F	165	M	690	12	5.4	20	4.5E-09	5.7E+05	2.6
	A13	13	F	160	M	695	7	5.6	25	7.3E-09	4.9E+05	2.2
	A14	14	F	158	M(T)	430	29	7.0	16	8.8E-09	4.6E+05	2.1
	A15	15	F	180	M	705	20	8.5	23	2.9E-09	6.5E+05	3.0
	A16	16	F	170	M	620	25	7.2	20	2.6E-09	6.7E+05	3.0
	A17	17	F	173	M	580	18	6.2	24	4.5E-09	5.8E+05	2.6
	A18	18	F	175	M	730	16	8.9	18	3.5E-09	6.2E+05	2.8
	A19	19	F	155	M(T)	540	24	7.5	20	9.4E-09	4.5E+05	2.1
	A20	20	F	175	M(T)	680	20	7.6	18	3.1E-09	6.4E+05	2.9
Comparison example	A21	21	F	180	M	850	3	5.0	45	1.6E-08	3.8E+05	1.7
	A22	22	F	160	M	500	40	7.6	2	3.0E-09	1.5E+05	0.7
	A23	24	F	150	M	500	1	3.8	60	4.8E-08	2.4E+05	1.1
	B1	11	F	130	P	250	20	2.2	30	8.0E-08	2.2E+05	1.0
	B2	12	F	130	M	650	13	2.5	30	1.2E-08	4.2E+05	1.9
	B3	13	F	120	M	750	6	2.5	55	1.1E-08	4.2E+05	1.9
	B4	14	F	128	M(T)	300	25	6.1	30	2.5E-08	3.1E+05	1.4
	B5	15	F	155	B+M	350	15	3.5	35	2.0E-08	3.2E+05	1.4
	B6	16	F	145	P	260	15	1.8	28	5.0E-08	2.3E+05	1.0
	B7	17	F	125	B	300	18	1.5	35	2.0E-08	3.5E+05	1.6
	B8	18	B	250	-	-	-	-	-	4.5E-08	2.2E+05	1.0
	B9	19	F	120	M	450	25	1.2	30	1.5E-08	3.0E+05	1.4
B10	20	F	170	B	280	20	4.5	25	1.4E-08	4.0E+05	1.8	
B11	23	B	230	-	-	-	-	-	5.0E-08	2.3E+05	1.0	
B12	25	M	420	-	-	-	-	-	7.0E-08	1.8E+05	0.8	

*F: Ferrite, P: Pearlite, B: Bainite, M: Martensite, M(T): tempered martensite.

The microstructure was identified by polishing the cross-section of the steel plate in the rolling direction to a mirror finish, then bringing out the structure by NYTAL etching and REPELLA etching, observing the microstructure using an optical microscope and using the generated phase together with the later explained test results. Further, the hardness was measured by a load of 10 kg using a micro-Vickers hardness tester. The ratios of the phases, the aspect ratio, and the distances between layers were found by image analysis of an optical micrograph.

FIG. 1 is a view of a test piece used for a fatigue crack propagation test. The conditions of the fatigue crack propagation test were as follows:

Load application method: 3 point bending
Stress ratio: 0.1
Environment: room temperature air
Crack length measurement: DC potential difference method

FIG. 2 is a view of a test piece used for a welded joint fatigue test. The welding was CO₂ gas arc welding by a heat input of 18 kJ/min. The fatigue test conditions were as follows:

Load application method: axial force
Stress ratio: 0.1
Environment: room temperature air
Test stress range: 150 MPa

Test Steel Nos. A1 to A20 were all steel produced from slabs of chemical compositions of the present invention in accordance with the requirements of the present invention. They satisfied the microstructure requirements, had a fatigue crack propagation speed da/dN when a stress intensity factor range ΔK is 20 MPa $\sqrt{m}$ of not more than 10^{-8} m/cycle, and had a welded joint fatigue life of at least two times that of the comparative example of Test Steel No. B1, that is, had superior fatigue properties.

On the other hand, the Test Steel Nos. A21 to A23 satisfied the requirements of production of the present invention, but were outside the range of limitation of the chemical composition. The Test Steel Nos. A21 and A23 had a ferrite-martensite microstructure, but the martensite fraction was small or the interlayer distance was large, so the propagation speed at the time of $\Delta K=20$ MPa $\sqrt{m}$ was 10^{-8} m/cycle or more. Therefore, the welded joint fatigue life was not more than 2 times that of the comparative example of Test Steel No. B1 and the fatigue properties were inferior to those of the steel of the present invention. Further, the Test Steel No. A22 had an excessive martensite fraction, so the toughness greatly deteriorated and brittle cracks occurred during the fatigue test, and the welded joint fatigue life was remarkably inferior to the steel of the present invention. Further, the interlayer distance was too small, so the propagation properties were also inferior to the steel of the present invention.

Further, the Test Steel Nos. B1 to B10 satisfied the range of limitation of chemical composition of the present invention, but were outside the production requirements. The Test Steel Nos. B1, B6, B7, B8, and B10 had second phases other than martensite. With microstructure other than martensite, it is hard to effectively block crack progression, so the fatigue crack propagation resistance is degraded compared with the steel of the present invention. The welded joint fatigue life also failed to be improved. The Test Steel Nos. B2 and B3 had second phases of martensite, but the ferrite hardness was small and the internal stress was not increased. In addition, the aspect ratio was small or the interlayer distance was large, so the frequency of striking the martensite at the time of crack progression was small and effective inhibition was not possible. Accordingly, the fatigue crack propagation resistance was degraded compared with the steel of the present invention

and the welded joint fatigue life was also not improved.

5 The Test Steel No. B4 had a high tempering
temperature and eased internal stress, so crack
progression failed to be inhibited. The fatigue
10 properties were inferior to those of the steel of the
present invention. The Test Steel No. B5 had a high
finishing rolling start temperature and accelerated
cooling start temperature and the majority of the second
15 phase comprised of bainite, so crack progression was not
blocked and the fatigue properties were inferior to those
of the steel of the present invention. The Test Steel No.
B9 had a second phase comprised of martensite, but the
finishing cumulative reduction ratio was small and the
20 aspect ratio was extremely small, so the frequency of
encountering the martensite at the time of crack
progression was small and the progression could not be
effectively blocked. Therefore, the fatigue properties
were inferior to those of the steel of the present
invention. Further, the Test Steel Nos. B11 to B12 did
not satisfy the range of limitation of the present
invention in either chemical composition or production
method, so were remarkably inferior in fatigue properties
compared with the steel of the present invention.

WHAT IS CLAIMED IS

1. A steel plate superior in fatigue crack propagation resistance containing, by wt%,

5 C: 0.03 to 0.2%,
Si: 0.01 to 1.6%,
Mn: 0.5 to 2%,
P: 0.02% or less,
S: 0.02% or less,
Al: 0.001 to 0.1%, and
10 N: 0.001 to 0.008%
and a balance of Fe and unavoidable impurities,
wherein a microstructure of the base material is a
lamellar structure having ferrite with a Vickers hardness
of at least 150 as a matrix phase and having martensite
15 with a Vickers hardness of 400 to 900, an area fraction
of 5 to 30%, and an aspect ratio (long axis/short axis)
of 3 or more as a second phase, and an average interlayer
distance between the ferrite and martensite in the plate
thickness direction is 3 to 50 μm , and a fatigue crack
20 propagation speed da/dN when a stress intensity factor
range ΔK of a stress ratio of 0.1 is 20 $\text{MPa}\sqrt{\text{m}}$ is not more
than 10^{-8} m/cycle.

2. A steel plate superior in fatigue crack propagation resistance as set forth in claim 1 further
25 containing, by wt%, one or more of

Cu: 0.1 to 2.5%,
Ni: 0.1 to 5%,
Cr: 0.01 to 1.5%,
Mo: 0.01 to 1.5%,
30 W: 0.01 to 1.5%,
Ti: 0.001 to 0.05%,
Nb: 0.005 to 0.2%,
Zr: 0.005 to 0.2%,
V: 0.005 to 0.2%, and
35 B: 0.0002 to 0.005%.

3. A steel plate superior in fatigue crack propagation resistance as set forth in claim 1 or 2,

further containing, by wt%, one or more of

Mg: 0.0005 to 0.01%,

Ca: 0.0005 to 0.01%, and

REM: 0.005 to 0.05%.

5 4. A method of production of steel plate superior
in fatigue crack propagation resistance wherein a
microstructure of a base material is a lamellar structure
having ferrite with a Vickers hardness of at least 150 as
a base phase and having martensite with a Vickers
10 hardness of 400 to 900, an area fraction of 5 to 30%, and
an aspect ratio (long axis/short axis) of 3 or more as a
second phase, an average interlayer distance between the
ferrite and martensite in the plate thickness direction
is 3 to 50 μm , and a fatigue crack propagation speed
15 da/dN when a stress intensity factor range ΔK of a stress
ratio of 0.1 is 20 $\text{MPa}\sqrt{\text{m}}$ is not more than 10^{-8} m/cycle,
comprising heating a slab containing the ingredients
described in any one of claims 1 to 3 to a temperature of
at least an A_{c3} transformation temperature to not more
20 than 1350°C, then rolling by a cumulative reduction ratio
of 10 to 80% in an austenite single phase region of the
 A_{r3} transformation temperature to 1250°C, then finishing
rolling by a cumulative reduction ratio of 40 to 90% in a
two phase region of austenite-ferrite of a rolling start
25 temperature of not more than the A_{r3} transformation
temperature and a rolling end temperature of not less
than 600°C.

5. A method of production of a steel plate
superior in fatigue crack propagation resistance as set
30 forth in claim 4, further comprising accelerated cooling
by a cooling rate of 5 to 80°C/s to 20 to 400°C after said
finishing rolling.

6. A method of production of a steel plate
superior in fatigue crack propagation resistance as set
35 forth in claim 4 or 5, further comprising tempering in a
temperature range of 300 to 500°C.