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(54) STEEL EXCELLING IN TOUGHNESS AT REGION AFFECTED BY WELDING HEAT

STAHL MIT HERVORRAGENDER HÄRTE DES DURCH SCHWEISSUNGSHITZE BETROFFENEN BEREICHS

ACIER EXCELLANT PAR SA DURETÉ DANS LES ZONES AFFECTÉES PAR LA CHALEUR DE LA SOUDURE

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Description

[0001] The present invention relates to steel excellent in toughness of the weld heat affected zone (HAZ) in small heat input welding to medium heat input welding and a method of production thereof.

[0002] The HAZ toughness of a low alloy steel is governed by various factors such as (1) the size of the crystal grains, (2) the state of dispersion of hard phases such as high-carbon martensite (M*), upper bainite (Bu), and ferrite sideplate (FSP), (3) the state of precipitation hardening, (4) the presence of any intergranular embrittlement, and (5) the micro-segregation of the elements. These factors are known to have a large effect on the toughness. Many technologies are being commercialized in order to improve the HAZ toughness.

[0003] It is safe to say that such toughness inhibiting factors are caused by additive elements. Reduction of the alloy element content increases the toughness. However, higher strength is always being sought in structural steel. Because of that, the addition of alloy elements is necessary. That is, the demands of strength and toughness are contradictory from the viewpoint of the alloy element content. Toughness increasing technology which does not depend on alloy elements has been sought.

[0004] As particularly excellent technology, it is known to use steel which does not substantially include any A1 so as to make the microstructure finer and in addition correctly balance the Ti, O, and N to suppress the precipitation of TiC and reduce precipitation hardening and thereby improve the toughness (JP-A 5-247531). In this case, the toughness of the weld heat affected zone is determined by the balance of the effects of the microstructure and the effects of the hardened layer which includes M*. In the prior art, this was solved by improving the toughness of the base material matrix by Ni and the like. However, the addition of large amounts of Cu, Ni, and other expensive alloy elements necessary for the realization of this technology invited an increase in the production costs. This became an obstacle in producing high strength steel excellent in CTOD property.

[0005] The point of the steel according to the prior invention not substantially including any Al and Nb is made use of in the present invention as well. However, in the prior invention, the C content is high, so the problem of the drop in toughness when increasing the Mn content remains unsolved. Further, there was a concern over the impurities Nb and V having a detrimental effect on the toughness.

[0006] Further, JP-A-2003-147484 follows the thinking of JP-A-5-247531 and, while making use of Ti oxides, adds Nb and raises the Mn content. This causes the austenite-ferrite transformation start temperature to drop to thereby suppress the formation of the hard phases and simultaneously to obtain a suitable microstructure to thereby satisfy the -10°C CTOD property. However, the invention of this JP-A-2003-147484 did not sufficiently satisfy the required CTOD property of weld joints at the much tougher level of -40°C or less. JP 2001 355 039 discloses an ultrahigh strength steel tube excellent in the low temperature toughness of the weld metal, but which contains relatively high amounts of niobium and molybdenum.

[0007] The present invention provides technology which inexpensively produces high strength steel excellent in toughness in multi-layer welding of small to medium heat input. The steel produced by the present invention is extremely good in the CTOD property of multi-layer weld zones of small to medium heat input among the levels of weld heat affected zone toughness. The gist of the present invention is as follows.

(1) A steel excellent in toughness of a weld heat affected zone characterized by containing, by mass%, C: 0.02 to 0.06%, Si: 0.05 to 0.30%, Mn: 1.7 to 2.7%, P: 0.015% or less, S: 0.010% or less, Ti: 0.005 to 0.015%, O: 0.0010 to 0.0045, N: 0.0020 to 0.0060% and optionally one or two of Cu: 0.25% or less and Ni: 0.50% or less and comprising a balance of iron and unavoidable impurities, having an amount of intermixture of impurities limited to Al: 0.004% or less, Nb: 0.003% or less, and V: 0.030% or less, and having a CeH represented by formula (A) in the range of 0.04 or less:

$$CeH = C+1/4Si-1/24Mn+1/48Cu+1/32Ni+1/0.4Nb+1/2V \qquad (A)$$

where, C, Si, Mn, Cu, Ni, Nb, and V show steel compositions (mass%).

(2) A steel excellent in toughness of a weld heat affected zone as set forth in (1), characterized in that the CeH is in the range of 0.01 or less.

(3) A method of production of steel excellent in toughness of a weld heat affected zone characterized by heating a slab satisfying the steel ingredients and CeH of (1) to a temperature of 1100°C or less, then treating it by thermo-mechanical control process.

[0008] The invention is described in detail in conjunction with the drawings, in which:

FIG. 1 is a view showing the relationship of a cooling time of 800 to 500°C and an M* fraction, and
FIG. 2 is a view showing the relationship of the CeH and CTOD properties.

[0009] According to the research of the present inventors, the CTOD property of the HAZ at the time of small to medium heating input (1.5 to 6.0 kJ/mm with a sheet thickness of 50 mm) welding (CTOD property at temperature of -40°C or less) is governed by the toughness of extremely local regions. Control of the microstructure of this portion and reduction of the embrittlement elements are important. In other words, the CTOD property is not the average property of the material, but is governed by the local embrittlement zones. If there are regions which cause embrittlement, even in just parts of the steel material, the CTOD property of the steel sheet will be remarkably impaired.

[0010] Specifically, the local regions which exert the greatest effects on the CTOD property are M*, ferrite sideplate (FSP), and other hard phases. In order to suppress the formation of this kind of hard phase, in the past it had been necessary to keep the hardenability of the steel low. This became a factor inhibiting higher strength.

[0011] The present invention is characterized by the following discoveries and their embodiment in a steel of a high HAZ toughness. Specifically,

1) In a small to medium heat input welded HAZ, generally the cooling time after welding is within 60 seconds. The inventors discovered that under such cooling conditions, if the C content is sufficiently low, by adequately controlling other embrittlement elements, even if adding Mn to 27%, M* which exerts a negative effect on toughness is no longer formed. FIG. 1 shows the M* fraction when changing the amount of Mn from 1.7% to 2.7% with 0.05%C-0.15% Si. It is learned that even if the Mn content changes, if the cooling time of 800 to 500°C is within 60 seconds or so, the M* fraction becomes very small. As a result, it becomes possible to raise the content of Mn for which addition in a large quantity had been thought to be impossible in the past due to causing deterioration of the toughness.
2) The inventors discovered that the steel ingredients could be made suitable in an Al-less based steel.
3) The inventors eliminated the unexpected factors reducing toughness by limiting the Al, Nb, and V present as impurities in the steel to certain limits or less.

[0012] That is, by employing Al-less based steel, it became possible to reliably form TiO and effectively improve the toughness.

[0013] By combining these three points, it became possible to realize a good CTOD property under difficult temperature conditions of -20°C or less in a small to medium heat input welded HAZ which could not be achieved until now.

[0014] Even when very little M* is formed, control of the embrittlement elements C, Si, Cu, Ni, Nb, V, and the like is essential. Specifically, it is essential to control the value (CeH) of C+1/4Si-1/24Mn+1/48Cu+1/32Ni+1/0.4Nb+1/2V to a predetermined range.

[0015] FIG. 2 shows the results when producing 20 kg of steel of the steel ingredients of 0.05% C-0.15%Si-1.7 to 2.7%Mn by vacuum melting, rolling it to steel sheet, imparting a heat history of an actual welded joint three times by a simulated thermal cycle device, then running a CTOD test.

[0016] $T\delta c$ 0.1 (670.9 CeH-67.6) is the temperature when the lowest value of three CTOD test values at different test temperatures is 0.1 mm. There is a clear trend for the $T\delta c$ 0.1 (CTOD property) to excellent substantially linear behavior as the CeH drops. If the CeH drops to around 0.01, it is learned that the $T\delta c$ 0.1 reaches -60°C.

[0017] That is, by satisfying the requirements of the present invention steel and controlling the CeH, the intended CTOD property can be obtained. With the present invention steel, control of the value of CeH according to the required CTOD property is one of the characterizing features of the invention. In addition to the control of the value of CeH, rectifying the contents of the other alloy elements is required for realizing steel provided with both high strength and a superior CTOD property. Below, the ranges of limitation and the reasons will be explained.

[0018] C has to be 0.02% or more in order to obtain strength, but if over 0.06%, it degrades the toughness of the welding HAZ and does not allow satisfaction of a good CTOD property, so 0.06% is made the upper limit.

[0019] Si inhibits the HAZ toughness, so a smaller amount is preferable in order to obtain a good HAZ toughness. However, with the invention steel, no Al is added, so addition of 0.05% or more is necessary for deoxidation. However, if the content is over 0.30%, the HAZ toughness is harmed, so 0.30% is made the upper limit.

[0020] Mn is an inexpensive element with a large effect of rectifying the microstructure and lowers the CeH, so addition does not harm the HAZ toughness of small to medium heat input, therefore it is desirable to make the content large and obtain a high strength. However, if over 2.7%, it promotes the segregation of the slab and facilitates formation of Bu harmful to toughness, so the content was made to an upper limit of 2.7%. Further, if less than 1.7%, the effect is small, so the lower limit was made 1.7%. Note that from the viewpoint of toughness, over 2.0% is more preferable.

[0021] P and S should both be small in amount from the viewpoints of base material toughness and HAZ toughness, but there are limits to their reduction in industrial production. 0.015% and 0.010%, preferably 0.008% and 0.005%, were therefore made the upper limits.

[0022] Al is not deliberately added in the present invention, but inclusion as an impurity in the steel is unavoidable. This forms Al oxides which inhibit the formation of Ti oxides, so a smaller content is desirable, but there are limits to its reduction in industrial production. 0.004% is therefore the upper limit.

[0023] Ti forms Ti oxides and makes the microstructure finer, so greatly contributes to improvement of the toughness, but if the content is too great, it forms TiC. This degrades the HAZ toughness, so 0.005 to 0.015% is a suitable range.

[0024] O is necessary for the formation of a large amount of oxides of Ti. If less than 0.0010%, the effect is small, while if over 0.0045%, it forms coarse Ti oxides and sharply degrades the toughness, so the range of content was made 0.0010 to 0.0045%.

[0025] N is necessary to form fine Ti nitrides and improve the base material toughness and HAZ toughness, but if less than 0.002%, the effect is small, while if over 0.006%, surface defects are formed at the time of billet production, so the upper limit was made 0.006%.

[0026] Further, Nb and V are inherently embrittlement elements. As shown by the large coefficient in formula (A), their presence causes the CeH to greatly rise and made the HAZ toughness remarkably fall, so these are not deliberately added in the present invention. Even when included as impurities in the steel, to secure toughness, Nb has to be limited to 0.003% or less. Further, V has to be limited to 0.030% or less, preferably 0.020% or less.

[0027] Cu and Ni result in little deterioration of the HAZ toughness due to their addition, have the effect of increasing the strength of the base material, and are effective for the further improvement of the properties, but increase the production costs, so the upper limits of the contents when added were made Cu: 0.25% and Ni: 0.50%.

[0028] Even if limiting the ingredients of the steel in the above way, if not forming a suitable structure by a suitably method of production, the desired effects cannot be exhibited. Due to this, the production conditions also have to be considered.

[0029] The present invention steel is preferably produced industrially by continuous casting. The reasons are that the solidification cooling rate of the molten steel is fast and it is possible to form fine Ti oxides and Ti nitrides in large amounts in the slab. When rolling the slab, the reheating temperature has to be made 1100°C or less. If the reheating temperature exceeds 1100°C, the Ti nitrides becomes coarser, the toughness of the base material decreases, and the effect of improvement of the HAZ toughness cannot be expected.

[0030] Next, the method of production after reheating requires treatment by thermo-mechanical control process. The reason is that even if a superior HAZ toughness is obtained, if the toughness of the base material is inferior, the steel product is insufficient. As methods of treatment by thermo-mechanical control process, 1) controlled rolling, 2) controlled rolling-accelerated cooling, 3) direct quenching-tempering after rolling, etc. can be mentioned, but the preferred methods are controlled rolling-accelerated cooling and the direct quenching-tempering after rolling.

[0031] Note that after producing the steel, even if reheating to a temperature of the Ar3 transformation point or more for the purpose of dehydrogenation etc., the characterizing features of the present invention are not impaired.

[0032] Further, the above method is one example of a method of production of the present invention steel. The method of production of the present invention steel is not limited to the above method.

EXAMPLES.

[0033] Thick-gauge steel sheets of various steel ingredients were produced by the converter-continuous casting-thick-gauge sheet process. The base material strength was determined and a CTOD test of the weld joints was run. The welding was performed by the submerged arc welding (SAW) method, generally used for test welding, with a welding heat input of 4.5 to 5.0 kJ/mm at the K groove so that the weld fusion line (FL) became perpendicular. The CTOD test was run by a sheet of a size of t (sheet thickness) x 2t notched by introducing a 50% fatigue crack in the FL location. Table 1 shows examples of the present invention and comparative examples.

[0034] The steel sheets produced by the present invention (Invention Steels 1 to 20) had yield strengths (YS) of 430 N/mm^2 or more and exhibited good breaking toughness of CTOD values at -20°C, -40°C, and -60°C all of 0.27 mm or more.

[0035] As opposed to this, Comparative Steels 21 to 26 had strengths and CTOD values inferior to the invention steels and did not possess the properties necessary as steel sheet used under harsh environments. Comparative Steel 21 had Nb added, therefore the Nb content of the steel sheet became too great. The value of CeH also became high, so the CTOD value was a low value. Comparative Steel 22 had too great a C content and also too great a value of CeH, so the CTOD value was a low value. Comparative Steels 23 and 24 had low CeH's, but the Al content was too high, Ti oxides were insufficiently formed, and the microstructure was not sufficiently made finer. Comparative Steel 25 had a CeH of about the same extent as the invention steel, but the C was too low and the O was too great, so the base material strength was low and the CTOD value was a low value. Comparative Steel 26 had an excessively large amount of Nb mixed in as an impurity, so despite CeH being low, the base material strength and CTOD value were low values.

Table 1

Steel class		C	Si	Mn	P	S	Cu	Ni	Nb	V	Ti	Al	N	O	CeH
Inv. ex.	1	0.021	0.13	2.65	0.005	0.002	0.24	0.42	<0.001	<0.001	0.010	0.003	0.0042	0.0023	-0.039
	2	0.023	0.10	2.57	0.006	0.003			0.001	<0.001	0.009	0.004	0.0035	0.0025	-0.057
	3	0.025	0.11	2.47	0.004	0.003			0.003	<0.001	0.011	0.003	0.0043	0.0026	-0.043
	4	0.025	0.15	2.39	0.005	0.002	0.15	0.24	<0.001	<0.001	0.011	0.002	0.0035	0.0023	-0.026
	5	0.031	0.08	2.38	0.005	0.008	0.15	0.30	0.002	<0.001	0.009	0.003	0.0033	0.0031	-0.031
	6	0.032	0.09	2.30	0.006	0.002			<0.001	0.020	0.009	0.003	0.0036	0.0027	-0.031
	7	0.036	0.11	2.27	0.012	0.003		0.35	0.001	<0.001	0.011	0.004	0.0040	0.0022	-0.018
	8	0.037	0.12	2.28	0.005	0.004	0.23		0.001	<0.001	0.009	0.003	0.0044	0.0033	-0.021
	9	0.038	0.12	2.16	0.006	0.005			<0.001	<0.001	0.011	0.002	0.0038	0.0018	-0.022
	10	0.040	0.15	2.13	0.009	0.003			0.002	0.025	0.011	0.003	0.0041	0.0020	0.006
	11	0.040	0.08	2.06	0.005	0.007			<0.001	<0.001	0.012	0.003	0.0043	0.0028	-0.026
	12	0.043	0.11	2.03	0.010	0.002			0.002	<0.001	0.010	0.002	0.0033	0.0032	-0.009
	13	0.044	0.10	1.94	0.007	0.001			0.003	<0.001	0.013	0.003	0.0035	0.0021	-0.004
	14	0.045	0.14	1.99	0.006	0.002			<0.001	0.020	0.008	0.003	0.0025	0.0038	0.007
	15	0.048	0.11	1.87	0.004	0.001			0.001	<0.001	0.010	0.004	0.0031	0.0025	0.000
	16	0.048	0.09	1.85	0.006	0.002			0.002	<0.001	0.009	0.003	0.0040	0.0024	-0.002
	17	0.050	0.12	1.80	0.006	0.003			<0.001	<0.001	0.011	0.002	0.0036	0.0017	0.005
	18	0.054	0.11	1.76	0.005	0.008			0.003	0.027	0.010	0.003	0.0030	0.0023	0.029
	19	0.057	0.19	1.78	0.006	0.002			0.001	0.015	0.009	0.003	0.0033	0.0026	0.018
	20	0.059	0.13	1.73	0.006	0.003	0.13	0.15	<0.001	<0.001	0.010	0.002	0.0042	0.0022	0.027
Comp. e	21	0.051	0.14	1.85	0.006	0.003			0.042		0.010	0.002	0.0041	0.0030	0.114
	22	0.094	0.12	1.88	0.008	0.004			0.026	0.023	0.011	0.003	0.0038	0.0032	0.122
	23	0.045	0.16	2.18	0.007	0.004			0.015		0.013	0.024	0.0036	0.0010	0.032
	24	0.043	0.11	2.11	0.006	0.002			0.018		0.009	0.031	0.0033	0.0038	0.028
	25	0.016	0.13	2.20	0.009	0.004			0.017		0.010	0.003	0.0031	0.0008	-0.001

(continued)

Steel class		C	Si	Mn	P	S	Cu	Ni	Nb	V	Ti	Al	N	O	CeH
x.	26	0.048	0.14	2.00	0.008	0.004			0.004		0.010	0.003	0.0031	0.0024	0.010

Table 2

Steel class		Production conditions		Base material properties			Welded joint toughness, δc (mm)	
		Slab reheating temperature (°C)	Working heat treatment method	Sheet thickness (mm)	Yield strength (MPa)	Tensile strength (MPa)	-40°C	-60°C
Inv. ex.	1	1050	ACC	45	531	610		0.83
	2	1050	ACC	50	454	543		0.78
	3	1100	DQ	50	452	543		0.51
	4	1100	ACC	65	448	541		0.48
	5	1050	ACC	60	493	570		0.56
	6	1050	ACC	50	465	553		0.43
	7	1100	ACC	50	495	568		0.49
	8	1050	ACC	60	471	562		0.58
	9	1100	ACC	55	467	559		0.56
	10	1100	ACC	60	450	552		0.41
	11	1050	ACC	65	442	530		0.46
	12	1050	CR	50	451	545		0.31
	13	1100	ACC	55	479	565	0.62	
	14	1050	ACC	60	464	567	0.49	
	15	1050	ACC	55	495	582	0.53	
	16	1000	ACC	60	496	594	0.67	
	17	1050	DQ	50	538	619	0.57	
	18	1100	ACC	60	437	528	0.30	
	19	1050	ACC	60	455	551	0.35	
	20	1100	ACC	60	446	547	0.42	
Comp. ex.	21	1150	ACC	50	463	567	0.04	
	22	1100	ACC	50	540	646	0.03	
	23	1100	ACC	60	435	542	0.06	
	24	1150	ACC	60	421	513	0.08	
	25	1100	ACC	60	379	469	0.09	
	26	1100	ACC	50	433	521		0.06

Working heat treatment methods: CR: controlled rolling (rolling at temperature region optimal for strength and toughness) ACC: Accelerated cooling (water cooling down to temperature region of 400 to 600°C after controlled rolling) DQ: Direct quenching-tempering after rolling

[0036] The steel produced by the present invention is high in strength, has an extremely good CTOD property of the FL part where the toughness degrades the most at the time of welding, and exhibits superior toughness. Due to this, production of a high strength steel product that can be used in offshore structures, earthquake resistant buildings, and other harsh environments became possible.

Claims

1. A steel excellent in toughness of a weld heat affected zone **characterized by** containing, by mass%, C: 0.02 to 0.06%, Si: 0.05 to 0.30%, Mn: 1.7 to 2.7%, P: 0.015% or less, S: 0.010% or less, Ti: 0.005 to 0.015% O: 0.0010 to 0.0045, N: 0.0020 to 0.0060% any optionally one or two of Cu: 0.25% or less and Ni: 0.50% or less and comprising a balance of iron and unavoidable impurities, having an amount of intermixture of impurities limited to Al: 0.004% or less, Nb: 0.003% or less, and V: 0.030% or less, and having a CeH represented by formula (A) in the range of 0.04 or less:

$$CeH = C+1/4Si-1/24Mn+1/48Cu+1/32Ni+1/0.4Nb+1/2V \qquad (A)$$

   where, C, Si, Mn, Cu, Ni, Nb, and V show steel ingredients (mass%).

2. A steel excellent in toughness of a weld heat affected zone as set forth in claim 1, **characterized in that** the CeH is in the range of 0.01 or less.

3. A method of production of steel excellent in toughness of a weld heat affected zone **characterized by** heating a slab satisfying the steel compositions and CeH of claim 1 to a temperature of 1100°C or less, then treating it by thermo-mechanical control process.

Patentansprüche

1. Stahl mit ausgezeichneter Zähigkeit einer Schweißwärmeeinflusszone, **dadurch gekennzeichnet, dass** er in Masse-% enthält: 0,02 bis 0,06 % C, 0,05 bis 0,30 % Si, 1,7 bis 2,7 % Mn, höchstens 0,015 % P, höchstens 0,010 % S, 0,005 bis 0,015 % Ti, 0,0010 bis 0,0045 % O, 0,0020 bis 0,0060 % N und optional höchstens 0,25 % Cu und/oder höchstens 0,50 % Ni sowie als Rest Eisen und unvermeidliche Verunreinigungen aufweist, eine Verunreinigungsbeimischungsmenge hat, die auf höchstens 0,004 % Al, höchstens 0,003 % Nb und höchstens 0,030 % V begrenzt ist, und einen durch Formel (A) dargestellten CeH-Wert im Bereich von höchstens 0,04 hat:

$$CeH = C + 1/4\ Si - 1/24\ Mn + 1/48\ Cu + 1/32\ Ni + 1/0,4\ Nb + 1/2\ V \qquad (A)$$

   wobei C, Si, Mn, Cu, Ni, Nb und V Stahlbestandteile (Masse-%) angeben.

2. Stahl mit ausgezeichneter Zähigkeit einer Schweißwärmeeinflusszone nach Anspruch 1, **dadurch gekennzeichnet, dass** der CeH-Wert im Bereich von höchstens 0,01 liegt.

3. Verfahren zur Herstellung eines Stahls mit ausgezeichneter Zähigkeit einer Schweißwärmeeinflusszone, **gekennzeichnet durch** Erwärmen einer die Stahlzusammensetzungen und den CeH-Wert von Anspruch 1 erfüllenden Bramme auf eine Temperatur von höchstens 1100 °C und anschließendes Behandeln derselben durch ein thermomechanisches Steuerverfahren.

Revendications

1. Acier excellent dans la ténacité d'une zone affectée par la chaleur de soudage **caractérisé en ce qu'**il contient, en % en masse, C : 0,02 à 0,06 %, Si : 0,05 à 0,30 %, Mn : 1,7 à 2,7 %, P : 0,015 % ou moins, S : 0,010 % ou moins, Ti : 0,005 à 0,015 %, O : 0,0010 à 0,0045, N : 0,0020 à 0,0060 % et éventuellement un ou deux de Cu : 0,25 % ou moins et Ni : 0,50 % ou moins et comprenant un complément de fer et d'impuretés inévitables, ayant une quantité de mélange d'impuretés limitée à Al : 0,004 % ou moins, Nb : 0,003 % ou moins, et V : 0,030 % ou moins, et ayant CeH représenté par la formule (A) dans la plage de 0,04 ou moins :

$$CeH = C+1/4Si-1/24Mn+1/48Cu+1/32Ni+1/0,4Nb+1/2V \qquad (A)$$

   où C, Si, Mn, Cu, Ni, Nb et V indiquent les ingrédients de l'acier (% en masse).

2. Acier excellent dans la ténacité d'une zone affectée par la chaleur de soudage selon la revendication 1 **caractérisé en ce que** le CeH est dans la plage de 0,01 ou moins.

3. Procédé de production d'un acier excellent dans la ténacité d'une zone affectée par la chaleur de soudage **caractérisé par** le chauffage d'une brame satisfaisant les compositions d'acier et CeH selon la revendication 1 à une température de 1100°C ou moins puis son traitement par un procédé de contrôle thermo-mécanique.

Fig.1

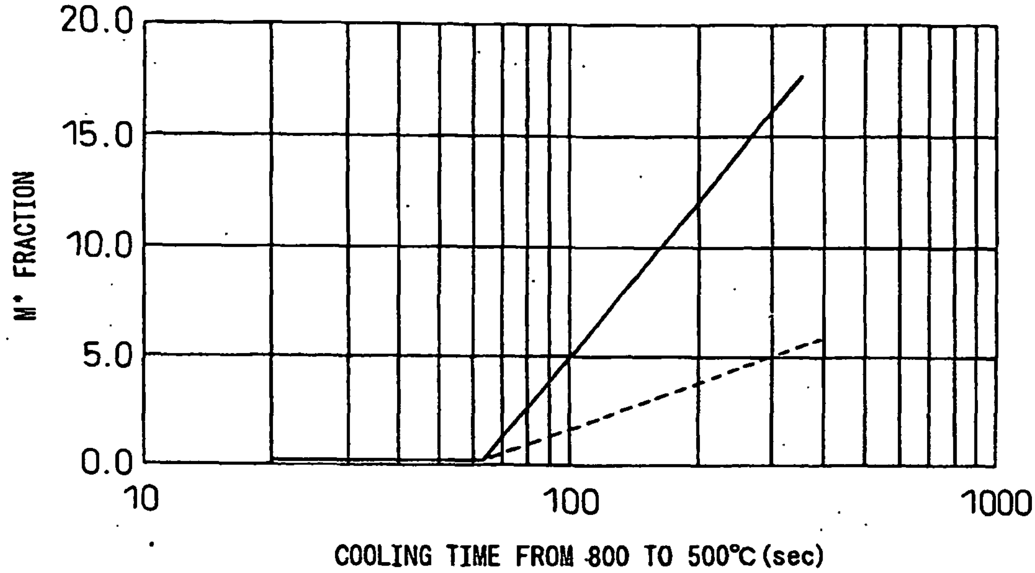
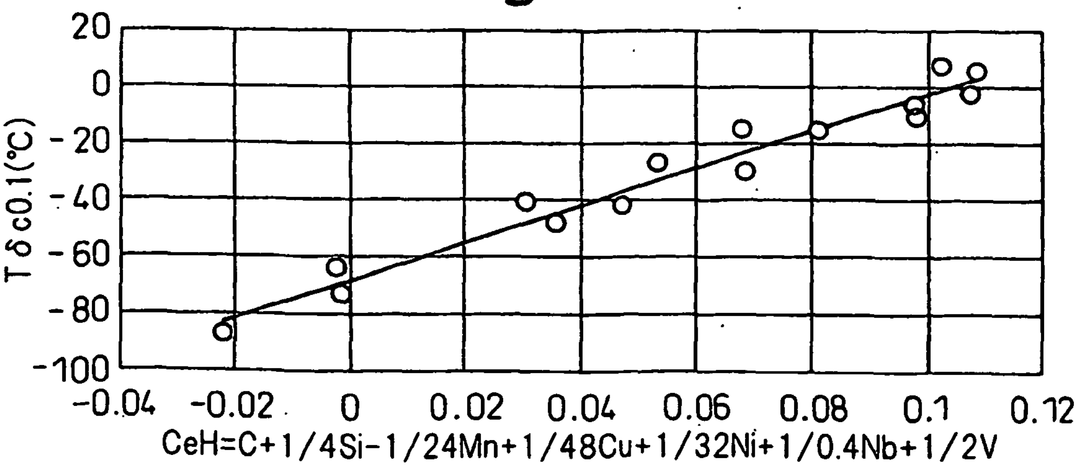

20.0
15.0
10.0
5.0
0.0
M* FRACTION
10
100
1000
COOLING TIME FROM 800 TO 500℃ (sec)

Fig.2


20
0
-20
-40
-60
-80
-100
Tδc0.1(℃)
-0.04
-0.02
0
0.02
0.04
0.06
0.08
0.1
0.12
CeH=C+1/4Si-1/24Mn+1/48Cu+1/32Ni+1/0.4Nb+1/2V

REFERENCES CITED IN THE DESCRIPTION

Patent documents cited in the description

- JP 5247531 A **[0004] [0006]**
- JP 2003147484 A **[0006]**
- JP 2001355039 A **[0006]**