UDC669.15-194.2-462

Development of API X100 UOE Line Pipe

Yoshio Terada*1 Hiroshi Tamehiro*1
Mitsugu Yamashita*2 Naoshi Ayukawa*3
Takuya Hara*1

Abstract:

With the aim of developing API grade X100 line pipe with excellent low-temperature toughness of both base material and heat-affected zone (HAZ) and also excellent field weldability, chemical compositions of base material and thermomechanical control process (TMCP) conditions were studied on two basic chemical systems, that is, a Nb-Mo system and a Nb-B system. Based on the experimental results, X100 line pipe was manufactured on a large scale and its properties were examined. This trial pipe exhibited excellent field weldability as well as sufficient strength and good low-temperature toughness.

1. Introduction

There has been a gradual increase in the demand for an ultra-high-strength X100 line pipe in order to further improve transportation efficiency through an increase in operation pressure, and to improve the installation efficiency of field welding through a decrease in thickness. Line pipes of API grade X80 have already been put into practical use, and line pipes of API grade X100 are now in demand. Generally, the fracture toughness required for pipeline safety of pipeline increases with increases in strength[1]. However, as the strength of steel increases, it becomes remarkably difficult to attain good low-temperature toughness. Furthermore, the addition of alloys to improve strength deteriorates field weldability and usually requires preheating or post-weld heat treatment during field welding. Many difficult metallurgical problems have to be overcome in the development of ultra-high-strength of X100-class line pipe with good low-temperature toughness and field weldability.

This paper describes the metallurgical design of the API grade X100 line pipe with good low-temperature toughness and field weldability, and the effects of the chemical composition and processing conditions on the mechanical properties of the API grade X100 line pipe steels. Finally, the properties of trial-manufactured UOE line pipe are presented.

2. Metallurgical Design of the X100 Line Pipe

2.1 Characteristics of 780 N/mm² class steel

Fig. 1 shows the relationship between the transformation temperature and tensile strength of continuously cooled steels[2].

The microstructure of the X100 line pipe is classified into the region of upper bainite and is inferior in low-temperature tough-

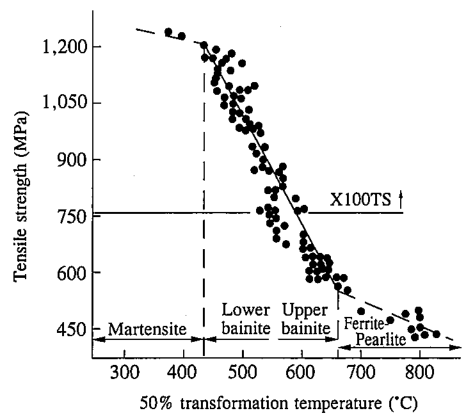


Fig. 1 Relationship between 50% transformation temperature and tensile strength of continuously cooled steels[2]

*1 Technical Development Bureau
*2 Kimitsu Works
*3 Head Office

ness as a result of the large effective grain size and the formation of high-carbon island martensite (M*). In order to improve the low-temperature toughness of the upper bainite, it is essential to refine prior austenite (γ) to as small a grain size as possible. Low-temperature toughness is improved when the microstructure can be changed from upper bainite to lower bainite or martensite through the addition of alloying elements, but a large addition of alloy causes remarkable deterioration in field weldability. In order to improve the field weldability, decrease in C and carbon equivalent is indispensable. Therefore, the key to develop X100 line pipe is the improvement of the low-temperature toughness of the upper bainitic microstructure without sacrificing the field weldability.

2.2 Strengthening mechanism and concept of improving low-temperature toughness of X100 line pipe

There are three basic strengthening mechanisms for steel: (1) precipitation hardening, (2) solid solution hardening, and (3) transformation strengthening. An ultra-high-strength steel such as X100 line pipe is accomplished with a combination of these mechanisms. Transformation strengthening is the most important among them, and it is achieved by the micro alloying and application of thermomechanical control process (TMCP) technology.

The improvement of toughness in the heat-affected zone (HAZ) and in the base material is an important consideration in developing the X100 line pipe. In order to improve the low-temperature toughness of the base material, it is necessary not only to refine the austenitic structure by thermomechanical rolling, but also to inhibit the austenite grain coarsening during slab-reheating by titanium nitride (TiN) particles. It is also important to make the steel clean and keep the center segregation of the continuously cast slabs to a low level.

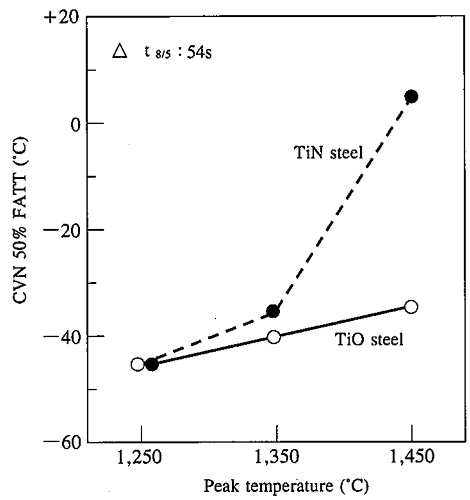
The grain refinement in the HAZ is effective in improving the HAZ toughness as well, because it suppresses the formation of M* harmful to toughness. Finely dispersed TiN particles inhibit the austenite grain coarsening in the HAZ. However, when the HAZ is reheated to a temperature above 1,400°C, the TiN particles coarsen or dissolve and lose their effect in inhibiting the austenite grain coarsening. Therefore, a technology was developed to improve HAZ toughness with the aid of titanium-oxide (TiO) particles finely dispersed in steel[3,4]. In this new type of steel (TiO steel), intragranular ferrite (IGF) grows radially from TiO particles as nuclei existing inside γ grains when the HAZ transforms from γ to α, refining the coarse γ grains to a great extent.

Fig. 2 shows the change in Charpy V-notch impact transition temperature with varying peak temperatures in the simulated HAZ. In the conventional TiN steel, the transition temperature increases remarkably when the peak temperature exceeds 1400°C, as TiN particles begin to dissolve at temperatures above that level. On the other hand, the transition tenperature of the TiO steel changes little with peak temperature, indicating that the TiO particles are chemically stable even at temperatures as high as 1,450°C, and the IGF develops irrespective of peak temperature. In the TiO steel also, an excess of the titanium not contained in the TiO combines with nitrogen to form TiN. These TiN particles inhibit the coarsening of austenite grains at and below a peak temperature of 1,350°C.

2.3 Thermomechanical control process (TMCP) technology

TMCP technology has enabled the production of excellent steel plate with high-strength and good low-temperature toughness. TMCP is roughly divided into two processes, as shown in **Fig. 3**. Type I, the so-called direct quenching and tempering process, involves the water cooling of plates to close to room temperature at a high cooling rate. Tempering treatment is required to obtain the appropriate ductility and toughness. Type II is an interrupted acclerated cooling process in which the plate is subjected to water cooling at a moderate cooling rate only in the transformation temperature region, and is then subjected to air cooling. As a result of the self-tempering effect of this process, excellent ductility and toughness are obtained without tempering. Since strength and low-temperature toughness depend on various process conditions, the processing conditions in TMCP have to be optimized in order to obtain high-strength and low-temperature toughness of the plate .

From the viewpoint of grain refinement, the slab-reheating temperature and rolling in the unrecrystallized region of the



Chemical composition (mass %)

Steel	C	Si	Mn	P	S	Nb	Ti	Al	N
TiO	0.07	0.25	1.81	0.008	0.001	0.039	0.017	0.003	0.002
TiN	0.07	0.25	1.81	0.008	0.001	0.039	0.017	0.019	0.003

Fig. 2 Change in Charpy V-notch impact transition temperature of simulated HAZ with peak temperature

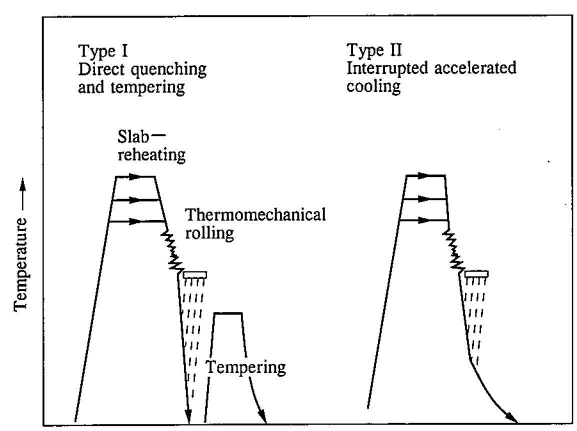
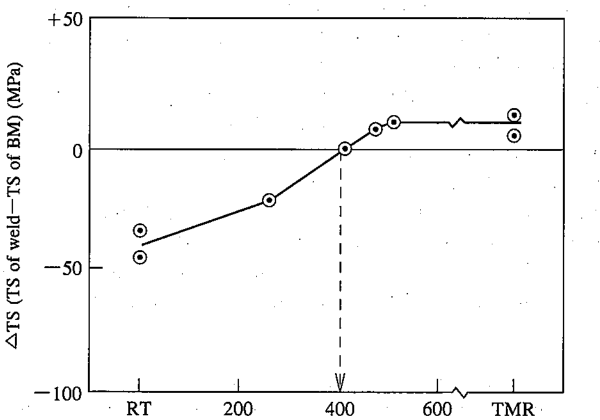


Fig. 3 Schematic illustration of accelerated cooling after controlled rolling

austenite are especially important. The lowering of the slab-reheating temperature decreases the γ grain size during reheating and contributes significantly to the refinement of the final microstructure.

The low-temperature toughness of the bainitic microstructure has a good correlation with the interval between apparent elongated prior austenite grain-boundaries including deformation bands and ferrite grain-boundaries, and the low-temperature toughness is improved as the interval is reduced. An increase in the cumulative reduction in the unrecrystallized region of austenite and the lowering of rolling temperatures can decrease apparent elongated prior autenite grain boundaries.



Note)
1. Nb-V steel (Plate thickness 22mm)
2. Tensile test specimen : 10mm ϕ round bar specimen (Gage length 50mm)

Fig. 4 Effect of stop temperature on difference in tensile strength between the base material and weld

In accelerated cooling, the strength is strongly dependent on the start and stop temperatures of the water-cooling . In interrupted accelerated cooling, low-temperature toughness and HAZ softening are affected by the stop temperature of the water-cooling. The strength increases with decreases in the stop temperature, and an extreme decrease in the stop temperature causes the formation of a large amount of M*, a deterioration in low-temperature toughness, an increase in HAZ softening, and a decrease in the tensile strength at welded joints.

Fig. 4 shows the effect of the stop temperature on the difference in tensile strength between the base material and the weld. The weld tensile strength is equal to that of the base material at a stop temperature of about 400°C, but drops below that of the base material at stop temperatures below 400°C. Therefore, to keep the strength of the welded portion higher than that of the base material, the temperature should be kept above 400°C . However, since a round bar specimen of the welded portion provides a conservative estimate compared to that provided by a full-thickness tensile specimen, a slightly lower temperature is actually allowable. The lower limit of stop temperature to prevent excess of HAZ softening depends on the chemical composition of the plate, the plate thickness, the cooling rate of the accelerated cooling, etc..

2.4 Alloy design of X100 line pipe

Experimental results for Nb-Mo system and Nb-B system, which together make up the basic chemical composition of the X100 line pipe, are presented here. **Table 1** shows the chemical composition of the plates. **Table 2** shows the TMCP conditions of the plates. **Fig. 5** shows the effect of molybdenum content on the mechanical properties of the Nb-Mo TMCP plate (Steel A1 to A4). The stop temperature of water cooling was constant at 450°C. The tensile strength increased with increases in the

Table 1 Chemical composition of steel plates (mass %, *ppm)

Steel	C	Si	Mn	P*	S*	Mo	Nb	Ti	B*	Others	C_{eq} 1)	P_{CM} 2)
A1	0.056	0.27	1.95	60	30	-	0.044	0.012	-	Ni,Cu	0.42	0.18
A2	〃	〃	〃	〃	〃	0.11	〃	〃	-	〃	0.44	0.19
A3	〃	〃	〃	〃	〃	0.22	〃	〃	-	〃	0.47	0.20
A4	〃	〃	〃	〃	〃	0.31	〃	〃	-	〃	0.48	0.20
B	0.061	0.29	1.92	80	30	0.10	0.042	0.015	-	〃	0.42	0.18
C1	0.067	0.26	2.00	40	10	0.10	0.036	0.019	-	〃	0.43	0.18
C2	〃	〃	〃	〃	〃	〃	〃	〃	16	〃	0.43	0.19
D	0.043	0.26	1.98	130	39	0.11	0.038	0.019	11	〃	0.40	0.16
E1	0.060	0.24	1.60	40	8	0.20	0.045	0.012	7	〃	0.41	0.19
E2	〃	〃	〃	〃	〃	0.29	〃	〃	9	〃	0.43	0.19
E3	〃	〃	〃	〃	〃	0.38	〃	〃	8	〃	0.45	0.20
E4	0.60	0.06	1.90	50	7	0.20	0.032	0.017	7	〃	0.47	0.20
E5	〃	〃	〃	〃	〃	0.29	〃	〃	6	〃	0.49	0.21
E6	〃	〃	〃	〃	〃	0.39	〃	〃	7	〃	0.50	0.21

1) C_{eq} = C + Mn/6 + (Cu+Ni)/15 + (Cr + Mo + V)/5
2) P_{CM} = C + Si/30 + (Mn +Cu + Cr)/20 +Ni/60 + Mo/15 + V/10 +5B

Table 2 TMCP conditions

	Steel A	Steel B	Steel C	Steel D	Steel E
1.Slab-reheating temperature (°C)	1,150	1,150	1,150	950-1,250	1,150
2.Cumulative reduction below 930°C (%)	75	75	75	75	63
3.Finish rolling temperature (°C)	740	740	730	720	880
4.Cooling rate of ACC (°C/S)	20	20	25	20	25
5.Stop temperature of ACC (°C)	450	RT-440	410	450	RT
6.Tempering condition	-	-	-	-	550°C×20min
7.Plate thickness (mm)	20	20	18	20	18

molybdenum content, and a tensile strength equivalent to that of API grade X100 was obtained when 0.2% or more molybdenum was added. Although the low-temperature toughness shows a tendency to decrease with increases in the molybdenum content, it remains on a good level even at a molybdenum content of 0.3% . The microstructure changes from ferrite-bainite to a single phase of the bainite consisting of bainitic ferrite and finely dispersed M*. The yield strength of bainite is relatively low because the yielding phenomenon disappears in the stress-strain curve in tensile test. The yield strength increases following the manufacture of UOE pipe.

Fig. 6 shows the effect of the stop temperature of accelerated cooling on the mechanical properties of Nb-Mo plate (Steel B)[5]. While the tensile strength increases with a decrease in this temperature, the yield strength and low-temperature toughness exhibit complicated behavior corresponding to changes in the microstructure and the stress-strain curve. The low-temperature toughness reaches its maximum at a stop temperature of 400°C, and is sufficient regardless of stop temperature. When the stop temperature is at 300°C or below, the tensile strength meets the requirements for X100 grade. However, it should be noted that the lowering of the stop temperature causes major HAZ softening when the seam weld is subjected to relatively large heat input.

The strengthening and toughening of the Nb-B steel are accomplished through the combined addition of niobium and boron. Boron is known to be an element that enhances hardenability during quenching. In order to make full use of its effects, it is essential to reduce the carbon content and optimize the TMCP conditions. **Table 3** shows a comparison of the mechanical properties of boron-free steel and boron-alloyed steel. The addition of boron increases the tensile strength by more than 100 MPa and the application of the accelerated cooling further increases it to as high as 50MPa. This increase in strength is achieved with little loss in low-temperature toughness.

Fig. 7 shows the effect of the slab-reheating temperature on the mechanical properties of Nb-B TMCP plates. As the slab-reheating temperature is lowered, the tensile strength decreases, but the BDWTT property is significantly improved. When the slab-reheating temperature is 1,150°C, high-strength and low-temperature toughness are obtained for X100 grade. This behavior can be explained by the microstructural change, the reduced amount of niobium in the solution during reheating, and so on.

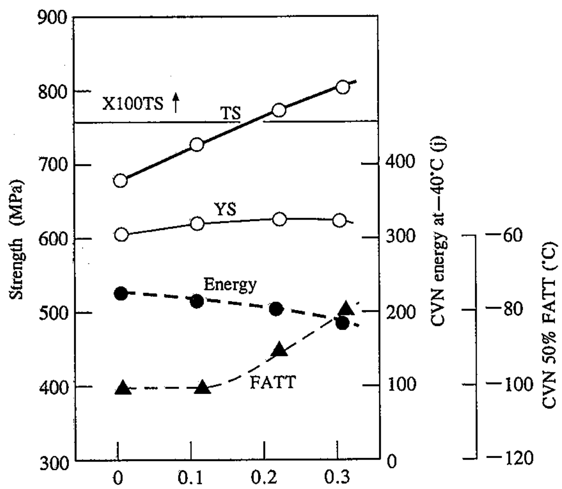


Fig. 5 Effect of molybdenum content on mechanical properties of Nb-Mo TMCP steel plate (Steel A1 to A4)

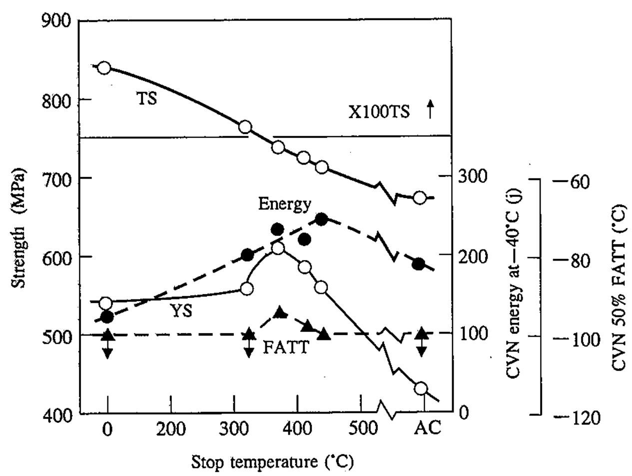


Fig. 6 Effect of the stop temperature of accelerated cooling on mechanical properties of Nb-Mo plates (Steel B)[5]

Table 3 Mechanical properties of Nb-B plates

Steel	Process	Tensile properties			CVN impact properties	
		YS (MPa)	TS (MPa)	El (%)	Energy at -40°C (J)	50% FATT (°C)
C1	TMR[1]	437	619	47	247	-110
	ACC[2]	550	670	44	368	-110
C2	TMR	537	746	36	249	-100
	ACC	633	805	34	254	-100

[1] TMR : Thermomechanical rolling
[2] ACC : Accelerated cooling after rolling

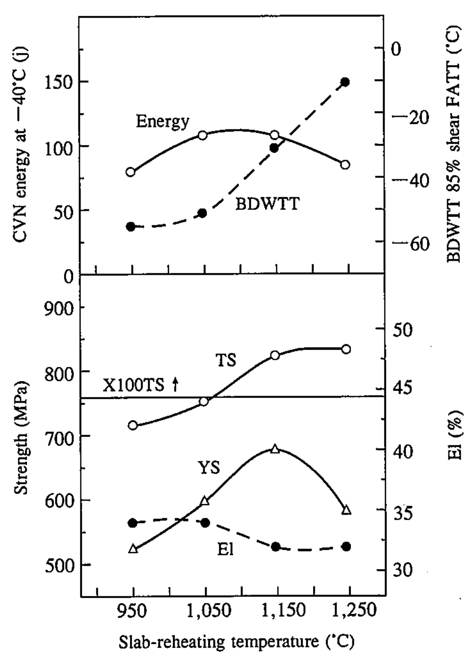


Fig. 7 Effect of slab-reheating temperature on mechanical properties of Nb-B plates (Steel D)

Since Nb-B steel can obtain high-strength with a comparatively small alloy element content, field welding without preheating is believed to be possible even for X100 line pipe .

Fig. 8 shows the effect of the silicon and molybdenum content on the simulated HAZ toughness and the M* formation of Nb-B steel. When the silicon content is high, the simulated HAZ toughness is improved with an increase in the molybdenum content. This is believed to be caused by a change of microstructure from the upper bainite to the lower bainite with an increase in the molybdenum content. On the other hand, when the silicon content is low, good HAZ toughness is obtained irrespective of the molybdenum content. In particular, when the molybdenum content is 0.2% to 0.3%, the improvements are remarkable. This is reason why a decrease in the silicon content reduces formation of M* in the upper bainite. In order to improve the HAZ toughness, it is desirable to have as small a silicon content as possible.

Fig 9 shows the relationship between the P_{CM} value for base material and the caluculated critical preheating temperature to prevent cold cracking in the TEKKEN test. This calculation is made based on the use of a high-cellulosic welding rod. In Nb-B steel with a low P_{CM} value, the critical preheating temperature is as low as 40°C and field welding without preheating can be realized. The critical preheating temperature of the Nb-Mo steel is also as low as approximately 100°C.

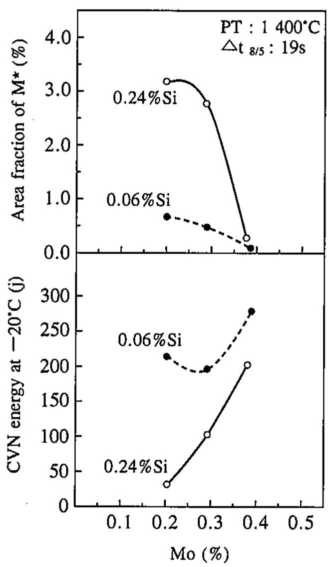


Fig. 8 Effect of silicon and molybdenum content on simulated HAZ toughness and formation of M*

3. Properties of X100 UOE Line Pipe

The manufacturing results for trial-manufactured X100 UOE line pipe are described here. **Table 4** shows the chemical composition of the manufactured trial-manufactured X100 line pipe. From the viewpoint of field weldability, Ceq and P_{CM} were set at 0.45% and 0.19%, respectively. In the pretreatment of hot metal, the hot metal was subjected first to desiliconization, and then to dephosphorization and desulphurization. After reheating, the continuously cast slab was thermomechanically rolled and then subjected to interrupted accelerated cooling in the heavy plate mill. UOE pipe was manufactured using this plate.

Table 5 shows the UOE pipe size and the mechanical properties of the UOE pipe. Although strengths of both the base material and welded joint meet the API specification for X100, the strength of the welded joint is lower since the stop temperature was 300°C or lower and the fracture location was in HAZ. To prevent HAZ softening, an increase in the stop temperature of accelerated cooling is necessary. Based on the results shown in

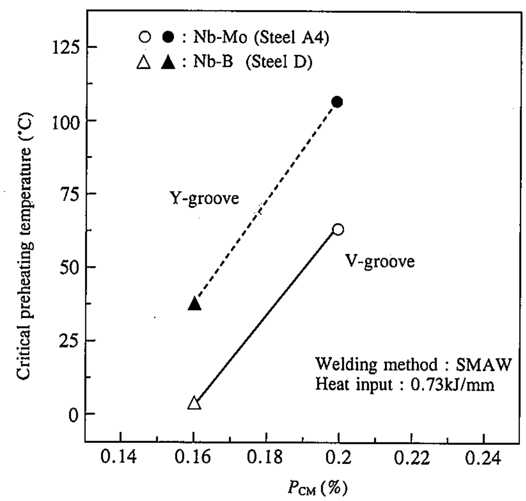


Fig. 9 Relationship between P_{CM} value of base material and calculated critical preheating temperature in y-groove (TEKKEN) cracking test

Table 4 Chemical composition of trial manufactured X100 line pipe

C	Si	Mn	P*	S*	Ni	Cu	Mo	Nb	Ti	C_{eq}	P_{CM}
0.060	0.22	1.96	60	26	0.39	0.17	0.11	0.045	0.013	0.45	0.19

Table 5 Mechanical properties of X100 line pipe

Pipe size		Base material					Seam welded joint				
OD (mm)	WT (mm)	Tensile properties*1)			CVN impact properties*2)	BDWTT properties	Tensile properties*3)		CVN impact properties*4)		
		YS (MPa)	TS (MPa)	El (%)	Energy at -20°C (j)	85% shear FATT (°C)	TS (MPa)	Location of fracture	vE_{-10}(j)		
									WM	HAZ1	HAZ2
762	19.05	710	848	30	133	-15	793	HAZ	144	71	130

*1) API 5L specimen taken from transverse direction
*2) Full-size specimen taken from transverse direction
*3) Transverse weld specimen with reinforcement
*4) Average value, notch position
WM : Weld metal centerline
HAZ1 : Fusion line
HAZ2 : Fusion line + 2mm

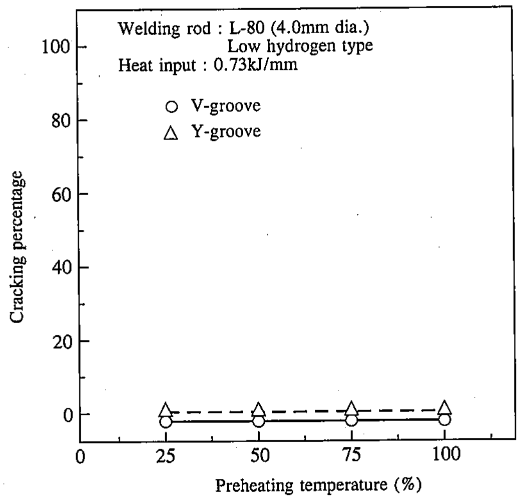


Fig. 10 TEKKEN test results

Fig. 5, in this case, a molybdenum addition of approximately 0.2% is required to ensure the base material strength. The toughness of the base material also shows good values. Further improvement of the low-temperature toughness can be expected with an increase in the stop temperature.

Fig. 10 shows the TEKKEN test results when a low-hydrogen welding rod was used. In addition to the y type groove, the V type groove was used to take into account route pass during field welding. No cracking was occurred with preheating at 25°C since Ceq and P_{CM} were suppressed at lower levels.

4. Conclusions

In order to develop X100 line pipe with excellent low-temperature toughness and superior field weldability, the chemical composition of the base material and TMCP conditions of the Nb-Mo system and the Nb-B system, which together make up the basic chemical composition for the X100 line pipe, were investigated. It was found that optimization of the chemical composition and TMCP conditions ensures high-strength and good low-temperature toughness for X100 without sacrificing field weldability.

Based on our experimental results, an X100 UOE pipe was trial-manufactured at our mill. This pipe showed high-strength and low-temperature toughness, as well as excellent field weldability.

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