KAWASAKI STEEL TECHNICAL REPORT

No.29 (November 1993)

Special Issue on Iron Power

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420 MPa and 500 MPa Yield Strength Steel Plates with High HAZ Toughness Produced by TMCP for Offshore Structures*


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1 Introduction

Along with the active oil and gas exploitation, offshore structure regions are expanding into deeper and deeper waters, and from mild climate zones to the Arctic Ocean. This has brought about changes to steel plates toward higher strength and heavier gages, with the safety at low temperature of −50 °C and thereabout focussed as special requirement for applications in the Arctic Ocean.

The most important quality requirement for offshore structure plates is a safety assurance as weld structure, with sufficient toughness of heat affected zone (HAZ) in particular. To this end, lowering of carbon equivalent is known to be effective, but this requirement is not generally compatible with the two requirements mentioned above, namely higher strength and heavier gages. The multipurpose accelerated cooling system (MACS) is an effective method for achieving high strength and toughness of a plate even with lower carbon equivalent by controlled rolling and the subsequent accelerated cooling.

This paper describes the influence of effective elements for good HAZ toughness, the appropriate design of chemical composition, and the properties of three types of steel plate that were produced. The steel plates were produced using continuous casting and the MACS process, and the properties of their welded joints were evaluated.

2 Aimed Properties of the Developed Steel Plates

At present, the high strength steel plates generally used for offshore structures are of 360 MPa yield strength, and the maximum thickness of the plates is about 100 mm.[1] The fracture toughness is normally

* Originally published in *Kawasaki Steel Giho*, 25(1993)1, 13-19

Table 1 Target properties of the steel plates and the welded joints

Steel	Plate thickness (mm)	Steel plate			Welded joint		
		YS (MPa)	TS (MPa)	V Charpy absorbed energy (J)	Welding method	Heat input (kJ/mm)	CTOD value (mm)
A	101.6	≥414	≥517	$_{V}E_{-40}\geqq48$	SAW	≤4.5	≥0.38 at −10°C
B	60	≥400	≥530	$_{V}E_{-60}\geqq48$	SAW	5~19	≥0.10 at −50°C
C	50	≥500	≥570	$_{V}E_{-40}\geqq48$	SAW	≤4.5	≥0.25 at −10°C

required in terms of crack tip opening displacement (CTOD) measured at a temperature of −10°C.

The aimed properties of the developed steel plates are listed in **Table 1**, which exceed those of the above mentioned steel plates in both yield strength and toughness at lower temperature. Steel A is supposed to give a yield strength of over 414 MPa for a 101.6-mm-thick plate. Steel B is to achieve a similar strength to that of steel A with a sufficiently great CTOD value at −50°C for icy water regions. Steel C is to give a yield strength of 500 MPa class.

The aimed prorerties of steel A were determined according to API specification 2W grade 60 steel,[2)] while those of steel B were based on API 2W grade 60 together with Supplementary Requirement S-2 for a lower service temperature. Those of steel C were based on API 2W, but with a yield strength raised to 500 MPa, because there is no existing specification for a steel over grade 60.

3 Study on Chemical Composition and Manufacturing Process

Table 2 lists the metallurgical considerations for producing steel plates which would provide the aimed properties. In order to assure high toughness of the welded joints, improvements were tried on chemical composition, including reduction in the carbon equivalent and Si and N contents, addition of rare earth metal (REM) and Ti. The MACS was also adopted.

There are two aspects concerning the HAZ toughness. One is the toughness deterioration caused by grain coarsening in HAZ that results from high heat input welding. The other is the local brittle zones (LBZ) which strongly depend on the multilayered weld pass making method involving heat input and interpass temperatures.

The recent studies[3,4)] have shown that four LBZs exist: coarse grain HAZ (CGHAZ), inter-critical HAZ (ICHAZ), inter-critically reheated coarse grain HAZ (ICCGHAZ), and subcritically reheated coarse grain HAZ (SCCGHAZ). **Figure 1** shows the appearance of these four LBZs in HAZ and their thermal cycles.

Table 2 Metallurgical countermeasures for improving the toughness of local brittle zones (LBZs)

LBZ	Metallurgy	Countermeasure
CG HAZ	Grain size Utilizing pinning effect by insoluble fine precipitate	· Fine dispersion of TiN · REM addition
	Microstructure Reduction of the MA constituent in upper bainite structure	· Lowering C_{eq} · Lowering C · Lowering N
	Nucleation of fine ferrite-pearlite structure	· Lowering C_{eq} · REM addition
	Matrix Reduction of free N and sol. Ti	· Lowering N · Control of Ti/N
IC HAZ	Improving the toughness of base metal	· Utilizing the TMCP
ICCG HAZ	Decreasing the MA constituent Decomposing the MA to ferrite and cementite	· Lowering C, Si, and P
SCCG HAZ	Minimizing the detrimental effect of precipitation hardening on toughness	· Restricting Nb and V contents

3.1 Study on CGHAZ Toughness

CGHAZ is defined as the zone in which base metal is subjected to a single heat cycle with a peak temperature above 1 350°C as shown in Fig. 1. The following measures are proposed for improving the CGHAZ toughness:

(1) To prevent the coarsening of austenite grains during the welding heat cycle by the pinning effectes[5,6)] of REM-oxysulfides and TiN.

(2) To form a fine ferrite-pearlite structure by reducing the carbon equivalent and nitrogen content.

3.2 Study on ICHAZ Toughness

ICHAZ is defined as the zone in which the base metal is subjected to a single heat cycle whose peak temperature exists between Ac_3 and Ac_1. Grain coarsen-

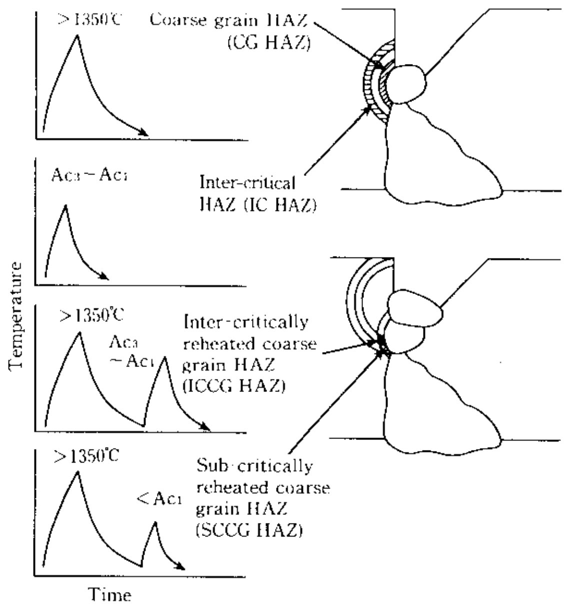


Fig. 1 Four local brittle zones and their thermal cycles

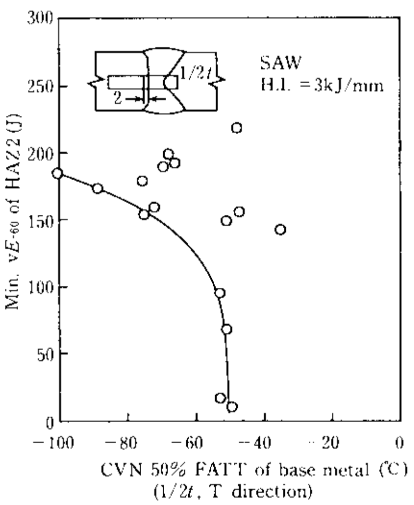


Fig. 2 Relation between toughness of base metal and that of HAZ 2 mm

ing hardly occurs in this region. However, a harder phase such as martensite, which is generated from the re-transformed austenite through a thermal cycle whose peak temperature exists between Ac_3 and Ac_1, can be formed. This results in a deterioration in toughness.

Figure 2 gives the relationship between the Charpy absorbed energy at a point 2 mm away from the fusion line toward the base metal which corresponds to ICHAZ and the fracture appearance transition temperature (FATT) of the base metal. The result indicates that the ICHAZ toughness is strongly affected by the toughness of the base metal. It has also been proved[3] that a

(a) ICCG HAZ (⇦ M-A constituents)
(b) ICCG HAZ tempered at 450°C (← Ferrite cementite aggregates)

Photo 1 Scanning electron microscopic observation of simulated HAZs

toughness increase in the base metal due to grain refining is effective for improving the ICHAZ toughness. Thermo-mechanical controlled process (TMCP) is one of the best ways to obtain a fine-grained microstructure.

3.3 Study on ICCGHAZ Toughness

ICCGHAZ is a region in which CCHAZ is reheated to a temperature between Ac_3 and Ac_1 by a subsequent welding thermal cycle. The re-transformed austenite, in which carbon is concentrated, transforms to a martensite-austenite (MA) constituent, ICCGHAZ becomes an LBZ because of the existence of MA. **Photo 1**(a) shows an example of MA in ICCGHAZ. A recent study[7] has shown that the toughness of ICCGHAZ is governed not only by the volume fraction of MA but also by its morphology and dispersion such as the aspect ratio and spacing.

Since ICCGHAZ is formed by a double welding thermal cycle, the LBZ problem tends to appear with low heat input multi-pass welding.

In multi-pass welding, ICCGHAZ is generally influenced by the third thermal cycle. If the peak temperature of the third thermal cycle exceeds about 450°C, the ICCGHAZ toughness is recovered[3]. This is due to the decomposition of MA into ferrite and cementite as shown in Photo 1(b).

Figure 3 shows the effect of the third reheating peak temperature on the ICCGHAZ toughness. The ICCGHAZ toughness of NM steel, which has high C and Si contents, slightly increases when the third peak temperature is above 500°C, but remains at a very low level even when reheated at 600°C. On the other hand, the ICCGHAZ toughness for TM1 and TM2 steels, each of which has a lower C content than that of NM steel, increases with increasing peak temperature of the third thermal cycle. The recovery of ICCGHAZ toughness starts at 350°C for the TM1 steel (lowC–lowSi). This recovery starting temperature is lower than that for TM2 steel (lowC–highSi) by 50°C. These results suggest that recovering ICCGHAZ toughness by low-temperature, short-term tempering strongly depended on the chemical composition.

(mass%)

Steel	C	Si	Mn	P	S	Al	Cu	Ni	Nb
TM1	0.07	0.11	1.45	0.010	0.001	0.035	0.26	0.73	0.012
TM2	0.08	0.38	1.46	0.004	0.001	0.033	0.20	0.20	0.025
NM	0.16	0.37	1.49	0.011	0.003	0.017	-	0.16	0.027

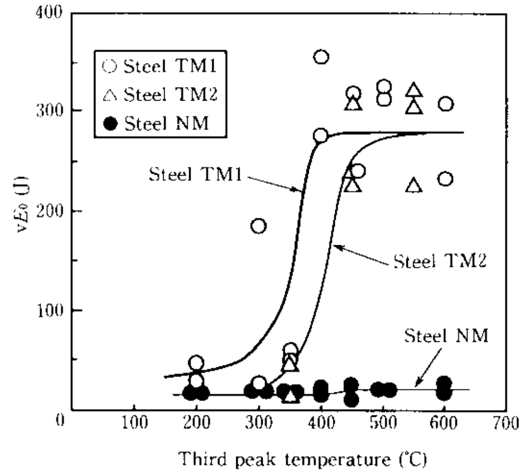


Fig. 3 Effect of the third reheating peak temperature on the toughness of the ICCGHAZ (The first and second reheating peak temperatures of simulated ICCHGAZs were 1 400°C and 800°C, respectively.)

Figure 4 shows the effects of C, Si and Nb on the ICCGHAZ toughness and that tempered at 450°C. The ICCGHAZ toughness is low, and the effect of chemical composition is not significant. Tempering at 450°C, however, improves the ICCGHAZ toughness. Reducing the Si content was effective for improving the toughness. The MA was observed in ICCGHAZ with any quantity of Si in the steel plates. When the Si content was reduced to less than 0.2% (mass %), furthermore, MA was decomposed into ferrite and cementite by tempering at 450°C. This fact is attributed to the decomposition of MA which starts at a relatively low temperature in a low-Si steel.[8] The result also indicates that the addition of C up to about 0.10% and of Nb up to 0.03% does not cause the toughness degradation of tempered ICCGHAZ.

3.4 Study on SCCGHAZ Toughness

SCCGHAZ is also an LBZ produced by a double welding thermal cycle, and is a region in which CGHAZ is reheated at a temperature below Ac_1 by the second welding thermal cycle. Fine Nb(C,N), which is in solution in CGHAZ, precipitates during reheating at around 600°C and results in deteriorated SCCGHAZ toughness. **Figure 5** shows the effect of Nb content on the SCCGHAZ toughnes, having little influence up to 0.03%.

On the basis of these studies on LBZ toughness, the chemical composition was determined. A REM and Ti-added low C–low Si–Cu–Ni–Nb system was selected for the basic composition. The decrease in strength due to the reduced carbon equivalent was compensated by adding Cu and Ni, and also by using TMCP. Both Cu

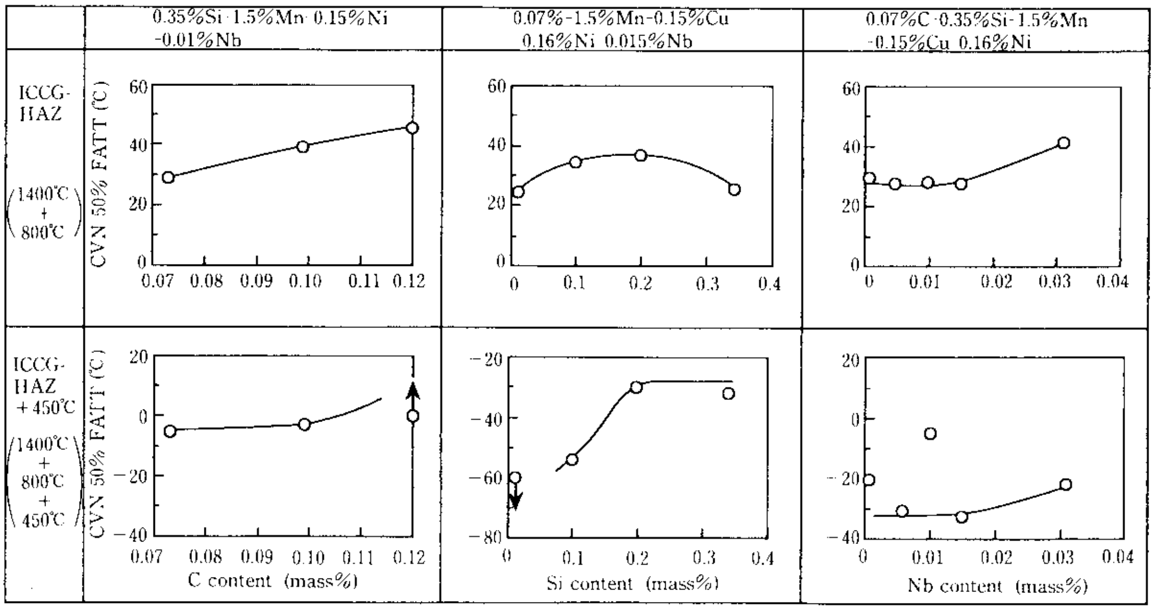


Fig. 4 Effects of C, Si, and Nb on toughness of ICCGHAZ and that tempered at 450°C

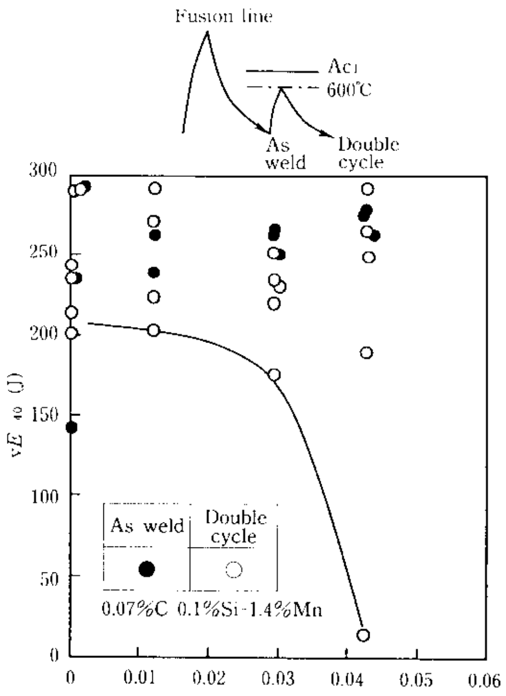


Fig. 5 Effect of Nb content on toughness of SCCG-HAZ

and Ni were added as the element to increase strength without reducing HAZ toughness.

4 Properties of the Steel Plates

4.1 Chemical Composition and Manufacturing Process

The chemical composition of each developed steel plate is shown in **Table 3**. The C_{eq} value varies from 0.38% to 0.41%, and the P_{cm} value was controlled at less than 0.20% for good weldability without preheating. A schematic diagram of the manufacturing process for each steel plate is shown in **Fig. 6**.

In the steelmaking process, impurities such as phosphorus and sulphur were reduced to sufficiently low

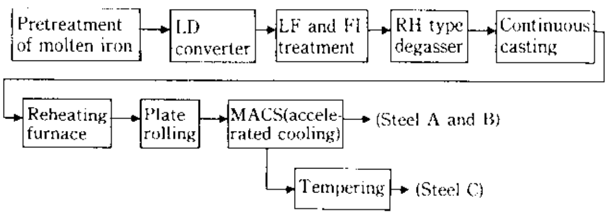


Fig. 6 Manufacturing process of steel plates

levels and REM was added. The steel was then continuously cast into 310-mm-thick slabs. In the plate rolling process, each slab was reheated to a moderately high temperature, controlled rolled to produce a plate mainly in the non-recrystallized austenite region, and accelerated cooled. Steel C of 500 MPa grade was tempered after accelerated cooling to ensure adequate strength. Steel A with a thickness of 101.6 mm was rolled with a specially high reduction per pass to eliminate any porosity at mid-thickness of the continuously cast slab.

4.2 Mechanical Properties of Base Metal

The optical microstructure of the base metal is shown in **Photo 2**, indicating a fine ferrite and bainite mixed structure for each steel plate.

The tensile properties and 2-mm V-notched Charpy impact properties of the base metals are listed in **Table 4**. Each plate satisfies the aimed values for strength and toughness shown in Table 1, including at the mid-thickness of the plates. The 101.6-mm-thick steel A had good resistance to lamellar tearing in spite of its heavy section, that is, the reduction of area in the through-thickness direction was more than 77%.

The brittle fracture initiation properties of the base meteal were investigated by a three-point bend CTOD test. The geometry of the CTOD test specimen was of $B \times 2B$ for steels B and C, and $B \times B$ for steel A (where B is the full thickness of the plate). All CTOD values exceeded 1.7 mm at the test temperature of $-10°C$ for steels A and C and at $-50°C$ for steel B.

4.3 Weldability

The sensitivity to weld cracking at low temperatures

Table 3 Chemical compositions of steel plates (mass%)

Steel	C	Si	Mn	P	S	Cu	Ni	Nb	Al	N	C_{eq}*1	P_{cm}*2	Note
A	0.08	0.18	1.53	0.004	0.001	0.19	0.40	0.024	0.029	0.0036	0.38	0.18	REM-Ti treated
B	0.07	0.11	1.45	0.010	0.001	0.26	0.73	0.012	0.035	0.0025	0.38	0.17	
C	0.10	0.15	1.50	0.004	0.002	0.26	0.49	0.030	0.037	0.0038	0.41	0.20	

*1 $C_{eq}=C+\frac{Mn}{6}+\frac{V+Mo+Cr}{5}+\frac{Cu+Ni}{15}$

*2 $P_{cm}=C+\frac{Si}{30}+\frac{Mn+Cu+Cr}{20}+\frac{Ni}{60}+\frac{Mo}{15}+\frac{V}{10}+5B$

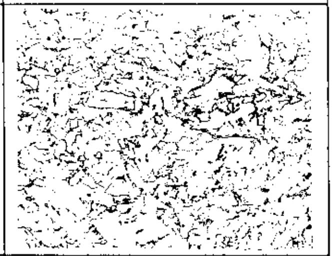
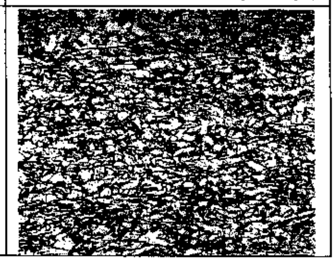
Steel	Thickness (mm)	25μm
A	101.6	
B	60	
C	50	

Photo 2 Microstructure of steel plates

was investigated by the Y-groove cracking test specified in JIS Z3158.

The preheating temperature without cracking was lower than 25°C for steels A and B, and 25°C for steel C as shown in Table 4. These results indicate that all the developed steels can be welded without preheating under actual welding conditions.

4.4 Properties after Strain-Aging

The change in toughness caused by strain aging was investigated on steel A by the Charpy impact test. The material was subjected to a maximum strain of 5% and then aged at 250°C for 1 h. The results in **Table 5** show

Table 5 Effect of strain aging on Charpy impact properties

Steel	Plate thickness (mm)	Strain and aging*1	Charpy impact properties*2		
			Absorbed energy (J)		50% FATT (°C)
			−40°C	−80°C	
A	101.6	No	316	273	−99
		3%+aging	312	156	−105
		5%+aging	329	305	−100

*1 Aging: 250°C×1 h

*2 Specimen location and direction: subsurface, transverse

Table 4 Mechanical properties of steel plates

Steel	Plate thickness (mm)	Location	Direction	Tensile properties			Charpy impact properties				Preheat temp. without cracking*2 (°C)
							Absorbed energy (J)			50% FATT (°C)	
				YP (MPa)	TS (MPa)	El (%)	−40°C	−60°C	−80°C		
A	101.6	1/4t	L	439	535	32	421	382	130	−73	≦25
			T	451	556	31	430	364	277	−92	
			Z	—	527	77*1	—	—	—	—	
		1/2t	L	428	530	33	424	421	397	−86	
			T	420	528	32	385	368	292	−84	
			Z	—	524	77*1	—	—	—	—	
B	60	1/4t	L	480	539	31	—	305	312	−114	≦0
			T	480	549	28	—	315	329	−114	
		1/2t	L	495	549	30	—	307	213	−115	
			T	485	554	27	—	295	250	−106	
C	50	1/4t	L	515	591	28	279	268	213	−105	25
			T	530	603	27	211	214	169	−95	
		1/2t	L	504	596	27	242	185	101	−77	
			T	523	598	26	176	143	86	−68	

*1 Reduction of area (%)

*2 JIS Z3158

that the changes of 50% FATT and absorbed energy at −40°C caused by strain aging were very small.

5 Welded Joint Properties

5.1 Welding Conditions

The welding conditions for the developed steel plates are shown in **Table 6**. The plates were welded by multipass submerged arc welding (SAW) with a heat input ranging from 3 to 5 kJ/mm. In addition, steel B for arctic use was welded by an efficient SAW method using one pass for each side with a high heat input of 19.3 kJ/mm.

5.2 Basic Properties of Welded Joints

The results of tensile tests and Charpy impact tests at various notch positions on the welded joints are shown in **Table 7**. The tensile strengths of all the welded joints are sufficiently higher than the aimed properties shown in Table 1. The Charpy absorbed energy values exceeded 128 J and are sufficiently higher than the aimed properties at every notch position in each welded joint with a heat input of less than 5 kJ//mm.

Table 6 Welding conditions

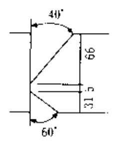
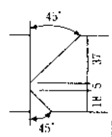
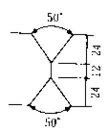
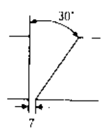
Steel	Plate thickness (mm)	Welding method	Wire × Flux	Groove shape (mm)	Electrode	Current (A)	Voltage (V)	Speed (mm/min)	Heat input (kJ/mm)	Preheat temp. (°C)	Interpass temp. (°C)
A	101.6	Multipass SAW	KW101B × KB100		—	550	32	350	3.0	100	max. 250
					L T	550 550	30 32	455	4.5	min. 250	min. 250
B	60	Multipass SAW	KW30T × KB100		—	600	32	230	5.0	RT	max. 150
		Each side one pass SAW	US255 × PFI50LT		L T1 T2	1200 1100 1150	38 43 45	450	19.3	RT	—
C	50	Multipass SAW	KW50CM × KB100		—	550	30	280	3.5	RT	max. 250
					L T	550 550	30 32	455	4.5	RT	max. 250

Table 7 Mechanical properties of welded joints

Steel	Plate thichness (mm)	Welding method	Heat input (kJ/mm)	Tensile test		V-Charpy impact test					
				TS (MPa)	Location of rupture	Testing temperature (°C)	Absorbed energy (J)				
							WM	FL	HAZ1	HAZ3	HAZ5
A	101.6	SAW	3.0	575	BM	−40	162	234	317	453	441
			4.5	566	BM	−40	145	217	295	442	434
B	60	SAW	5.0	569	BM	−60	177	231	276	342	345
			19.3	588	BM	−60	88	57	103	158	289
C	50	SAW	3.5	634	BM	−60	107	128	165	232	223
			4.5	623	BM	−60	83	134	183	239	251

In the case of steel B being welded with a heat input of 19.3 kJ/mm, the absorbed energy met or exceeded 57 J even at the notch position on the fusion line (which includes 50% welded metal and 50% HAZ).

5.3 CTOD Test Results for the Welded Joints

The CTOD test was carried out according to BS 5762 on specimens with the same dimensions as those of the base plate. In addition, API RP2Z[9] was applied to test and evaluate K-bevel welded joints made by multi-pass SAW.

The results of the CTOD test are shown in **Fig. 7**. The CTOD values for the coarse-grained HAZ in the welded joints made by mutli-pass SAW with a heat input of less than 5 kJ/mm satisfy API specification 2W. The minimum values were 0.61 mm for steel A at the test temperature of −10°C, 0.54 mm for steel B at −50°C, and 0.42 mm for steel C at −10°C. After the test, each specimen was sectioned to confirm that each fatigue crack tip included the coarse-grained HAZ over at least 16% of the specimen thickness to satisfy the API RP2Z rule.

The welded joints of steel B, which were made with a heat input as high as 19.3 kJ/mm, gave a CTOD value on the fusion line exceeding 0.14 mm at −50°C. Many concepts have been proposed to apply the CTOD value for evaluating the integrity of structures. API RP2Z for offshore structure use specifies the minimum CTOD value as 0.25 mm for welded joints of plates of less than 75 mm thickness, and as 0.38 mm for thicker plates. Since this specification, however, does not specify the conditions under which the material is to be used, the specified CTOD values do not seem to have a firm basis for structural integrity.

On the other hand, according to the investigation by Nakano et al.[10] and Yajima et al.,[11] a CTOD value of 0.1 mm at the design temperature (the minimum temperature at which the structure is to be used) is high enough to ensure the integrity of welded joints in an ordinary structure which has no components with a significantly high stress concentration. These studies suggest that steel B, even if high heat input welding is applied, can be used for an ordinary offshore structure at a temperature of −50°C.

5.4 Surface-Notched, Wide Plate Tensile Test of Welded Joint

The surface-notched, wide plate tensile test was performed at −10°C using the welded joints of steel A with a heat input of 4.5 kJ/mm. The specimen geometries are shown in **Fig. 8**. The notch, which was machined along the fusion line and fatigue precracked before test, was of the semi-elliptical surface type. The notch size was varied in the range between 23 and 60 mm in terms of the effective defect parameter, $\bar{a}$, as defined in BSI PD6493.[12] The CTOD value was measured by using a double clip gage method,[10] in which two clip gages were mounted on the knife edges attached to the specimen at two different heights from the specimen surface along the notch as illustrated in Fig. 8. Two sets of double clip gages were placed at two different locations.

Photo 3 shows an example of the microstructure of the cross section at the brittle fracture initiation point. A ductile crack was initiated from the fatigue crack tip located in the grain-coarsened HAZ, and brittle fracture occurred following the ductile crack growth in the HAZ. All of the specimens fractured in a same manner.

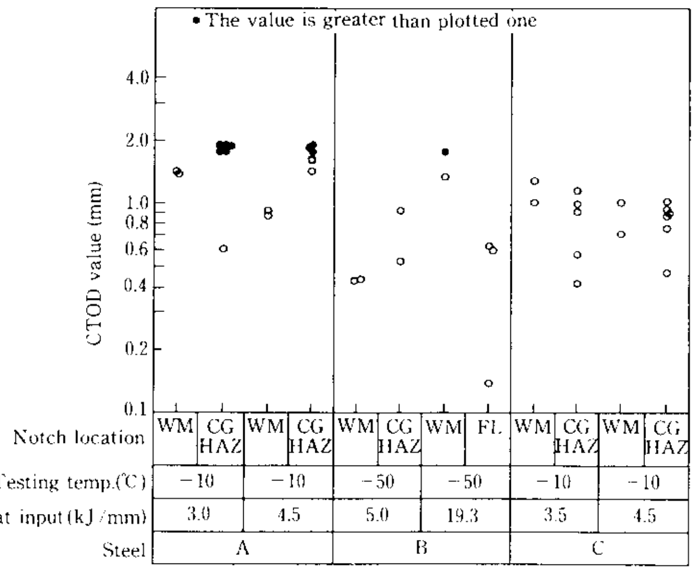
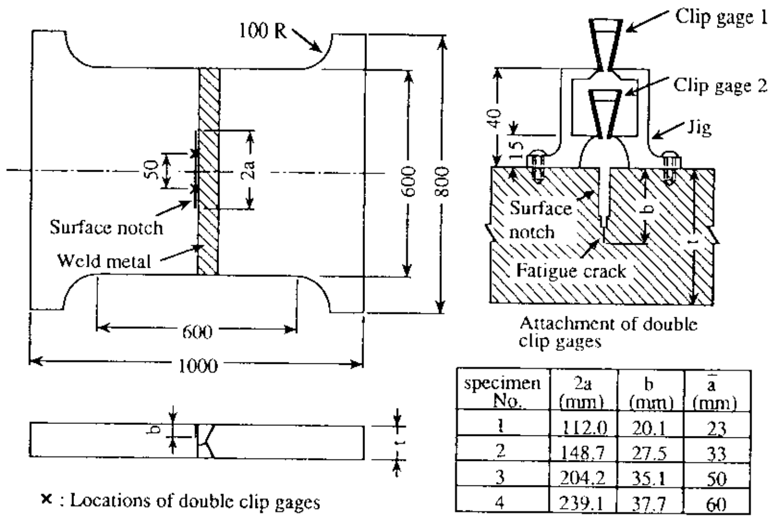


Fig. 7 CTOD test results of SAW welded joints



specimen No.	2a (mm)	b (mm)	$\bar{a}$ (mm)
1	112.0	20.1	23
2	148.7	27.5	33
3	204.2	35.1	50
4	239.1	37.7	60

Fig. 8 Geometry of surface notched, wide plate tensile test specimen

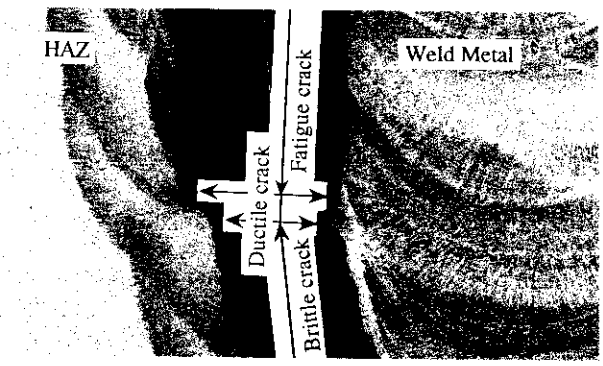


Photo 3 Microstructure at fracture initiation (Specimen 4)

Table 8 Test results of surface notched, wide plate tensile test

Specimen No.	Dimensions of specimen			Fracture stress		Critical CTOD δ_c (mm)	Strain	
	Thickness t (mm)	Width W (mm)	Effective defect parameter $\bar{a}$ (mm)	Gross σ_g (MPa)	Net σ_n (MPa)		GL= 250mm	GL= 500mm
1	104.2	600	23	563	578	≧6.20	≧0.067	≧0.082
2	103.5	600	33	554	585	≧8.50	0.113	0.081
3	103.5	594	50	477	528	4.29	0.038	0.025
4	103.6	600	60	430	487	1.74	0.011	0.006

Table 8 summarizes the test results of the surface notched, wide plate tensile test. The net fracture stress, the critical CTOD value and the strain at brittle fracture decreased with an increase in the effective defect parameter. Even for specimen 4 whose effective defect parameter was as large as 60 mm, the net fracture stress was larger than the yield strength of the steel plate at −10°C. The critical CTOD value for specimen 4 of 1.74 mm was almost the same as the lowest value obtained by the three point bend test.

6 Conclusions

On the basis of fundamental studies to improve the HAZ toughness of welded joints, three types of steel plate for offshore structures were produced with appropriate compositions involving low C-low Si-Cu-Ni-Nb with REM and Ti addition. They were produced by reducing the contents of such impurities as phosphorus and sulphur to sufficiently low levels and by applying both the continuous casting and TMCP processes. The HAZ toughness of their welded joints was evaluated after submerged arc welding with a heat input mainly ranging from 3 to 5 kJ/mm. The main conclusions are as follows:

(1) The 101.6 mm-thick YP 420 MPa steel plate had sufficiently high strength and toughness for API 2W Grade 60. The CTOD value in HAZ of the welded joints exceeded 0.61 mm at −10°C.

(2) The 60-mm-thick YP 420 MPa steel plate for arctic use had sufficiently high strength and toughness for API 2W Grade 60 at lower service temperatures. The CTOD value in HAZ of the welded joints exceeded 0.54 mm at −50°C.

(3) The 50-mm-thick YP 500 MPa steel plate had sufficiently high strength and toughness. The CTOD value in HAZ of the welded joints exceeded 0.42 mm at −10°C.

(4) The surface notched, wide plate tensile test proved that the welded joint of the 101.6 mm thick YP 420 MPa steel plate made with a heat input of 4.5 kJ/mm gave sufficiently high net fracture stress and CTOD value at −10°C.

The authors thank the staff of Mitsubishi Heavy Industries, Ltd. their cooperation in the development and evaluation of the YP 420 MPa steel plate for arctic use.

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